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Hybrid composite laminates for a new generation of aerospace structures

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*“Nothing takes place in the world whose meaning
is not that of some maximum or minimum.”*

Leonhard Euler

“I always liked numbers.”

Terence Tao

Abstract

The inherent brittleness of high-performance Carbon-Fiber-Reinforced Polymers (CFRPs) remains a barrier to their widespread use in safety-critical structures, as their catastrophic failure mode offers no warning and limits damage tolerance. This thesis addresses this challenge by investigating the design, manufacturing, and failure mechanisms of thin-ply hybrid composites engineered to fail in a safe, progressive, and pseudo-ductile manner.

The research investigated different hybrid architectures, beginning with an experimental campaign on producing highly-dispersed, all-carbon intrayarn hybrids through an air-jet co-mingling process. This campaign highlighted the considerable challenges in achieving a uniform, filament-level dispersion, as the process resulted in a side-by-side yarn structure rather than a homogeneously intermingled yarn, revealing the potential of this material as a precursor for creating intralayer hybrid configurations. The primary investigation of this work is therefore focused on the more readily manufacturable interply hybrid architecture, for which a systematic design methodology based on analytical damage mode maps was established and analyzed.

The main contribution of this work is the *post-mortem* investigation of the three-dimensional internal damage state in failed all-carbon Double-Edge Notched Tension (DENT) specimens using high-resolution micro-Computed Tomography (μ CT). This volumetric analysis provided direct evidence for the toughening mechanisms that were previously inferred from surface observations. The results confirmed the presence of a complex internal network of stable fragmentation of the low-elongation plies, 0° split cracks, and extensive fiber pull-out, which together govern the material's improved toughness and pseudo-ductile response.

The research demonstrates that strategically designed interply hybrids offer a robust and practical solution to overcome the brittleness of advanced composites, establishing a validated pathway for the design of a new generation of safer aerospace lightweight structures.

Keywords: Thin-Ply Hybrid Composites, Pseudo-ductility, Micro-Computed Tomography (μ CT), Damage Mechanisms, Interply Hybrids, Air-jet Co-mingling.

Resumo

A fragilidade intrínseca dos Polímeros Reforçados com Fibra de Carbono (CFRPs) de alto desempenho continua a ser uma barreira à sua utilização generalizada em estruturas em que a segurança seja um fator crítico. Por apresentar um modo de falha do tipo catastrófico, não oferece aviso prévio e exibe, por essa razão, uma limitada tolerância ao dano. Esta tese explora esta limitação cobrindo aspectos como o design, o fabrico e os mecanismos de falha de compósitos híbridos de camadas finas, concebidos para falhar de um modo seguro, progressivo e pseudo-dúctil.

A investigação cobre diferentes arquiteturas híbridas, começando com uma campanha experimental focada na produção de híbridos com uma configuração fibra-a-fibra totalmente em carbono – e por essa razão, com elevada dispersão –, através de um processo de mistura por jato de ar. Esta campanha evidenciou desafios consideráveis na obtenção de uma dispersão uniforme ao nível do filamento, tendo os resultados mostrado uma estrutura de fios lado a lado, em vez de um fio único com fibras homogeneamente entrelaçadas. Porém, este resultado revelou o potencial de se usar o material como base para a criação de configurações híbridas intracamada. Assim, a investigação principal deste trabalho foca-se na arquitetura de híbridos camada-a-camada, de fabrico mais acessível, para a qual foi estabelecida e analisada uma metodologia de design baseada em mapas analíticos de modos de dano.

A principal contribuição deste estudo é a investigação *post-mortem* do estado de dano interno tridimensional em provetes DENT (Double-Edge Notched Tension) de compósitos totalmente em carbono, utilizando microtomografia computadorizada (μ CT) de alta resolução. Esta análise volumétrica forneceu evidências diretas sobre os mecanismos que afetam a ductilidade e a falha e que eram anteriormente inferidos a partir de observações na superfície dos componentes. Os resultados confirmaram a presença de uma rede interna complexa de fragmentação estável das camadas de baixa elongação, fissuras longitudinais a 0° e uma superfície de fratura muito irregular, mecanismos que, em conjunto, justificam a resposta pseudodúctil do material (maior tenacidade).

A investigação demonstra que os híbridos interlaminares quando estrategicamente combinados proporcionam uma solução robusta e prática para superar a fragilidade dos compósitos avançados, sendo por isso uma hipótese a considerar em futuras estruturas aeroespaciais.

Palavras-Chave: Compósitos Híbridos de Camada Fina, Pseudoductilidade, Microtomografia Computorizada (μ CT), Mecanismos de Dano, Híbridos Interlaminares, Co-mistura por Jato de Ar.

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List of Abbreviations

2D	Two-Dimensional
3D	Three-Dimensional
3DP	Binder Jetting
4D	Four-Dimensional (3D + Time)
AD	Areal Disorder
AE	Acoustic Emission
AFP	Automated Fiber Placement
APS	Advanced Photon Source
AR	Areal Ratio
ATL	Automated Tape Laying
ATP	Automated Tape Placement
CAI	Compression-After-Impact
CF	Carbon Fiber
CFD	Computational Fluid Dynamics
CFRP	Carbon-Fiber-Reinforced Polymer
CL	Computed Laminography
CLPT	Classical Laminated Plate Theory
CLT	Classical Laminate Theory
CNTs	Carbon Nanotubes
COD	Crack Opening Displacement
CP	Cross-Ply
CSD	Crack Shear Displacement
CT	Computed Tomography
CTE	Coefficient of Thermal Expansion
DENT	Double-Edge Notched Tension
DIC	Digital Image Correlation
DL	Deep Learning
DVC	Digital Volume Correlation
ECM	Experimental Compliance Method
ENF	End Notched Flexure
EPFL	Ecole Polytechnique Fédérale de Lausanne

ESRF	European Synchrotron Radiation Facility
ESRF-EBS	European Synchrotron Radiation Facility - Extremely Brilliant Source
EuXFEL	European X-ray Free Electron Laser
FAW	Fiber Areal Weight
FBP	Filtered Back-Projection
FE	Finite Element
FPX	Flat Panel Extension
FRP	Fiber-Reinforced Polymer
GFRP	Glass-Fiber-Reinforced Polymer
GLS	Global Load-Sharing
GMT	Glass Mat Thermoplastics
HE	High-Elongation
HiPerDiF	High Performance-Discontinuous Fiber
HiPerDuCT	High Performance Ductile Composite Technologies
HM	High Modulus
HT	High Tensile
IFS	Inter-Fiber Spacing
ILSS	Interlaminar Shear Strength
IM	Intermediate Modulus
INEGI	Institute of Science and Innovation in Mechanical and Industrial Engineering
LCLS	Linac Coherent Light Source
LCM	Liquid Composite Molding
LE	Low-Elongation
LEFM	Linear Elastic Fracture Mechanics
LLS	Local Load-Sharing
LPAC	Laboratory for Processing of Advanced Composites
LUT	Lookup Table
MCCFC	Mitsubishi Chemical Carbon Fiber and Composites
NASA	National Aeronautics and Space Administration
NCF	Non-Crimp Fabric
NTPT	North Thin Ply Technology
OHC	Open-Hole Compression
OHT	Open-Hole Tension
OM	Optical Microscopy
OoA	Out-of-Autoclave
PA	Polyamide
PAN	Polyacrylonitrile
PE	Polyethylene
PEEK	Polyetheretherketone

PETRA	Positron-Elektron-Tandem-Ring-Anlage
PP	Polypropylene
PPS	Polyphenylene Sulfide
PPDM	Probabilistic Progressive Damage Model
QI	Quasi-Isotropic
RaaD	Resolution at a Distance
RFI	Resin Film Infusion
ROM	Rule-of-Mixtures
RTM	Resin Transfer Molding
RVE	Representative Volume Element
SCCR	Short Circuit Current Rating
SCF	Stress Concentration Factor
SEM	Scanning Electron Microscopy
SLS	Selective Laser Sintering
SMC	Sheet Molding Compound
SPRing-8	Super Photon Ring - 8 GeV
SQRTM	Same-Qualified-Resin-Transfer-Molding
SRCL	Synchrotron Radiation Computed Laminography
SRCT	Synchrotron Radiation Computed Tomography
TEM	Transmission Electron Microscopy
UD	Unidirectional
UHM	Ultra-High Modulus
VARTM	Vacuum-Assisted Resin Transfer Molding
VIP	Vacuum Infusion Processing
XFEL	X-ray Free Electron Laser
μCT	Micro-Computed Tomography

List of Symbols

Latin Symbols

a, a_0	Crack length, Initial crack/notch length
$[A], [B], [D]$	Laminate stiffness matrices (extensional, coupling, bending)
AD	Areal Disorder
AR	Areal Ratio (for hybridization analysis)
A_{fiber_i}	Cross-sectional area of fiber i
$\bar{A}$	Mean area of Delaunay tessellation elements
$A_{Voronoi_i}$	Area of the Voronoi cell for fiber i
C_x, C_y	Coordinates of the overall yarn's center of gravity
C_{xA}, C_{yA}	Coordinates of the center of gravity for fiber component A
c_p	Specific heat capacity
CSD	Crack Shear Displacement
CV	Coefficient of Variation
d	Feature size (in phase-contrast imaging)
d_f	Filament diameter
d_{ij}	Euclidean distance between the centers of fibers i and j
D_F	Nominal diameter of a fiber
E	Energy of incident X-ray photons
E_1, E_2	Lamina moduli in the longitudinal and transverse directions
E_{1T}, E_{2T}	Measured longitudinal and transverse tensile moduli
E_b	Flexural modulus
$\hat{E}$	Equivalent modulus for orthotropic fracture mechanics
E_f, E_m	Young's modulus of the fiber and matrix
E_{fT}	Fiber transverse Young's modulus
E_h	Elastic modulus of a hybrid composite
E_H, E_L	Elastic modulus of the HE and LE constituents
E_{th}	Theoretical hybrid modulus
FAW	Fiber Areal Weight
G_{12}	In-plane shear modulus
G_{app}	Apparent transverse fracture toughness

G_{fLT}	Fiber in-plane shear modulus
G_I	Mode I energy release rate
G_{Ic}, G_{IIc}	Mode I and Mode II critical energy release rates
H	Degree of Hybridization
I, I_0	Transmitted and initial X-ray intensity
IFS	Inter-Fiber Spacing
$ILSS$	Interlaminar Shear Strength
k	Thermal conductivity
K	Constant in the attenuation coefficient formula
K_{IC}	Mode I fracture toughness (material property)
K_I	Mode I stress intensity factor
K_t	Stress concentration factor
L, W	Specimen gauge length and width
l, w	Specimen length and width (in DENT analysis)
l_c	Critical stress recovery length
l_{fpz}	Fracture process zone length
L_t	Lateral distribution index (for co-mingled yarn)
m	Weibull modulus
M, M_m, M_f	General composite, matrix, and fiber moduli (Halpin-Tsai)
$\{M\}, \{N\}$	Moment and force resultant vectors
$\{M^{thermal}\}, \{N^{thermal}\}$	Thermal moment and force resultant vectors
N_{FA}, N_{FB}	Number of fibers of type A and B
n	Total number of plies
P, P_{MAX}	Applied load and maximum load
P_u	Ultimate/peak load (in DENT analysis)
$[Q], [\bar{Q}]$	Reduced and transformed reduced stiffness matrices
$\mathcal{R}, \mathcal{R}_0$	R-curve (crack resistance) of laminate and 0° plies
$\mathcal{R}_{90}$	R-curve (crack resistance) of 90° plies
R_a	Average displacement of all filaments from the yarn's center
R_A, R_B	Average displacement of components A and B from the yarn's center
R_d	Radial distribution index (for co-mingled yarn)
R_i, R_j	Radii of fibers i and j
s_{ij}	Components of the compliance matrix
S_{12}	Measured in-plane shear strength
t	Ply or specimen thickness
t_0, t_{90}	Total thickness of 0° and 90° plies
t_H, t_L	Total or half-thickness of HE and LE layers
$t_{outlife}, t_{storage}$	Out life and storage life of prepreg
T_{cure}, T_g	Cure temperature and glass transition temperature
T_o	Final/room temperature

$[T], [T']$	Stress and strain transformation matrices
V, V_f	Volume and fiber volume fraction
$V_{f,local}$	Local fiber volume fraction
W_f, W_m	Fiber and matrix weight fraction
x	Path length through a material
X_T, Y_T	Measured longitudinal and transverse tensile strengths
X_T^L, X_C^L	Laminate plain tensile and compressive strengths
X_T^{OHT}, X_C^{OHC}	Laminate open-hole tensile and compressive strengths
z, z_k	Through-thickness coordinate, coordinate of ply k ; or Propagation distance (in phase-contrast imaging)
$z_{mid,k}$	Mid-plane z-coordinate of ply k
Z	Atomic number

Greek Symbols

α	Modulus ratio (E_L/E_H); or Relative crack length (a/w)
α_1, α_2	Lamina longitudinal and transverse CTEs
α_f, α_m	CTE of fiber and matrix
α_{fL}, α_{fT}	Fiber longitudinal and transverse CTEs
β	The LE to HE layer thickness ratio, t_L/t_H
γ	LE proportion in a hybrid, $t_L/(t_L + t_H)$
γ_{12}	Engineering shear strain
δ	Crack opening displacement (COD)
Δa	Crack increment
ΔT	Change in temperature
ε, ε	Strain
ε^0	Mid-plane strain vector
ε_f	Elongation at break / Failure strain
$\varepsilon_{pd}, \varepsilon_{py}$	Pseudo-ductile and pseudo-yield strain
$\varepsilon_H, \varepsilon_L$	Failure strain of the HE and LE constituents
ζ	Rescaled aspect ratio for DENT analysis; or Transition point in bili-near hybrid strength model
η	Parameter in Halpin-Tsai equations
θ_k	Orientation angle of ply k
κ	Curvature vector; or DENT correction factor
λ	X-ray wavelength; or Compliance ratio for DENT analysis
μ	Linear X-ray attenuation coefficient
ν_{12}	Major Poisson's ratio
ν_{21}	Minor Poisson's ratio
ν_f, ν_m	Poisson's ratio of fiber and matrix
ν_{fLT}	Fiber longitudinal-transverse Poisson's ratio

ξ	Geometric aspect ratio for DENT analysis (w/l); or Reinforcement factor in Halpin-Tsai equations
ρ	Material density; or Orthotropy parameter for DENT analysis
σ	Stress; or Standard deviation
σ_L	Average strength of LE fiber bundle
σ_b	Flexural strength
σ_f, σ_m	Strength of the fiber and matrix
σ_f	Final failure stress
σ_h	Longitudinal tensile strength of a hybrid composite
σ_H, σ_L	Strength of the HE and LE constituents
σ_l	Longitudinal tensile strength (general)
σ_{py}	Pseudo-yield stress
σ_y	Matrix tensile yield strength
$\sigma_{LF}, \sigma_{del}, \sigma_{HF}$	Critical stresses for LE fragmentation, delamination, and HE failure
σ_Y^h	Stress at damage onset in a hybrid composite
τ_{12}	In-plane shear stress
τ_y	Interlaminar shear strength or matrix shear yield strength
ϕ	Simplified DENT correction function ($\xi \rightarrow 0$)
Φ	DENT correction function/matrix

Chapter 1

Introduction

1.1 Motivation

The increasing concentration of greenhouse gases in the atmosphere presents a global challenge that requires transformative changes across all sectors of industry and society. International commitments, such as the Paris Agreement, have established ambitious targets to limit global warming to considerably below 2°C. However, as illustrated in Figure 1.1, current climate policies are projected to result in a warming of around 2.7°C by 2100, exceeding this goal. Even if all current national pledges are met, warming is still expected to reach 2.1°C, which highlights a gap between existing commitments and the necessary reductions. To stabilize the climate, global emissions – which have risen rapidly over the last half-century and have not yet peaked – must be reduced to net-zero [172].

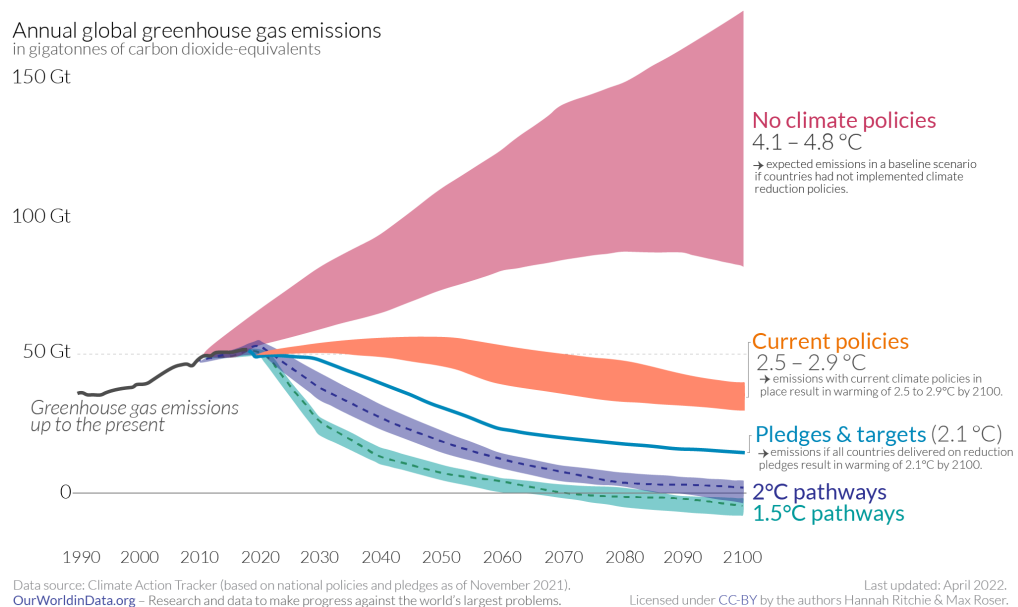


Figure 1.1: Global greenhouse gas emissions pathways and projected warming by 2100 under different policy scenarios (reproduced from [172]).

The transportation sector is a primary focus for decarbonization efforts, standing as the second-largest source of global greenhouse gas emissions after electricity and heat, as shown in Figure 1.2. Focusing on carbon dioxide (CO₂), the most prevalent greenhouse gas, transport is responsible for approximately one-quarter of direct CO₂ emissions from fuel combustion [171]. Within this sector, aviation is a significant and particularly challenging area to decarbonize. Although it accounts for 11.6% of transport CO₂, its global share is approximately 2.5% of total CO₂ emissions. However, its total contribution to climate change extends beyond carbon dioxide. Aircraft emissions at high altitudes, such as nitrogen oxides and water vapor, trigger additional atmospheric effects, with the formation of persistent condensation trails (contrails) having a significant warming impact. To quantify this total impact, climate scientists use the metric of Effective Radiative Forcing (ERF), which measures the net change in the Earth's energy balance. When these non-CO₂ effects are included, aviation's contribution is estimated to be responsible for approximately 3.5% of the total warming from human activities [170].

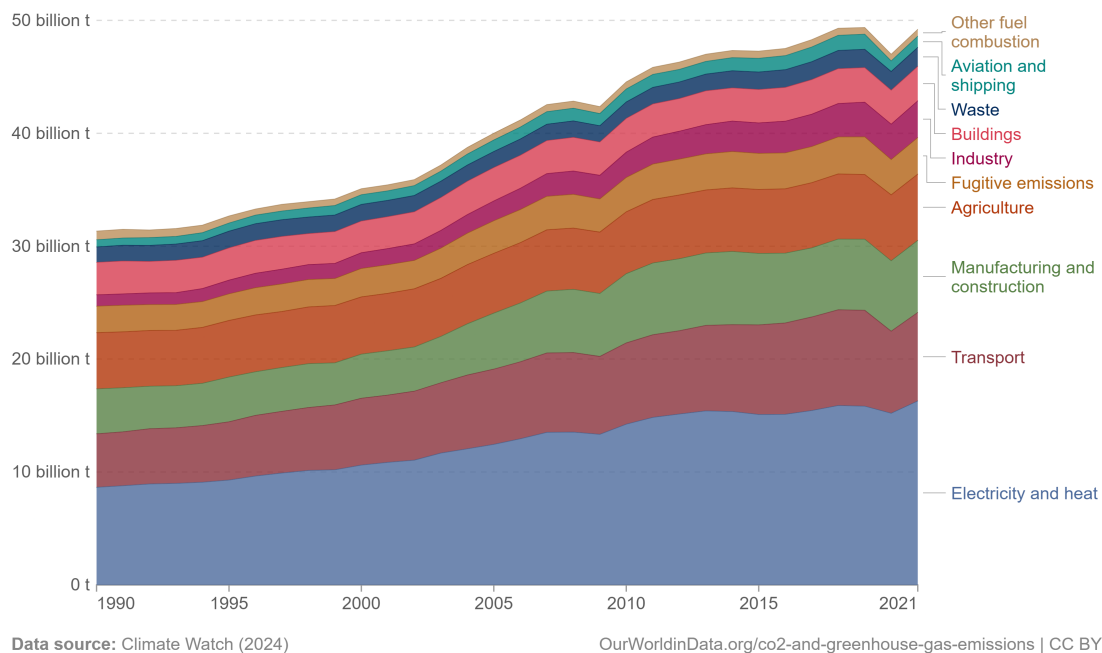


Figure 1.2: Greenhouse gas emissions by sector, World, 1990 to 2021. The chart shows the contribution from transport alongside other sectors. Emissions are measured in tonnes of carbon dioxide-equivalents and do not include land-use change (reproduced from [162]).

Given this climate impact and the industry's reliance on high-energy-density liquid fuels, improving efficiency is a primary mitigation strategy. As the fundamental physics of flight inseparably link an aircraft's mass to its fuel consumption and emissions, reducing structural weight is one of the most direct strategies for improving operational efficiency and reducing the sector's environmental footprint. Consequently, the aerospace industry has widely adopted advanced lightweight materials like Carbon-Fiber-Reinforced Polymers (CFRPs). These materials offer an exceptional combination of high strength coupled with low density, as illustrated by their position in the Ashby chart in

Figure 1.3. This results in superior specific properties that enable the construction of lighter and more efficient airframes.

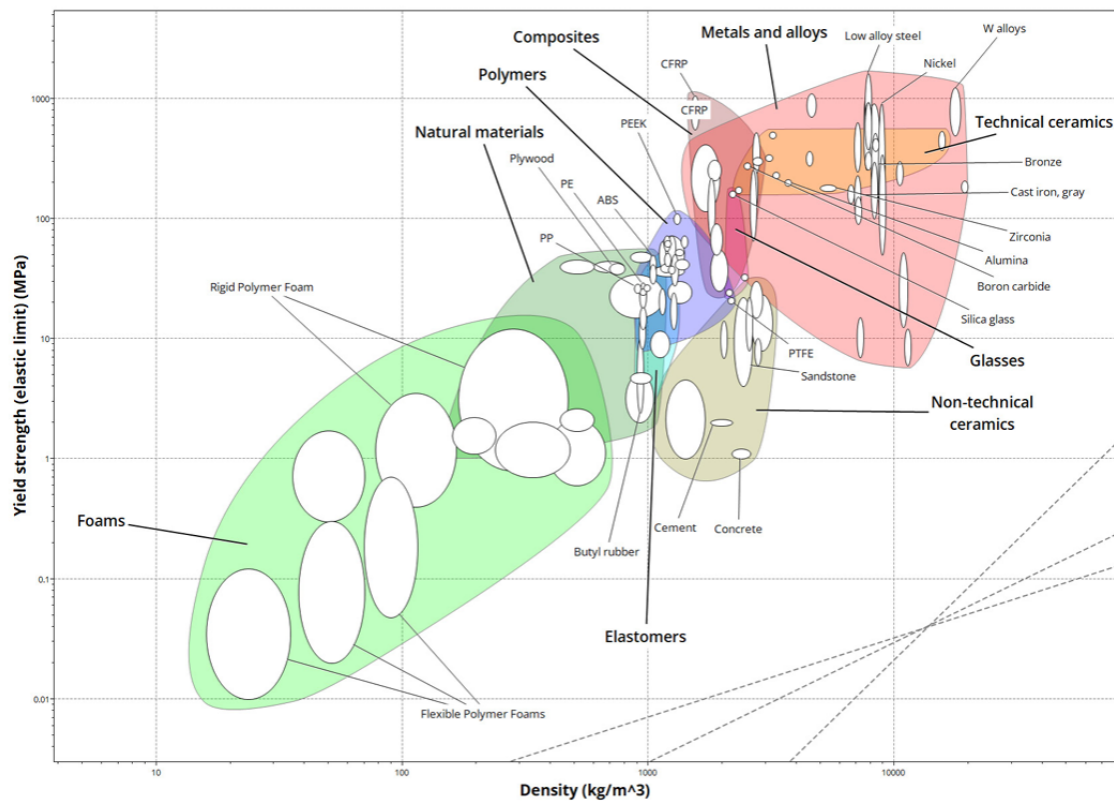


Figure 1.3: Ashby chart for yield strength versus density, illustrating the higher specific strength of composites compared to other material families (adapted from [12]).

Despite their widespread use, a fundamental limitation prevents their unrestricted application in primary aircraft structures: their inherent brittleness. Unlike metallic alloys that exhibit ductile yielding, high-performance composites tend to fail catastrophically with little to no warning, requiring conservative designs that can offset some of their weight-saving potential. This research is motivated by the need to engineer a new class of composite materials that retain the performance of CFRPs while incorporating a safe, progressive, and pseudo-ductile failure mechanism.

Research aimed at mitigating this brittleness has explored innovative material architectures, primarily thin-ply technology and fiber hybridization. While thin-ply composites improve strength by suppressing premature damage, this can lead to an even more abrupt final failure. Fiber hybridization, which combines different fiber types, offers a path to control the failure process. Different configurations are explored in the field, ranging from highly-dispersed intrayarn concepts to more structured interlayer designs, each with a theoretical potential to control damage progression. However, a significant gap often exists between these idealized architectures and what is practically manufacturable at a high and consistent quality, as well as a deep understanding of their internal behavior.

1.2 Objectives

Based on the identified research gaps, the objectives of this dissertation are:

1. To conduct a comprehensive review of the state of the art in composite materials, focusing on thin-ply technology, fiber hybridization strategies, the principles of pseudo-ductility, and advanced volumetric imaging methods.
2. To investigate the manufacturing feasibility of producing filament-level, all-carbon hybrid yarns using an air-jet co-mingling process and to evaluate the challenges associated with achieving a high degree of fiber dispersion.
3. To establish and apply a design methodology based on analytical damage mode maps for selecting interply hybrid configurations predicted to exhibit a pseudo-ductile failure response.
4. To perform a novel, three-dimensional investigation of the internal damage mechanisms in failed interply hybrid specimens using micro-Computed Tomography (μ CT), correlating the observed damage with the material's architecture and surface-based macroscopic fracture behavior.

1.3 Dissertation Structure

Following this introductory chapter, the dissertation is organized into four main chapters:

Chapter 2 provides a comprehensive review of the state of the art. It begins with the fundamentals of composite materials and Fiber-Reinforced Polymers, before examining the two primary technologies investigated in this work. The first section details the development, performance, and trade-offs of thin-ply composites. The second section covers the principles of fiber hybridization, including different architectural configurations, the origins of the hybrid effect, and a review of mechanical properties. These concepts are combined in a discussion on pseudo-ductility and the specific approach of using thin-ply hybrids. The chapter concludes with an overview of composite manufacturing technologies and a review of the principles and applications of X-ray Computed Tomography as an advanced method for volumetric damage analysis.

Chapter 3 documents the initial experimental campaign focused on producing an ideal, fiber-by-fiber dispersed hybrid material. This chapter details the manufacturing principles of air-jet co-mingling, a textile-based process for creating intrayarn hybrids. It describes the experimental setup and presents the results of a trial to co-mingle all-carbon fiber tows.

Chapter 4 presents the core experimental investigation of this thesis, focused on the more practical interply hybrid architecture. The first section establishes a systematic methodology for designing pseudo-ductile, all-carbon interply hybrids using analytical damage mode maps and other predictive models. This framework is then used to analyze a benchmark experimental study on equivalent

hybrid systems, reviewing their longitudinal, transverse, shear, and fracture properties. The main contribution of the chapter is the final section, which presents a novel, post-mortem volumetric analysis of failed Double-Edge Notched Tension specimens from the benchmark study using high-resolution micro-Computed Tomography (μ CT).

The dissertation concludes with **Chapter 5**, which synthesizes the findings from both the co-mingling investigation and the detailed analysis of interply hybrids. This final chapter presents the main conclusions of the research and provides some recommendations for future work, including potential research directions that build upon this dissertation.

Chapter 2

State of the art

2.1 Introduction to Composite Materials

For centuries, materials made from more than one constituent have been observed in nature, yet it was only in the mid-20th century that their deliberate design and manufacture became widespread with the rise of modern composites.

Their fundamental differences from the traditional material families of metals, ceramics, and polymers soon led to their classification as a distinct class of materials. One of their main advantages is the ability to adjust their properties, combining physical or chemical characteristics that conventional materials cannot replicate.

Composites offer a unique combination of properties, including high strength, stiffness, and toughness, coupled with low density for significant weight reduction. This results in excellent specific modulus and specific strength, as illustrated by their position in the Ashby chart in Figure 2.1. They also provide good resistance to corrosion and abrasion, a longer fatigue life, and, with proper design, greater impact toughness. These attributes make composites highly valuable in specialized applications that require demanding material performance, such as the aerospace, underwater structure, bioengineering, and transportation industries [36].

Other, less well-known, advantages include part count reduction, electromagnetic transparency, erosion and wear resistance, acoustic and vibration damping, thermal or acoustic insulation, low thermal expansion, tunable thermal conductivity, self-healing, adjustable permeability, fire resistance and retardancy, protection from lightning strikes, and magnetoelectric response [13].

Notable examples of some of these engineered composite materials include reinforced concrete, ceramic matrix composites which integrate ceramic and metallic phases and fiber-reinforced polymers. Among the latter, Carbon-Fiber-Reinforced Polymer (CFRP) and Glass-Fiber-Reinforced Polymer (GFRP) are the most widely used and studied examples, with CFRP being the main focus of this thesis and further detailed in the next sections.

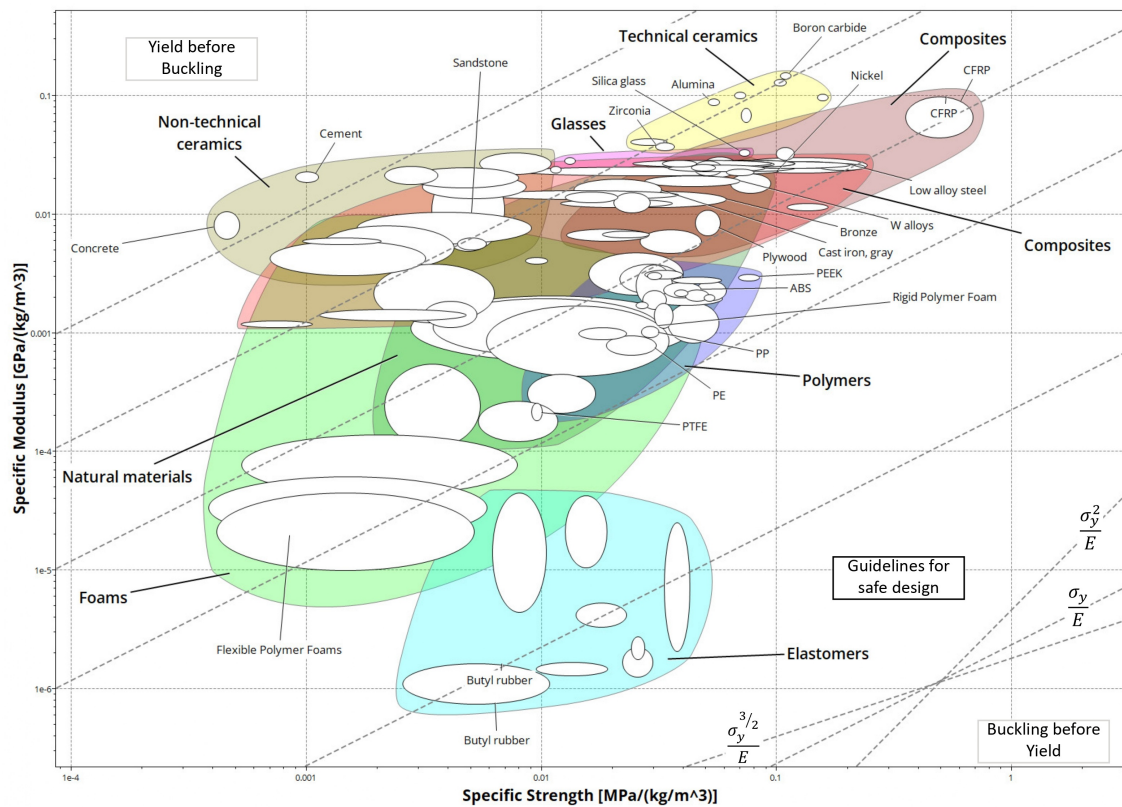


Figure 2.1: Ashby chart for specific modulus versus specific strength (adapted from [11]).

2.1.1 Fiber-reinforced polymers

Fiber-Reinforced Polymers (FRPs) consist of fibrous reinforcement embedded within a polymeric matrix where fibers such as carbon, glass, and aramid are frequently employed due to their exceptional strength, stiffness and fatigue resistance while polymer matrices bind these fibers ensuring correct alignment as well as enabling stress transfer and offering protection from mechanical abrasion and environmental factors such as moisture, ultraviolet radiation and chemical attacks.

The polymer matrices can be broadly classified into thermosetting and thermoplastic polymers. Thermosetting polymers, such as epoxy resins, unsaturated polyesters, and vinyl esters, are widely used due to their favorable mechanical properties, chemical resistance, and thermal stability [99]. These polymers require curing, a process that transforms the resin from a liquid into a solid through chemical cross-linking, creating a rigid three-dimensional structure. Epoxy resins specifically exhibit strong adhesion to reinforcing fibers, good mechanical strength, and chemical resistance, retaining a high percentage of their mechanical properties even at elevated temperatures (up to approximately 150°C), making them particularly suitable for structural applications. However, thermosets typically show limitations such as brittleness, moisture sensitivity, limited shelf life, and demanding manufacturing requirements, including restrictive curing cycles involving high temperatures and extended processing times [130, 151].

As for thermoplastic matrices such as polypropylene, polyamide (nylon), and polyethylene, they are known for their recyclability, ability to be reshaped, and high impact resistance. Unlike thermosetting polymers, thermoplastics do not undergo curing, as they do not chemically cross-link. As a result, they can be repeatedly melted and reshaped due to their reversible thermal behavior [99]. Despite these advantages, thermoplastics have several processing challenges, including high viscosity during melting, difficulty in fiber impregnation, uneven resin distribution, residual stresses, and changes in crystallinity related to varying cooling rates, all of which affect their mechanical properties and the quality of the final composite components. Consequently, the manufacturing knowledge for thermoplastics is less established than for their counterpart. For this reason, despite having generally better toughness and impact resistance, thermoplastics are not used as frequently in high-performance structural applications [130, 151].

However, the properties of fiber-reinforced composites do not depend exclusively on the material of the constituents but also on fiber arrangement, orientation, and concentration within the matrix. Continuous fibers are typically aligned in one direction, optimizing strength and stiffness along that direction, whereas discontinuous fibers may exhibit various orientations, including aligned, random, or partially oriented configurations. Figure 2.2 illustrates these different arrangements.

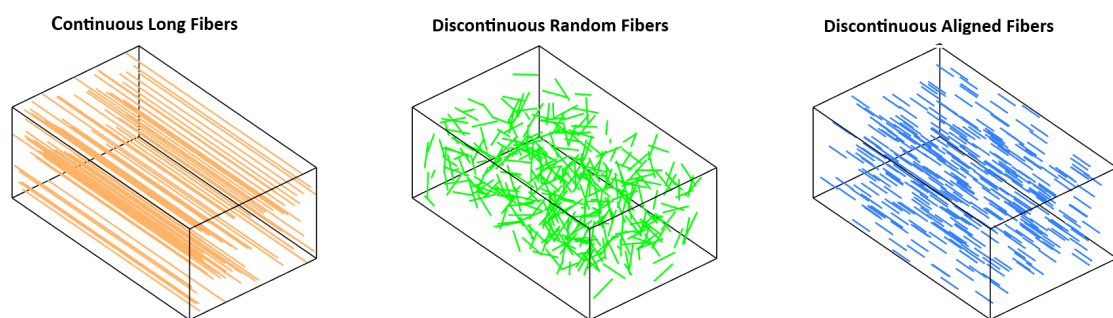


Figure 2.2: Schematic reinforcement arrangements in composites: (a) continuous long fibers, (b) discontinuous randomly oriented fibers, and (c) discontinuous aligned fibers.

For discontinuous fiber composites, properties strongly depend on the length, orientation, and amount of fibers. While short fibers allow for easier manufacturing of complex parts they generally result in lower strength and stiffness compared to composites reinforced using continuous fibers. However, if short fibers are aligned within the matrix, the composite can achieve mechanical properties approaching those of continuous fiber composites. The manufacturing methods for aligned discontinuous fiber composites generally follow two approaches: aligning loose short fibers or introducing controlled breaks into continuous fiber materials. Recent advancements in these techniques show their potential. For example, a 2014 study [233] on the HiPerDiF (High Performance-Discontinuous Fiber) method, which aligns loose fibers, concluded that it can produce composites with mechanical properties competitive with their continuous fiber counterparts. The study showed that these composites can achieve a tensile strength of 1509 MPa and a modulus of 115 GPa at a 55% fiber volume fraction. Despite such promising results, a subsequent review [198]

noted that while these systems have clear applications, further development is still required to better quantify and categorize their properties and processes.

When short fibers are randomly oriented, the mechanical properties become more uniform in all directions (quasi-isotropic), but the overall strength and stiffness typically becomes lower. Additionally, short fibers may form clusters or uneven distributions within the composite, leading to local weaknesses and affecting the damage development [68, 99].

A combination of continuous fibers and nanoscale discontinuous reinforcements, such as Carbon Nanotubes (CNTs), is also possible. This approach primarily improves matrix-related properties. For instance, adding CNTs into composite materials reinforced with continuous fibers can improve interlaminar shear strength and resistance to impact damage [20]. Also, attaching CNTs directly onto continuous fibers can improve the bond between fibers and matrix, leading to improvements in both the composite's resistance to delamination and fracture [28].

The structural design of a composite extends beyond the arrangement of individual fibers. Often, materials are structured as composite laminates, which consist of multiple distinct layers with varying fiber orientations to optimize performance. To properly analyze these complex structures, the standard approach is to consider them at different structural scales, as detailed by Daniel and Ishai [51] and illustrated in Figure 2.3:

- **Macroscopic:** At this scale, composites are studied in terms of their overall structural performance and behavior in applications.
- **Mesoscopic:** This scale involves studying the composite's internal structure, such as layers or woven fabric arrangements, which significantly influences the macroscopic properties.
- **Microscopic:** This scale focuses on the individual fiber-matrix interactions and the detailed mechanisms governing strength and durability.

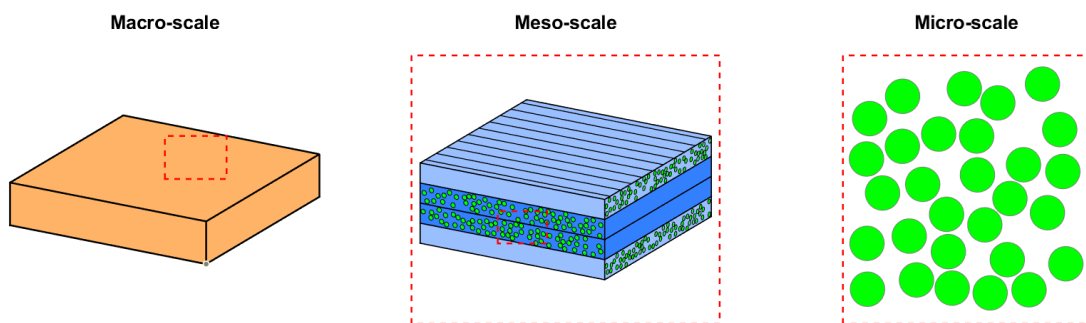


Figure 2.3: Schematic of the different structural scales.

In this multi-scale perspective, the meso-scale acts as the bridge between microscopic phenomena and macroscopic performance. Recent advancements have therefore focused heavily on engineering the ply itself, leading to the development of thin-ply composites.

2.2 Thin-Ply Composites

Within the study of composites at the meso and microscopic scale, thin-ply laminates have emerged as a prominent alternative to standard composites. These materials are distinguished by their substantially reduced ply thickness, broadly defined as below $100\ \mu\text{m}$ with corresponding Fiber Areal Weights (FAW) typically below $100\ \text{g}/\text{m}^2$. Common carbon fiber thin plies fall in the range of $20\text{--}60\ \mu\text{m}$ thickness, representing a significant reduction compared to the standard $125\ \mu\text{m}$ ply. Conversely, standard plies might also be considered *thick* when exceeding $200\ \mu\text{m}$ [71]. From the design perspective, the ability to add more layers per unit thickness allows greater freedom in selecting ply orientations, and thus finer property tuning and optimization [9, 71].

2.2.1 Historical Development and Manufacturing

While early work in the 1980s by researchers such as Bailey et al. [26] and Crossman and Wang [42] had already reported a positive effect of ply thickness on delaying transverse cracking, the concept of reducing ply thickness below the conventional was first demonstrated by Kawabe et al. in 1997 [115], who produced $20\text{--}50\ \mu\text{m}$ carbon/epoxy prepreg with a pneumatic tow-spreading process. Despite the innovation, high equipment cost, slow production rates and concerns over fiber damage limited any immediate commercial applicability.

In the literature, this air-jet tow-spreading process is also known as the Fukui method. As detailed by El-Dessouky [55], this pneumatic technology uses a combination of low-pressure air jets and a vacuum to make the tow momentarily lose tension. In this tension-free state, the filaments can be spread uniformly over a support plate without breakage, resulting in a thin, wide tape with a highly regular microstructure [55], as illustrated in Figure 2.4. While this pneumatic-vacuum process is common, other tow-spreading technologies have been developed, including mechanical methods (using rollers and bars), electrostatic charging, and vibration-acoustic systems [13, 32]. Using this method, Sasayama et al. [180] prepared laminae about $50\ \mu\text{m}$ thick, stacked them into quasi-isotropic laminates of arbitrary overall thickness and studied first-ply failure of this type of materials. Results of their tensile tests with acoustic-emission monitoring showed an improvement in the initial failure stress as lamina thickness decreases [180]. This approach has since been validated as an effective method for identifying specific damage modes, such as fiber fragmentation and delamination, in thin-ply composite systems [65].

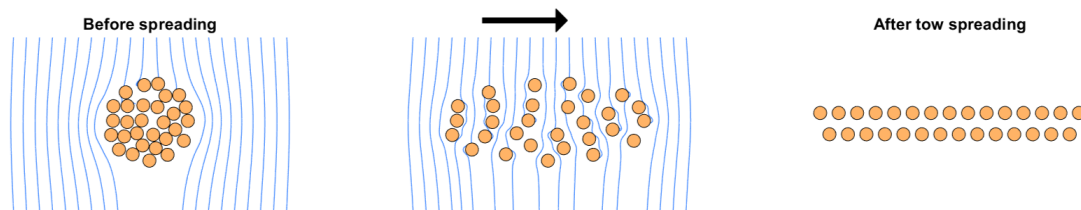


Figure 2.4: Schematic of the fiber tow spreading process.

Building on these findings, a comprehensive study was carried out by Sihm et al. in 2007 [193]. They evaluated 40 μm plies produced via tow-spreading, reporting a complex set of performance trade-offs. The detailed findings from this work are discussed in the relevant performance sections below. Notably, co-authors of this study, Tsai and Kawabe, had earlier filed a patent related to thin ply laminates, published as US 2006/0093802A1 [215]. This patent defined thin plies, provided details on hybrid lay-ups that mix thin and conventional plies, quantified the resulting gains in stiffness, and mentioned improvements in delamination resistance.

These advantages were then corroborated by Yokozeki et al. [232] in 2008, who repeated the experiments with a toughened-epoxy system. They reported a 9% increase in Open-Hole-Compression (OHC) and an 8% increase in Compression-After-Impact (CAI) strength for their thin-ply laminates compared to the standard-ply counterparts.

Advances in various tow-spreading technologies have led to the commercial availability of these materials from manufacturers such as North Thin Ply Technology (NTPT), Oxeon (TeXtreme), and Hexcel (PrimeTex), with the thinnest unidirectional prepregs now reaching down to 15 g/m^2 FAW [158].

2.2.2 In-Situ Effect, Microstructure, and Failure Mode Shift

A fundamental aspect of thin-ply behavior is the *in-situ* effect, where plies embedded within a laminate exhibit higher transverse and shear strength than when tested alone. This well-known effect arises because the constraining effect of neighboring plies – whose stiffness and location in the laminate also influence the magnitude of the effect – reduces the elastic energy available for crack propagation within the embedded ply, thus preventing crack growth and delaying the onset of matrix cracking [17]. This concept is further supported by the observation that the tensile strength of unidirectional (UD) thin-ply laminates shows little to no improvement over thick-ply counterparts—and can even decrease slightly due to minor fiber damage from the tow-spreading process—confirming that the primary benefits arise from ply constraint rather than an inherent material improvement [32].

Theoretical models based on Linear Elastic Fracture Mechanics (LEFM) initially predicted that this *in-situ* transverse strength should scale inversely with the square root of the ply thickness ($1/\sqrt{t}$) [17]. However, experimental measurements on damage onset, for instance using acoustic emission, have revealed that this scaling is not universally observed. Some material systems exhibit a more linear increase in damage initiation stress as ply thickness decreases, a trend which is illustrated in Figure 2.5 [7, 32]. The plot compares their experimental data with predictions from both the theoretical *in-situ* strength model and a more closely fitting empirical linear model.

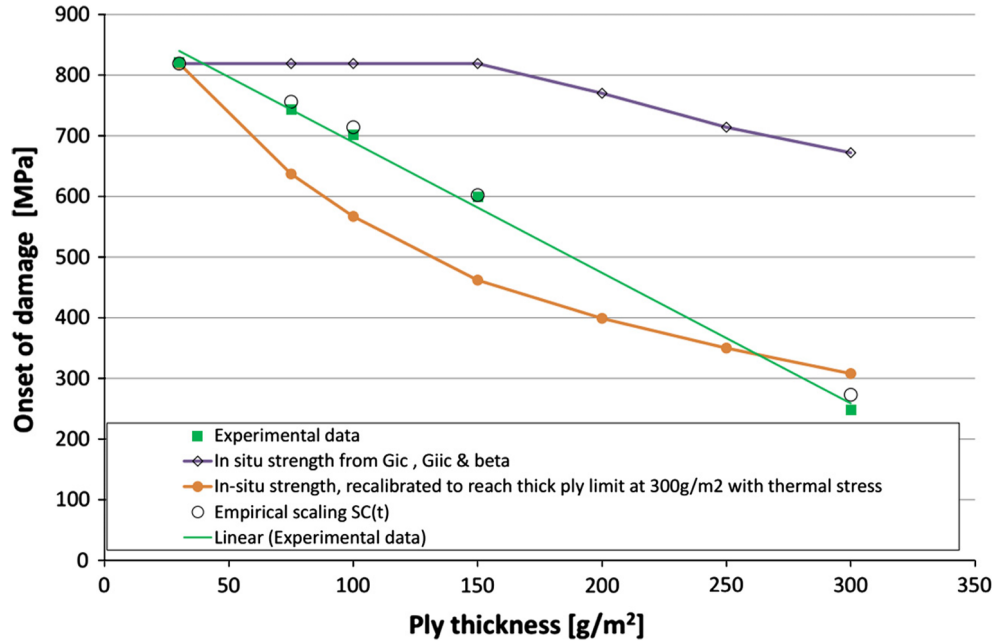


Figure 2.5: Comparison of theoretical and experimental scaling for damage onset stress as a function of ply thickness (reproduced from [7]).

Based on this observation, Amacher et al. [7] developed an empirical linear scaling function to predict these *in-situ* strength gains which are normally poor for thick laminates due to their inability to capture delamination. Their experimental data, plotted against ply thickness, revealed a clear linear increase in the stress at damage onset, which contrasted with the asymptotic curve predicted by traditional fracture mechanics models.

A potential explanation for these varying observations includes a duality of parameters in the ply thickness effect. While the ply's *in situ* strength increases due to better constraint, its apparent transverse fracture toughness (G_{app}) might simultaneously decrease, particularly in very thin plies. This toughness reduction is thought to occur if the crack opening displacement becomes too small relative to the microstructural constituents to enable the formation of a fully developed fracture process zone. This may have an effect potentially more pronounced in toughened matrices. This mechanism was experimentally confirmed by Cugnoni et al. [43], who measured a significant decrease in apparent transverse fracture toughness as ply thickness was reduced. Their work also identified the material prerequisites for this beneficial constraint to exist: a sufficiently ductile matrix (e.g., >2.5% elongation-to-failure) is needed to prevent premature fracture, while a high fiber modulus intensifies the constraining effect (driven by the ply's orthotropy, E_1/E_2), allowing it to appear even at larger ply thicknesses. This duality between constraint-induced strength increase and potential toughness reduction dictates the specific ply thickness scaling observed for a given material system [32, 43]. While this effect is well-established for tension and shear, a micro-mechanical modeling study by Arteiro et al. [16] was the first to identify a significant *in situ* effect for transverse compression as well, with the model predicting an increase in strength as ply thickness decreases [16].

In parallel to the *in-situ* effect, the reduction in ply thickness also results in a refined microstructure characterized by a more uniform fiber dispersion and reduced resin-rich zones [71]. As shown in Figure 2.6 for unidirectional, cross-ply (CP), and quasi-isotropic (QI) laminates, this microstructural refinement – and the corresponding reduction in defects – is evident as the ply thickness decreases from 200 μm down to 20 μm .

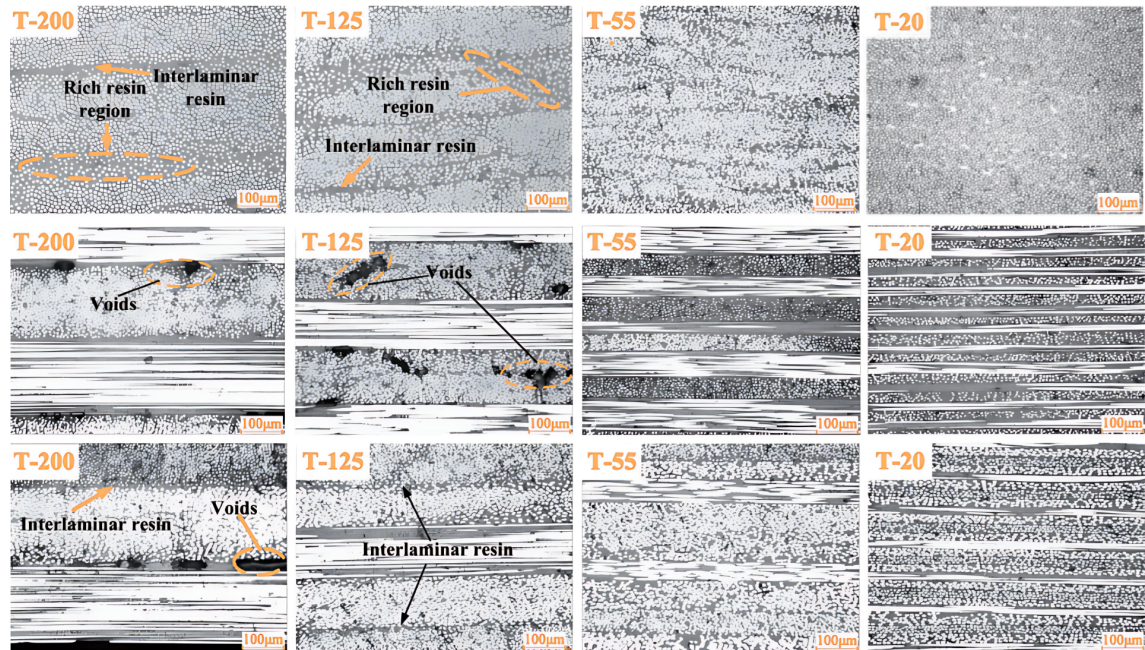


Figure 2.6: Effect of ply thickness (T , in μm) on the microstructure of various CFRP laminates (adapted from [71]).

Taken together, the suppression of matrix cracking via the *in-situ* effect (regardless of the exact scaling) and the refined microstructure cause a significant shift in failure mechanisms. Unlike thick plies, which often fail progressively through extensive matrix cracking and delamination, thin plies suppress these failure modes. This leads to a more catastrophic, fiber-dominated failure that occurs near the ultimate failure strain [71]. A study by Amacher et al. [6] clearly demonstrated this transition: their thick-ply quasi-isotropic specimens failed progressively, while the thin-ply counterparts failed in a quasi-brittle manner from a single macroscopic crack with almost no preceding damage.

2.2.3 Mechanical Performance Profile

The refined microstructure and altered damage mechanisms of thin-ply composites translate into notable changes across various material properties. The following subsections examine these changes, covering static strength, fatigue and durability, impact performance, and the trade-offs in fracture behavior.

2.2.3.1 Static Strength (Tension, Compression, Shear, and Bearing)

Under static loading, thin-ply composites exhibit mostly improvements when compared to their standard-ply counterparts. A clear summary of these gains across various properties is provided in Table 2.1, which is taken from the work of Galos et al. [71]. It is important to note that the method of stacking plies to increase laminate thickness influences whether such benefits are gained. The importance of this choice was illustrated in earlier work by Wisnom et al. [224], who found that increasing a quasi-isotropic laminate's thickness via sublamine repetition led to a 10% strength increase, whereas using ply-blocking resulted in a 62% strength reduction. Amacher et al. [7] likewise confirmed that ply-block stacking of thin plies offered only a minor performance improvement over a standard thick ply [7].

Table 2.1: Comparison of strength properties for thin, standard, and thick-ply CFRP laminates (data reproduced from [71]).

Ply thickness	Laminate type	Tensile strength (MPa)	Compression strength (MPa)	Compression after impact (~30 J) (MPa)	Interlaminar shear strength (MPa)
Thin	Quasi-isotropic	940 ± 48	770 ± 6	240	80 ± 1
	Unidirectional	2740 ± 130	1250	–	80 ± 1
	Cross-ply	–	–	–	87 ± 1
Standard	Quasi-isotropic	812 ± 5	664 ± 3	220	71 ± 1
	Unidirectional	2740 ± 130	1150	–	76 ± 1
	Cross-ply	–	–	–	53 ± 2
Thick	Quasi-isotropic	850 ± 41	606 ± 164	140	50 ± 4
	Unidirectional	2740 ± 130	1100	–	67 ± 2
	Cross-ply	–	–	–	55 ± 2

Tensile Strength

For unnotched tensile strength, Table 2.1 shows a notable increase for quasi-isotropic laminates from 812 MPa to 940 MPa (an improvement of approximately $\approx 16\%$). This aligns with and quantifies the benefits reported in the previously introduced study by Sihm et al. [193], who originally demonstrated a 10% increase in un-notched tensile strength by suppressing matrix micro-cracking and delamination through the use of dispersed thin plies. Other studies have shown even clearer improvements. For instance, Amacher et al. [7] reported a +42% increase in unnotched tensile strength and, more drastically, a +230% increase in the stress at the onset of damage when comparing 30 g/m² and 300 g/m² plies.

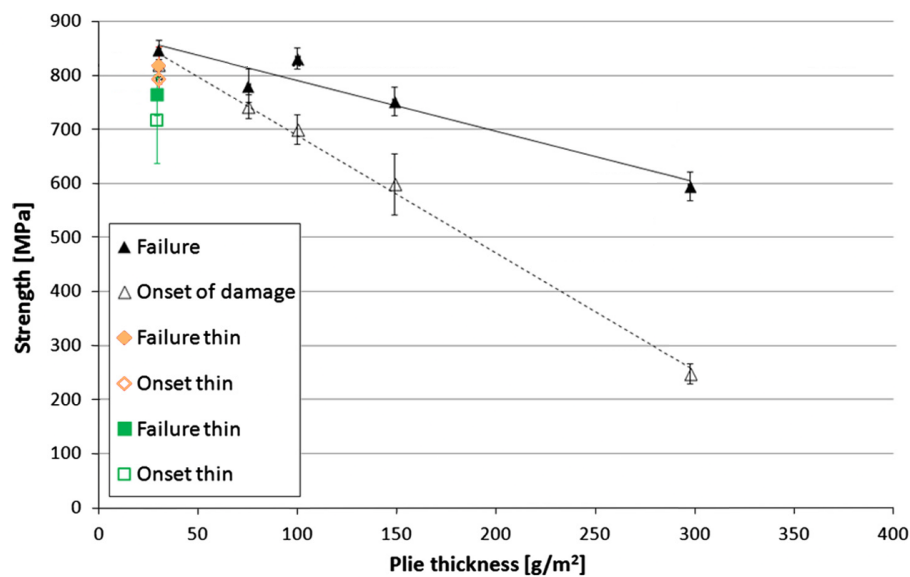


Figure 2.7: Onset of damage and ultimate strength in untouched QI specimens (adapted from [7]).

This trend is illustrated in Figure 2.7, where it clearly shows that as ply thickness decreases, the stress for damage initiation rises much more steeply than the ultimate failure stress, nearly closing the gap between the two for the thinnest plies.

However there is a trade-off, in this 2014 work, Amacher et al. also presented findings from a previous investigation by the same research group [6], which used the same M40JB/80EP constituents and ply thicknesses. The results revealed that in quasi-static Open-Hole Tension (OHT), the thin-ply laminate had a lower ultimate strength (380 MPa) than its thick-ply counterpart (545 MPa), despite having a much higher damage onset stress, 352 MPa and 255 MPa, respectively. This change in performance was linked to the shift in failure mode above mentioned [6, 7].

This fundamental difference in failure mode leads to distinct and opposing scaling behaviors known as the hole size effect. As detailed in the review by Arteiro et al. [17], thick-ply laminates can exhibit an inverse hole size effect, where the notched strength paradoxically increases with the hole diameter. This occurs because the extensive delamination that forms around the notch acts as a stress-relief mechanism that alleviates the stress concentration, and a larger hole provides more volume for this to develop. In contrast, thin-ply laminates, with their suppressed delamination, behave according to conventional fracture mechanics. Thus, their failure is governed by the stress concentration at the notch tip, leading to a decrease in strength as the hole size increases [17].

This behavior however is not universal, a study by Tariq et al. [207] on quasi-isotropic laminates from 50 to 200 μm found that after an initial strength reduction from 50 to 100 μm , a slight increase in strength was noted for 150 μm and 200 μm plies with the authors suggesting this phenomenon may be attributed to stress redistribution mechanisms [207].

Compressive Strength

In contrast to the mixed results in tension, the benefits in compression extend to both unnotched and notched specimens. For instance, Table 2.1 highlights a significant increase in compressive strength for quasi-isotropic laminates, from 664 to 770 *MPa* [71].

These gains – a nearly 24% increase for UD material, from 848 *MPa* for thick plies to 1052 *MPa* for thin plies [6] – are attributed to the more uniform microstructure and reduction of resin-rich areas. Improved fiber alignment and reduced waviness inherent in thin plies make the laminate less prone to failure via fiber kinking and micro-buckling instabilities [7, 17].

The benefits in compression also extended to notched quasi-isotropic laminates. In open-hole compression tests, Amacher et al. [7] found that thin-ply specimens exhibited an 18% higher ultimate strength than their thick-ply counterparts (255 *MPa* vs. 216 *MPa*). The stress at the onset of damage showed a similar improvement of approximately 18% [7].

This trend of increasing compressive strength with decreasing ply thickness was also validated in a study by Tariq et al. [207] who examined plies from 50 to 200 μm and confirmed a monotonic increase [207]. The underlying mechanics of this improvement were first explained in a micro-mechanical modeling study by Arteiro et al. [16], who identified a significant *in situ* effect for transverse compression. Their model predicted that the constraint from neighboring plies suppresses localized matrix failure in thinner plies, forcing a transition to a more dispersed damage mechanism that combines wedge cracking and ply fragmentation. This shift in failure mode at the micro-scale results in a higher apparent transverse compressive strength for the embedded ply, contributing to the improved performance of the overall laminate [16].

Shear and Bearing Strength

Shear and, notably, bearing strengths are also improved using thin-ply laminates. Earlier work on quasi-isotropic laminates had already shown significant gains, with thin plies increasing the ultimate bearing strength by 23% in standard conditions and 138% in hot-wet conditions [7]. The authors attributed this substantial gain to the suppression of matrix-dominated failure modes with the matrix properties degraded by heat and moisture.

More recent bolted-joint experiments in hybrid thin/thick stacks recorded monotonic increases in onset and ultimate bearing loads with thin-ply content, with a fully thin-ply laminate delivering a 47% rise in bearing-damage onset over a conventional UD lay-up [134]. These gains were associated with a shift from matrix splitting to fiber-controlled failure around the hole and are now captured by validated mesoscale Finite Element (FE) models [134].

Another benefit, also highlighted in Table 2.1, is the significant improvement in Interlaminar Shear Strength (ILSS). For all laminate types shown, the thin-ply configuration exhibits a higher ILSS than both standard and thick-ply counterparts, with the thin cross-ply laminate reaching 87 *MPa*, compared to 53 *MPa* for the standard-ply equivalent. This suggests a more robust interface less prone to delamination under shear loading.

The mechanics behind this ILSS improvement were investigated in detail by Huang et al. [96]. They demonstrated that the benefits are from two related effects. First, thin-ply laminates exhibit a more uniform microstructure, characterized by fewer voids, reduced resin-rich areas between plies, and more consistent fiber dispersion. Second, this microstructural refinement leads to a more uniform interlaminar shear stress distribution through the laminate's thickness.

This transition is illustrated in Figures 2.8, which shows the interlaminar shear stress profiles for QI laminates with ply thicknesses ranging from 200 μm down to 20 μm .

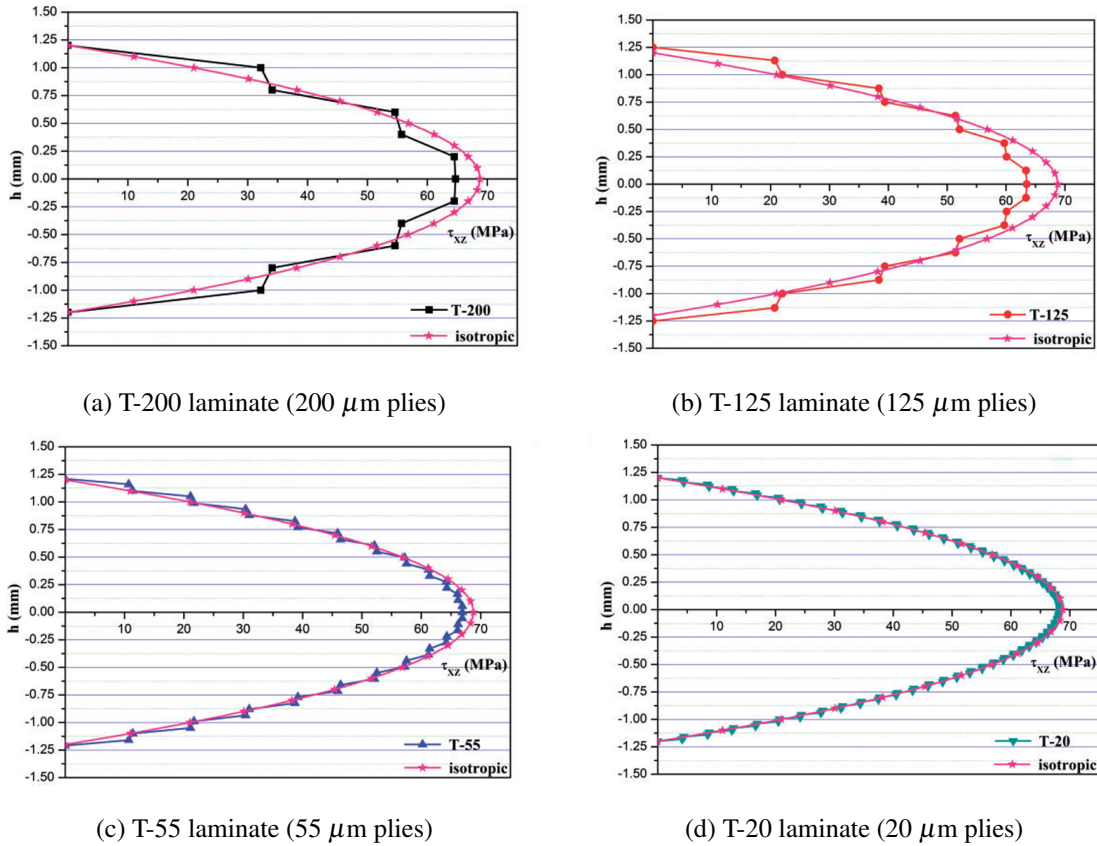


Figure 2.8: Distribution of interlaminar shear stress for thicker-ply laminates: (a) 200 μm (b) 125 μm (c) 55 μm and (d) 20 μm (reproduced from [96]).

The profile for the thick-ply laminate (T-200) is visibly irregular, with sharp discontinuities and stress peaks at the ply interfaces. As the ply thickness decreases through the standard (T-125) and thin (T-55) configurations, these peaks are progressively smoothed out. For the thinnest-ply laminate (T-20), the stress distribution becomes a nearly perfect parabolic curve, closely matching the ideal profile of a homogeneous, isotropic material. As the plies become thinner, the sharp stress concentrations that typically occur at the interfaces between plies of different orientations are smoothed out. The authors describe this as a *pseudo-isotropic* behavior, identifying the mitigation of interlaminar stress peaks as a primary reason for the increased shear strength [96].

2.2.3.2 Fatigue and Durability

The suppression of sub-critical matrix cracks directly translates to substantial benefits in fatigue resistance. By delaying damage initiation and growth, thin plies maintain stiffness over extended cycling and exhibit significantly improved endurance limits [7, 71]. In their 2007 study, Sihn et al. [193] reported a significant improvement in fatigue resistance. After 50,000 cycles, the thin-ply laminates showed negligible degradation, retaining over 97% of their initial stiffness and even showing a slight increase in strength, while the thick-ply counterparts suffered a significant 17.7% loss in stiffness and a 16.8% loss in strength.

In OHT fatigue tests, the difference was even clearer, as illustrated in Figure 2.9. At a peak stress of 316 MPa, the thick-ply laminate failed before 20,000 cycles, while the thin-ply laminate survived over 1 million cycles with negligible stiffness loss. This high endurance suggests the existence of a clear fatigue limit, which the authors identified at a stress level very close to the static onset of damage, below which the material could achieve a quasi-infinite life [7, 8].

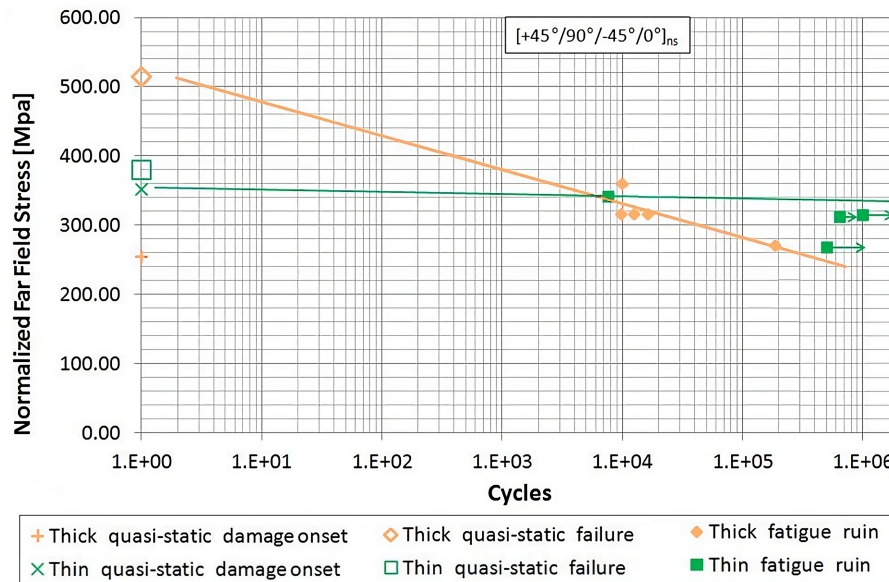


Figure 2.9: Fatigue results in open-hole tension for thick-ply (300 g/m^2) and thin-ply (30 g/m^2) quasi-isotropic laminates, showing static results and S-N curves (adapted from [8]).

2.2.3.3 Impact Performance and Damage Tolerance

The response to impact loading is also favorably affected, but the early literature presented a complex and sometimes contradictory picture, particularly for damage tolerance after impact. While some findings show that thin-ply laminates can exhibit a 10-20% increase in Compression-After-Impact (CAI) strength compared to standard-ply versions [71], other studies report decreasing damage tolerance. However, this is understandable given that the performance is highly dependent on the specific materials, impact energy, and the resulting failure modes.

For moderate energy levels, the smaller intrinsic flaw sizes and refined interfaces often result in reduced delamination areas, contributing to higher residual compressive strength. However, at very

high impact energies, the almost complete suppression of delamination can potentially lead to more localized fiber damage if energy cannot be effectively dispersed through interlaminar failure [17].

The study by Amacher et al. [7] provided a particularly detailed analysis of this behavior, revealing a clear transition in impact failure modes. The research identified an optimum ply thickness of 100 g/m^2 that minimized the projected damage area for a given impact energy, as shown by the plots in Figure 2.10. This intermediate-ply laminate dissipated impact energy through a more effective combination of delamination and fiber fracture, outperforming both the thick and thin-ply extremes. This finding showed that for impact damage resistance, thinner was not necessarily better.

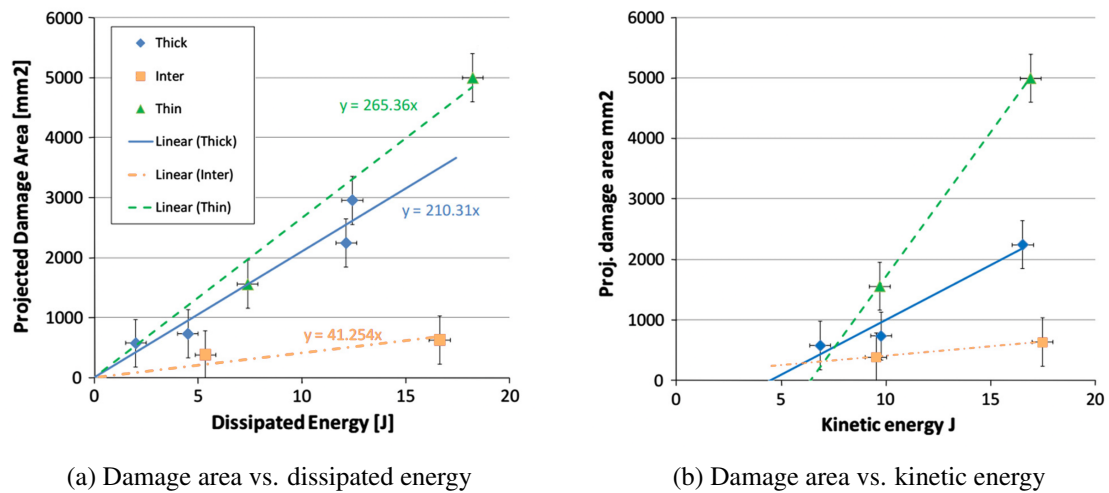


Figure 2.10: Projected impact damage area for thin (30 g/m^2), intermediate (100 g/m^2), and thick (300 g/m^2) ply laminates as a function of (a) dissipated and (b) potential impact energy (adapted from [7]).

This understanding has since been advanced by research on optimized systems incorporating toughened matrices and interleaves. Early work by Yokozeki et al. [232] on a toughened-epoxy system already demonstrated an 8% increase in Compression-After-Impact (CAI) strength and a shift in the damage from a circular to a narrower, elliptical delamination for thin plies [232]. A more recent and comprehensive study by Cugnoni et al. [43] provided a better method, showing that while matrix toughening helps, interlayer toughening is a valid strategy, boosting CAI strength by 44% [43]. They found a strong and positive ply thickness effect, with $67 \mu\text{m}$ plies achieving an 84% higher CAI than $268 \mu\text{m}$ plies. This reversal of the previous trend is attributed to a change in the subsurface damage mechanism: in thin-ply toughened laminates, the ligaments between delaminations remain thicker and more stable, resisting buckling more effectively. This work, in a way, resolves much of the earlier ambiguity, showing that for properly designed systems, thinner plies can improve damage tolerance.

2.2.3.4 Fracture Behavior

Despite these advantages, significant considerations regarding fracture behavior arise from the suppression of damage mechanisms. The reduction in splitting and local delamination near notches or holes means thin-ply laminates exhibit less stress redistribution, leading to more brittle net-section failures potentially at slightly lower ultimate loads compared to thicker laminates where progressive damage provides stress relief [17, 193].

Furthermore, this behavior extends to interlaminar fracture toughness. While Mode II toughness appears largely unaffected by ply thickness, the picture for Mode I is more nuanced. Fracture toughness at crack *initiation* is practically independent of ply thickness, but the steady-state propagation toughness (the $\mathcal{R}$ -curve) decreases by as much as 50% in very thin plies [67]. This reduction is attributed to differences in fiber bridging. The uniform microstructure of thin plies creates a flatter crack front that prevents the formation of energy-absorbing bridging fiber bundles, which are more common in the heterogeneous microstructure of thicker plies [67].

The translaminar fracture toughness is also highly dependent on ply thickness. It is not an uniquely material property but also it has been considered as an *in-situ* one that scales with the thickness of the 0° ply-block [211]. Teixeira et al. [211] experimentally demonstrated this using compact tension tests on cross-ply laminates with 0° ply-blocks ranging from 30 to 120 μm . As shown in Figure 2.11, their results revealed a strong linear increase in both initiation and propagation toughness with increasing thickness. Quantitatively, the propagation toughness increased significantly from 49 to 160 kJ/m^2 , even when delamination was suppressed. This was attributed to an increase in the energy dissipated by fiber pull-out mechanisms, as the height of the pulled-out fibers was observed to increase with the ply-block thickness [211].

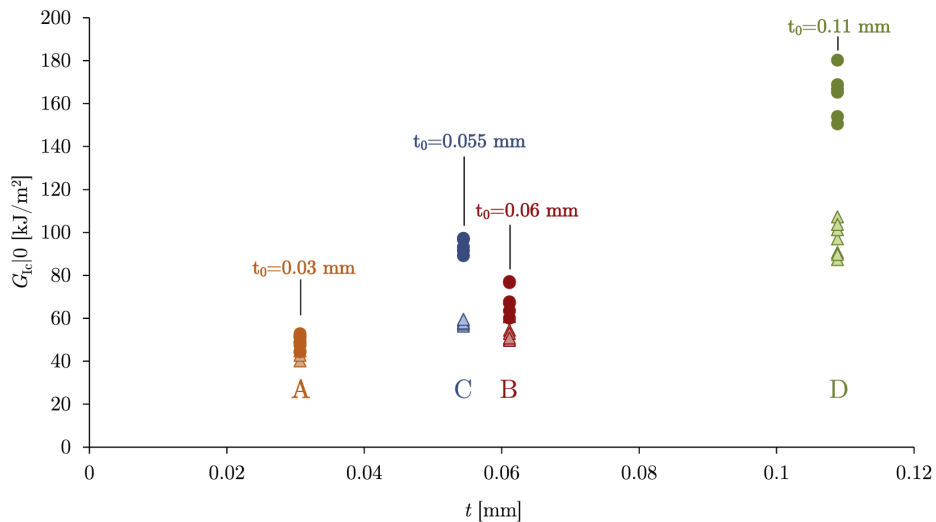


Figure 2.11: Translaminar fracture toughness as a function of 0° ply-block thickness for thin-ply cross-ply laminates. Initiation and propagation marked with $\triangle$ and $\bullet$, respectively (reproduced from [211]).

Figure 2.12 summarizes these dependencies across all major fracture modes.

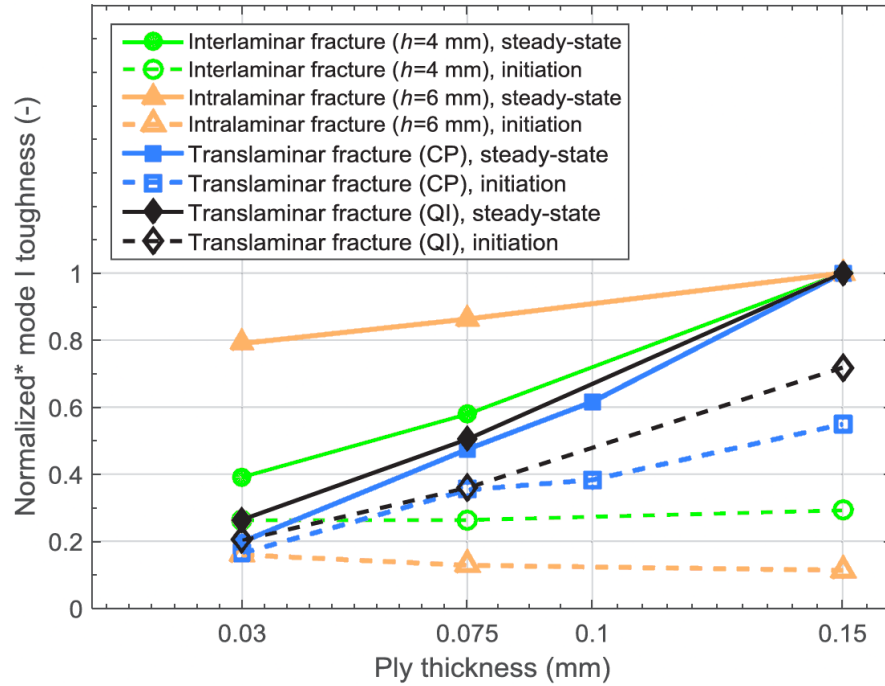


Figure 2.12: Summary of normalized fracture toughness for inter-, intra-, and translaminar modes as a function of ply thickness, showing the distinction between initiation (dashed lines) and steady-state (solid lines) behavior (adapted from [67]).

This reduction in translaminar toughness is a concern because it is the property that governs damage tolerance [32]. Frossard quantified this trade-off directly: for the same material system, reducing the ply thickness from 150 to 30 μm resulted in a 75% reduction in translaminar fracture toughness, which accompanied only a 35% increase in strength [67]. This finding shows that the loss in toughness can be proportionally much larger than the gain in strength, potentially negating the benefits of thin plies in damage-critical applications.

2.2.4 Multifunctional Properties

Beyond purely structural improvements, the refined microstructure of thin-ply composites can lead to other multifunctional benefits. For instance, the reduction in the size and number of interlaminar resin-rich zones has been shown to improve electrical resistivity in the through-thickness direction. This improved electrical performance also translates to a notable improvement in lightning strike resistance since the lower electrical anisotropy allows the current from a strike to be distributed more effectively through multiple plies rather than being concentrated on the surface, which reduces localized bulk damage and leads to higher residual strength [17]. Furthermore, boosts to the vibration damping characteristics have also been reported for thin-ply Non-Crimp Fabrics (NCFs), expanding their potential applications [71].

2.2.5 Performance Summary and Trade-offs

The complex performance profile of thin-ply composites, with its numerous advantages and trade-offs discussed in the preceding sections, is summarized in Table 2.2.

Table 2.2: Summary of Composite Properties in Thin vs. Standard/Thick Plies

Property	Thin Plies	Standard/Thick Plies
<i>General Design & Microstructure</i>		
Design Freedom / Optimization	Better	Worse
Microstructure Uniformity (fewer defects)	Better	Worse
Manufacturability	Worse	Better
<i>Damage Initiation & Failure Modes</i>		
Transverse Cracking Onset	Better	Worse
Damage Onset Stress (General)	Better	Worse
Failure Mode (Progressive vs. Catastrophic)	Worse	Better
<i>Static Strength</i>		
Tensile Strength (Unnotched, QI)	Better	Worse
Tensile Strength (Unnotched, UD)	Neutral	Neutral
Tensile Strength (Open-Hole, OHT)	Worse	Better
Compressive Strength (Unnotched & Notched)	Better	Worse
Bearing Strength	Better	Worse
Interlaminar Shear Strength (ILSS)	Better	Worse
<i>Durability & Damage Tolerance</i>		
Fatigue Resistance / Endurance Limit	Better	Worse
Impact Damage Resistance	Conditional	Conditional
Compression-After-Impact (CAI)	Conditional	Conditional
<i>Fracture Behavior</i>		
Apparent Transverse Fracture Toughness (G_{app})	Worse	Better
Mode I Interlaminar Fracture Toughness	Worse	Better
Mode II Interlaminar Fracture Toughness	Neutral	Neutral
Translaminar Fracture Toughness	Worse	Better
<i>Multifunctional Properties</i>		
Through-Thickness Electrical Performance	Better	Worse
Lightning Strike Resistance	Better	Worse
Vibration Damping	Better	Worse

The preceding analysis demonstrates that while thin-ply technology offers a path to high-strength composites by suppressing damage, this same mechanism can lead to performance trade-offs. This illustrates that altering the meso-structure alone has limitations. An alternative approach is to also modify the material at the constituent level through fiber hybridization.

2.3 Hybrid composites

Fiber-hybrid composites represent a specific category within the broader field of composite materials, defined as materials reinforced by a combination of at least two distinct types of fibers within the same matrix.

By combining different fiber types, the objective is to leverage their respective advantages while minimizing their drawbacks, allowing for the creation of materials whose properties are highly customizable. This strategy enables several possibilities: reducing cost by partially replacing expensive fibers such as carbon with more economical options such as glass, increasing toughness or impact resistance, achieving a more balanced overall set of properties such as the stiffness-toughness trade-off, controlling the failure mode to be more gradual, and incorporating specific functional properties such as electrical conductivity or vibration damping [126, 199, 201].

Historically, research into fiber-hybrid composites gained attention in the 70s and 80s, significantly driven by the high cost of early carbon fibers, making carbon/glass combinations promising [126, 199]. The concept was debated as early as the first International Conference on Composite Materials in 1975, following early work from Bunsell and Harris [35] on carbon-glass interlayer hybrids. In their study, they compared the tensile response of hybrids where the layers were intentionally left unbonded versus those that were well-bonded. The unbonded specimens failed with a single, sharp drop in load as soon as the carbon layer fractured. In contrast, the bonded specimens exhibited a "saw-tooth" stress-strain curve, characterized by a series of smaller, repeated load drops. Bunsell and Harris attributed this progressive failure to the multiple fragmentation of the brittle carbon layer, a mechanism made possible by the continuous load transfer across the strong interlayer bond. Their research provided the first experimental evidence that a strong interlayer bond could transform the material's failure from catastrophic to progressive due to the gradual fragmentation of the carbon fibers [32, 35].

Concurrently, Hayashi [91] made the important observation that carbon fibers in carbon/glass hybrids could achieve 40% higher failure strains compared to non-hybrid carbon laminates. This phenomenon, where the hybrid performs differently than a simple rule-of-mixtures prediction, became known as the *hybrid effect*, which is discussed in Section 2.3.2.

Despite this early interest, progress was slowed by technical limitations. The minimum achievable ply thickness at the time restricted the potential mechanical improvements, mainly due to energy-related constraints [53]. However, interest re-emerged, driven by the modern commercial availability of thin plies and the continuing need for materials with improved toughness [199, 201].

This renewed focus is exemplified by major research initiatives such as the UK's High Performance Ductile Composite Technologies (*HiPerDuCT*) program, a large-scale collaboration between the University of Bristol and Imperial College London aimed at overcoming the "inherent brittleness" of conventional composites [226]. Running from 2011 to 2018, the project systematically investigated a wide range of strategies to induce a ductile or pseudo-ductile response, including the use of thin-ply, novel hybrid configurations, and discontinuous fibers. This body of modern research initiated a new field focused on creating composites that fail more gradually, leading

to a more systematic understanding of how fiber arrangement influences performance and the identification of three main hybridization techniques.

2.3.1 Hybrid composite configurations

Hybrid composites can be arranged in various configurations, distinguished by how the different types of fiber, which can be continuous or discontinuous, are spatially arranged within the matrix. This spatial arrangement, commonly described in terms of dispersion, is a key parameter because it directly affects fiber interactions, overall mechanical properties, failure mechanisms, and the magnitude of any hybrid effects [199, 201].

A higher degree of dispersion implies more intimate mixing between the different fiber types. To formalize this concept, Swolfs et al. [201] proposed a quantitative measure for dispersion based on the average proximity of low-elongation fibers to their nearest high-elongation neighbors. The formula for this metric, along with other microstructural parameters, is provided in Section 2.7.1.4.

Based on the level of dispersion, three principal configurations are generally recognized [126, 199, 234], as illustrated in Figure 2.13.

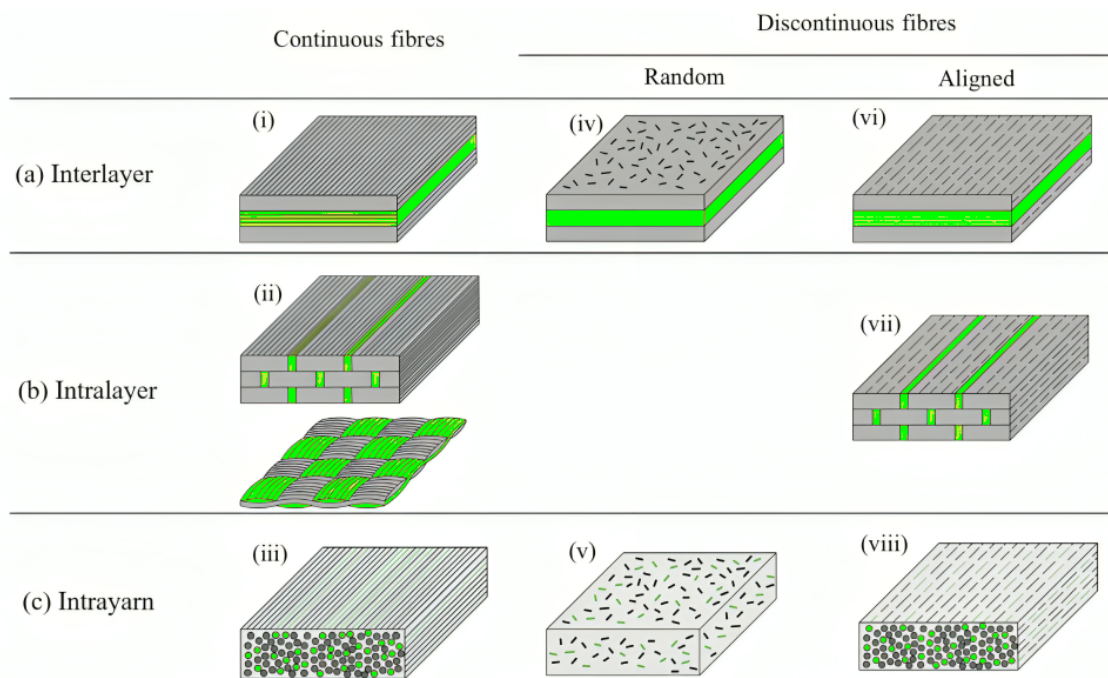


Figure 2.13: Hybrid configurations for continuous and discontinuous fiber reinforced composites, for different levels of dispersion and alignment (adapted from Yu et al. [234]).

- The **interlayer** (or ply-by-ply) configuration involves stacking distinct layers, each containing only one fiber type. This results in the lowest dispersion level and is often the simplest manufacturing approach. While hybridization is traditionally defined by combining different fiber types, modern approaches also include creating hybrids by strategically mixing plies

of the same material but with different thicknesses, such as standard-ply and thin-ply, to optimize performance and control failure modes [135].

- The **intralayer** (or tow-by-tow/yarn-by-yarn) configuration features the mixing of different fiber tows or yarns within the same layer, for instance, through co-weaving or placing bundles adjacently. This approach results in an intermediate dispersion.
- The **inrayarn** (or fiber-by-fiber) configuration involves different fibers thoroughly mixed within the same tow, often via processes like commingling. This approach achieves the highest dispersion level (also described as *intimately mixed* [126] or as intermingled [234]).

The level of dispersion can be further refined within these principal configurations. For instance, an interlayer hybrid can be made with two thick layers (low dispersion) or with many thin, alternating ones (higher dispersion). Similarly, an intralayer hybrid can use different bundle sizes. Figure 2.14 illustrates several of these specific microstructures, showing a progression from simple layered arrangements to a fully random mix of fibers.

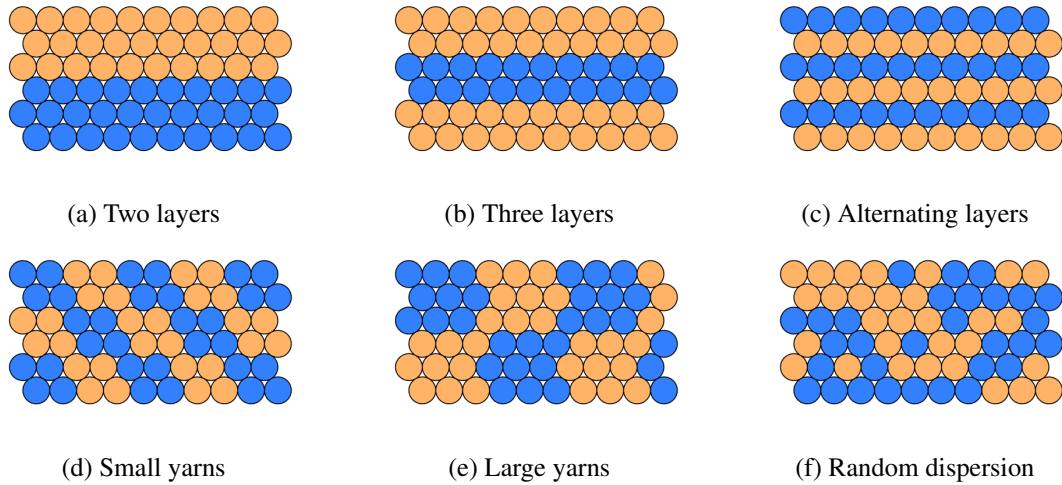


Figure 2.14: Examples of different hybrid microstructures with varying degrees of fiber dispersion.

Additionally, more complex designs can be created by combining multiple principal configuration types. One example, provided in [199], is an inrayarn hybrid yarn woven with another homogeneous yarn.

Beyond the configuration type, the properties of a hybrid composite are often described by the proportion of the low-strain material, denoted as γ . For an interlayer hybrid, this fraction is calculated using the formula:

$$\gamma = \frac{t_L}{t_H + t_L} \quad (2.1)$$

where t_L represents the total nominal thickness of LE material plies and t_H represents the total nominal thickness of the high-strain material plies within a defined ply-block [32].

2.3.2 The Hybrid Effect

One of the most significant benefits of fiber hybridization is the ability to tailor material properties beyond what is achievable with a single fiber type. The term *hybrid effect* specifically describes the phenomenon where the actual properties of a hybrid composite deviate from a Rule-of-Mixtures (ROM) prediction. The ROM serves as a baseline model, estimating the overall properties of a composite (such as its elastic modulus or strength) by taking a weighted average of the properties of its individual constituents, typically based on their volume fractions. A deviation from this predicted behavior can be synergistic (positive), leading to improved performance, or antagonistic (negative) [126, 199]. Beyond such deviations, hybridization also allows for the deliberate tuning of effective properties such as thermal expansion, fracture toughness, fatigue/impact performance, and vibration damping by leveraging the characteristics of each fiber type [199, 201].

Historically, the experimental focus of hybrid research has predominantly been on the positive effect related to the increase in the failure strain of low-elongation (LE) fibers. When embedded within a composite containing high-elongation (HE) fibers, the LE fibers can sustain greater strain before fracture compared to their performance in a non-hybrid composite. This failure strain increase, often reported in the range of 10-50% [199], is a valid indicator of a positive hybrid effect. Figure 2.15 shows these two definitions: the deviation from the ROM and the increase in LE fiber failure strain.

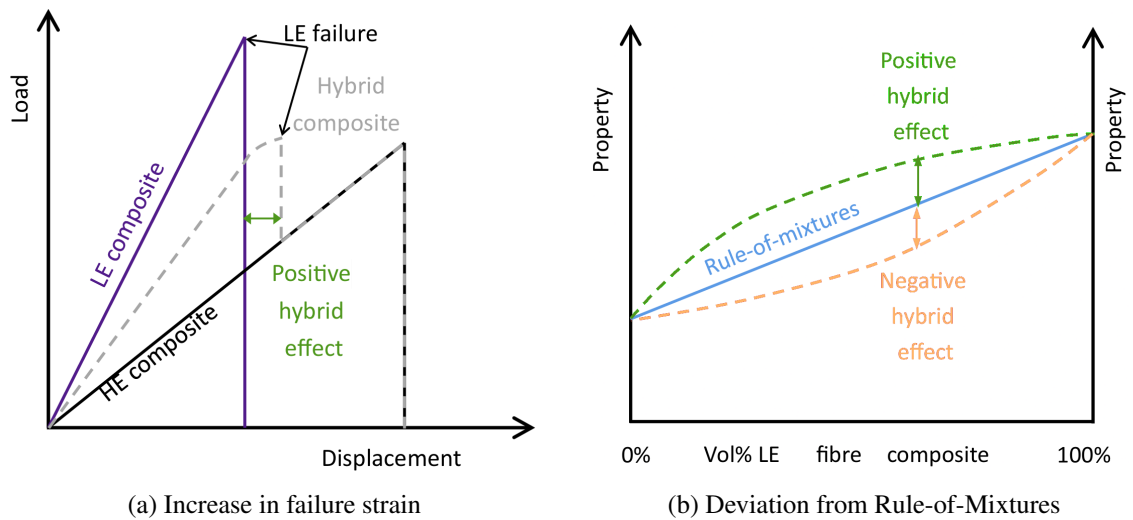


Figure 2.15: Two definitions of the hybrid effect: (a) the apparent increase in failure strain of the LE fibers and (b) the deviation from the ROM for a given property (adapted from [199]).

2.3.2.1 Hybrid Effect Origins

The origins of the positive hybrid effect, particularly the LE fiber strain increase, are generally attributed to a combination of different mechanisms. These are typically ranked by their perceived importance, starting with two that are considered secondary contributors before moving to the dominant mechanism.

Thermal Residual Stresses

Thermal residual stresses arise from the mismatch in coefficients of thermal expansion (CTE) between different fiber types and the matrix during cooling from processing temperatures. In common carbon/glass hybrids, the higher CTE of glass fibers can induce compressive residual stress in the carbon (LE) fibers, thereby increasing the applied tensile strain required to reach their failure point. However, calculations often show this mechanism alone cannot account for the full observed strain increase [199]. Strong early evidence for this conclusion was highlighted by Kretsis [126], who noted that a positive hybrid effect was still observed in carbon/Kevlar hybrids even though the residual thermal stresses in the carbon fibers were tensile, which would theoretically lower their failure strain.

Micromechanical modeling, particularly by Guerrero et al. [81], has since confirmed that this mechanism is secondary. While thermal stresses have a negligible effect on the subsequent formation of fiber break clusters, they can influence damage initiation. Specifically, inducing compressive residual stress in the LE fibers can delay their failure, increasing the hybrid's yield and failure strength. Conversely, inducing compression in the HE fibers can increase the final failure and ductile strains, enabling the possibility of slightly adjusting the final properties [81]. This aligns with recent high-fidelity modeling which confirms that the primary role of thermal stresses is to induce an initial pre-stress that effectively shifts the stress-strain curve but does not fundamentally alter the subsequent damage progression [165].

Dynamic Effects

Another potential mechanism involves dynamic effects, which are associated with the rapid energy release during a fiber break. In a hybrid composite, the behavior of these dynamic stress waves is altered by the differing properties of the constituent fibers. While modeling suggests these effects can change the formation patterns of fiber break clusters, their influence on the final failure properties is generally considered limited [199]. Therefore, like thermal stresses, dynamic effects are classified as a secondary contributor to the overall hybrid performance.

A detailed computational investigation by Guerrero et al. [81] specifically isolated this phenomenon and confirmed this conclusion. Their model showed that while dynamic effects cause larger fiber break clusters to form earlier in the loading process, they ultimately have a negligible impact on the final failure strength and strain of the composite [81].

Statistical Effects and Stress Redistribution

The dominant contribution to the hybrid effect is believed to come from statistical effects and the redistribution of stress around fiber breaks. The base concept is that the presence of HE fibers delays the formation of the critical clusters of LE fiber breaks that would otherwise lead to failure. This is achieved through two potential sub-mechanisms: first, by altering stress concentrations to "shield" nearby LE fibers, and second, by physically bridging cracks that originate from LE fiber breaks. The effectiveness of this overall statistical interaction is also linked to the inherent

strength variability (scatter) of the LE fibers, as a higher scatter provides more opportunity for these mechanisms to develop. Size effects related to the statistical strength distribution may also contribute to the positive hybrid effect, because the reduced volume of LE fibers within the hybrid composite lowers the probability of containing a strength-limiting flaw compared to a non-hybrid LE composite [199].

However, detailed finite element modeling of the micromechanics has provided a more nuanced understanding of these sub-mechanisms, particularly challenging the notion of stress shielding. A study by Swolfs et al. [200], using simulations with realistic random fiber packings, first revealed that common modeling assumptions, such as assuming equal radii for different fiber types, can lead to significant errors in predicting local stresses [200]. More importantly, their work found that in a carbon/glass hybrid, the stress concentration factors (SCFs) on the intact carbon fibers near a break were not significantly reduced compared to those in a non-hybrid carbon composite. The hybrid volume fraction and the stiffness of the hybridization fiber also had only a minor influence on the stress concentrations. This suggests that the dominant statistical mechanism is the physical bridging of the crack by the high-elongation fibers, which arrests its growth, rather than a significant reduction of the stress overload on neighboring low-elongation fibers ("shielding" effect) [200].

2.3.2.2 Factors Influencing the Hybrid Effect

While the underlying mechanisms provide a physical basis for the hybrid effect, its magnitude is controlled by several factors, including how the effect is measured, the loading conditions, and the composite's microstructural design.

A primary challenge lies in accurately defining the baseline for comparison. For instance, any deviation from an appropriate ROM prediction can be considered a hybrid effect, but accurately quantifying this deviation is non-trivial. Defining this baseline requires choosing both the appropriate form of the ROM and the correct compositional parameters. The model's form can be linear for elastic modulus or potentially bilinear for strength [199], while the most common compositional parameter is the fiber volume fraction. Applying an incorrect model or parameter can therefore lead to misleading conclusions about the presence or magnitude of a hybrid effect.

The importance of these details can be better explained by specific examples. A simple volumetric ROM, for instance, is inadequate for predicting the flexural modulus of interlayer hybrids because the stacking sequence significantly influences bending stiffness, requiring a more advanced model like classical laminate theory [126, 199]. As for the compositional parameters, Kretsis [126] demonstrated that for the same material, choosing the volume fraction of the constituent fibers instead of the volume fraction of the constituent plies could turn a calculated +3% synergistic deviation into a -6% antagonistic one [126].

This ambiguity extends from theoretical models to experimental practice. As Kretsis [126] also pointed out, it was often unclear whether the baseline failure strain of the low-elongation material should be measured from a monolithic specimen of the same total thickness as the hybrid, or from a specimen with a thickness equal to that of a single ply within the hybrid, which can yield very different results [126].

Beyond the challenges of quantification, the hybrid effect is fundamentally governed by physical factors. While a positive effect is often the design objective, the deviation from the ROM is not always synergistic. In fact, micromechanical studies have revealed that antagonistic effects on strength can occur under specific conditions. In a detailed micromechanical analysis, Mishnaevsky and Dai [109] provided an example, demonstrating that a carbon/glass hybrid under uniform displacement (strain-controlled) loading can have a lower tensile strength than either a pure glass or a pure carbon composite. They attributed this to a catastrophic "chain event": under uniform strain, the much stiffer carbon fibers carry a disproportionate amount of the load, and upon their fracture, this high stress is suddenly transferred to the adjacent, less-stiff glass fibers, triggering their immediate failure. Their work also highlighted that this negative effect is highly sensitive to the loading and microstructure since it disappeared under force-controlled loading and could be mitigated by segregating fibers into pure clusters rather than intimately mixing them.

Several material and microstructural properties are known to promote a positive hybrid effect. A larger difference between the failure strains of the LE and HE fibers, coupled with higher strength variability (scatter) in the LE fibers, generally improves synergy. A lower volume fraction of LE fibers is also beneficial, as each LE fiber receives more effective support from surrounding HE fibers [126, 199].

The degree of fiber dispersion, however, has a more complex influence than often assumed. While it is a common belief that maximizing fiber intermingling produces the strongest effect, detailed modeling has revealed important nuances. For instance, simulations by Swolfs et al. [202] on 50/50 carbon/glass hybrids, summarized in Figure 2.16, showed that while a randomly dispersed (inrayarn) configuration yielded a significant 9% failure strain increase, an idealized single-fiber, layer-by-layer (interlayer) architecture achieved an even greater increase of 16%.

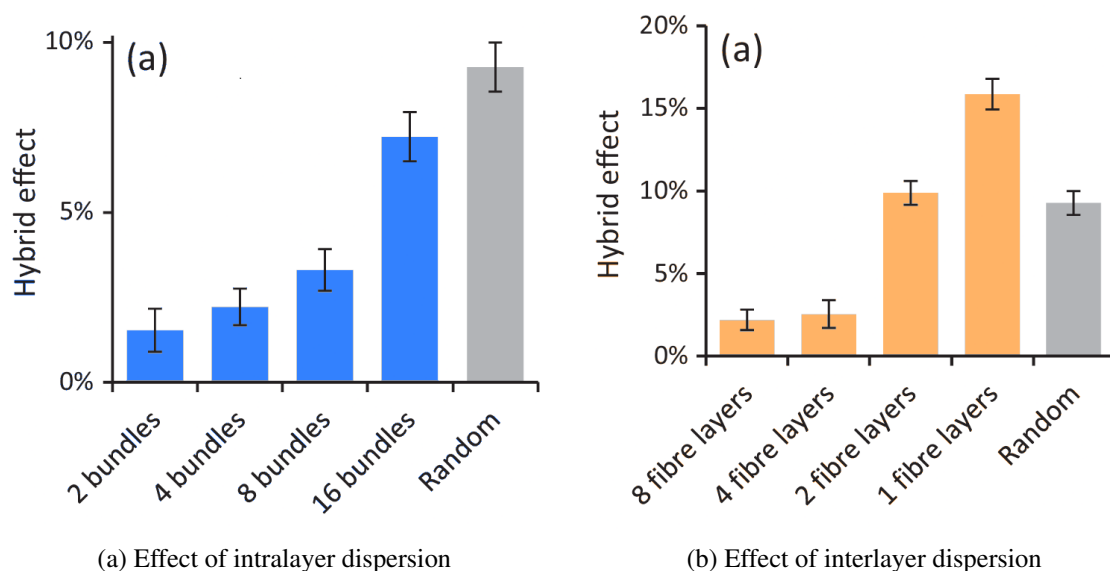


Figure 2.16: Influence of the degree of dispersion on the hybrid effect for 50/50 carbon/glass composites. (a) Intralayer hybrids with different bundle sizes (b) Interlayer hybrids with different layer thicknesses (adapted from [202]).

This counter-intuitive result was attributed to the failure mechanism. In the layered structure, a cluster of broken fibers has fewer pathways to grow because each fiber has only two neighbors of the same type, restricting damage propagation to a single direction. In a random configuration, however, a fiber has more neighbors of the same type, offering multiple pathways for the cluster to expand. This restriction of damage pathways in the layered design more effectively delays the development of a critical-sized cluster and postpones final failure. This finding shows that maximum dispersion is not always the best strategy, as architectures like thin-ply interlayer hybrids can outperform fully mixed designs.

2.3.3 Influence of Size Effects on Hybrid Performance

The statistical nature of fiber failure means that hybrid composite performance is also subject to size effects, where a material's strength changes with its volume. A detailed micromechanical investigation by Guerrero et al. [81] on carbon-carbon hybrids revealed a important trade-off associated with this phenomenon. Their simulations showed that while increasing a specimen's volume (both length and cross-section) enhances the relative positive hybrid effect, it simultaneously leads to an earlier onset of failure, reducing the composite's absolute strength and ductility.

The mechanism driving this size-dependent behavior is the formation of significantly larger clusters of broken fibers in larger specimens, which causes failure. The study clarified that this strength degradation was much milder when scaling the cross-section compared to scaling the length. The improved relative hybrid effect in larger samples was attributed to the fact that the LE fibers are better supported by the hybrid architecture, making their strength less sensitive to scaling compared to their behavior in a non-hybrid composite [81].

2.3.4 Mechanical Properties

The mechanical behavior of hybrid composites has been extensively documented, particularly in comprehensive reviews by Kretsis [126] and Swolfs et al. [199, 201].

2.3.4.1 Tensile properties

The tensile behavior of hybrid composites is typically characterized by three main properties, which are discussed in detail below: the elastic modulus, the onset of damage, and the ultimate tensile strength. A key parameter for describing the composition of a hybrid composite is the low-elongation volume fraction, denoted by γ . For an interlayer hybrid laminate, this is defined as the ratio of the total thickness of the low-strain material (t_L) to the total thickness of the composite:

$$\gamma = \frac{t_L}{t_L + t_H} \quad (2.2)$$

where t_H is the total thickness of the high-strain counterpart.

Tensile Modulus

Under longitudinal tensile loading, the elastic modulus (E_h) generally conforms well to a linear Rule-of-Mixtures (ROM) based on constituent moduli (E_L , E_H) and LE volume fraction (γ):

$$E_h = E_L \gamma + (1 - \gamma) E_H \quad (2.3)$$

The validity of this model depends on the accurate determination of the volume fraction (V_f) for each constituent. Consequently, a higher hybrid modulus is achieved by using stiffer constituent fibers or by increasing the volume fraction of the stiffer fiber type.

Consistent with experimental observations by Amacher et al. [7] on unidirectional composites, which showed no significant variation in longitudinal tensile modulus across different ply thicknesses (30 to 300 g/m²), this property is generally considered independent of ply thickness. Any observed deviations from the linear ROM are typically attributed to extrinsic factors, such as fiber misalignment or errors in V_f estimation [7].

Tensile Onset of Damage

The initiation of damage in UD laminates is a statistical process involving individual fiber breakage and subsequent stress redistribution. Due to the inherent variability in fiber strength (that is often described by a Weibull distribution), some fibers break at low strains, concentrating stress on their neighbors. This can lead to the unstable growth of fiber-break clusters and cause premature laminate failure [32].

Hybrid laminates generally exhibit a greater tolerance to these initial breaks, as the HE fibers can effectively hinder the development of critical LE fiber break clusters. This mechanism can delay the onset of significant damage, representing a positive hybrid effect on the LE fiber failure strain (ϵ_L). The stress at damage onset (σ_Y^h) can thus be estimated from the improved LE failure strain:

$$\sigma_Y^h = E_{\text{hybrid}} \epsilon_L \quad (2.4)$$

This positive effect is typically more pronounced at lower LE volume fractions (γ), while high LE content can produce an antagonistic effect. The damage onset strain can be maximized by:

- Using LE fibers with greater strength variability (i.e., a lower Weibull modulus).
- Increasing HE fiber stiffness (E_H) to better mitigate local stress concentrations.
- Ensuring a sufficient failure strain ratio between the fibers ($\epsilon_H/\epsilon_L \geq 2$) to prevent simultaneous failure.

However, this behavior does not translate directly to quasi-isotropic (QI) laminates, where damage may be initiated by transverse cracking or delamination in off-axis plies before the failure of longitudinal fibers [32].

Tensile Strength

The longitudinal tensile strength (σ_h) of hybrid composites frequently diverges from a linear ROM and is better represented by a bilinear model that reflects the sequential failure process:

$$\sigma_h = \begin{cases} \sigma_H(1 - \gamma) & \text{for } \gamma \leq \zeta \\ \sigma_L\gamma + \varepsilon_L E_H(1 - \gamma) & \text{for } \gamma \geq \zeta \end{cases} \quad (2.5)$$

where the transition point, ζ , is a function of the constituent strengths and moduli:

$$\zeta = \frac{\sigma_H - \varepsilon_L E_H}{\sigma_H + \sigma_L - \varepsilon_L E_H} \quad (2.6)$$

When the LE volume fraction is low ($\gamma \leq \zeta$), ultimate strength is governed by the failure of the HE fibers. Conversely, at high LE volume fractions ($\gamma \geq \zeta$), strength is determined by the failure of the LE fibers, plus the residual load carried by the intact HE fibers at that strain level. A positive hybrid effect can improve the composite's strength above this bilinear prediction, particularly in regimes where LE failure is the dominant mode [32].

2.3.4.2 Compressive properties

While the compressive modulus of a hybrid composite is often nearly equal to the tensile modulus and generally follows a linear ROM, the compressive strength is more complex and generally lower than the tensile strength, particularly for composites containing fibers susceptible to buckling, such as carbon. In a 1987 review, Kretsis [126] noted that while compressive strength was, on average, 80% of the tensile strength, it exhibited a non-linear trend not seen in tension. Specifically, a minimum strength value appeared at compositions with a low volume fraction of the stiffer fiber (e.g., carbon fibers in a carbon/glass hybrid) [126].

This non-linear behavior is a consequence of the micromechanics of fiber stability. A unidirectional composite's compressive strength is dictated by the lateral support the matrix provides to the fibers, which prevents failure by microbuckling or kinking. The outcome of hybridization depends on how the constituent materials alter this support. A synergistic effect can occur when stiff, LE fibers are added to a more flexible, HE composite, improving overall stability and strength. Conversely, the inverse scenario leads to an antagonistic effect. This reduces lateral support for the stiff fibers, leading to the experimentally observed drop in overall strength – an effect more pronounced at higher HE fiber volume fractions and where there is a significant mismatch in constituent moduli [126, 199]. Thus, while hybridization affects these micromechanisms of failure, it does not typically introduce fundamentally different macroscopic failure modes compared to non-hybrid composites.

The role of uniform microstructural support is also highlighted by non-hybrid examples, such as the influence of ply thickness. For instance, Amacher et al. [7] observed a 24% increase in UD compressive strength when reducing the ply thickness of a carbon fiber composite, which they attributed to improved microstructural homogeneity and more uniform fiber support.

This explains the central challenge of hybridization in compression: the antagonistic effect is a more common outcome than the synergistic one. Therefore, in contrast to the strain increases seen in tension, a net positive hybrid effect in compression is rare.

2.3.4.3 Flexural properties

Flexural properties, both modulus and strength, are dominated by the laminate's stacking sequence, particularly in interlayer hybrids. The governing principle is that placing stiffer plies furthest from the neutral axis maximizes bending stiffness, as these plies must withstand the highest stresses. Consequently, the flexural strength is determined by the tensile and compressive strengths of these outermost plies. While this behavior suggests predictability, established models like Classical Laminar Theory (CLT) may not be sufficient. In a study on S-2 glass/TR30S carbon hybrids, Dong et al. [60] found that CLT significantly underestimated the flexural modulus and provided inconsistent predictions for flexural strength. Furthermore, while a distinct hybrid effect is absent for flexural modulus – which decreases linearly as glass content increases as shown in Figure 2.17a – it is established for flexural strength. As seen in Figure 2.17b, the strength peaks with a glass fiber fraction of approximately 37%, resulting in a laminate stronger than either of the non-hybrid baselines. This is attributed to the placement of the compressively stronger glass fibers on the outer surface, which delays the dominant compressive failure mode observed in these composites [60].

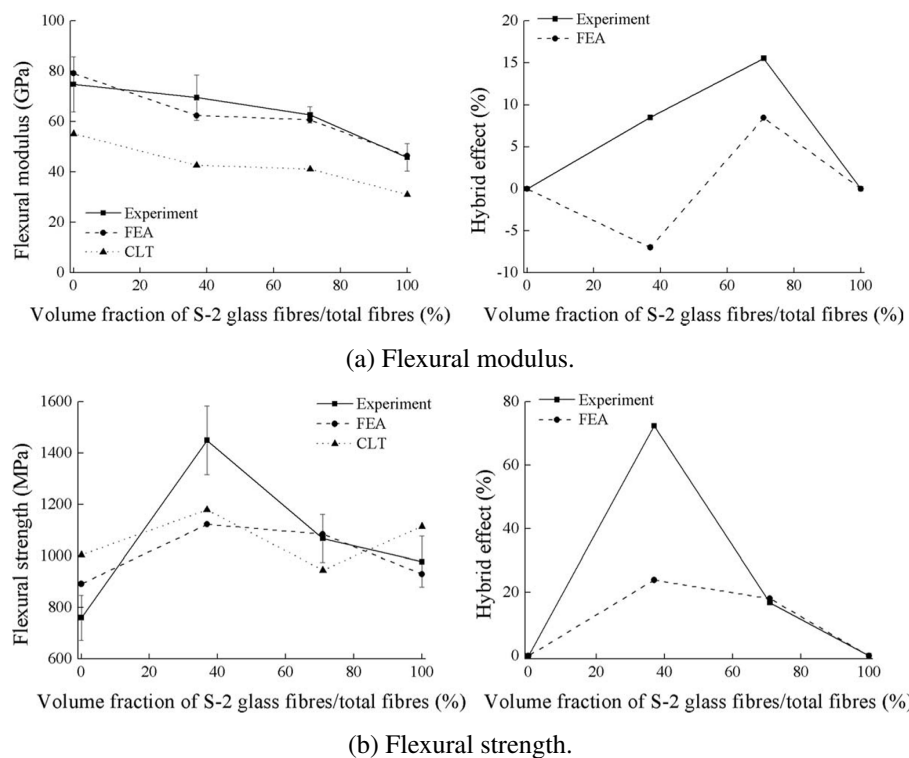


Figure 2.17: Flexural properties and corresponding hybrid effect for S-2 glass/TR30S carbon hybrids as a function of glass fiber volume fraction (adapted from [60]).

2.3.4.4 Shear properties

The characterization of shear properties in hybrids is less comprehensive than for other loading modes, with literature reviews noting significant challenges in obtaining consistent data. Early reviews, for instance, attributed these problems primarily to the test methodologies of the era [126]. Inconsistent results were obtained from different methods, such as short-beam bending and $\pm 45^\circ$ tensile tests. The $\pm 45^\circ$ tensile test was noted as particularly unsuitable for hybrids because, much like flexural specimens, the stacking sequence is an intrinsic part of the material definition, which complicates the interpretation of results.

These measurement challenges are reflected in the ambiguous data reported for shear behavior. For in-plane shear, while strength appears to follow the rule-of-mixtures, the shear modulus (G_{12}) exhibits deviations. The complexity is even greater for interlaminar shear strength, where the interface between dissimilar plies means that failure may not occur at the expected mid-plane, further complicating accurate measurement. Given these, there remains a lack of convincing evidence for any synergistic hybrid effect in shear.

2.3.4.5 Impact resistance

Improving impact resistance is often a primary motivation for hybridization, achieved by combining stiff fibers like carbon with tougher fibers such as glass, aramid, or polyethylene [126, 199]. The fundamental principle is to use the stiff fibers to distribute the impact load over a larger area, while the tougher fibers dissipate energy via deformation. This energy absorption is increased by activating additional failure modes, including extended delamination and fiber pull-out, which are associated with the tougher constituents [199].

The effectiveness of these mechanisms is controlled by the composite's architecture. Generally, higher fiber dispersion leads to better damage tolerance with the stacking sequence being, once again, very relevant. In this case, placing tougher layers on the outer surfaces can improve resistance to impact penetration, while positioning stiff LE fibers toward the laminate's core has also been found to be effective [199, 201]. Although positive synergistic effects on impact energy absorption are frequently observed, the specific outcome depends heavily on the chosen materials, configuration, impact velocity, and geometry [126, 199, 201].

2.3.4.6 Fatigue resistance

Hybridization can significantly improve fatigue performance, particularly under tension-tension loading, through a primary mechanism of crack bridging. In this process, intact HE fibers span fatigue cracks that initiate from broken LE fibers, which slows damage accumulation and extends the composite's life, especially with good interfacial adhesion and high fiber dispersion.

For example, studies on carbon/PE hybrids have resulted in flatter S-N curves, indicating slower strength degradation with cycles, and reduced scatter in fatigue life compared to non-hybrids [199]. In separate work, a life prediction model by Shan and Liao [191] for interply glass-carbon hybrids quantified the fatigue life across different fiber volume fractions, as illustrated in Figure 2.18.

Another finding from their model is that the overall fatigue response becomes dominated by the more fatigue-resistant carbon fiber component, even at low carbon contents, which also explains the improved environmental fatigue resistance observed in the same study [191].

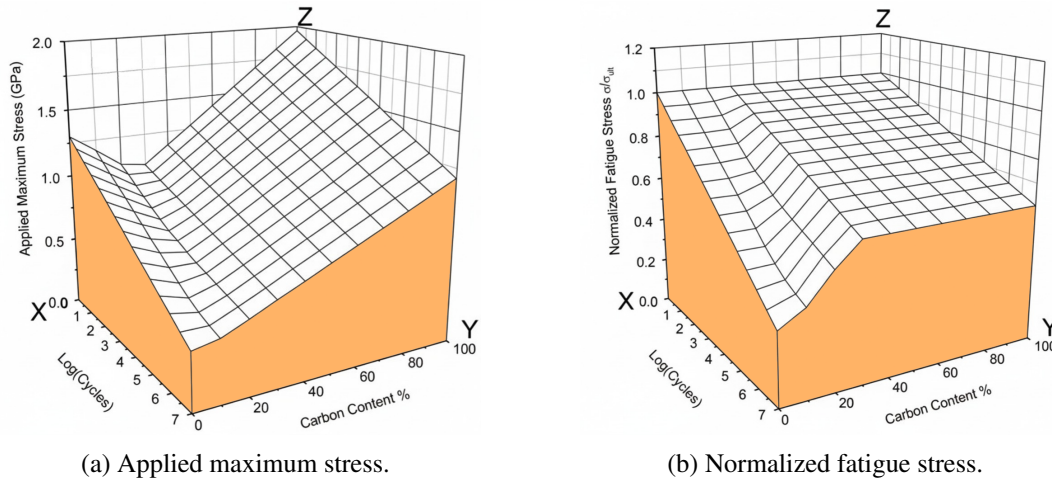


Figure 2.18: Model predictions of fatigue curves for glass-carbon laminated interply hybrid composites (adapted from [191]).

These benefits, however, do not always extend to loading cycles that involve compression, where fatigue performance becomes limited by the compressive strength and resistance of the fibers or the interface. In such cases, the addition of brittle LE fibers like carbon to a composite such as GFRP can lead to an antagonistic effect, potentially reducing the fatigue life compared to the non-hybrid baseline [199, 201].

2.3.4.7 Vibration damping

While academic research often focuses on strength and stiffness, hybridization is frequently used in industry to adjust functional properties, one of the most important of which is vibration damping [201]. High-stiffness fibers like carbon typically exhibit poor material damping, meaning they have low internal energy dissipation, which can lead to undesirable vibrations and noise in a structure. The damping performance can be improved by incorporating fibers with higher intrinsic damping, such as glass, basalt, or natural fibers like flax, into a carbon-based hybrid.

This approach allows for a balance between high stiffness, provided by the carbon fibers, and improved damping from the secondary constituent. These properties are useful for applications such as tennis rackets, bicycle frames, and other structures where vibration mitigation is required [201].

2.3.4.8 Translaminar fracture toughness

Translaminar fracture toughness characterizes a material's resistance to crack growth through the ply thickness, perpendicular to the fibers. Figure 2.19 illustrates this crack path and contrasts it with an interlaminar crack, which is discussed in the following section.

Translaminar toughness is typically dominated by energy dissipation mechanisms such as fiber-matrix debonding and subsequent fiber pull-out. In a hybrid, high-elongation fibers can bridge cracks that initiate in the low-elongation constituent, carrying load after the LE fibers have failed. This process increases the work of fracture and can lead to rising $\mathcal{R}$ -curves, which indicate increasing crack resistance with crack growth.

Hybridization can directly influence these toughening mechanisms, for instance by creating more complex crack paths [59] or by altering the effective pull-out length. A study on woven carbon/glass hybrids [203] demonstrated this latter effect, finding that specific layups can result in either positive (+15%) or negative (-15%) synergistic effects on toughness, with positive results explicitly linked to longer pull-out lengths. However, the literature also reports conflicting findings, with some studies attributing toughness improvements to greater pull-out and others to secondary damage modes such as delamination [32, 161].

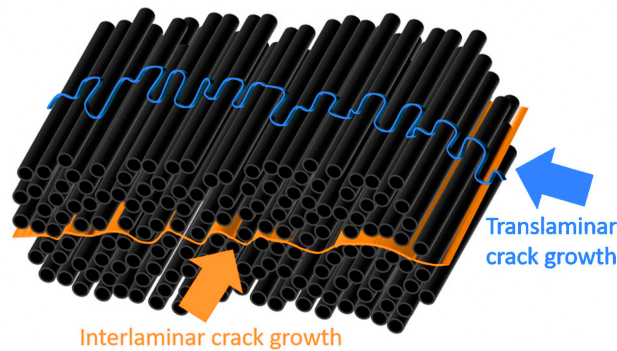


Figure 2.19: Distinction between interlaminar and translaminar crack (adapted from [201]).

2.3.4.9 Interlaminar Fracture Toughness

While translaminar fracture deals with cracks growing through a ply, interlaminar fracture toughness governs the resistance to cracks growing between plies. This property is characterized by the Mode I (opening) and Mode II (sliding) critical energy release rates, G_{Ic} and G_{IIc} respectively.

In hybrid composites, the interface between dissimilar materials is a potential weak point, and its fracture toughness is very dependent on the loading direction. A study by Sapozhnikov et al. [179] provided the first direct comparison of these properties for a hybrid interface against its constituent non-hybrid interfaces, discovering two contrasting effects [179]:

- **Mode I (G_{Ic}):** A negative hybrid effect was observed, where the toughness of the hybrid interface was governed by, and nearly equal to, the weaker constituent interface.
- **Mode II (G_{IIc}):** A positive effect was observed, where the toughness of the hybrid interface was nearly double the average of its constituents and close to their sum.

This increased shear toughness makes the hybrid interface more resistant to delamination, allowing the controlled, gradual failure [179]. This type of failure response is highly desirable and is explored in greater detail in the following section.

2.4 Pseudo-ductility

Achieving pseudo-ductility is an objective in the development of advanced composites, defining the ability of an otherwise brittle material to exhibit a non-linear stress-strain response with inelastic strain accumulation greater than 0.5% before failure [199, 201]. This behavior, which resembles the yielding of ductile metals, increases toughness and improves resistance to stress concentrations while providing a warning before catastrophic failure [41, 199, 201]. Cyclic loading tests confirm the response is pseudo-ductile because it is driven by damage accumulation rather than plastic deformation, which manifests as a permanent reduction in modulus from fiber fragmentation without the stiffness recovery or residual strain seen in truly ductile materials [222].

From early experimental work by Cz  l and Wisnom [47], the terms pseudo-yield stress and pseudo-ductile strain were established to describe this behavior. As illustrated in the schematic stress-strain curve in Figure 2.20, this response is defined by the following points [64]:

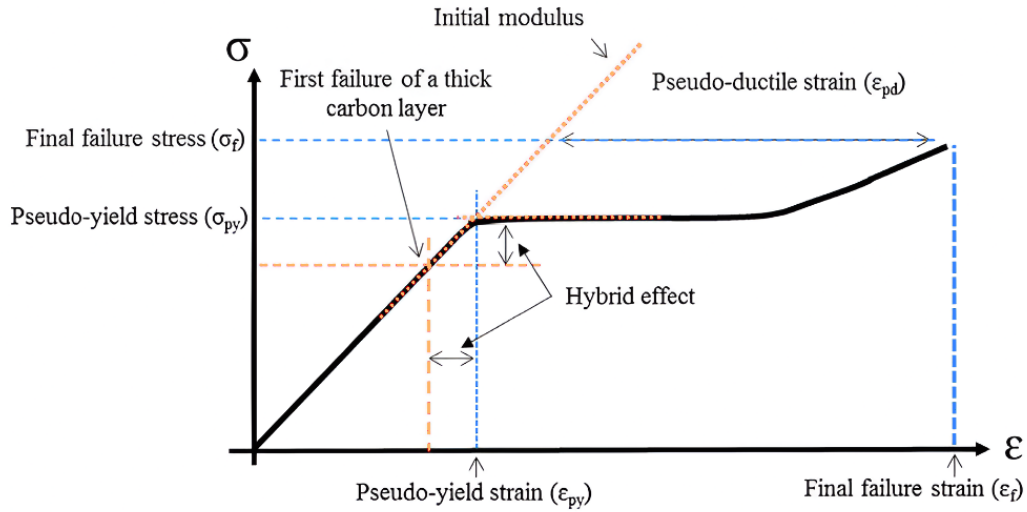


Figure 2.20: Schematic of a stress–strain graph of a pseudo-ductile hybrid (adapted from [64]).

- The *pseudo-yield stress* (σ_{py}): The stress where the response deviates from the initial linear elastic behavior.
- The *pseudo-yield strain* (ϵ_{py}): The corresponding strain, which can be determined from the intersection of lines fitted to the stress-strain curve before and after the "knee" signifying the onset of LE fiber fragmentation.
- The *pseudo-ductile strain* (ϵ_{pd}): The additional strain gained from the gradual failure process, defined as the strain difference between the final failure point and the initial elastic slope extended to the final failure stress.
- The final failure strain (ϵ_f) and stress (σ_f): The coordinates of the point where the HE material ultimately fails.

2.4.1 Mechanisms and Design Strategies

The mechanism for pseudo-ductility typically involves the sequential fragmentation of LE fibers, a process that relies on the hybrid effect to repeatedly increase the LE fiber failure strain before the high-elongation fibers fail [64, 199, 201]. An optimization study by Conde et al. [41] using a spring-element model with a genetic algorithm concluded that high fiber dispersion is necessary because low-dispersion layouts with clustered LE fibers fail simultaneously [41]. In such configurations, a failing LE fiber is surrounded only by other LE fibers that cannot carry the sudden stress increase, causing a large load drop that disrupts the gradual failure process. Pseudo-ductile behavior can be achieved through several design strategies, including the use of aligned discontinuous fibers. Yu et al [234] demonstrated this approach by comparing a high-strength carbon/E-glass hybrid with a high-modulus carbon/E-glass hybrid, where the latter combination exhibited a pseudo-ductile strain up to 1.1% at a relative carbon volume of 0.25.

A second strategy uses continuous fibers, although controlling the damage process in conventional thick plies is difficult after the initial fracture of LE plies. The energy released often causes unstable delamination or the formation of fiber break clusters, leading to abrupt failure before non-linear strain can accumulate [199, 201]. Sapozhnikov et al. [177] showed that pseudo-ductility is still achievable in thick-ply (110 to 220 μm) all-carbon hybrids by leveraging a different mechanism of stable, progressive delamination at the LE/HE interface, which itself creates the pseudo-ductile plateau [177]. This method required a high interfacial fracture toughness to ensure delamination propagated in a controlled manner after a single fracture of the LE ply, presenting an alternative to thin-ply approaches.

Another study by Sapozhnikov et al. [178] analyzed the notch sensitivity of these pseudo-ductile, thick-ply, all-carbon hybrids, showing that quasi-isotropic laminates with a central open hole were highly notch-insensitive under tension and preserved their pseudo-ductile stress-strain behavior. The calculated notch sensitivity factors for ultimate strength were below 1.15, approaching the ideal value of 1.0. Digital Image Correlation (DIC) and Acoustic Emission (AE) showed that the pseudo-ductile mechanisms localized around the hole, which blunted the stress concentration and prevented premature failure [178].

While these above mentioned strategies can induce a pseudo-ductile response, an alternative approach focuses on using the inherent properties of thin plies. This method is explored in detail in Section 2.5.

2.4.2 Modeling Approaches

Following early experimental results, analytical and numerical models were developed to predict pseudo-ductile behavior by building on statistical failure theories like the chain-of-bundles model and exploring different load-sharing assumptions, such as local (LLS) and global (GLS), to account for stress transfer around broken fibers [69, 89, 237]. Computationally efficient analytical models represent one end of this spectrum, including the probabilistic progressive damage model (PPDM), which predicts damage progression through averaging methods and Weibull statistics, alongside

approaches that predict the failure sequence by calculating stress thresholds for fragmentation and delamination [103, 104, 165]. At the high-fidelity end are micromechanical simulation models that use Representative Volume Elements to explicitly model the material's microstructure, including individual, randomly distributed fibers with stochastic strengths. These numerical models can incorporate matrix plasticity and damage through thermodynamically consistent constitutive laws and use cohesive laws for the fiber-matrix interface, allowing them to capture local failure phenomena that analytical models average out [18, 78, 142, 143, 208]. An example from Gonçalves et al. [78], shown in Figure 2.21, illustrates how these models can predict failure paths within an RVE and capture the influence of manufacturing effects such as residual stresses from curing. For instance, they can predict the formation of damage clusters from the stress concentrations around an initial fiber break (Figure 2.22) to the final failure paths influenced by manufacturing effects (Figure 2.21) [78, 208].

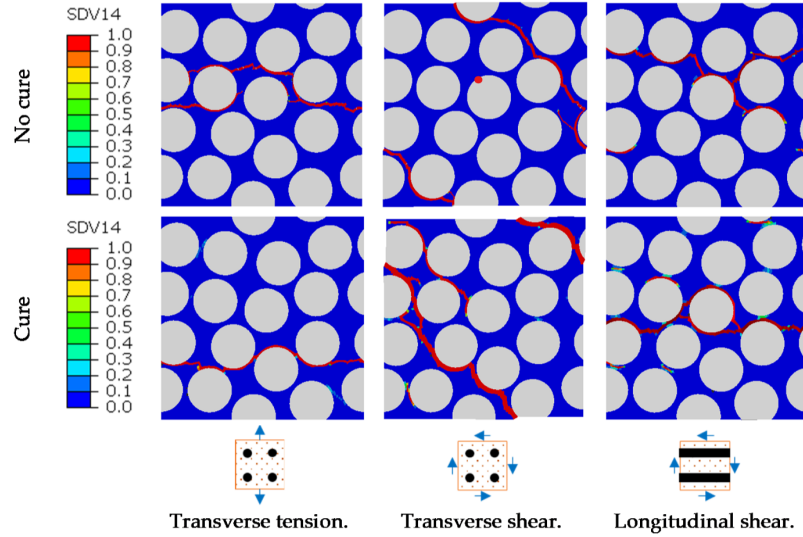


Figure 2.21: Micromechanical prediction of failure patterns with and without cure-induced residual stresses under different loading conditions (reproduced from [78]).

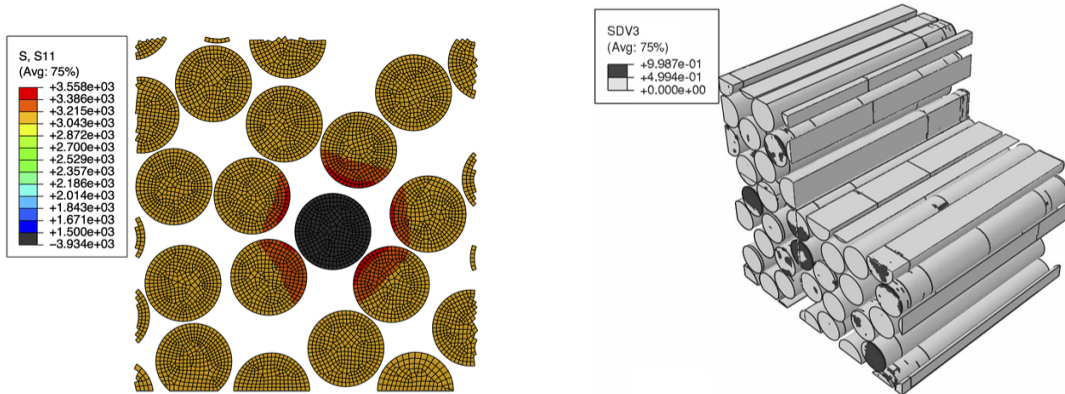


Figure 2.22: Micromechanical simulation showing stress concentrations around an initial fiber break (left) leading to a damage cluster (right) (adapted from [208]).

2.5 Thin-ply hybrid composites

Integrating the distinct strategies of thin-ply technology and fiber hybridization creates a material architecture specifically designed to transform the response from the abrupt failure typical of conventional thin-ply to a progressive, non-linear pseudo-ductile behavior, as illustrated by the schematic stress-strain curves in Figure 2.23b [48].

The success of this approach relies on promoting the desired mechanism of progressive layer fragmentation while simultaneously suppressing unstable delamination which can develop at two distinct scales (presented in Figure 2.23a). The first is intra-sublaminar delamination, which occurs within a single unidirectional (UD) hybrid building block if the low elongation material releases more energy at its first fracture than the interface toughness can withstand. The second is inter-sublaminar delamination, which occurs between neighboring hybrid sublaminae due to free-edge stresses caused by their different orientations in a multi-directional stack [49].

The challenge of intra-sublaminar delamination was first addressed by Czél and Wisnom [47], who established that the performance of a UD hybrid system is controlled by the absolute thickness of the LE ply. Their energy-based fracture mechanics model predicted a critical LE layer thickness below which the energy released from a fracture would be insufficient to cause unstable delamination. This prediction was confirmed by experiments showing that specimens with thin carbon layers of approximately $29\text{ }\mu\text{m}$ and $58\text{ }\mu\text{m}$ failed in a progressive, pseudo-ductile manner, while thicker layers of approximately $87\text{ }\mu\text{m}$ and $116\text{ }\mu\text{m}$ failed abruptly [47].

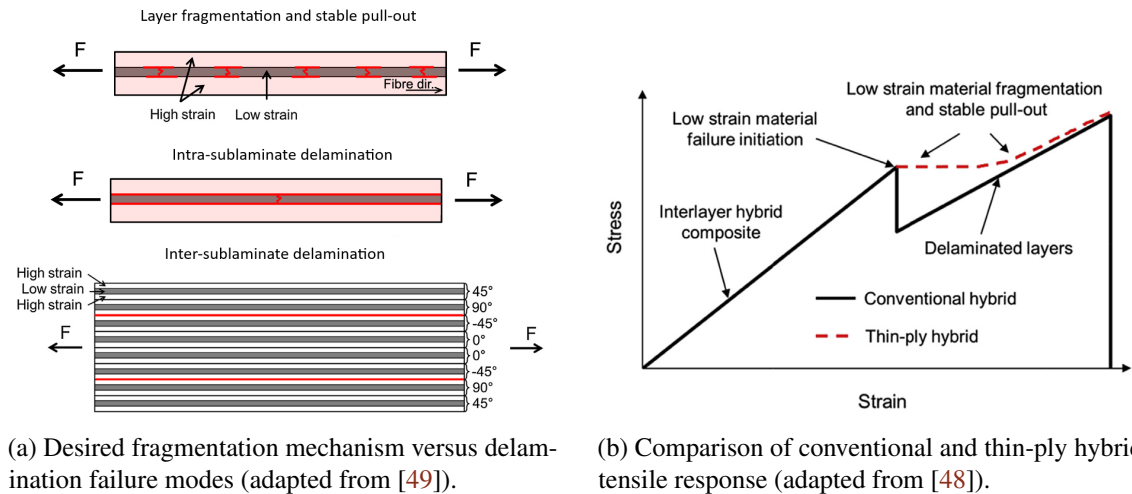


Figure 2.23: Schematics illustrating (a) the underlying damage mechanisms to be promoted or suppressed and (b) the targeted pseudo-ductile behavior of thin-ply hybrids.

2.5.1 Development and Underlying Principles

This thickness-dependent behavior was identified as a manifestation of the hybrid effect, a phenomenon investigated by Wisnom et al. in work that was first presented at a conference [227] and later published in detail [221]. Their study on carbon/glass hybrids showed a fiber strain increase

of up to 20% was observed only when the carbon layers were very thin, with the effect becoming negligible for plies thicker than approximately $80\text{ }\mu\text{m}$. The authors separated the phenomenon into two components: an "initiation effect" that delays the first critical cluster of fiber breaks, and a "fragmentation effect" where strain is accumulated before a stiffness drop occurs. Both mechanisms depend on the constraining effect of adjacent HE plies at the micro-level, which is only active when the LE layer is sufficiently thin [221].

The "initiation effect" is further explained by the micromechanics of the layered structure. As shown by Swolfs et al. [202], interlayer hybrids are efficient at delaying failure because they constrain the growth of fiber break clusters [202]. A break in a thin LE layer has fewer adjacent LE fibers to propagate to compared to a more dispersed, random configuration. This "pathway restriction" limits the development of critical clusters and promotes controlled fragmentation, providing a micromechanical basis for the performance of thin-ply hybrids over more intimately mixed architectures (e.g., comingled) [201, 202].

With a mechanistic understanding established, the focus shifted to systematic engineering of these materials. Czél et al. [45] conducted a comprehensive study on a wide range of S-glass/carbon thin-ply hybrids, experimentally confirming the trade-off between pseudo-yield stress and pseudo-ductile strain, where configurations yielding higher pseudo-ductile strains often correspond to lower initial yield stresses. This work demonstrated that material properties could be deliberately tuned by selecting different grades of LE carbon fiber, resulting in a family of materials from a balanced hybrid with a 60 GPa modulus and 1.44% pseudo-ductile strain, to a high-stiffness version achieving a 124 GPa modulus and 2.64% pseudo-ductile strain [45]. This design process was supported by the application of damage mode maps, which graphically represent expected failure mechanisms as a function of the LE layer's absolute and relative thickness, enabling the identification of optimal configurations [103, 226] (In subsequent sections, this methodology is further detailed and used to construct damage mode maps to provide the rationale for the thin-ply hybrid configurations selected).

The design framework was then extended to all-carbon interlayer hybrids to produce composites with much higher initial stiffness, a concept first presented by Czél et al. [46, 48]. This approach involved sandwiching central layers of high-modulus (M55) or ultra-high-modulus (YSH-70A, XN-80) carbon thin plies between outer layers of high-strength (T1000) carbon. The resulting composites achieved initial moduli up to 245 GPa , a pseudo-yield stress of 990 MPa and a pseudo-ductile strains up to 0.94%.

This work also identified that the pseudo-yield strain of the high-modulus fibers was elevated beyond their nominal failure strain, a direct result of the thin-ply hybrid architecture [48]. However, achieving a synergistic effect is not always guaranteed. A study by Danzi et al. [54] on thin-ply carbon-carbon hybrids confirmed that while certain configurations achieved pseudo-ductility and increased fracture toughness over a non-hybrid baseline, this increase was mainly attributable to the ply-thickness effect itself rather than hybridization synergy [32, 54].

A subsequent objective was to address another known trade-off specific to thin-ply laminates, which often exhibit brittle failure in notched configurations despite their high unnotched strength.

Two distinct hybridisation strategies have been developed to mitigate this issue. The first strategy transfers the pseudo-ductile mechanism from unidirectional materials into multi-directional laminates. As detailed by Czél et al. [49], this was achieved by using pseudo-ductile UD hybrids as sublaminates "building blocks" within a quasi-isotropic stack, with the experimental tensile responses for both the UD building block and the final QI laminate shown in Figure 2.24.

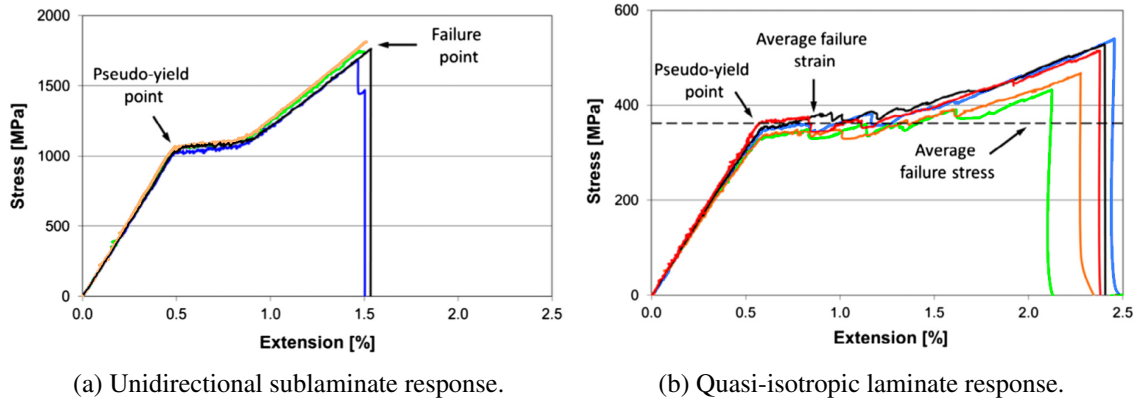


Figure 2.24: Experimental tensile response demonstrating the transfer of pseudo-ductility from (a) the unidirectional T1000/XN-80 "building blocks" to (b) the final quasi-isotropic laminate constructed from them (adapted from [49]).

However, an initial "orientation-blocked" design, where plies of the same orientation were grouped, proved susceptible to free-edge delamination. This was resolved by implementing an "orientation-dispersed" concept, where individual material plies were distributed throughout the stacking sequence, which suppressed delamination while maintaining the pseudo-ductile response [64]. Both configurations demonstrated a high degree of notch insensitivity, as the controlled fragmentation of the 0° plies blunted the local stress concentration to such an extent that the net-section strength of notched specimens exceeded the un-notched strength of the laminate [49, 226]. The robustness of this orientation-dispersed design was further examined under off-axis loading, where Fotouhi et al. [66] determined that the pseudo-ductile response was maintained for loading angles up to 20° , although the achievable pseudo-ductile strain decreased as the angle increased due to more pronounced matrix cracking.

An alternative strategy, known as selective ply-level hybridisation, was presented by Furtado et al. [70]. This approach involves combining thin off-axis plies with thicker 0° plies directly within the laminate. The thin plies suppress undesirable damage modes like delamination, while the thicker 0° plies are more effective at crack blunting, resulting in an increased open-hole tensile strength over both pure thin-ply and standard-ply baselines. This method was subsequently validated by Lovejoy et al. [135] using an aerospace-grade system, confirming its effectiveness at mitigating the typical notched brittleness of thin-ply composites.

Further investigations have provided a better understanding of the fragmentation process. Cyclic loading tests, for example, were used to distinguish the post-damage behavior of different laminate types [222]. For UD hybrids, these tests revealed a permanent modulus loss upon

unloading, as illustrated in Figure 2.25a, confirming that the non-linearity is a direct result of fragmentation damage. This response contrasts with that of thin angle-ply laminates, which, as shown in Figure 2.25b, exhibit non-linearity from matrix deformation and fiber re-orientation, resulting in permanent strain but no significant loss of reloading stiffness. This damage-induced modulus loss with little residual strain is what classifies the hybrid's behavior as pseudo-ductile, while the angle-ply response is considered ductile [222].

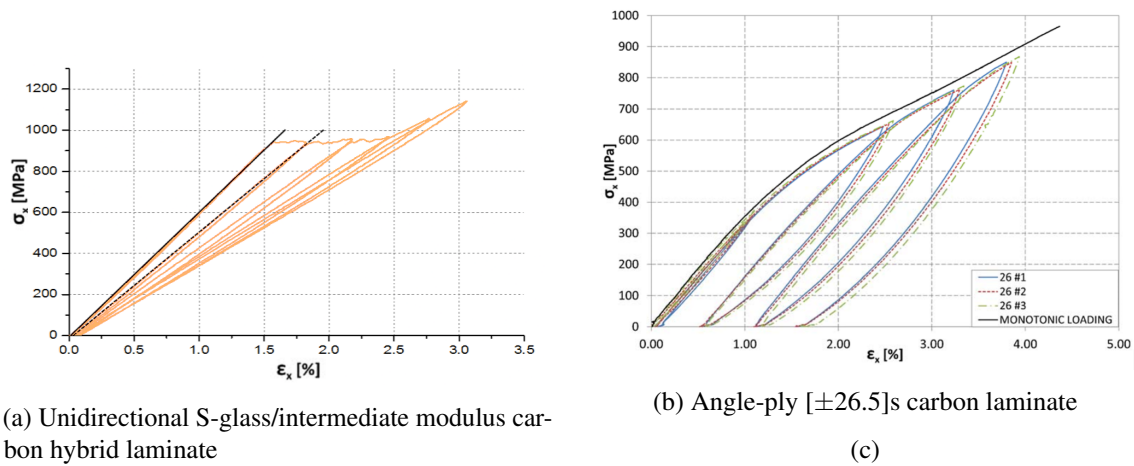


Figure 2.25: Comparison of loading-unloading-reloading responses, illustrating the difference between (a) the damage-based pseudo-ductility in UD hybrids and (b) the deformation-based ductility in angle-ply laminates (adapted from [222]).

Complementing these experimental findings, numerical modeling has been used to explain the specific shape of the tensile stress-strain curve. A parametric study by Sheibanian et al. [192] identified the Weibull modulus of the LE fibers as the parameter that determines the rising slope of the pseudo-ductile plateau, where a lower modulus signifying greater strength variability results in a steeper slope. This modeling work revealed that the overall mechanical response is largely independent of the specific spatial pattern of fragmentation sites, provided the statistical strength distribution is correctly represented [192]. Beyond tensile behavior, the fragmentation mechanism has also proven effective in compression and bending, yielding a non-linear response and preventing catastrophic failure in those loading scenarios [226].

The consequences of fragmentation and the statistical parameters that shape the response have been investigated in more detail. Cyclic loading tests clarified the post-damage behavior, revealing a permanent modulus loss upon unloading which distinguishes this response from true ductility, where permanent strain is retained instead of stiffness being reduced [222]. Complementing these experimental findings, numerical modeling has explained the specific shape of the tensile stress-strain curve. A parametric study by Sheibanian et al. [192] identified the Weibull modulus of the LE fibers as the parameter that determines the rising slope of the pseudo-ductile plateau, as illustrated in Figure 2.26. This modeling demonstrates that a lower modulus, which signifies greater strength variability, results in a steeper plateau slope. The work also revealed that the overall mechanical response is largely independent of the specific spatial pattern of fragmentation sites,

provided the statistical strength distribution is correctly represented [192]. Beyond tensile loading, the fragmentation mechanism has also proven effective in compression and bending, yielding a non-linear response and preventing catastrophic failure in those scenarios [226].

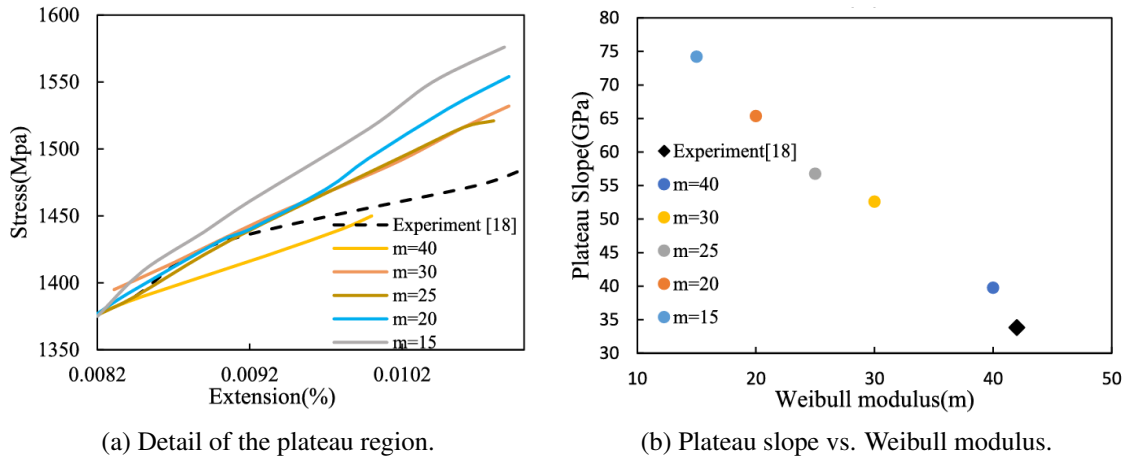


Figure 2.26: Numerical estimation of the effect of the Weibull modulus (m) on the stress-strain response of a T1000/M55 hybrid laminate, showing (a) a detailed view of the plateau region and (b) the direct relationship between the chosen Weibull modulus and the resulting plateau slope (reproduced from [192]).

2.5.2 Manufacturing Challenges

Despite the mechanical performance demonstrated by experiments and models [103, 105, 106, 201], the widespread adoption of thin-ply hybrid composites is constrained by manufacturing challenges. The difficulties begin with the raw material, as achieving consistent prepreg quality for low areal weight materials can introduce defects like fiber waviness and gaps, which in turn complicates processing. The manufacturing process is further constrained by the high number of plies required, potentially over six times that of a standard laminate, which increases layup time and elevates the risk of trapped air and higher void content that may require extra steps like intermediate debulking [135]. This sensitivity to prepreg quality is reflected in measured void contents, which have been reported to be nearly four times higher in laminates made from lower-quality thin plies (0.45%) compared to higher-quality ones (0.12%) [135].

In addition to these general challenges, creating intimate hybrid architectures such as intrayarn configurations introduces the difficulty of achieving a uniform micro-level mixture of different fiber types, a problem for which alternative methods like calendering are being explored [15]. While these materials demonstrate considerable performance potential, their practical application is dependent on addressing these production issues.

2.6 FRP Manufacturing

The mechanical performance of a fiber-reinforced polymer depends on the properties of its constituent fibers and matrix, but it is the manufacturing process that arranges these components into a final microstructure. While this process is designed to control desired features like fiber volume fraction and packing density, it can also introduce defects such as matrix micro-voids, fiber-matrix debonding, and matrix cracking [32, 52]. Among these defects, voids are a common issue that degrades component integrity, and as a review by Mehdikhani et al. [141] highlights, predicting this degradation requires not only to consider void content alone but also the shape, size, and location of the voids. This dependency on morphology is demonstrated in studies by Hamidi et al. [86], which show that consolidation pressure more readily eliminates circular voids between fiber tows than the irregular voids trapped within them. Ultimately, the presence of these defects negatively affects properties such as inter-laminar shear strength (ILSS), tensile strength, and fatigue life [141].

Beyond the process itself, the resin matrix selection – either thermoset or thermoplastic – influences both processing and handling characteristics. Thermoset resins, predominantly epoxy, undergo an irreversible curing reaction, yielding the high fatigue resistance and thermal stability required for performance-intensive applications. Thermoplastics, in contrast, can be reshaped upon reheating, making them suitable for high-throughput production cycles and recycling [130].

2.6.1 Overview of Manufacturing Processes

This overview of FRP manufacturing technologies is based on a review of information from [13, 130, 137, 146, 197], which categorizes the available processes into open-mold, closed-mold, and additive manufacturing. To illustrate these methods, all figures presented in this section are adapted from the review by Lunetto et al. [137].

2.6.1.1 Open-Mold and Bagging Processes

Open-mold processes utilize a single-sided tool and are often applied in prototyping and the production of large structures.

Hand Lay-up and Spray-up

For low-volume production, basic techniques include hand lay-up, where fiber reinforcements are placed into a mold by hand and saturated with resin, and the related spray-up process, where a mixture of chopped fibers and resin is sprayed onto the mold surface. Both methods are often consolidated with a hand roller to ensure good impregnation and remove air bubbles.

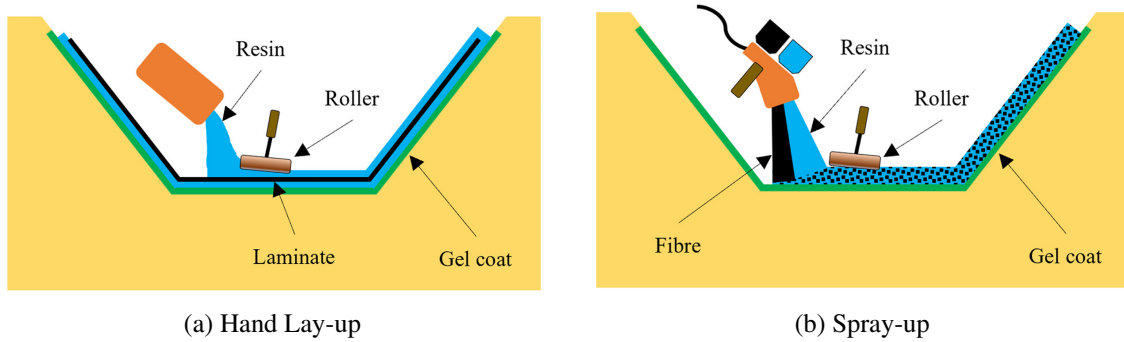


Figure 2.27: Diagrams of open-mold manual processes (reproduced from [137]).

The quality of parts made with these manual methods can be improved through vacuum bagging, where the laminate is sealed under a flexible bag and a vacuum is applied to draw out excess resin and trapped air for better consolidation and a higher fiber volume fraction.

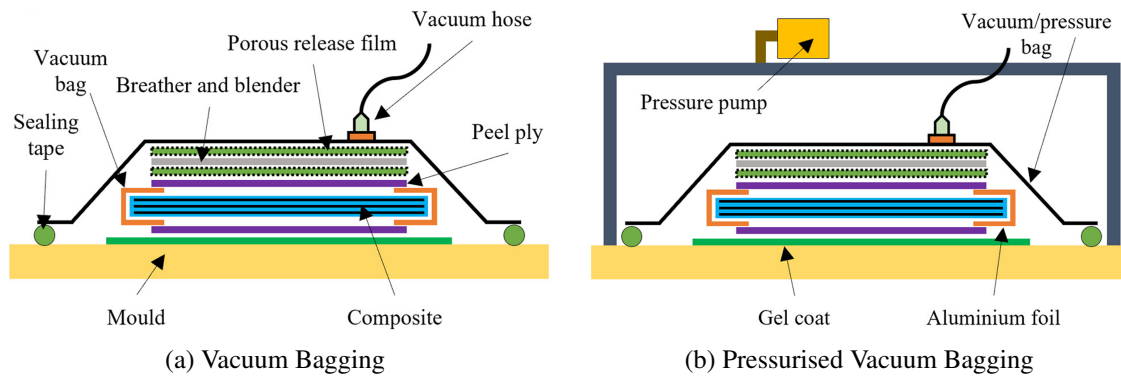


Figure 2.28: Diagram of consolidation methods (reproduced from [137]).

Pultrusion

Another specialized open-mold technique is pultrusion, a continuous process for manufacturing parts with a constant cross-section, such as beams or rods. Fibers are pulled through a resin bath for impregnation and then guided through a heated steel die that both shapes and cures the composite into its final, rigid profile.

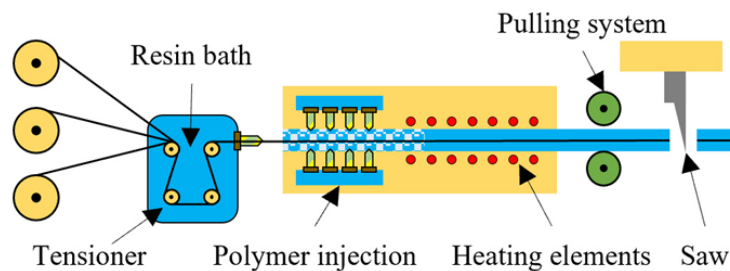


Figure 2.29: Diagram of the Pultrusion process (reproduced from [137]).

Filament Winding

Whereas pultrusion creates components with a constant linear profile, filament winding is a process used for creating cylindrical or hollow, axisymmetric structures. This process involves winding continuous, resin-impregnated fibers under tension onto a rotating mandrel in predetermined patterns to build the component layer by layer.

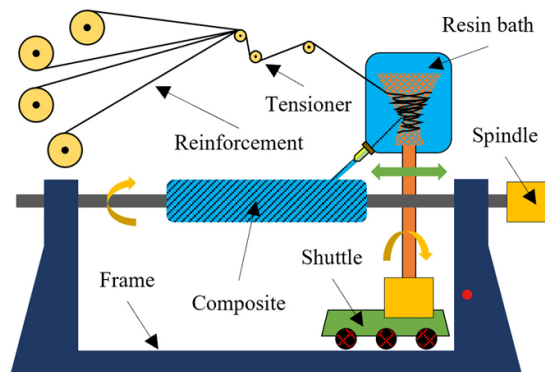


Figure 2.30: Diagram of the wet Filament Winding process (reproduced from [137]).

2.6.1.2 Closed-Mold Processes

For higher production volumes and better dimensional accuracy, closed-mold processes are used since they employ a matched toolset that defines all part surfaces.

Liquid Composite Molding (LCM)

The LCM family of processes involves injecting a liquid thermoset resin into a mold containing a dry fiber preform. In Resin Transfer Molding (RTM), resin is injected under pressure into a rigid, two-part mold, whereas in the Vacuum-Assisted RTM (VARTM) variation, a vacuum bag replaces the top mold half to help draw the resin through the fibers.

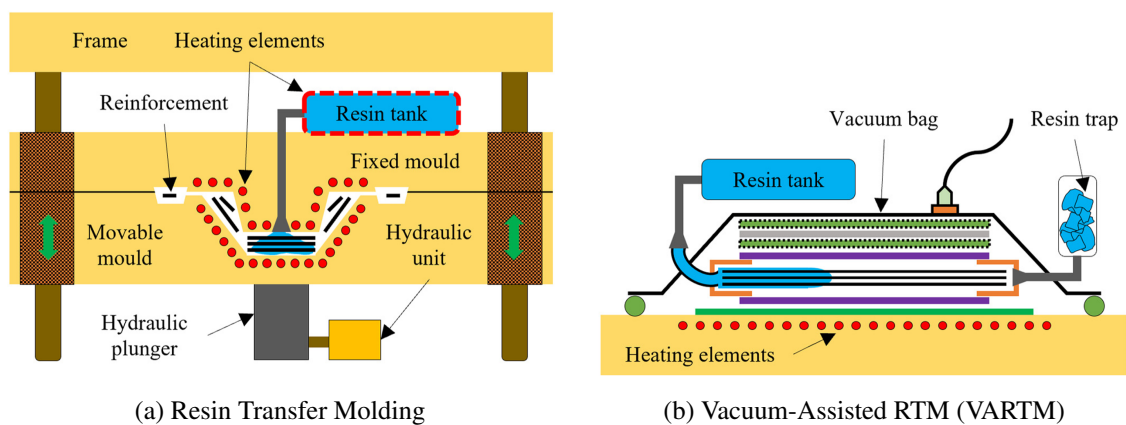


Figure 2.31: Diagrams of LCM methods (reproduced from [137]).

Variants of these infusion technologies include Vacuum Infusion Processing (VIP), Resin Film Infusion (RFI), and specialized high-performance Out-of-Autoclave (OoA) techniques like Same-Qualified-Resin-Transfer-Molding (SQRTM), which combines prepreg materials with liquid infusion to create aerospace-grade components without an autoclave.

Compression Molding

Different from LCM, which begins with dry fibers, compression molding uses pre-compounded materials like Sheet Molding Compound (SMC) or preforms placed in a heated mold. A press then applies high pressure and heat to shape and cure the part, while a variant known as cold press molding uses lower pressure with room-temperature curing resins.

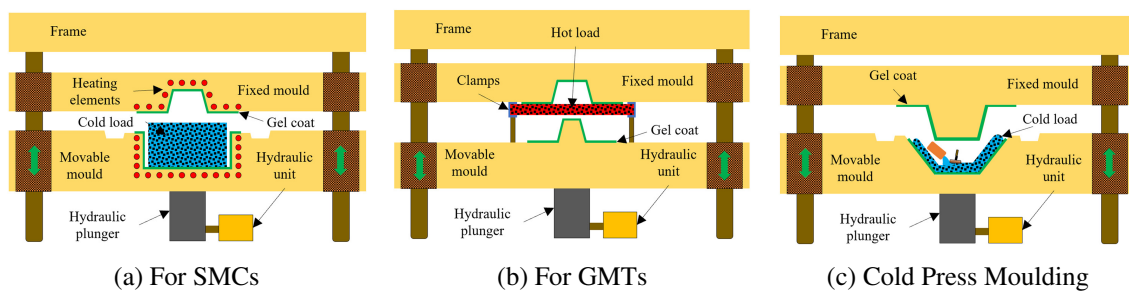


Figure 2.32: Diagrams of Compression Moulding processes (reproduced from [137]).

2.6.1.3 Additive Manufacturing

Additive manufacturing has recently emerged as a method for fabricating composites with highly complex geometries. Among the available techniques are powder bed processes, as illustrated in Figure 2.33. These methods build parts layer by layer from a powder mixture of polymer and short fibers, using either a laser for sintering (SLS) or a liquid binding agent (3DP).

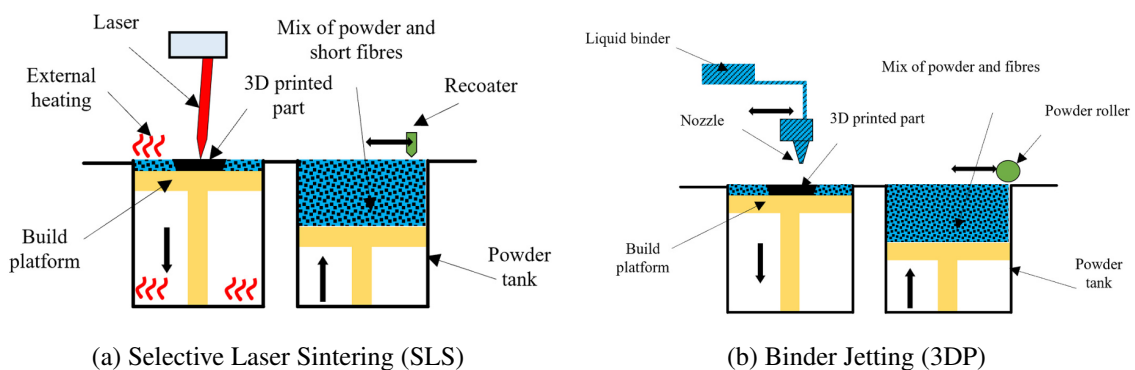


Figure 2.33: Diagrams of powder bed additive manufacturing processes (reproduced from [137]).

While these methods offer a wide range of capabilities, the highest mechanical performance is typically achieved using pre-impregnated materials (prepregs), which are processed using dedicated layup and curing techniques as detailed in the following section.

2.6.2 Prepreg technology

Prepregs are composite materials where reinforcing fibers are pre-impregnated with a partially cured polymer resin, creating a standardized intermediate building block for fabrication. This format removes the need for manual resin application and is supplied in two primary forms: unidirectional tapes that offer optimal mechanical performance by aligning fibers with load paths, and woven fabrics that provide superior drapability for forming complex geometries [13, 149]. To produce them, a common method is hot-melt impregnation, which involves pressing fibers and a resin film together under heat and pressure to achieve a consistent resin distribution. Another technique, solution dipping, passes fibers through a resin solution, followed by roller pressing to obtain a uniform resin content. The final product is typically wound into rolls and stored at low temperatures, often around 0°C, to slow the curing reaction and extend its shelf-life.

The use of prepregs allows for high control over resin-to-fiber ratios, an outcome difficult to obtain with wet lay-up techniques. This control leads to greater strength, uniformity, and repeatability while also reducing material waste. Prepregs therefore feature consistent resin distribution, low void content, and predictable mechanical properties, along with good drapability for forming complex geometries. Despite these benefits, prepregs also have limitations. They are generally more costly than their individual constituents and require refrigerated storage to maintain their viability. Furthermore, many prepregs are formulated for autoclave curing, a process that requires both high-pressure and high-temperature conditions [13].

The autoclave technique, illustrated in Figure 2.34, involves applying pressures between 3 – 7 bar and elevated temperatures from 120 – 180°C for epoxy systems within a specialized pressure vessel. Inside, the prepreg stack is covered with breather and bleeder cloths and sealed within a vacuum bag. This assembly allows the external pressure to compact the laminate, conform it to the tool surface, and collapse any voids that form during the cure [97]. The temperature cycle initiates the polymerization reaction, transforming the resin into a solid, cross-linked network. Consequently, components cured via autoclave typically exhibit high fiber volume fractions, minimal porosity, superior mechanical properties, and improved reproducibility [32, 63].

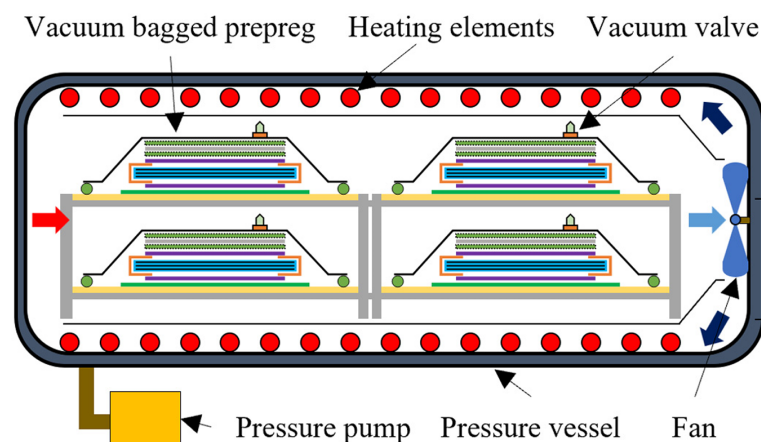


Figure 2.34: Diagram of the Autoclave moulding process for prepregs (reproduced from [137]).

To address this limitation, specialized Out-of-Autoclave systems, such as semi-prepregs, have been developed. These materials often consist of a resin film interleaved with dry fabric, a design that allows air to be evacuated through the dry fiber network before the resin flows, resulting in low-void laminates without the need for high autoclave pressure [63, 98, 139].

To improve production efficiency and part consistency over manual layup, automated technologies are used for prepreg processing. An advantage of combining prepregs with autoclave curing is the suitability of the layup process for automation, which ensures microstructural accuracy and consistency [32].

The two main technologies, illustrated in Figure 2.35, are Automated Tape Laying (ATL), which places wide prepreg tapes for large components with simple geometries, and Automated Fiber Placement (AFP), which uses a band of narrow, individually controlled tapes, or tows.

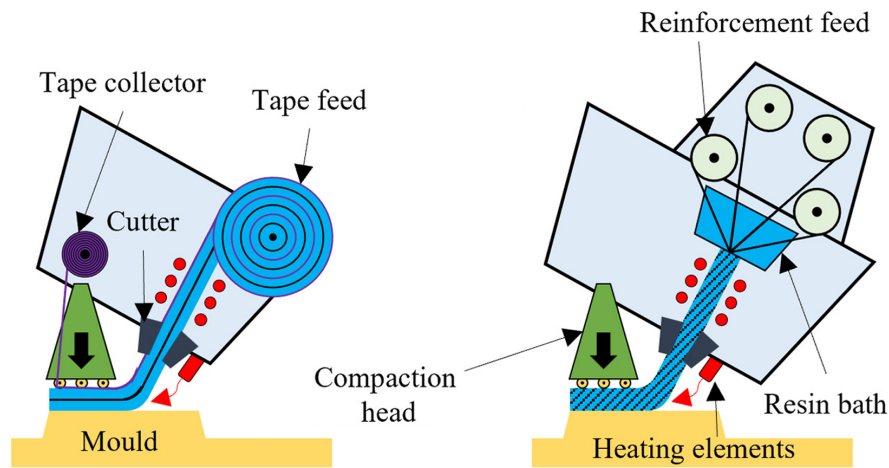


Figure 2.35: Diagram of (left) Automated Tape Laying (ATL) and (right) Automated Fiber Placement (AFP) (reproduced from [137]).

The use of narrow tows allows AFP to fabricate parts with complex curvature. For example, Smith et al. [194] experimentally demonstrated that $\frac{1}{8}$ -inch tows achieve a minimum defect-free steering radius less than half that of wider $\frac{1}{4}$ -inch tows. These automated processes can be supplemented by techniques like hot drape forming to shape the prepreg stack before the final cure. While automation improves repeatability, it also introduces its own characteristic manufacturing defects, such as gaps, overlaps, and wrinkles [136].

With this, prepregs have established themselves as the standard material for primary aircraft structures, exemplified by their extensive use in modern airframes such as the Airbus A350 and Boeing 787, where prepregs constitute over 50% of structural composition and their adoption has steadily increased across aerospace, wind energy, automotive, and other sectors [13].

2.6.3 Thin-Ply Prepregs Manufacturing

The manufacturing of thin-ply prepregs centers on tow spreading, a process that transforms standard high-count fiber tows (typically >12K filaments) into thinner, wider unidirectional tapes. This method can reduce the areal weight by up to 500%, allowing for the production of materials with areal weights as low as 30 g/m^2 , as used in some manufacturing trials [13, 87].

Industrial adoption of tow spreading has been growing since around 2006, supported by specialized equipment and improved processes, resulting in reduced costs and increased research interest [13]. Due to this, there are now many commercial solutions from a wide range of companies including North Thin Ply Technology (NTPT), Oxeon, Chomarat, Hexcel, Sakai Ovex, Mitsuya, Mitsubishi Rayon, TPCM, SK Chemicals, and Microtex, among others [32, 71].

The manufacturing line, shown schematically in Figure 2.36, is a multi-stage operation designed for continuous production. The process begins at a fiber feeding stage, where bobbins of fiber tows are loaded onto a creel stand. This unit is responsible for unwinding the tows and guiding them to subsequent stages with a consistent and controlled tension and flow rate, which is necessary to ensure proper fiber alignment in the final product. Following this, the tows enter the main spreading stage where they are de-compact and transformed into a flat, wide tape of individual filaments. The spread fibers then proceed to an impregnation stage, where a resin system, often in the form of a thin film, is applied and distributed evenly throughout the reinforcement with the assistance of rollers to achieve full saturation. To meet the width tolerances required for subsequent automated layup processes, the edges of the freshly made prepreg tape are often trimmed by rotary blades. Before the final take-up stage where the material is wound onto bobbins, it passes through an inline quality control station, typically a light table, for visual inspection to identify defects such as tow misalignments, width variations, or unimpregnated areas [13].

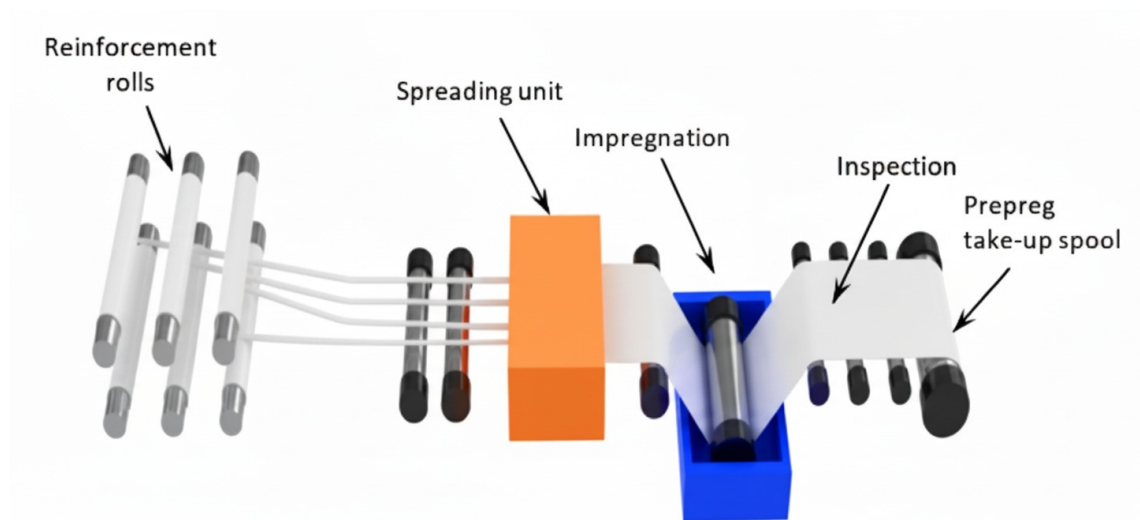


Figure 2.36: Schematic of a typical thin-ply prepreg manufacturing line (adapted from [13]).

The spreading methods themselves can be classified into two main categories: passive methods, which rely on tension and movement over guides, and active methods, which employ external

energy sources [13, 32]. A more detailed breakdown of these technologies is presented in Table 2.3. This overview is adapted from Broggi, who draws upon the comprehensive patent literature review by Irfan et al. [101] to categorize the specific methods and their characteristics.

Table 2.3: Overview of tow-spreading technologies, reproduced from Broggi [32] (the patent review was adjusted to consider patents after the year 2000).

Technology	Method	Description	Patents	References
Mechanical	Passive	Fibers are typically spread over 3 to 5 spreading bars, heated up to 150°C, and a pre-tension is applied up to 20N. Limited spreading capability and tend to damage the fibers.	[82, 119, 127, 132]	[101, 138]
Pneumatic-vacuum	Active	An airflow is used to separate and spread the tows. Increased spreading capability, but performances depend on fiber sizing. Fibers are less likely to be damaged.	[10, 114]	[39, 56, 116, 155, 193, 230]
Electrostatic	Active	Fibers are electrostatically charged to achieve spreading.	[153]	
Vibrations and acoustic	Hybrid	Similar to the mechanical methods with the addition of acoustic or mechanical vibrations improve the spreading.	[5, 206]	

The specifics of industrial spreading equipment are often proprietary. However, both Argypoulos [13] and Broggi [32], in a partnership between NTPT and the Laboratory for Processing of Advanced Composites (LPAC) at the Ecole Polytechnique Fédérale de Lausanne (EPFL), were able to share some insights into this manufacturing process. They highlighted that tow-spreading parameters must be correctly tuned for stable filament alignment, including tension, air velocity, and spreader geometry. If the tow is drawn too tightly through the airflow region, filaments may not separate thoroughly. Conversely, if it is too slack, the tape may not maintain a consistent width. Once the thin tapes are formed, they can be combined with thermoset or thermoplastic resins and partially consolidated into stable sheets before the final autoclave cure or infusion cycle.

Among the active methods, pneumatic techniques are common because they use an airflow or vacuum region to relax tension on the fibers, which allows filaments to separate with minimal friction and results in uniform ply quality [193] (as detailed in Section 2.2).

Because a significant number of plies are required to achieve a conventional laminate thickness, automation becomes essential for efficient production. However, applying standard automated fiber placement to thin-ply tapes presents distinct manufacturing challenges not found with standard thickness materials. As demonstrated by Harik et al. [88], process parameters such as heating, compaction force, and tow feed rate must be carefully re-optimized for the more delicate material.

The extreme thinness of the tapes makes the process highly sensitive to material inconsistencies, such as variations in tow width, which can lead to defects like gaps and overlaps. In some cases, modifications to the AFP hardware itself may be necessary to handle the material without causing damage like tow jamming or curling [88]. Following layup, a consolidation step using autoclave curing or vacuum-assisted resin infusion is employed to minimize voids and resin-rich pockets, which improves mechanical properties, particularly compressive strength [7].

2.6.4 Hybrid prepegs manufacturing

Hybrid prepreg manufacturing combines different fiber types to use their complementary properties, which modifies the mechanical characteristics and cost of the final composite. The hybridization of prepegs can be achieved at different scales, a factor that directly influences the manufacturing complexity, as illustrated in Figure 2.37.



Figure 2.37: Scales of composite hybridization.

The most direct methods involve arranging standard, single-material prepegs. Interlayer hybridization is achieved by stacking plies of different prepreg types, while intralayer hybridization involves alternating different prepreg tows within a single layer, a task often performed with Automated Fiber Placement (AFP). In contrast, intra yarn hybridization requires the prepreg itself to be hybrid, with dissimilar fibers co-mingled within each tow. This last approach presents a considerable manufacturing challenge, which consists in achieving a high level of fiber dispersion in a controlled and repeatable manner, especially when combining materials with dissimilar sizes or surface conditions. A review of existing techniques by Argyropoulos [13] indicates that many prior methods failed to produce a material that was simultaneously a thin-ply, a prepreg, and a well-dispersed hybrid, as these attempts were often limited by partial fiber mingling or required secondary processes like resin infusion [13].

One method developed to create these highly dispersed intra yarn prepegs is a calendering process, as illustrated in Figure 2.38. In this approach, two distinct single-material thin-ply prepegs are guided through a set of heated, counter-rotating rolls. The applied pressure forces the fibers from the separate prepegs to interpenetrate, creating a single co-mingled hybrid prepreg tape with a high level of fiber co-dispersion. Adjusting the calendering parameters directly, such as temperature

and pressure, impacts the microstructural and mechanical performance of the resulting composite materials [14].

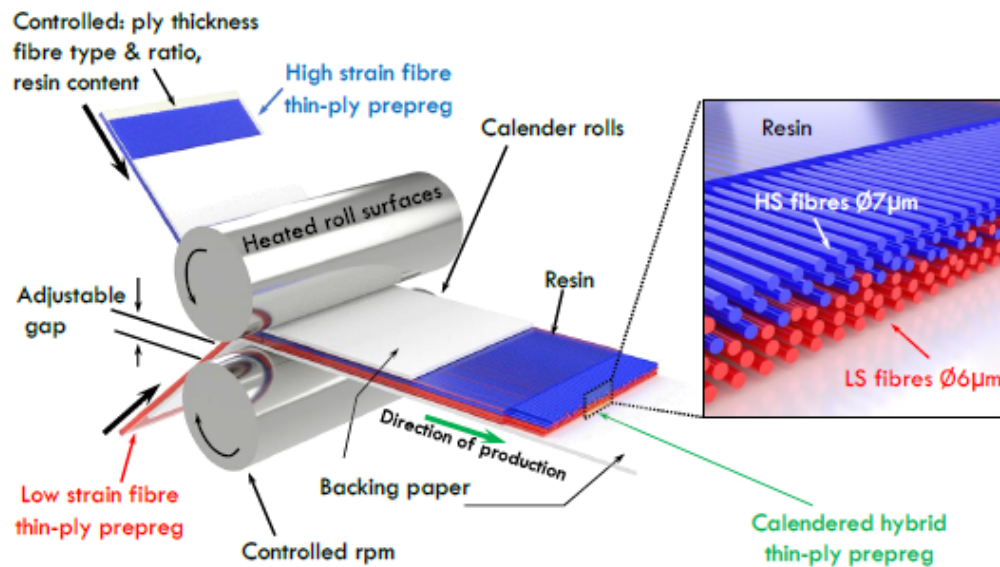


Figure 2.38: Diagram of a calendaring process for creating ply-level hybrid thin-ply prepregs (reproduced from [13]).

While this process allows for customized material properties, the scaling up of thin-ply production is more resource-intensive compared to standard methods and can prolong layup times without automation. Nevertheless, the resulting improvements often justify the complexity for aerospace, automotive, and other high-performance applications [7, 193].

2.6.5 Virtual manufacturing frameworks

The resulting microstructure from the manufacturing process shows that design approaches assuming an ideal, defect-free material do not accurately mirror reality. To address this gap, "virtual manufacturing" frameworks have emerged that incorporate the manufacturing process directly into the analysis. These frameworks use multiscale simulations to predict how the material evolves during the curing cycle by connecting phenomena across different length scales [97, 190]. They can capture the buildup of micro-residual stresses that appear from the interplay between chemical shrinkage and the thermal expansion mismatch of the constituents. These models demonstrate that significant internal stresses and damage, such as matrix cracking, can develop within the composite before it is ever subjected to an external load [50, 52]. This results in an as-manufactured composite with reduced transverse strength, a direct consequence of the micro-residual stresses [78]. Therefore, including the manufacturing process in simulations is a necessity for the correct design of composites, a task that in turn depends on robust methods for the analysis and characterization of the post-manufactured microstructure.

2.7 Microstructural analysis and characterization

Microstructural analysis serves to verify the integrity and quality of the manufactured composite laminates, especially hybrid composites which incorporate multiple fiber types within a common matrix. This includes evaluating aspects such as homogeneity, the quality of resin impregnation, fiber volume fraction, void content, the achieved degree of hybridization and the presence of any resin-rich or fiber-rich areas. Furthermore, it is used to examine the levels of fiber mixing achieved through various manufacturing processes and to assess the impact of production parameters on the final quality and microstructural features of hybrid thin-ply laminates.

While various advanced techniques such as scanning electron microscopy (SEM) and transmission electron microscopy (TEM) are available for this type of analysis, optical microscopy (OM) remains a prevalent choice. This preference for OM is based on its ease of use, accessibility, non-destructive nature, high-resolution imaging capabilities and the facility for real-time visualization. The utility of OM can be significantly amplified by automated image analysis techniques. These computational methods enable detailed characterization of microstructural features, including the determination of fiber volume fractions, fiber orientations and, more importantly, of the relationship between the spatial arrangement of fibers and the material's overall behavior and properties [13].

Logically, the success of the automated analysis depends on the images used, which in their turn depend on the sample used and the preparation routine selected. One of the most common routines used to obtain a sample cross-section involves wet cutting with water-cooled saws to prevent damage, followed by casting the samples in resin, subsequent multi-stage grinding using progressively finer abrasive papers (such as P400 down to P1200 silicon carbide paper) and fine polishing. It is often expected that polishing hybrid samples will demand longer durations compared to non-hybrid materials due to the differing properties of the constituent fibers. To further refine the analysis, specific image improvements like the removal of scratch marks, the digital separation of adjacent fibers that might be wrongly identified as a single entity and the application of edge-recognition algorithms can be employed to improve boundary detection and get accurate measurements.

2.7.1 Hybrid Parameters Calculation

As mentioned, due to the presence of multiple fiber types which can be difficult to distinguish if their physical characteristics are similar, robust quantitative characterization methods required where the calculations are made over multiple regions of interest within the composite sample defined according to the nominal fiber diameters.

To provide additional context for several of these methods, it is important to understand two geometric analysis techniques: Delaunay tessellation and Voronoi tessellation. These are concepts used to analyze spatial point patterns, such as the distribution of fiber centers in a composite cross-section. The Voronoi diagram partitions the plane into polygonal regions, where each region encloses a single point and consists of all locations closer to that point than to any other. The

Delaunay triangulation connects points whose Voronoi regions share a common edge, forming a network of triangles.

Figure 2.39 provides a visual comparison of these two methods applied to the same set of randomly distributed fiber centers.

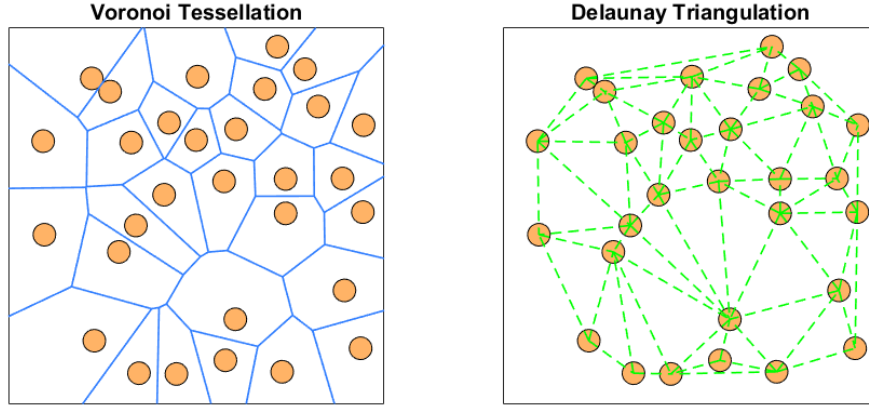


Figure 2.39: Visual comparison of Voronoi Tessellation (left) and Delaunay Triangulation (right)

An example of the application of these techniques is given by Katuim et al. [113] who utilized the Voronoi tessellation to evaluate fiber volume content and clustering in unidirectional CFRP tapes. The most common metrics derived from such analyses, along with visual representations of some of these concepts, are presented next.

2.7.1.1 Areal Disorder (AD)

This parameter quantifies the regularity of the fiber packing. It is determined using a Delaunay tessellation, as defined by Bray et al. [31]:

$$AD = \left(\frac{\bar{A}}{\sigma} + 1 \right)^{-1} \quad (2.7)$$

where $\bar{A}$ is the mean area of the tessellation elements and σ is the standard deviation of these areas.

A lower AD value represents a more regular and uniform fiber distribution conversely, a higher AD, suggests a more irregular arrangement. AD can be calculated globally for all fibers or for individual fiber types within the hybrid. Essentially, AD measures how much the local packing density varies from the average, based on the Delaunay triangle areas. A perfectly regular, crystal-like arrangement would have very consistent tessellation areas, leading to a low standard deviation, σ , as seen in Figure 2.40.

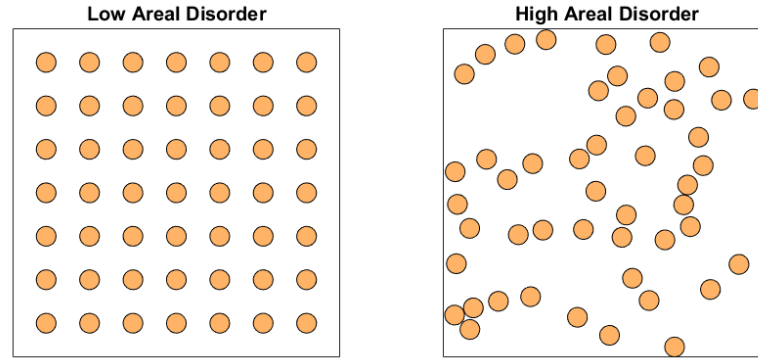


Figure 2.40: Microstructural parameter: Areal Disorder

2.7.1.2 Inter-Fiber Spacing (IFS)

The Inter-Fiber Spacing (IFS) quantifies the actual surface-to-surface distance between adjacent fibers. Similarly to the nearest neighbor distribution [166, 212, 216], the Delaunay triangulation, can also be used here to select the fiber pairs for IFS calculation. For any neighboring pair, i and j , with centers C_i and C_j and radii R_i and R_j respectively, the IFS is given by:

$$IFS_{ij} = d_{ij} - (R_i + R_j) \quad (2.8)$$

where d_{ij} represents the Euclidean distance between the centers of fiber i and fiber j .

$$d_{ij} = \sqrt{(x_i - x_j)^2 + (y_i - y_j)^2} \quad (2.9)$$

The resulting distribution of IFS_{ij} values – better visualized as a histogram – characterizes the spectrum of matrix gap sizes. Positive IFS values indicate the presence of the matrix between the fibers, while $IFS \leq 0$ suggests fiber-to-fiber contact and possible damage. This distribution is therefore useful when searching for local stress concentrations in the matrix, potential for crack initiation and resin flow paths.

The schematic representation can be seen in Figure 2.41.

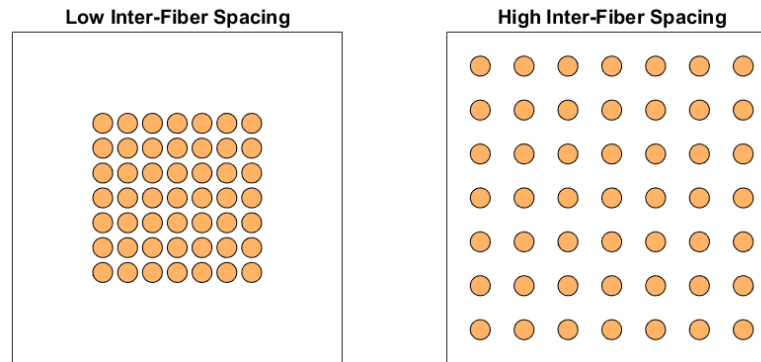


Figure 2.41: Microstructural parameter: Inter-Fiber Spacing

2.7.1.3 Local Fiber Volume Fraction ($V_{f,local}$)

The local fiber volume fraction, $V_{f,local}$, provides a measure of fiber concentration in the vicinity of each fiber or within small sub-regions of the composite. For a given fiber i , its local fiber volume fraction can be calculated differently.

One approach to assess local fiber arrangement involves the use of Voronoi cells associated with each fiber, a method detailed in the work by Pyrz [166]. This approach examines the areas of these cells, $A_{Voronoi_i}$, to describe the fiber distribution with each cell area reflecting the local fiber packing density.

Specifically, smaller cell areas indicate higher local fiber concentration, while larger areas signify sparser packing or regions with increased matrix content. The local fiber volume fraction is then conceptually understood by relating a fiber's cross-sectional area to the area of its corresponding Voronoi cell.

$$V_{f,local_i} = \frac{A_{fiber_i}}{A_{Voronoi_i}} \quad (2.10)$$

where A_{fiber_i} is the ratio of the cross-sectional area and $A_{Voronoi_i}$ is the corresponding Voronoi cell area.

Other explicit methods for determining local V_f are also described in the literature. Savvas et al. [181] detail a method for computing local volume fraction variation that uses a moving window technique. In this, the volume fraction is computed in each window using image analysis tools. Similarly, Mikkelsen et al. [145, 150] describe a dataset where local fiber volume fraction is determined within a moving averaging box, with the box size often scaled to the fiber diameter.

The statistical distribution of $V_{f,local}$ values quantifies the spatial heterogeneity of fiber concentration. A narrow distribution indicates a relatively homogeneous dispersion of fibers in terms of local density. Conversely, a broad distribution can signify the existence of either fiber-rich or resin-rich regions.

2.7.1.4 Dispersion

This metric measures how well different fiber types are mixed at a local level by considering the proximity of one fiber type to its neighbors of another type. Swolfs et al. [201] provided an early definition that related dispersion to the reciprocal of the smallest repeating unit in a patterned arrangement.

This parameter can be quantitatively calculated from:

$$\text{Dispersion} = \left\langle R_{LE_j} \left(\frac{\sum_{i=1..6} d_i(LE_j \rightarrow HE_i)}{6} \right)^{-1} \right\rangle_{\text{all } j} \quad (2.11)$$

Here, R_{LE_j} is the nominal radius of the j -th Low Elongation fiber (LE_j). The term $d_i(LE_j \rightarrow HE_i)$ represents the distance from this j -th Low Elongation fiber to the i -th of its six closest neighboring High Elongation fibers (HE_i). The denominator represents the average of these six

distances to the nearest HE neighbors. This entire ratio, calculated for each LE fiber j , is then averaged over all LE fibers in the sample, as indicated by the angle brackets $\langle \dots \rangle_{\text{all } j}$.

A high dispersion value means that, on average, fibers of one type are found very close to fibers of the other type, indicating good local intermingling. Conversely, a low dispersion value suggests that fibers tend to be surrounded by fibers of the same type, leading to larger segregated regions or clusters. This is conceptually shown in Figure 2.42.

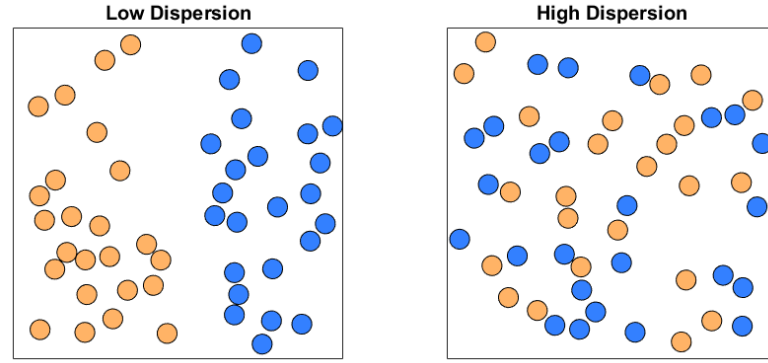


Figure 2.42: Microstructural parameter: Dispersion

The effectiveness of dispersion can also be influenced by the relative proportions of the fiber types for instance, if one fiber type forms very thick layers, the average distance from a fiber within that layer to the other fiber type might be larger, lowering the calculated dispersion for that dominant type. Calculations can be performed referencing either LS or HS fibers to understand effects in anisotropic arrangements [201].

2.7.1.5 Degree of Hybridization (H)

This parameter compares the local fiber composition of the actual microstructure to that of an idealized, completely random distribution of the same fibers. Diao et al. [57] introduced this term for glass-carbon fiber composites, assessing the extent of intermingling by comparing analyses from experimental micrographs with theoretical models of randomly distributed fibers.

The local Areal Ratio (AR) is first determined within a square moving window:

$$AR = \frac{N_{F_A} D_{F_A}^2}{N_{F_A} D_{F_A}^2 + N_{F_B} D_{F_B}^2} \quad (2.12)$$

where N_{F_A} and N_{F_B} are the numbers of fibers of type A and B within the window, and D_{F_A} and D_{F_B} are their nominal diameters.

The window size is typically chosen based on characteristic lengths like fiber diameters, with Diao et al. [57] recommending the size to be six times the nominal diameter.

The degree of hybridization, H , is then defined as the ratio of the coefficient of variation of the sample's AR distribution, CV_{Sample} , to that of a randomly generated microstructure, CV_{Random} , expressed as a percentage:

$$H = \frac{CV_{Sample}}{CV_{Random}} \times 100\% \quad (2.13)$$

The random microstructure serving as the baseline is chosen to have the same overall fiber type ratio, fiber number, and fiber positions (if on a lattice) or packing density (if randomly placed) as the experimental sample, but with a random assignment of fiber types.

In other words, H indicates the statistical randomness of the mixture. A higher H value (closer to 100%) indicates that the local compositional variation within the sample is very similar to that of a statistically random mixture. Therefore, a microstructure that is perfectly ordered and achieves excellent local mixing would exhibit high dispersion and low areal disorder, but it would have a low H . This is because its perfect order is distinctly non-random when compared to a statistically generated random baseline.

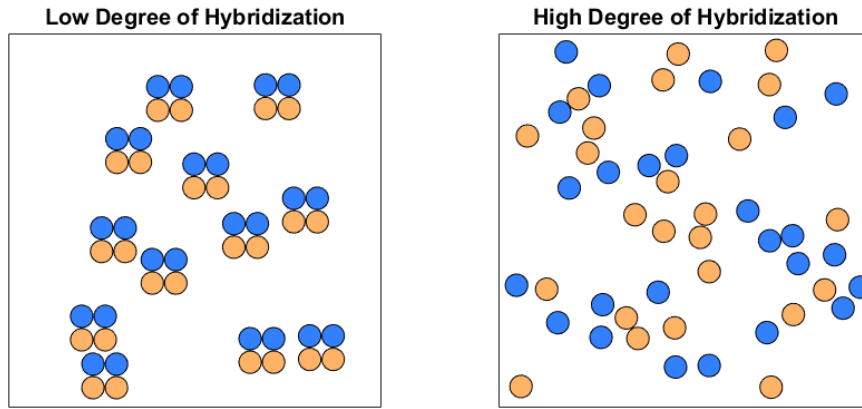


Figure 2.43: Microstructural parameter: Degree of Hybridization

In summary, these are complementary metrics and it is possible, for example, to achieve a low Areal Disorder, constant inter-fiber spacing, high Dispersion concurrently with a high Degree of Hybridization, as is conceptually illustrated in Figure 2.44.

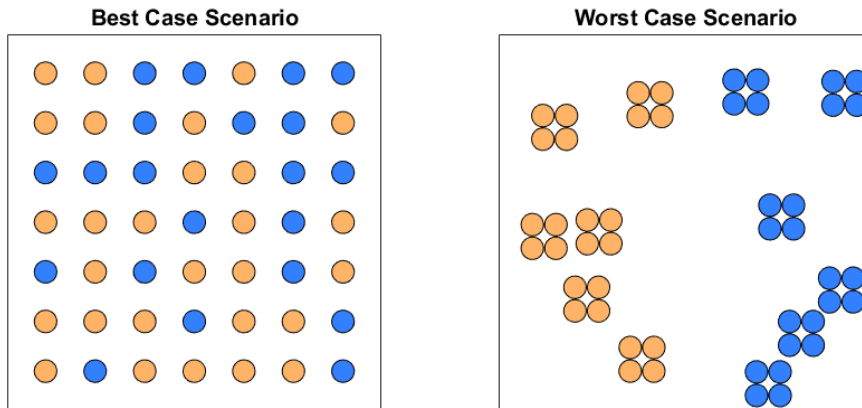


Figure 2.44: Visual illustration of the best and worst scenarios considering the microstructural parameters

2.8 X-ray Computed Tomography

The first application of computed tomography in the early 1970's has since evolved into a technique capable of resolving structural detail at sub-micron resolution in 3D with little to no alteration of the material under test. Previously, such data could only be acquired through the successive sectioning of a material, a destructive process that imposed a limit on the section thickness resolution [229].

While 2D microscopy is essential for characterizing the as-manufactured state, it is insufficient for understanding failure, which is a three-dimensional (3D) process. The complex, interacting failure modes that govern the tensile response of composite laminates – including splitting, fiber fragmentation, and delamination – involve initiation and propagation through the laminate thickness, phenomena that are obscured in 2D analysis. To overcome these limitations, X-ray CT has become the state-of-the-art method for this type of volumetric analysis. [229].

This section will first introduce the principles of the primary CT modalities before reviewing their application in the literature for the analysis of CFRPs.

2.8.1 Principles of X-ray Imaging Techniques

The core imaging modalities are distinguished by their X-ray source, which separates laboratory systems from synchrotron facilities (observable in Figure 2.45). This overview also covers computed laminography, practical data acquisition strategies, and common image artefacts.

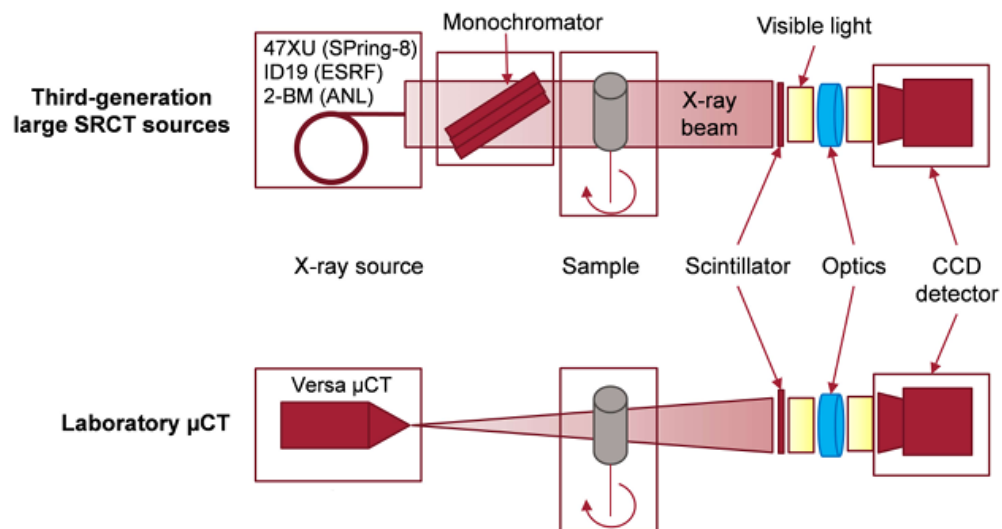


Figure 2.45: Illustrative comparison of synchrotron radiation (SR) and laboratory systems for X-ray CT scanning (adapted from [125]).

2.8.1.1 Laboratory-based X-ray Microscopy (μCT)

Laboratory-based X-ray microscopes operate on the principle of cone-beam computed tomography, a technique that reconstructs a 3D volume from a series of 2D projection images (radiographs). In

this setup, an X-ray source generates a diverging cone of photons that passes through a sample mounted on a high-precision rotation stage. As the sample rotates through a full 360° (or at least 180° plus the cone angle), a detector system – typically using a scintillator, which converts the transmitted X-rays into visible light, paired with a photodetector – captures hundreds or thousands of these radiographs from different angles [220, 229]. The complete workflow, from image acquisition through to quantitative analysis, is illustrated in Figure 2.46.

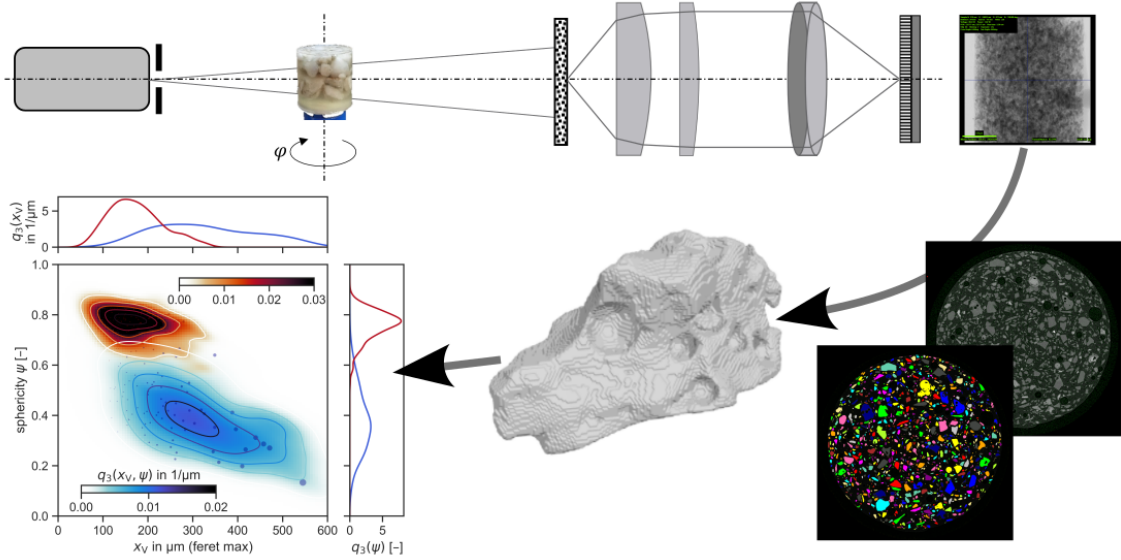


Figure 2.46: Exemplary workflow of a tomographic measurement (reproduced from [58]).

Each radiograph is essentially a map of X-ray attenuation, where the contrast is governed by the Beer-Lambert law:

$$I = I_0 e^{-\mu x} \quad (2.14)$$

Here, I is the transmitted intensity, I_0 is the initial intensity, and x is the path length through the material. The linear attenuation coefficient μ is a material property that describes the extent to which it absorbs X-rays. For the energies typically used in μ CT, the photoelectric effect dominates, and this coefficient is given by:

$$\mu = K \rho \frac{Z^4}{E^3} \quad (2.15)$$

where K is a constant, ρ is the material density, Z is the atomic number, and E is the energy of the incident photons [73, 229].

To create the final 3D volume, the series of 2D projections acquired during the rotation is computationally processed using a reconstruction algorithm, most commonly Filtered Back-Projection (FBP). The "filtering" step is a necessary part of this process because a simple back-projection of the attenuation data would result in a blurred image. This blurring is corrected by

applying a sharpening filter to the projection data beforehand, which yields a clear reconstruction of the object's internal structure [220, 229].

While laboratory μ CT systems can achieve sub-micron spatial resolutions, they are often limited by longer acquisition times and lower image contrast compared to synchrotron sources [73]. To mitigate the issue of low contrast, particularly for damage features like matrix cracks, *ex-situ* scans can be improved by applying a radio-opaque dye penetrant to the specimen before imaging [168].

2.8.1.2 Synchrotron Radiation Computed Tomography (SRCT)

SRCT is a powerful imaging technique, in development since the early 1980s [213], that relies on a large-scale facility known as a synchrotron to generate extremely intense X-ray beams. The generation process, detailed by Wright [229], is a multi-stage acceleration of electrons. The process begins with electrons being pre-accelerated in a linear accelerator (linac), after which they are injected into a booster ring to be brought to near the speed of light. Finally, these high-energy electrons are circulated in a large storage ring. The layout of a typical synchrotron facility, illustrating this multi-stage process, is shown in Figure 2.47.

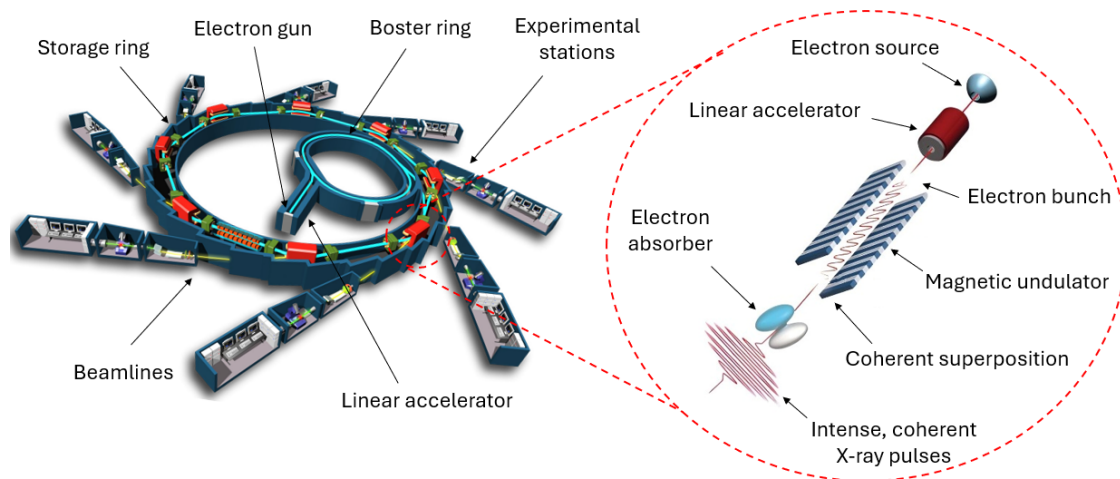


Figure 2.47: Schematic of a third-generation synchrotron radiation facility. The overall layout (left) shows the main components and the magnified view (right) illustrates the generation of X-rays within an insertion device (adapted from [111]).

The production of X-rays in these facilities relies on the precise manipulation of the electron path. Modern light sources are classified as third- or fourth-generation, distinguished by their increasing brightness and coherence. These facilities are built on one of two core architectures: circular storage rings (synchrotrons) or single-pass linear accelerators (XFELs). Table 2.4 provides a comparison of several of the world's largest facilities, ranked by their circumference or length.

Table 2.4: Comparison third- and fourth-generation light source facilities (adapted from [219]).

Facility Name	Type	Location	Country	Gen.	Energy (GeV)	Length/Circ. (m)	Commissioned
EuXFEL	Linear	Schenefeld	Germany	4th	17.5	3400	2017
LCLS	Linear	SLAC	US	4th	8	3000	2009*
PETRA III	Circular	DESY	Germany	3rd	6.0	2304	2009
PETRA II	Circular	DESY	Germany	3rd	12	2304	1995
SPring-8	Circular	RIKEN	Japan	3rd	8	1436	1997
APS	Circular	Argonne, IL	US	3rd	7.0	1104	1995
ESRF-EBS	Circular	Grenoble	France	4th	6	844	2020
ESRF	Circular	Grenoble	France	3rd	6	844	1992

*Commissioning date reflects the start of user operations; first light was achieved in April 2009.

The scale and technology of these facilities define their unique capabilities. The largest of these, the European X-ray Free Electron Laser (EuXFEL), operates as a 3.4 km long linear accelerator instead of a circular synchrotron [62]. This design allows it to generate up to 27000 X-ray flashes per second with a brilliance a billion times higher than that of conventional X-ray sources. Its circular counterpart in terms of scale is PETRA III at DESY, which, with a circumference of 2.3 km, is the world's largest storage ring light source. While PETRA III is notable for its size, the ESRF-EBS in France represents a significant technological leap as the world's first high-energy fourth-generation synchrotron. Through a revolutionary new storage ring design, it increased the brilliance and coherence of its X-ray beams by a factor of 100.

Although their designs vary, all such facilities employ a system of specialized magnets, collectively known as a magnetic lattice, to generate and control the beam for a given experiment type:

- **Bending magnets:** These force the electrons into a circular path. As the electrons are deflected by the Lorentz force, they emit synchrotron radiation tangentially to their trajectory.
- **Insertion devices (undulators and wigglers):** These are series of magnets placed in straight sections of the ring that cause the electrons to oscillate, generating far more powerful X-ray beams. Undulators are tuned to produce X-rays that interfere constructively for a high-intensity beam, while wigglers generate a wider spectrum of radiation.
- **Focussing magnets:** Also positioned in straight sections, these magnets are used to focus the electron beam, resulting in a brighter X-ray source.

This system of magnets produces an X-ray beam with the unique properties of high flux and coherence, which are discussed next.

High Flux and Monochromaticity

A synchrotron beam delivers a photon flux orders of magnitude higher than a laboratory source. This enables fast data acquisition, with a full tomogram collected in seconds or minutes rather than hours [73]. Recent advancements have made it possible to perform tens or even hundreds of consecutive scans during a single test, a technique termed "tomoscopy" [125]. This temporal resolution is fundamental for performing *in-situ* experiments, where damage can be observed and quantified in 3D as it evolves under mechanical load [148, 185]. For laboratory μ CT, the substantially longer scan times required for comparable resolution render such time-dependent studies impractical, in part due to material relaxation blurring the images [125, 156].

However, a potential challenge with high-flux synchrotron X-rays, particularly at lower energies (<30 keV), is the risk of inducing radiation damage in the polymer matrix. This can alter the material's mechanical properties and potentially influence damage mechanisms, a factor that should be considered when designing experiments and interpreting results [205].

High Spatial Coherence and Phase-Contrast Imaging

A significant advantage of SRCT for composite materials analysis is the high spatial coherence of its parallel beam, a property that enables Propagation-Based Phase-Contrast Imaging. In this technique, the coherent X-ray wavefront is distorted as it passes through the sample. By allowing the beam to propagate a sufficient distance (z , the propagation distance) to the detector, these phase shifts are converted into intensity variations through Fresnel diffraction phenomena [229]. The practical effect of this is the formation of a "bright fringe" around interfaces, which is particularly useful for identifying fine cracks [148]. The principle of this technique is illustrated in Figure 2.48.

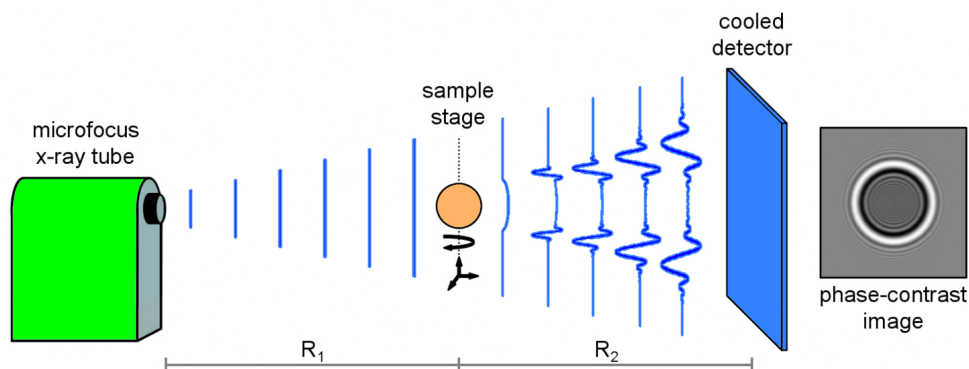


Figure 2.48: Schematic of an in-line phase-contrast imaging system. An X-ray beam propagates from a source to the sample and then to the detector. During this second propagation step, interference effects convert invisible phase shifts into measurable intensity fringes (adapted from [238]).

The resulting image is highly dependent on the propagation distance, creating distinct imaging regimes. These regimes, which are a function of the propagation distance (z), feature size (d), and X-ray wavelength (λ), are summarized in Table 2.5.

Table 2.5: Imaging regimes in propagation-based phase-contrast CT (adapted from [229]).

Regime	Condition (z)	Resultant Image Description
Contact	$z \approx 0$	Pure absorption image with no phase effects.
Near-field	$z < d^2/\lambda$	Strong interference fringes appear at interfaces (“edge-detection”).
Fresnel	$z \approx d^2/\lambda$	The image becomes a more complex diffraction pattern.
Fraunhofer	$z > d^2/\lambda$	The image is a Fourier transform of the object’s structure.

The *near-field regime* is particularly valuable for composite analysis, as the resulting edge enhancement allows for the visualization of features often invisible in absorption-contrast, such as fine matrix cracks and the precise tips of delaminations [73, 74, 202].

2.8.1.3 Computed Laminography

While conventional tomography is powerful, it is not well-suited for imaging objects with a high aspect ratio, such as plate-like composite specimens. The conventional method requires rotating the sample 180° or 360° around an axis perpendicular to the X-ray beam, which for a wide, flat plate, results in a very long attenuation path when the beam is parallel to the plate, leading to poor image quality [73].

Computed Laminography (CL) is a complementary technique designed specifically to overcome this limitation. Instead of a perpendicular rotation, CL rotates the sample around an axis inclined with respect to the beam, an arrangement illustrated in Figure 2.49. This ensures a consistent and relatively short X-ray path length through the sample’s thickness at all angles, enabling high-resolution 3D reconstruction of laterally extended objects. When this technique is implemented at a synchrotron facility, it is referred to as Synchrotron Radiation Computed Laminography (SRCL) [29, 73].

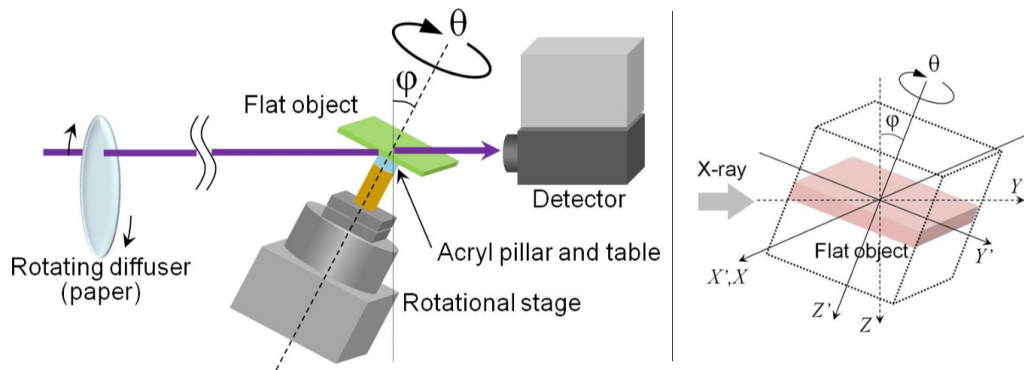


Figure 2.49: Schematic of the Computed Laminography (CL) setup. The flat object is rotated around an axis (θ) that is inclined by an angle (ϕ) with respect to the direction normal to the plate. This arrangement ensures a consistent and relatively short X-ray path length through the sample thickness (adapted from [95]).

2.8.1.4 In-situ vs. Ex-situ and Temporal Imaging Strategies

To study the evolution of damage over time, a time-lapse or 4D (3D + time) approach is employed. The primary methods are defined by the temporal relationship between the applied stimulus and scan acquisition, as illustrated in Figure 2.50. This can be conducted in two primary ways: *ex-situ* or *in-situ*. In an *ex-situ* analysis, a sample is scanned in a static state (e.g., post-mortem), often between stages of a mechanical test. Conversely, for an *in-situ* analysis, the sample is subjected to mechanical or environmental stimuli inside the scanner during imaging.

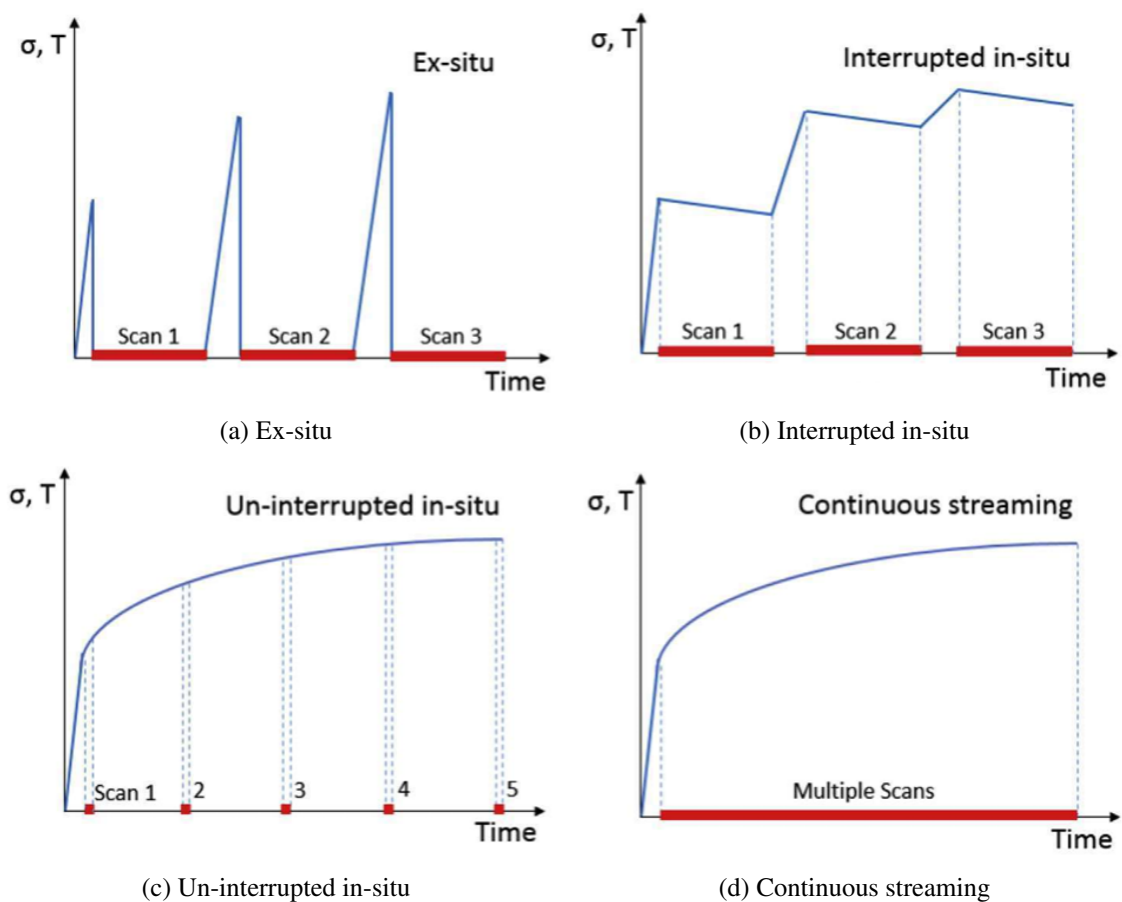


Figure 2.50: Comparison of four different CT-based approaches for scanning material coupon (adapted from [125]).

The *in-situ* approach itself can be further divided into two strategies. In an "interrupted" test, loading is paused to perform a scan at a static load step. In contrast, "continuous" imaging – a capability enabled by the high flux of synchrotrons – involves acquiring scans rapidly without pausing the experiment. For many damage mechanisms, *in-situ* testing is preferred because it avoids measurement artefacts such as crack closure, where damage features that are open under load may become invisible once the load is removed for an *ex-situ* scan [73].

2.8.1.5 Image Artefacts

While selecting the appropriate temporal strategy avoids measurement errors like crack closure, the final reconstructed CT images can also contain artefacts, which are distortions or features that do not correspond to the real structure of the object.

Motion artefacts, for instance, are a particular concern during the *in-situ* experiments just described, where any sample movement or deformation during a scan can degrade image quality. Wright [229] provides a comprehensive overview of the most common types, which include:

- **Beam-hardening:** This occurs with polychromatic lab-based sources. Lower-energy X-rays are absorbed more readily, causing the average energy of the beam to "harden" as it passes through the object. This differential hardening of the beam across various path lengths leads to characteristic distortions, such as streaking or a "cupping" effect, where the object's periphery appears brighter than its core. It is not an issue in monochromatic synchrotron sources.
- **Ring artefacts:** These appear as concentric rings centered on the axis of rotation and are typically caused by miscalibrated or defective pixels in the detector. During reconstruction, the persistent greyscale error from a single defective pixel is projected into a complete circle as the sample rotates relative to the detector.
- **Centring errors:** These artefacts are induced by a slight lateral drift of the specimen during its rotation, a problem that occurs when the specimen is not perfectly concentric with the axis of the rotation stage.
- **Streaking:** Beyond those caused by beam hardening, streaking artefacts can also result from full-scale X-ray attenuation, which occurs when an object is too dense for X-rays to penetrate. They also arise from edge gradient effects, where sharp interfaces between materials with large density differences cannot be accurately represented across adjacent detector pixels.
- **Motion artefacts:** Any movement of the sample during the scan, such as from mechanical instability or stress relaxation during *in-situ* loading, can cause blurring or streaking in the final image.

2.8.2 Literature Review on CT for CFRP Analysis

Over the past decades, X-ray CT has become an established and versatile tool for the analysis of CFRPs. Its applications are broad and can be categorized into three areas:

- Characterizing the as-manufactured microstructure (e.g., fiber orientation and void morphology) [140, 196].
- Generating high-fidelity geometries for image-based finite element modelling [189].
- Investigating material degradation and damage [72].

Of these areas, the application to damage analysis is particularly prominent, providing insights into failure under a wide range of conditions, from quasi-static tension and compression to dynamic fatigue and low-velocity impact [34, 73, 131]. These investigations now extend to exploring the unique failure mechanisms in advanced composite systems, such as thin-ply, hybrid, and notched laminates with complex reinforcement [124, 144, 157].

Historically, early applications of laboratory-based μ CT explored a range of polymer and composite systems. For instance, initial studies focused on materials like elastomeric matrix composites, with Bayraktar et al. [27] conducting *in-situ* fracture observations. The technique was quickly applied to fiber-reinforced polymers, with Schilling et al. [184] characterizing delamination and matrix cracking in glass and carbon fiber composites, and Symons [204] employing μ CT to characterize impact damage in CFRP laminates [229].

While numerous studies have utilized laboratory μ CT, a significant breakthrough for understanding damage evolution came with the application of *in-situ* Synchrotron Radiation Computed Tomography. This technique now represents the state-of-the-art analysis and is therefore the focus of the more recent literature discussed here.

One of the first studies to successfully apply this method was by Moffat et al. [148], who used "interrupted" *in-situ* SRCT on a notched $[90/0]_S$ laminate. Using phase-contrast imaging, they provided direct 3D observations of 0° split cracks originating from the notch root, and demonstrated the ability to quantify damage from measurements of crack opening displacement (COD) and crack shear displacement (CSD). Following this, the work of Scott et al. [188] provided direct experimental data for validating micromechanical strength models. They were able to observe and quantify the accumulation of individual fiber fractures in 3D deep within a commercial laminate as it was loaded, providing the data needed to validate the statistical failure theories that had, until then, been largely theoretical. This line of investigation was then extended to fatigue loading by Garcea et al. [74], who provided detailed 3D evidence of the role of toughening particles, identifying the mechanism of fatigue crack growth as the degradation of bridging ligaments formed in the crack wake. In turn, this fatigue analysis has since been extended to more complex architectures, such as 3D woven composites, often using an interrupted *ex-situ* approach to track damage accumulation over many cycles [83].

The ability to directly observe and quantify individual fiber fractures enabled a new wave of research focused on the underlying micromechanics of the fragmentation process. For instance, Swolfs et al. [202] used *in-situ* SRCT to provide quantitative data on fiber break density specifically for validating computational models. This was advanced by Rosini et al. [173], who conducted *in-situ* SRCT tests to failure on notched $[90_2/0_2]_S$ laminates, developing an automated image analysis framework to track thousands of individual fibers and link initial break locations to local morphology. Subsequently, the use of Digital Volume Correlation (DVC) with *in-situ* SRCT has been established as a powerful method to quantify the full-field strain redistribution around fiber breaks [133]. This DVC technique was applied to unidirectional composites by Schöberl et al. [185], who used BaTiO_3 nanoparticle markers to directly measure the strain recovery length in broken fibers, an important parameter in all micromechanical strength models [220].

Beyond tracking single fibers, CT is used to study the complex 3D interaction between different damage modes with the work of Bull et al. [33] on impact damage illustrating this. They combined laboratory μ CT to map the overall damage with high-resolution SRCT to resolve local details, revealing an interconnected network where delaminations formed a "spiral staircase" linking matrix cracks in adjacent plies. Similarly, SRCT has resolved the distinct micromechanisms of interlaminar fracture, such as fiber bridging and particle debonding, under both Mode I and Mode II loading [29].

Pushing the resolution further, nanoscopic SRCT has been used to uncover the fundamental origins of crack initiation. Watanabe et al. [218] tested specimens under opening-mode (Mode I) loading and concluded that crack initiation is a competition between two mechanisms: debonding at the fiber-matrix interface versus cracking within the resin itself. They found that the local resin pocket thickness was the deciding factor with "thin" resin regions promoting interface debonding, while "thick" regions favored in-resin cracking. This was extended by Kimura et al. [118], who tested specimens under mixed-mode loading. They concluded that damage progresses through a distinct sequence: it begins with interfacial debonding, which then leads to the formation of microcracks within the resin.

Recent research has focused specifically on the complex failure of thin-ply and hybrid composite systems. At the coupon level, investigations using laboratory CT have mapped the influence of ply thickness on the suppression of transverse cracking in thin-ply laminates [168]. At a higher resolution, in-situ SRCT has provided direct micromechanical insights. The work by Ni et al. [157], for instance, provided quantitative evidence of the damage suppression effect, showing that thin-ply laminates developed 6.5 times less damage surface area than standard-ply equivalents. Similarly, SRCT has been used to characterize the micromechanisms of the hybrid effect in notched carbon-glass specimens [144]. Methodologically, the field has also advanced to better capture these phenomena. Kopp et al. [124] implemented new canted loading rigs to improve 3D imaging, and AhmadvashAghbash et al. [3] developed new specimen geometries to enable detailed studies of translaminar fracture.

While these imaging techniques provide 3D data, a bottleneck remains in the analysis of the datasets they generate. The challenge arises because the datasets are on the order of 10 GB/mm^3 , yet the damage features of interest are sparse, occupying less than 1% of the scan volume, and are obscured by imaging artifacts [125]. The manual or semi-automated analysis of 4D data is time-consuming, often taking months or even years for a single experiment [156]. To address this, emerging research has applied deep learning (DL) to automate the segmentation of damage features and reduce the analysis time. A notable example is the work presented in the doctoral thesis of Kopp [125] (and associated journal publication [123]), which established a framework for this purpose while also investigating damage in composites. As artificial intelligence is expected to play an increasing role in materials characterization, such automated methods represent the future of the field, though they fall outside the scope of the current work.

Chapter 3

Experimental Investigation of Co-mingled Hybrid Yarn Production

This chapter details an experimental campaign that focused on manufacturing hybrid composites at the filament level. Although the primary objective was to produce a fiber-by-fiber dispersed hybrid yarn by co-mingling two types of high-performance carbon fibers, T800 and HR40, the experimental plan was adapted due to limitations in material supply. Instead, a single type of carbon fiber was processed to evaluate the co-mingling process. Using a single fiber type presents a limitation because it does not permit the study of interactions between dissimilar materials. Conversely, this approach offers the advantage of isolating the effects of the manufacturing process on the fibers. It allows for a direct assessment of the process's capability to open and intermingle carbon fiber tows and to quantify any process-induced damage without the variable of material incompatibility. The chapter begins by reviewing the principles governing fiber co-mingling, describes the experimental setup, presents the results, and discusses the implications of the outcome for alternative hybridization strategies.

3.1 Manufacturing Principles of Fiber Co-mingling

Fiber co-mingling is a textile-based manufacturing process designed to blend at least two different types of continuous filament fibers into a single, homogeneous yarn. This technique positions the constituent materials in close proximity throughout the yarn's cross-section, creating a uniform intermediate material for use in subsequent textile processes such as weaving, knitting, or braiding [121, 195]. While the technique was historically explored for combining reinforcement fibers like carbon, glass, or aramid with thermoplastic matrix fibers such as Polypropylene (PP), Polyamide (PA), or PEEK [121, 195], the method is equally applicable to creating hybrid reinforcements by mixing different structural fibers.

Co-mingling presents a method to achieve hybrid composites at a fiber-by-fiber level, a dispersion that micromechanical models suggest leads to more pronounced hybrid effects in the final composite's performance. This approach contrasts with manufacturing methods for thin-ply

hybrid prepregs that involve simultaneously spreading separate high-count fiber tows before resin impregnation [13]. Achieving a sufficient fiber-by-fiber mixture with simultaneous spreading at an industrial scale presents challenges, as Argyropoulos [13] highlighted that the process can be unreliable due to difficulties in the controlled mixing of dissimilar fibers. An alternative manufacturing route is to first produce a co-mingled hybrid yarn, which is then spread into a thin, wide tape and subsequently impregnated with a thermoset resin. Although this method increases the number of manufacturing operations, it may improve the quality of the final hybridized thin-ply prepreg by separating the mixing and spreading stages.

3.1.1 Air-Jet Co-mingling: Process and Equipment

Co-mingled yarns can be produced through numerous textile technologies, including online, ring, and wrap spinning, as well as braiding, parallel winding, and film stacking [195]. Among these options, air-jet texturing is a well-established method for this purpose. The technique, also known as intermingling, utilizes a turbulent air jet to blend dissimilar filaments, achieving a homogeneous dispersion of fibers throughout the yarn cross-section [195]. Its inherent flexibility allows for the production of hybrid yarns from numerous material combinations, such as those involving carbon and various thermoplastic fibers. This flexibility extends to producing more complex, three-component yarns, where two distinct reinforcement fiber types are co-mingled with a single thermoplastic matrix [90].

A typical air-jet co-mingling machine comprises several components:

- **Feeder Creels and Yarn Guides:** Bobbins of the individual filament yarns are mounted on creels, from which the yarns are guided towards the processing zone.
- **Delivery Godets (Feed Rollers):** The constituent filament yarns are fed into the machine by separate pairs of godets that control the speed and tension of each yarn component. These systems, often using individually motor-driven rollers, can provide stress-free yarn feeding to eliminate slippage and maintain consistent yarn evenness. When processing stiff filaments such as stainless steel that are prone to slippage, modifications like applying a polyolefin shrink sleeve to the godet surface have been used to improve grip [1].
- **Spreading Bars/Elements (Optional):** Before entering the air-jet nozzle, reinforcement fibers may pass over spreading bars or similar elements that help to open the filament bundle and, in the case of reinforcement fibers, break its sizing, preparing the filaments for more effective mixing [121, 195].
- **Heating Elements (Optional):** Some setups include heating elements, such as infrared heaters or specialized heated godet rolls, to pre-heat the fibers and achieve precise temperature control [40, 112].
- **Yarn Wetting Device (Optional):** Certain machine configurations may include a device for yarn wetting prior to texturing [40].

- **Air-Jet Nozzle and Mixing Box:** This is the core of the process, where the different yarns converge and pass through an air-jet nozzle housed within a mixing or sound box. A turbulent flow field generated by compressed air creates entanglements known as "nips" that provide coherence to the yarn [121, 182, 195]. The nozzle's design is a primary factor in the process's effectiveness and is discussed in more detail below.
- **Take-Off Godet (Drawing Roller):** After exiting the air-jet nozzle, the newly formed co-mingled hybrid yarn is drawn by a take-off godet or roller system.
- **Winding Unit:** Co-mingled yarn is wound onto a bobbin or spool.

3.1.1.1 Nozzle Design and Optimization

The geometry of the commingling nozzle determines the mixing effectiveness and the degree of potential fiber damage. A fundamental design distinction exists between texturing nozzles, where airflow is generally parallel to the yarn path, and intermingling nozzles, where airflow acts perpendicular to the filament bundles. Investigations by Choi et al. [40] showed that while intermingling nozzles produced a more homogeneous filament distribution, texturing nozzles were gentler on brittle carbon fibers.

Given the complexity of the internal airflow, Computational Fluid Dynamics (CFD) is a common tool for analyzing and optimizing nozzle geometries [217]. This optimization process involves simulating the effects of geometric parameters, such as the position, diameter, and angle of internal air channels, on the turbulent flow field. The simulation results are often validated experimentally with high-speed imaging to correlate the simulated flow with observed filament behavior. This development can lead to modular nozzle designs with distinct components like a rebound plate and a middle plate containing the flow channels [217].

Furthermore, the development of nozzles for new material combinations presents specific challenges. Conventional nozzle designs have been found to be unsuitable for processing dense and bend-resistant metal filaments [1]. To address this, novel nozzles have been engineered with geometries that generate airflow parallel to the fibers upon entry and maximize flow velocity at the outlet, which helps to manage the high stiffness of metal filaments during mixing [1].

3.1.2 Process parameters

The final structure and quality of the co-mingled yarn are governed by several process parameters that influence filament distribution and blending quality. Air pressure is a dominant factor, as higher pressures generally improve filament opening and blending, though excessively high pressure can damage brittle reinforcement fibers such as carbon [40, 182, 195]. For instance, Schäfer et al. [182] found that increasing air pressure from 5 to 6 bar for carbon/PPS yarns improved blending but also noted that high pressures can cause breakage [182]. Similarly, Choi et al. [40] observed that decreasing air pressure significantly reduced damage to carbon filaments during CF/PEEK

co-mingling [40]. Therefore, an optimal air pressure, typically reported between 3 to 6 bar (0.3 to 0.6 MPa), is normally selected based on the specific fibers being processed.

A principle of this process is the application of overfeed, which is achieved by running the delivery godets slightly faster than the take-off godet. This practice reduces filament tension within the nozzle, allowing the fibers to move freely and mix effectively [112, 195]. The overfeed ratios applied to each yarn component directly influence filament entanglement, with studies reporting values of 2-3% for carbon fibers and 3.5-7.5% for thermoplastic filaments like PEEK or PA 6,6, depending on the desired yarn structure [40, 112]. For multi-component systems, a strategy of differential overfeed can be applied, where higher overfeed rates are used for more elastic filaments to compensate for their elongation and ensure an even mixture with stiffer components [90]. Finally, the production speed affects the residence time of the filaments in the mixing zone and involves a trade-off between mixing quality and manufacturing productivity. While lower speeds generally improve mixing, common production speeds for dissimilar material systems range from 50 to 100 m/min, with examples including a take-up speed of 50 m/min used by Choi and a delivery speed of 90 m/min used by Kaplan [40, 112]. However, significantly higher production speeds of up to 200 m/min have also been successfully utilized for other material combinations [120]. Beyond these operational settings, factors such as the nozzle design and the properties of the raw materials, including filament denier, shape, and rigidity, also substantially influence the final yarn quality [182, 195].

The interplay of these parameters directly influences the migration of filaments within the yarn. For instance, in a carbon/PPS system, an increase in air pressure was found to cause the more flexible PPS filaments to migrate toward the yarn's core, while a combination of lower production speed and higher pressure produced the most effective overall blend [182].

3.1.3 Yarn Quality and Characterization

The quality of the co-mingled yarn is assessed by the homogeneity of fiber distribution, the degree of intermingling, and the extent of fiber damage. A primary technique for evaluating the yarn's internal structure involves microscopic analysis. This is typically done by embedding yarn samples in an epoxy resin, preparing a polished cross-section, and using optical microscopy combined with image processing software to map the positions of the individual filaments [182]. In addition to direct visual assessment, yarn quality is evaluated by characterizing its mechanical properties to assess process-induced damage. This includes tensile testing of the co-mingled yarn itself, as well as manufacturing and testing composite specimens to determine how the yarn's structure translates to final part performance.

From the filament position data obtained through microscopy, quantitative metrics can be calculated to describe the blending quality. For example, Schäfer et al. [182] utilized a radial distribution index (R_d) to indicate whether a fiber component (A) tends to migrate towards the center ($R_d < 1$) or the surface ($R_d > 1$) relative to another component (B). It is calculated as the ratio of the average displacement of the two components from the yarn's geometric center:

$$R_d = \frac{R_A}{R_B} \quad (3.1)$$

To measure the overall uniformity of the component distribution, a lateral distribution index (L_t) was also used. A value of $L_t = 0$ indicates a perfectly uniform distribution, where the center of gravity for a fiber component (C_{xA}, C_{yA}) coincides with the overall yarn's center of gravity (C_x, C_y). The index is normalized by the average displacement of all filaments from the center, R_a :

$$L_t = \frac{\sqrt{(C_{xA} - C_x)^2 + (C_{yA} - C_y)^2}}{R_a} \quad (3.2)$$

A challenge in co-mingling is to achieve this homogeneous distribution while minimizing damage. A poor distribution can lead to high void content and degraded mechanical properties in the final composite part. As illustrated in Figure 3.1, a non-uniform distribution can manifest through the compaction of reinforcement fibers into a core-sheath structure, a side-by-side positioning of the components, or the migration of thermoplastic filaments into the core, all of which contrast with the desired uniform blend of materials [195].

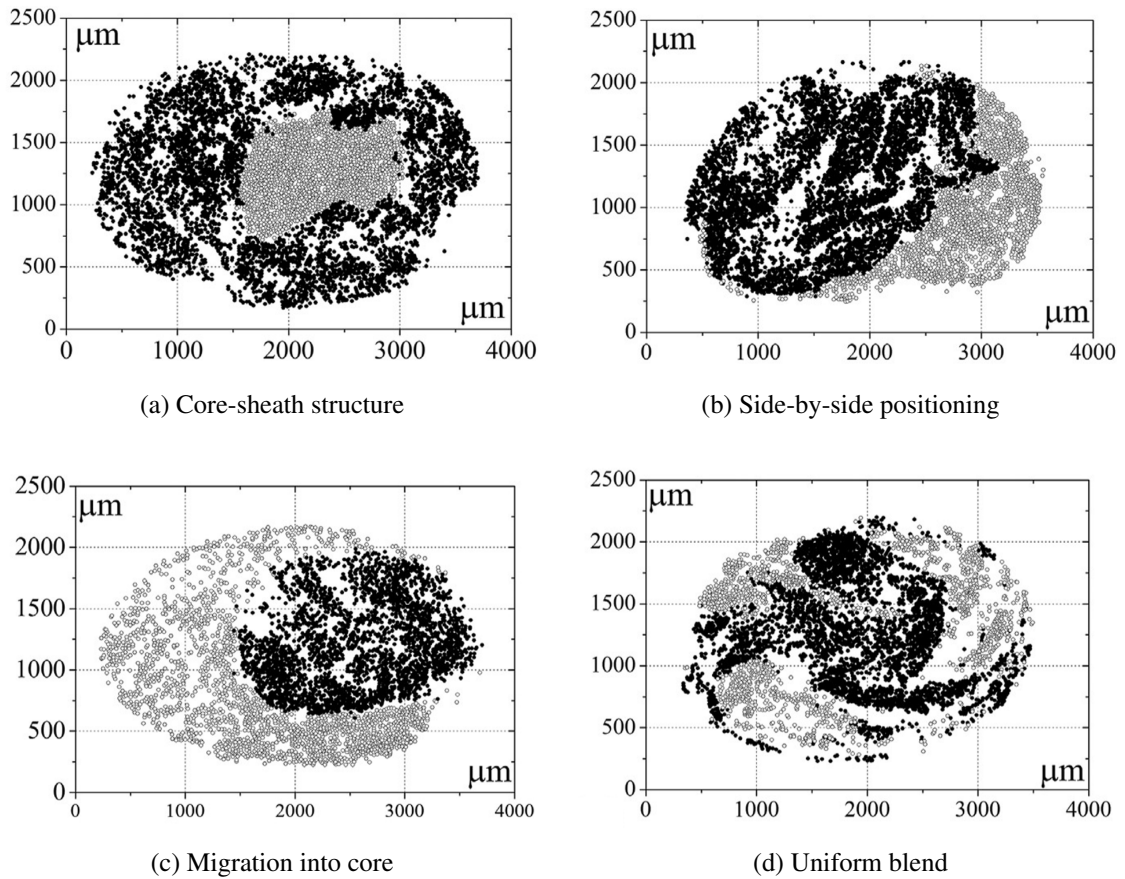


Figure 3.1: Examples of filament distribution in a co-mingled yarn cross-section. White circles represent reinforcement fibers and black circles represent thermoplastic fibers (adapted from [195]).

The brittle nature of carbon fibers makes them susceptible to breakage and surface abrasion

from the mechanical action of the air jet [40, 195]. Early investigations by Choi [40] into co-mingling carbon fibers with polyetheretherketone (PEEK) filaments highlighted difficulties in achieving complete intermingling due to the contrasting stress-strain behaviors of the stiff carbon and flexible PEEK fibers and the tendency for sized carbon filaments to adhere to each other. That work demonstrated that to protect the integrity of the reinforcement fibers, it is necessary to control the process parameters, finding for example that decreasing air pressure was necessary to reduce damage to the carbon filaments [40]. However, it was also noted that even with air-jet texturing of CF/PEEK, complete mixing was not achieved with their settings.

3.2 Experimental Setup

The experimental campaign was conducted on a Retech air-jet co-mingling machine, a model typically used in the textile industry for processing polymer yarns rather than high-performance carbon fibers. Figure 3.2 shows the experimental arrangement, which consists of the co-mingling machine on the left and a separate winding unit on the right.



Figure 3.2: The Retech air-jet co-mingling experimental setup.

The machine is equipped with five independent godet sets that were configured for processing two input tows. The feed side was arranged into two parallel paths, one for each incoming tow, where each path utilized two godet sets. In each path, the first godet set drew fiber from its supply spool while the second controlled the feed rate and overfeed into the air-jet nozzle. A final, fifth godet set served as the take-off unit, pulling the co-mingled yarn from the nozzle and directing it through a system of pulleys towards the automated winder. The machine's technical specifications are presented in Table 3.1.

Table 3.1: Technical Specifications of the Retech Co-mingling Machine.

Parameter	Specification
Manufacturer	Retech AG, Switzerland
Voltage	400Y / 230V AC
Phase	3
Frequency	50 Hz
Total Load Current	32 A
Largest Motor Current	4.3 A
Short Circuit Current Rating (SCCR)	10 kA

The experimental setup required modifications to accommodate the oversized carbon fiber spools, which did not fit the machine's standard vertical creel system. An alternative feed system was devised, using a custom-built horizontal stand to hold the input spools, as shown in Figure 3.3b. This stand featured an axle that rotated on two bearings with a clamp system to secure the fiber spool. Although the base was not entirely stable, the overall weight of the system was sufficient to allow for consistent unwinding of the fiber tow. This horizontal unwinding introduced an oscillating movement to the yarn path, as the extraction point traversed the length of the spool. This oscillation created variable tension in the input tow, which was further compounded by the fiber's inherent adhesion. To mitigate this inconsistency, the first set of godets was operated in an overfeed mode to act as a tension buffer, meaning that a portion of the initial overfeed was used to correct for setup issues rather than solely to optimize the filament mixing in the nozzle.

Consistent yarn tension was managed through each godet set, which features an integrated, angled cylindrical tensioning roller as depicted in Figure 3.3a, and stable performance was maintained by wrapping the yarn between five and seven times around this roller.



(a) Godet with integrated tensioning roller.



(b) Custom component for the winding unit.

Figure 3.3: Details of the machine setup, showing a standard Retech godet set (a) and the custom-built holder for the input spools (b).

The rotational position of the tensioners was adjusted to control the spacing between fiber

bundles, and for these trials, they were positioned to prevent the tows from contacting each other before the merging guide. This guide was a rudimentary component, made from a sheathed electrical wire bent into a spiral coil, which presented an uneven and potentially abrasive surface unsuitable for brittle fibers. The abrupt change in direction forced by this guide likely contributed to filament damage during processing. Furthermore, the experiments were conducted using the machine's default air-jet nozzle, and while air pressure was an adjustable parameter, the fixed internal geometry of the nozzle was not, a factor that may explain the limited filament dispersion observed in the results. Figure 3.4 provides a view of this processing zone, illustrating how the two tows from the feed godet sets are brought together at the merging guide before entering the air-jet box.



Figure 3.4: The air-jet mixing zone, showing the feed godet sets, the merging guide, and the open mixing box containing the nozzle.

A separate modification was required at the automated winding unit, shown in Figure 3.5. This system operates by rotating the take-up spool through frictional contact with a motor-driven, grooved drum. The helical grooves on the drum guide the yarn back and forth to distribute it across the take-up spool. To ensure the process could be initiated with an empty spool, which otherwise lacked sufficient friction to begin rotating, rubber O-rings were fitted to the extremities of the take-up spool. This ramp-up modification increased the grip between the two components, allowing

the take-up spool to be driven by the rotating drum and properly wind the yarn. The performance of the winding system was imperfect, however, likely due to the stiffness of the carbon fiber, which occasionally caused the yarn to jump out of the guiding grooves and led to a non-uniform build-up on the final spool.

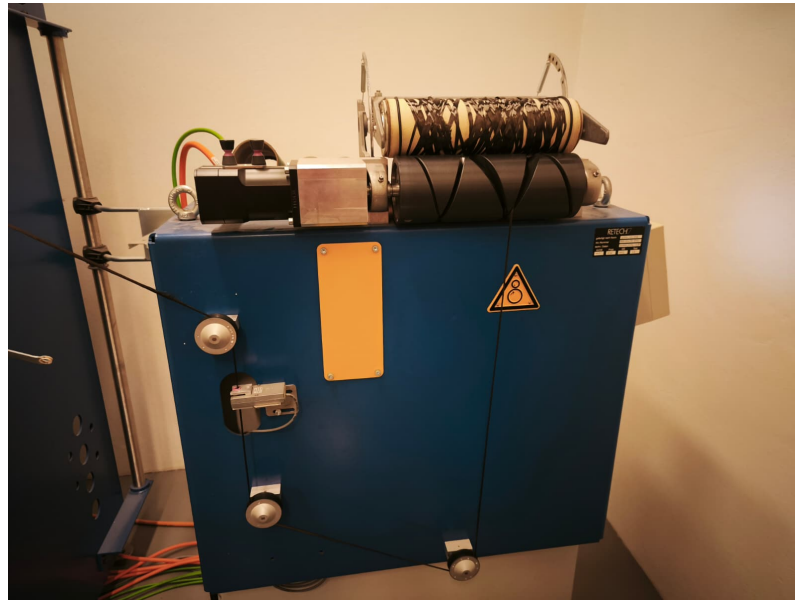


Figure 3.5: The automated winding unit with its grooved traversing cylinder and take-up spool.

A summary of the functional components and their configuration for this experimental work is provided in Table 3.2.

Table 3.2: Summary of Functional Components and Experimental Configuration.

Component	Function and Configuration
Feed System	Custom horizontal stand for oversized carbon fiber spools.
Godet Sets	Five independently controlled sets. Two sets used for each input tow (draw-off and feed), one for take-off.
Tensioning System	Angled cylindrical rollers integrated with godets. Tension controlled by 5 to 7 yarn wraps.
Merging Guide	Rudimentary guide consisting of a coiled, sheathed electrical wire to combine tows before the nozzle.
Air-Jet Nozzle	Standard machine-default nozzle. Air pressure controlled between 4 and 6 bar.
Winding Unit	Automated winder with a grooved drum for yarn traversing. Take-up spool modified with rubber O-rings to provide initial friction for start-up.

3.3 Results and Observations

As stated in the chapter's introduction, the initial plan to co-mingle T800 and HR40 carbon fibers was altered due to material availability. The experimental trials were therefore conducted using a single type of carbon fiber: TenaxTM-E UTS50 F24. This filament yarn, produced by Teijin, is a 24K tow with a nominal linear density of 1600 tex. The key properties of this fiber are listed in Table 3.3, which also includes the properties of the originally planned HR40 and T800 fibers for comparison.

Table 3.3: Material properties of the Tenax UTS50 fiber used in the experiment, compared with the originally planned HR40 and T800 fibers.

Property	Symbol	HR40 [80, 147]	T800 [214]	Tenax UTS50 [210]	Unit
Tensile strength	σ_l	4410	5880	5100	MPa
Tensile modulus	E_l	395	294	245	GPa
Elongation at break	ε_f	1.2	2.0	2.1	%
Density	ρ	1.82	1.80	1.78	g/cm ³
Filament diameter	d_f	6	5	7	μm
Sizing Type	–	–	–	PU	–
Sizing Level	–	–	–	2.0	%

The properties of the Tenax UTS50 fiber present specific characteristics relevant to the co-mingling process. Although its tensile modulus of 245 GPa is lower than that of T800 and HR40, its filaments have the largest diameter at 7 μm . This larger diameter leads to a high bending stiffness for each filament, a factor that counteracts the flexibility suggested by the lower modulus. Furthermore, the tow features a 2.0% polyurethane sizing, which promotes filament-to-filament cohesion to maintain the tow's integrity.

Before processing two tows simultaneously, initial trials were conducted with a single carbon fiber tow to assess the isolated effect of the air jet. Following this validation, the main trials with two tows were performed, during which several combinations of godet speeds and air pressures were tested to identify a stable configuration. In these two-tow trials, the air pressure was varied between 4 and 6 bar, but this did not produce any clear beneficial changes in the output yarn. The final parameters were therefore selected based on the godet speeds that offered the most consistent processing. For the final production run, the left feed path used a draw-off speed of 25.0 m/min (G1) and a feed speed of 24.0 m/min (G2), while the right path used a draw-off speed of 21.0 m/min (G3) and a feed speed of 21.2 m/min (G4). With the final take-off godet (G5) running at a speed of 20.0 m/min, this configuration resulted in overfeed ratios into the nozzle of 1.20 for the left tow and 1.06 for the right tow. The asymmetry in these speeds was a necessary compensation for unbalanced tension, as mechanical limitations prevented adjustment of the tensioning roller on the left-side godet sets. The layout of these godets is illustrated schematically in Figure 3.6.

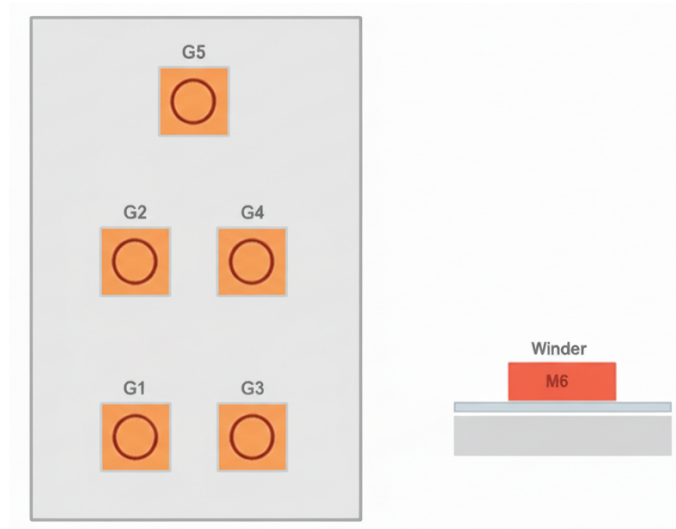


Figure 3.6: Schematic of the godet set layout on the co-mingling machine control interface.

Macroscopic examination of the produced yarn, shown in Figure 3.7, revealed that the constituent tows did not fully separate and intermingle as would be required for a fiber-by-fiber blend. Instead, the process resulted in a yarn where the constituent fiber bundles remained largely distinct and ran parallel to each other in a side-by-side arrangement. This outcome suggests that the selected process parameters and the machine's configuration were insufficient to overcome the cohesion of the individual carbon fiber tows. Factors such as the sizing applied to the fibers, which can cause filaments to adhere to one another, and the inherent stiffness of the carbon filaments likely contributed to the resistance to tow opening and filament mixing within the air-jet nozzle. Although the process successfully combined the two input tows into a single continuous yarn, the resulting internal structure was not a homogeneous, fiber-by-fiber blend.

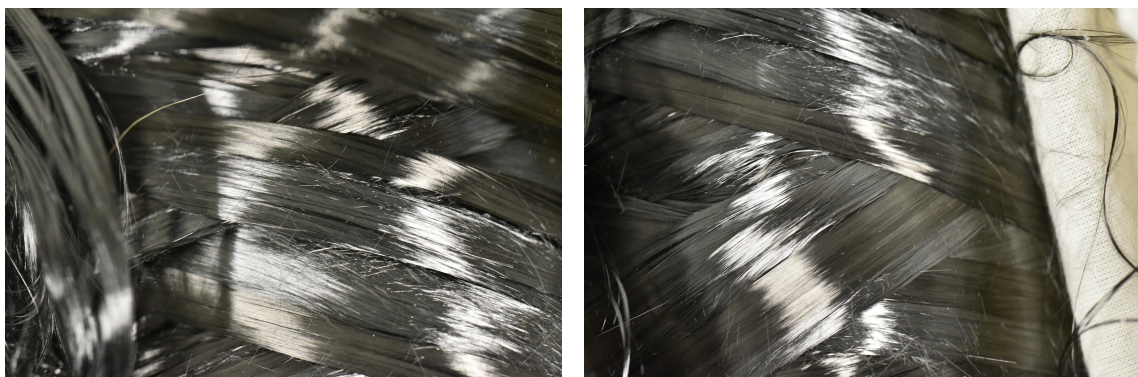


Figure 3.7: Macroscopic views of the produced carbon fiber yarn. The images clearly show the lack of filament dispersion, with the constituent tows remaining distinct in a parallel, side-by-side structure. Some loose surface filaments are also visible, indicating minor process-induced disruption that was insufficient to achieve full tow opening.

3.4 Discussion and Conclusions

The primary conclusion of this work is that the objective of creating a filament-level dispersed yarn was not achieved. This outcome can be attributed to a combination of factors, including the cohesive sizing of the selected fiber, the limitations of the chosen machine parameters, deficiencies in the experimental setup, and the use of a standard, unoptimized air-jet nozzle. The process instead yielded a yarn where the constituent tows remained largely distinct in a side-by-side arrangement, an alternative structure that may have a specific utility in other advanced manufacturing processes.

The side-by-side hybrid yarn could be a precursor material for producing intraply hybrid composites using automated manufacturing techniques like Automated Fiber Placement (AFP). In conventional intraply manufacturing, an AFP machine must frequently stop, cut the tow, and switch spools to place different materials adjacent to one another. This process of changing materials introduces the potential for manufacturing defects, such as gaps or overlaps between adjacent tows, which can affect the performance of the final part. By using a pre-combined side-by-side hybrid yarn containing both fiber types, a single tow could be placed continuously. This would eliminate the need for material changes during the layup of an intraply ply and could therefore mitigate the formation of associated defects.

The experimental outcome demonstrates that the combination of the carbon fiber's properties and the machine's configuration prevented the successful production of a co-mingled yarn. The high bending stiffness of the individual $7\ \mu\text{m}$ filaments, combined with the strong tow cohesion from the 2.0% polyurethane sizing, presented a resistance to tow opening that the air-jet process could not overcome with the tested parameters. Furthermore, the non-standard feeding system and rudimentary guides, which were necessary modifications for the setup, likely contributed to the minor fiber damage observed as loose surface filaments.

The viability of this proposed application depends on ensuring the manufacturing process did not induce significant damage to the fibers. The mechanical action of the air jet and the passage of the yarn through the machine's components could cause filament breakage or surface abrasion. Therefore, before this side-by-side yarn can be used as a precursor for high-performance intraply composites, further investigation is required to quantify the extent of any process-induced fiber damage and to characterize the mechanical properties of the yarn itself.

Chapter 4

Interply Hybrids

This chapter provides an analysis of interply hybrid composites. The chapter begins by establishing a methodology for the selection of constituent materials and configurations. This framework is then used to analyze the benchmark experimental study on interply hybrids conducted by Danzi [53], whose results are presented and discussed. Finally, a detailed post-mortem analysis of the double-edge notched tension (DENT) specimens from that study, based on micro-Computed Tomography (μ CT), is presented to investigate the damage mechanisms.

4.1 Selection Methodology for Interply Hybrid Configurations

This section proposes a systematic methodology for the selection of modern, pseudo-ductile hybrid composite configurations. The goal is to establish a clear framework for choosing constituent materials based on the requirements of sectors such as aerospace. To demonstrate the application of this methodology, it is then used to analyze the well-documented benchmark hybrid systems previously investigated by Danzi [53], which itself builds upon the work of Amacher [9] and Czél [46, 48, 106].

Logically, given the focus in aerospace application, the material's selection process also took into account the sector requirements. While material selection can be influenced by factors like cost, industrial relevance or availability, manufacturability (and related to this, the ease of spreading) as well as expected toughness improvement and the preservation of tensile properties with the last one related to onset of damage and strength [13, 32] the choice was to develop pseudo-ductile all-carbon hybrid composites specifically for aerospace applications, where materials with high specific stiffness and strength are required.

The main factors influencing this choice include:

1. **Focus on all-Carbon hybrid systems:** While earlier studies extensively investigated carbon/glass hybrids [45], the negative effect of glass fibers on specific stiffness led researchers to focus on all-carbon systems to comply with aerospace requirements. Furthermore, tough fibers like glass or aramid were avoided, as they can lead to a decrease in the overall strength

of the hybrid. Another factor is the positive hybrid effect reported for intermediate-modulus carbon fibers when used as high-strain fibers [32].

2. **Use of Predictive Tools:** The "damage mode map" methodology developed by Jalalvand, Czél, and Wisnom [103, 104, 106] is used to predict the sequence of failure modes based on the properties and thicknesses of the LE and HE layers. It is implemented to select hybrid configurations likely to exhibit the desired pseudo-ductile behavior.
3. **Alignment with thin-ply research:** The selection also aligns with findings from related thin-ply research such as that conducted at EPFL by Amacher and co-workers [44]. Their work screened potential thin-ply hybrid combinations including T800-HR40 and T800-Twaron. They identified the T800 fiber as a suitable baseline due to their good balance of modulus and strength and found that the T800-HR40 ply-level hybridization specifically yielded promising results in terms of increased translaminar fracture toughness and achieving pseudo-ductility. It is important to note, that the cost aspect is even more pronounced when it comes to thin-ply where part of the cost comes from the manufacturing related with the prepreg production line. [32]

4.1.1 Constituent Materials and Lamina Properties

Following standard convention, the fibers are categorized by failure strain into a low-elongation (LE) and a high-elongation (HE) type. The materials used in the experimental study are:

- Low Elongation (LE): PYROFIL™ HR40 carbon fiber from Mitsubishi Chemical Carbon Fiber and Composites (MCCFC)
- High Elongation (HE): T800 carbon fiber by Toray Industries

HR40 and T800 are PAN-based carbon fibers where the former has a very high stiffness and the latter has both high tensile strength and ductility [147, 214]. Table 4.1 lists the most relevant properties available from the manufacturers datasheets.

Table 4.1: Material properties of HR40 and T800 carbon fibers.

Property	Symbol	HR40 [80, 147]	T800 [214]	Unit
Tensile strength	σ_t	4410	5880	MPa
Tensile modulus	E_t	395	294	GPa
Elongation at break	ε_f	1.2	2.0	%
Thermal conductivity	k	45	11.3	W/mK
Density	ρ	1.82	1.80	g/cm ³
Specific heat	c_p	0.75	0.74	J/gK
Filament diameter	d_f	6	5	μm

The modulus–strain mismatch promotes controlled fragmentation of the HR40 plies and progressive load transfer to the tougher T800 layers, originating the "pseudo-ductile" stress–strain plateau.

As for the epoxy matrix system, however, it differs from the benchmark. This study considers prepregs manufactured with ThinPreg™ 80EP-402 resin [158] whereas Danzi [53] used ThinPreg™ 80EP-736. A comparison of the few neat resin properties available for both systems from the datasheet and the literature [54] is shown in Table 4.2. Additional relevant properties for the ThinPreg™ 80EP-402 system used in this work, obtained from its datasheet, are detailed in Table 4.3.

Table 4.2: Comparison resin properties of ThinPreg™ 80EP-402 and 80EP-736.

Property	Symbol	80EP-402 [158]	80EP-736 [54]	Unit
Density (Cured Resin)	ρ	1.21	1.24	g/cm^3
Tensile Modulus	E_m	3.44	3.53	GPa
Tensile Strength	σ_m	62	75.0	MPa

Table 4.3: Additional properties of ThinPreg™ 80EP-402 resin [158]

Property	Symbol	Value	Unit	Test Method
Typical cure range	T_{cure}	120–135	$^{\circ}C$	–
Max T_g (glass transition temp.)	T_g	$> 185^*$	$^{\circ}C$	–
Flexural modulus	E_b	3.57	GPa	ISO 178
Flexural strength	σ_b	148	MPa	ISO 178
Fracture toughness K_{IC}	K_{IC}	0.78	$MPa\sqrt{m}$	ISO 13586
Out life (18–22 $^{\circ}C$)	$t_{outlife}$	60	days	–
Storage life (–18 $^{\circ}C$)	$t_{storage}$	24	months	–

* Maximum T_g after post-cure at 160 $^{\circ}C$.

The differences in matrix properties, particularly potential impacts on T_g , toughness, and interfacial behavior, require the recalculation of both the laminate properties and the associated damage mode maps. More specifically, the matrix properties influence the resulting lamina stiffness and strength and the interlaminar fracture toughness (G_{CII}) which are inputs to the predictive model [103, 106].

The purpose of this recalculation is to validate the expected failure behavior for the specific material system (T800/HR40/80EP-402) confirming if the manufactured layups are predicted to fall within the desired pseudo-ductile failure region, where stable LE fragmentation occurs.

The longitudinal mechanical properties of a UD lamina depend directly on the properties of its constituents and their relative proportions within the composite. These proportions are typically represented by the fiber volume fraction (V_f). Since prepreg material data often specifies the resin

content by weight fraction, W_m , to use micromechanical models like the Rule of Mixtures the fiber volume fraction must be calculated from the resin weight fraction using the densities of the fiber, ρ_f , and matrix, ρ_m . The relationship, considering a voidless volume, is [51]:

$$V_f = \frac{V_{fiber}}{V_{composite}} = \frac{V_{fiber}}{V_{fiber} + V_{matrix}} = \frac{W_f/\rho_f}{W_f/\rho_f + W_m/\rho_m} = \frac{(1 - W_m)\rho_m}{(1 - W_m)\rho_m + W_m\rho_f} \quad (4.1)$$

It is important to note that because the fiber and matrix generally have different densities the fiber volume fraction is typically not equal to the fiber weight fraction.

Once the fiber volume fraction is known the longitudinal Young's Modulus (E_l) of the lamina can be estimated using the simple rule-of-mixtures:

$$E_l = E_f V_f + E_m (1 - V_f) \quad (4.2)$$

where E_f and E_m are the Young's moduli of the fiber and matrix respectively. The longitudinal strength can be estimated similarly often approximated by $\sigma_l \approx \sigma_f V_f$ where σ_f is the fiber tensile strength assuming the fibers dominate the longitudinal strength.

These calculations were performed to estimate the lamina properties for both the T800/80EP-402 (HE) and HR40/80EP-402 (LE) systems used in this study. The calculations used the fiber properties from Table 4.1 and the properties of the 80EP-402 matrix detailed in Table 4.2.

Since the matrix can be supplied with a resin content ranging from 32% to 42% by weight – variability present due to the manufacturing process – both cases are considered in the laminate properties calculations. In addition, it is also considered the medium value between these (37%) serving this as the reference value. These lamina property calculations were evaluated using a custom MATLAB script, which is available in Appendix A.1.

The resulting estimated lamina stiffness and strength values used as inputs for the damage mode map analysis are presented in Table 4.4.

Table 4.4: Lamina Properties for T800/80EP-402 and HR40/80EP-402.

W_m (%)	W_f (%)	V_f (%)		E_l (GPa)		σ_l^* (GPa)	
		T800	HR40	T800	HR40	T800	HR40
32	68	58.8	58.6	174.3	232.9	3.46	2.58
37	63	53.4	53.1	158.6	211.4	3.14	2.34
42	58	48.1	47.9	143.2	191.0	2.83	2.11

* Estimated using: $\sigma_l \approx \sigma_f V_f$

Although seemingly small these differences require evaluation of the predicted failure modes due to the potential variations in toughness impacting G_{IIc} . The primary purpose of generating these updated damage mode maps is to validate the expected failure behavior of the specific hybrid layup configurations.

4.1.2 Damage Mode Maps

To predict the failure behavior of the T800/HR40/80EP-402 hybrid system, damage mode maps were generated using the analytical methodology developed by Jalalvand et al. [103, 106]. These maps provide a visual guide to the expected dominant failure mode process based on the thicknesses of the constituent layers [104, 106], and this analytical approach is often used as a guideline in designing hybrid composites for specific failure behaviors [32]. The methodology typically relies on underlying mechanics models, such as shear-lag theory for interlayer hybrids [32, 103], to determine critical stress levels.

The method first determines the critical laminate stress levels required to initiate three fundamental damage modes [103]:

- σ_{LF} : Stress for fragmentation in the low-elongation (LE) layers.
- σ_{del} : Stress for interlaminar delamination between LE and high-elongation (HE) layers.
- σ_{HF} : Stress for failure of the high-elongation (HE) layers.

For which the governing equations, for these critical stress levels, are:

$$\sigma_{LF} = \sigma_L \frac{\alpha\beta + 1}{\alpha(\beta + 1)} \quad (4.3)$$

$$\sigma_{del} = \frac{1}{1 + \beta} \sqrt{\frac{(1 + \alpha\beta)}{\alpha\beta} \left(\frac{2G_{IIC}E_H}{t_H} \right)} \quad (4.4)$$

$$\sigma_{HF} = \frac{1}{1 + \beta} \frac{\sigma_H}{K_t \sqrt[m]{V}} \quad (4.5)$$

where:

- σ_L, σ_H are the nominal longitudinal tensile strengths of the LE and HE layers, respectively. The σ_L term in Eq. (4.3) technically represents the average strength of the LE layer's fiber strength distribution ($\bar{\sigma}_L$), which accounts for statistical variations and is typically higher than the weakest-link strength. However, following common practice when applying this model [32, 103], the nominal strength σ_L is used as an approximation for $\bar{\sigma}_L$.
- E_L, E_H are the longitudinal Young's moduli of the LE and HE layers.
- t_L, t_H are the half-thicknesses of the LE and HE layers. Schematic representation in Figure 4.1.

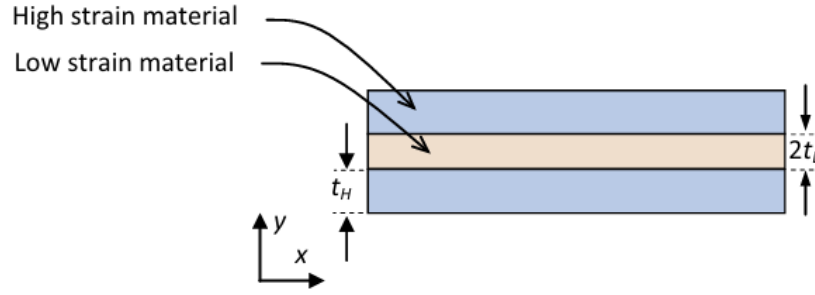


Figure 4.1: Hybrid laminate arrangement and thickness nomenclature (adapted from [103]).

- α is the ratio of LE to HE layer modulus:

$$\alpha = \frac{E_L}{E_H} \quad (4.6)$$

- β is the ratio of LE to HE layer half-thickness:

$$\beta = \frac{t_L}{t_H} \quad (4.7)$$

- G_{IIc} is the mode II interlaminar fracture toughness between the layers.
- m is the Weibull modulus for the HE material's tensile strength.
- V is the effective volume of the HE material under stress in the gauge section:

$$V = W L t_H \quad (4.8)$$

where W and L are the specimen gauge width and length, respectively.

- K_t is the stress concentration factor in the HE layer due to damage

Based on the calculated values and the relative order of these three critical stresses (σ_{LF} , σ_{del} , σ_{HF}), the model predicts which of the following four potential failure scenarios will occur after the initial crack in the low-strain material, as illustrated schematically in Figure 4.2

1. Premature failure of the high-strain material.
2. Catastrophic delamination followed by high-strain material failure.
3. Fragmentation (multiple fractures) in the low-strain material and then high-strain material failure.
4. Fragmentation in the low-strain material followed by dispersed delamination and then high-strain material failure.

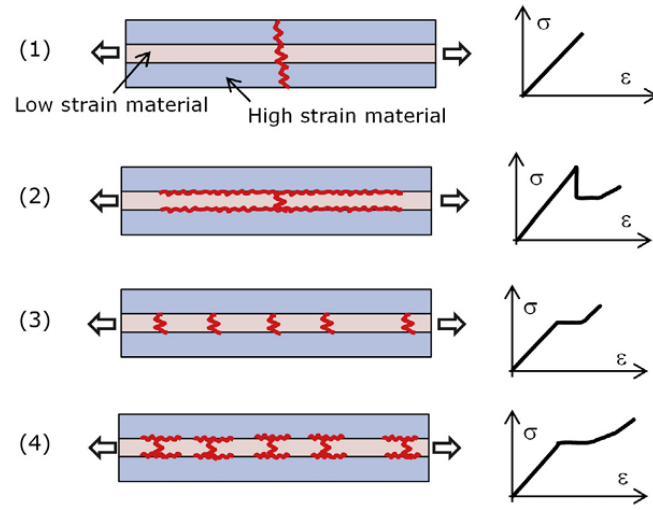


Figure 4.2: Different damage scenarios of UD hybrid laminates (adapted from [103]).

The damage mode map visually represents these predicted scenarios in a space defined by the laminate geometry. It is constructed by plotting the absolute LE layer half-thickness (t_L) against the relative LE layer thickness fraction (γ), known also as low-strain material volume fraction [106] (concept introduced in Section 2.3.2).

$$\gamma = \frac{t_L}{t_L + t_H} = \frac{\beta}{1 + \beta} \quad (4.9)$$

The map is divided into regions corresponding to the different predicted failure scenarios (1-4 above). The boundaries between these regions are determined by finding the conditions where the critical stresses for two competing fundamental modes are equal. For example, $\sigma_{LF} = \sigma_{del}$ determines the transition involving fragmentation and delamination.

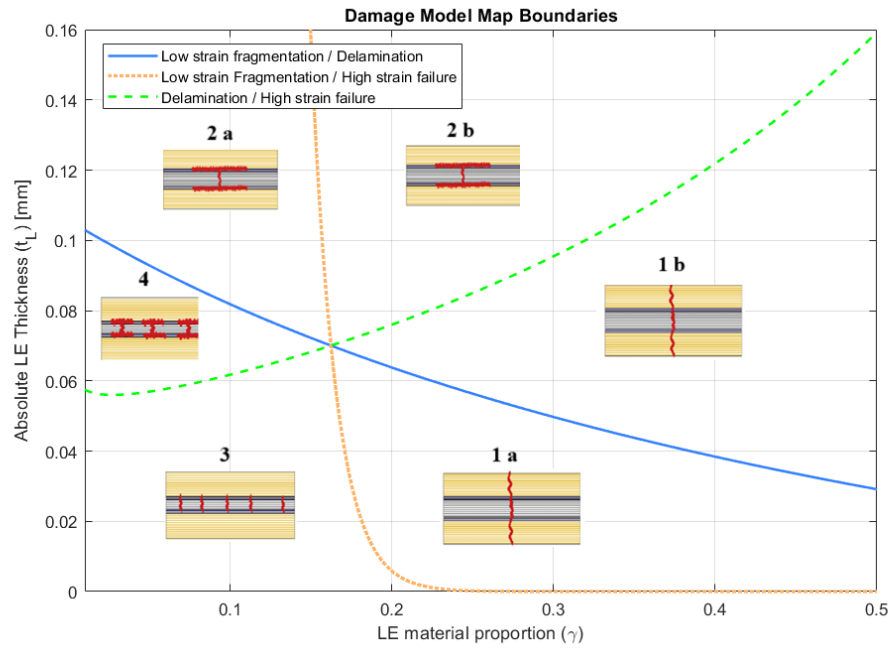


Figure 4.3: Damage mode map with associated failure scenarios.

For any given hybrid configuration (defined by t_L and γ), the relative order of these critical stresses dictates the expected sequence of damage events after the initial LE crack [103]. Table 4.5, adapted from [103], summarizes the relationship between the possible stress orderings and the resulting failure process, consolidating them into the four distinct scenarios listed previously.

Table 4.5: Summary of expected damage modes for different conditions after the first crack in the LE material (reproduced from [106]).

No.	Order of required stress for damage modes	Expected damage process after the initial crack in the low strain material
1a	$\sigma_{HF} < \sigma_{LF} < \sigma_{del}$	1. Failure in the high strain material
1b	$\sigma_{HF} < \sigma_{del} < \sigma_{LF}$	1. Failure in the high strain material
2a	$\sigma_{del} < \sigma_{LF} < \sigma_{HF}$	1. Catastrophic delamination 2. Failure in the high strain material
2b	$\sigma_{del} < \sigma_{HF} < \sigma_{LF}$	1. Catastrophic delamination 2. Failure in the high strain material
3	$\sigma_{LF} < \sigma_{HF} < \sigma_{del}$	1. Fragmentation of the low strain material 2. Failure in the high strain material
4	$\sigma_{LF} < \sigma_{del} < \sigma_{HF}$	1. Fragmentation of the low strain material 2. Dispersed delamination 3. Failure of the high strain material

Jalalvand et al. originally developed and applied this methodology to understand and predict the behavior of hybrid composites, particularly demonstrating its use for carbon/glass systems [103, 105]. Danzi [53] provides a review and further application of this approach, which offers valuable context for the current study.

The damage mode maps were regenerated using the methodology and equations presented above, implemented using a custom MATLAB script, which is available in Appendix A.2.

The specific not yet mentioned inputs were chosen to align with the benchmark study [53] where applicable. Each one is further detailed individually below.

4.1.2.1 Weibull Modulus

A significant distinction exists between the Weibull modulus (m) used to characterize the strength variability of single fibers and the much higher values applied to laminates in predictive failure models. These m values describe strength variability at different material scales, and the laminate-level modulus is not directly calculated from single fiber data.

The tensile strength variability of individual high-performance carbon fibers is typically quantified by a low Weibull modulus, often in the range of $m = 4$ to 8. This reflects the statistical nature of strength governed by microscopic flaws at the fiber level.

However, when fibers are embedded within a matrix to form a unidirectional (UD) composite ply or laminate, the failure process becomes more complex as seen, where the laminate's strength is not simply dictated by the single weakest fiber but rather the laminate failure under tension involves a multi-stage process as mentioned before.

This sequence averages out some of the extreme variability of individual fiber strengths. Consequently, the measured tensile strength of UD laminates generally exhibits significantly less scatter (a lower coefficient of variation) and is thus represented by a considerably higher effective Weibull modulus (m), often in the range of 20–40 for UD carbon/epoxy tensile strength [223].

In the Jalalvand failure map methodology, the Weibull modulus is specifically incorporated into the equation for the high-strain layer failure stress (σ_{HF}), within the $V^{-1/m}$ term. This accounts for the statistical size effect on the HE layer's strength as a whole. Using a laminate-level Weibull modulus is therefore appropriate for this ply-strength prediction.

Determining an appropriate laminate Weibull modulus for the specific T800/80EP-402 system used in this thesis requires experimental data showing strength variation with specimen volume. A review of available literature reveals a general scarcity of data for this exact material combination.

Focusing on T800 fibers, while some strength values are reported, many studies that investigate size effects or provide detailed strength data either maintain the same carbon and matrix configuration but test only a single volume (often following ASTM D3039 [25]) without systematic volume variation [93, 94, 117], or provide insufficient data points – lack of different volume specimens – to reliably derive m [129].

Other studies that do vary specimen volume and maintain constituent consistency may not focus on simple UD laminates, instead investigating quasi-isotropic (QI) or other configurations, making direct extraction of UD tensile m problematic [102]. Some research provides size effect data but for

non-rectangular geometries, such as Hwang's [100] work on composite pressure vessels, which is not directly comparable to UD flat coupon behavior.

Further complicating the direct extraction of m for T800-based systems, studies like that of Okabea and Takedab [160], while focusing on T800H/3631 unidirectional CFRP composites tested under specific conditions (0.4 mm/min crosshead speed, room temperature), did not present their findings in a format conducive to extracting the necessary statistical parameters for Weibull analysis.

Given this lack of specific, processable data for the target T800 material system, a simplification is adopted. Standard tensile strength size effect data for a different unidirectional carbon/epoxy system (IM7/8552) presented by Wisnom et al. [224] can be used as a representative case for typical carbon/epoxy behavior.

The fundamental relationship governing the strength-volume scaling according to Weibull theory relates the strength σ at volume V to a reference strength σ_0 at a reference volume V_0 :

$$\frac{\sigma}{\sigma_0} = \left(\frac{V}{V_0} \right)^{-\frac{1}{m}} \quad (4.10)$$

Taking the natural logarithm of Equation 4.10 yields a linear relationship between $\ln(\sigma)$ and $\ln(V)$:

$$\ln(\sigma) = -\frac{1}{m} \ln(V) + \left(\ln(\sigma_0) + \frac{1}{m} \ln(V_0) \right) = -\frac{1}{m} \ln(V) + \text{constant} \quad (4.11)$$

This predicts that experimental data should form a straight line when plotted on log-log axes, with the slope directly related to the Weibull modulus.

Wisnom et al. [224] scaled tensile tests on UD specimens made from IM7/8552 confirmed this expected size effect, with strength decreasing with volume [224]. The results are reproduced in Table 4.6.

Table 4.6: Summary of tests on scaled UD specimens (IM7/8552) (reproduced from [224]).

Thickness (mm)	Gauge length (mm)	Width (mm)	No. of specimens	Failure stress (MPa)	C.V. (%)
0.5	30	5	12	2806	4.2
1	60	10	12	2687	2.5
2	120	20	9	2553	3.8
4	240	40	6	2410	6.0

From this data the equation below is obtained and the corresponding plot can be seen in Figure 4.4:

$$\ln(\sigma) = -0.0244 \ln(V) + 8.0485 \quad (4.12)$$

From the slope, the Weibull modulus is calculated as $m = 1/0.0244 \approx 41$. Although, as noted by Wisnom, this value is higher than the 20–30 range typically reported for such materials [224].

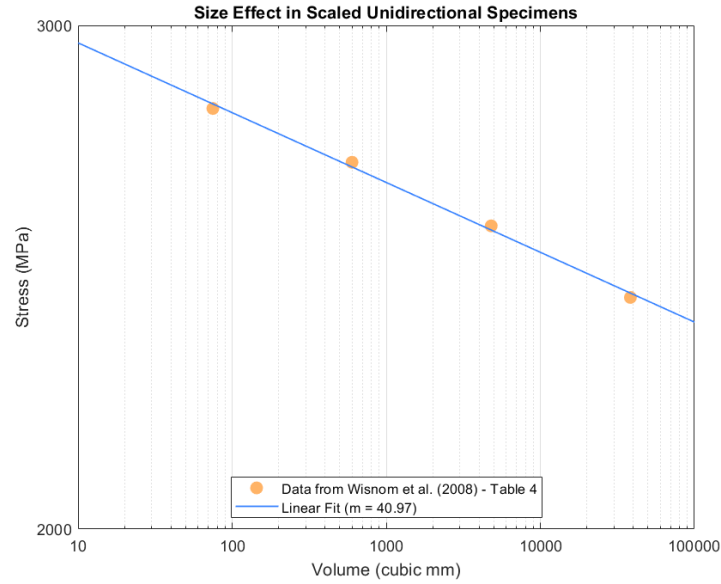


Figure 4.4: Size effect in scaled unidirectional specimens.

This adaptation is further justified by considering the strain rate sensitivity inherent in composite materials. Raju et al. [167] demonstrates that tensile properties, including strength, are sensitive to strain rate across various specimen volumes for materials like fiberglass and carbon. Crucially, they observed that the m tends to increase (indicating a diminishing size effect) as strain rate increases. This suggests that even if volume-strength data were experimentally obtained for the specific T800/80EP-402 system, the derived m value would be dependent on the testing rate selected. Given this inherent variability due to strain rate, using a single experimentally determined m for T800/80EP-402 might not necessarily offer greater precision for general modeling purposes compared to adopting a well-documented value from a similar carbon/epoxy system. Therefore, using $m = 41$ is considered a valid simplification.

4.1.2.2 Stress Concentration Factor

The Stress Concentration Factor (SCF), K_t , is utilized in the Jalalvand failure map methodology to account for the localized increase in stress within the HE layer following damage, such as cracking, in the adjacent LE layer. This factor is incorporated into the equation for the failure stress of the HE layers, σ_{HF} (Equation 4.5).

The selection of an appropriate K_t value is guided by the precedent established in the original development of this analytical model by Jalalvand et al. [103]. They reasoned that for simplicity and computational efficiency, a constant K_t derived from detailed micromechanical studies would be preferable to incorporating more complex, variable stress concentration analyses into the model.

The most relevant study is the work by van den Heuvel et al. [92]. This research presented a comprehensive 3D finite element (FE) analysis of stress concentrations around a single fiber break in unidirectional carbon fiber-reinforced epoxy composites with good fiber/matrix adhesion. The authors investigated three distinct epoxy matrix systems, that can be distinguished by their yield strengths:

- (100:0) matrix with $\sigma_y \approx 71$ MPa
- (70:30) matrix with $\sigma_y \approx 59$ MPa
- (50:50) matrix with $\sigma_y \approx 47$ MPa

The 80EP-402 resin employed in the current work has a reported tensile strength of 62 MPa (Table 4.2), suggesting its mechanical response, particularly regarding yielding, may be most similar to the (70:30) matrix system.

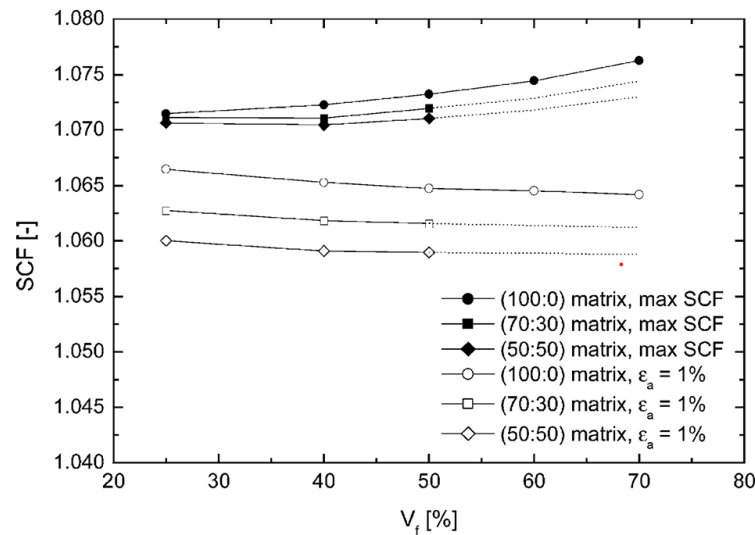


Figure 4.5: The SCF as a function of the fibre volume fraction (V_f) for the three matrix systems (reproduced from [92]).

Figure 4.5 displays the calculated SCF as a function of fiber volume fraction (V_f) for these three matrix types. Considering a representative V_f of approximately 58.8% (Table 4.4), examination of the curve corresponding to the (70:30) matrix in figure reveals a maximum SCF of approximately 1.073. More broadly, across all matrix systems and V_f values from 25% to 70%, van den Heuvel et al. [92] reported that the maximum SCFs remained consistently within the range of approximately 1.071 to 1.075.

Based on these, and rounding up, a K_f value of 1.08 is adopted.

4.1.2.3 Specimen Dimensions

Specimen dimensions were selected based on guidelines from the standard ASTM D3039/D3039M-14, which provides recommendations to achieve valid gage section failures for tensile property determination.

A specimen width of $W = 15$ mm is used, consistent with the standard's recommendation for 0° unidirectional polymer matrix composites. As for the specific gauge length it is set to $L = 80$ mm. This chosen length fits within the nominal 138 mm space available between tabs based on the standard's recommended overall length (250 mm) and tab lengths (56 mm each). This choice, present in the benchmark study, ensures that the measurement zone is correctly distanced from tab-end stress concentrations.

Table 4.7: 0° UD Specimen Geometry (data from ASTM D3039, Table2 [25])

Parameter	Value	Unit
Specimen Width	15	mm
Overall Length	250	mm
Specimen Thickness	1.0	mm
Tab Length	56	mm
Tab Thickness	1.5	mm
Tab Bevel Angle	7 or 90	$^\circ$

4.1.2.4 Interlaminar Toughness Sensitivity

The Mode II interlaminar fracture toughness, G_{IIc} , affects the critical stress for delamination, σ_{del} (Equation 4.4), and thus the predicted damage mode map boundaries. This toughness is commonly measured using the End Notched Flexure (ENF) test, which is standardized in ASTM D7905 [21]. The ENF test typically yields an initiation toughness value, and the standard often involves an initial test from a polymer film to generate a true precrack for the main tests, a process which normally gives a lower, more conservative G_{IIc} measurement. The determination of an appropriate G_{IIc} value is subject to various factors, including material constituents, interface characteristics, temperature, moisture, and specific details of specimen geometry and precracking [225].

A review of the literature provides extensive data for T800/epoxy systems. For this system, reported room temperature G_{IIc} values vary, for example, Gong et al. [79] found a mean value of 2.26 N/mm – using the experimental compliance method (ECM) – while Kageyama et al. [110] reported values for a toughened T800H/3900-2 system ranging from 0.55 N/mm (precracked) to 1.12 N/mm (no precrack). Other studies, such as Saidpour et al. [175], show Mode II values for some T800-based carbon/epoxy systems in the 1.0 – 1.1 N/mm range.

Despite information on the T800 constituent, specific experimental G_{IIc} data for the hybrid of the constituents interface select in this work could not be found. There are however, studies focusing on hybrid composites with T800 and HR40 fibers, such as those by Danzi et al. [54] using

T800/80EP-736 and HR40/80EP-736, and Cugnoni et al. [44] using T800-TP175 and HR40-TP175 plies, but these have primarily centered on Mode I intralaminar or translaminar fracture toughness rather than Mode II. For a visual distinction between these fundamental fracture mechanisms, refer to Figure 4.6, which illustrates the different Modes.

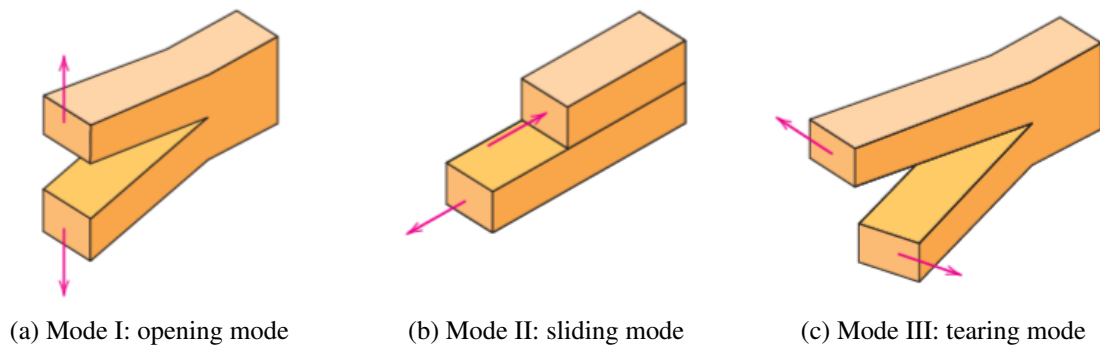


Figure 4.6: Modes of crack surface displacement (adapted from [36]).

As for specific hybrid carbon-carbon interfaces, Ribeiro et al. [169] reported an experimental G_{IIc} of 1.30 N/mm for a high strength carbon / ultra-high modulus carbon hand lay-up hybrid. Unfortunately, this system varies in constituent materials and manufacturing from the prepreg system in this study.

Given this, for the T800/HR40/80EP-402 system, a simplification is made by selecting a single, conservative G_{IIc} value, given the absence of direct experimental data. This estimation approach is similar to the methodology employed by Danzi [53] where a value of 0.5 N/mm was selected for damage mode maps of T800/HR40 based hybrids. Thus, in this work, the same value is selected with this choice aligning with the lower-bound value of 0.55 N/mm reported by Kageyama et al. [110] – reported that the use of data from precracked specimens is generally considered to yield more conservative toughness values [225].

4.1.2.5 Failure Map Plot

Furthermore, to ensure the accuracy of the generated maps, the custom MATLAB script was benchmarked using values provided in the reference work by Jalalvand et al. [103], Danzi [53] and Broggi [32]. It is also important to reiterate that the original formulation, and thus the inputs to equations, use t_H and t_L as the half-thicknesses of the high-elongation and low-elongation layers, respectively, and input values must be adapted accordingly to consider total ply thicknesses.

This recalculation yielded 9 distinct damage mode maps, covering the investigated parameter variations (three resin contents). These maps illustrate the predicted failure regions for the T800/HR40/80EP-402 system, all presented in Figure 4.8. The complete list of parameters can be observed in the following table (Table 4.8)

Table 4.8: Parameters used for the failure map.

Propert	Symbol	Value	Unit
Laminate Weibull modulus	m	41	-
Mode II critical ERR	G_{IIc}	0.5	$kJ \cdot m^{-2}$
Stress concentration factor	K_t	1.08	-
Sample length	L	80	mm
Sample width	W	15	mm

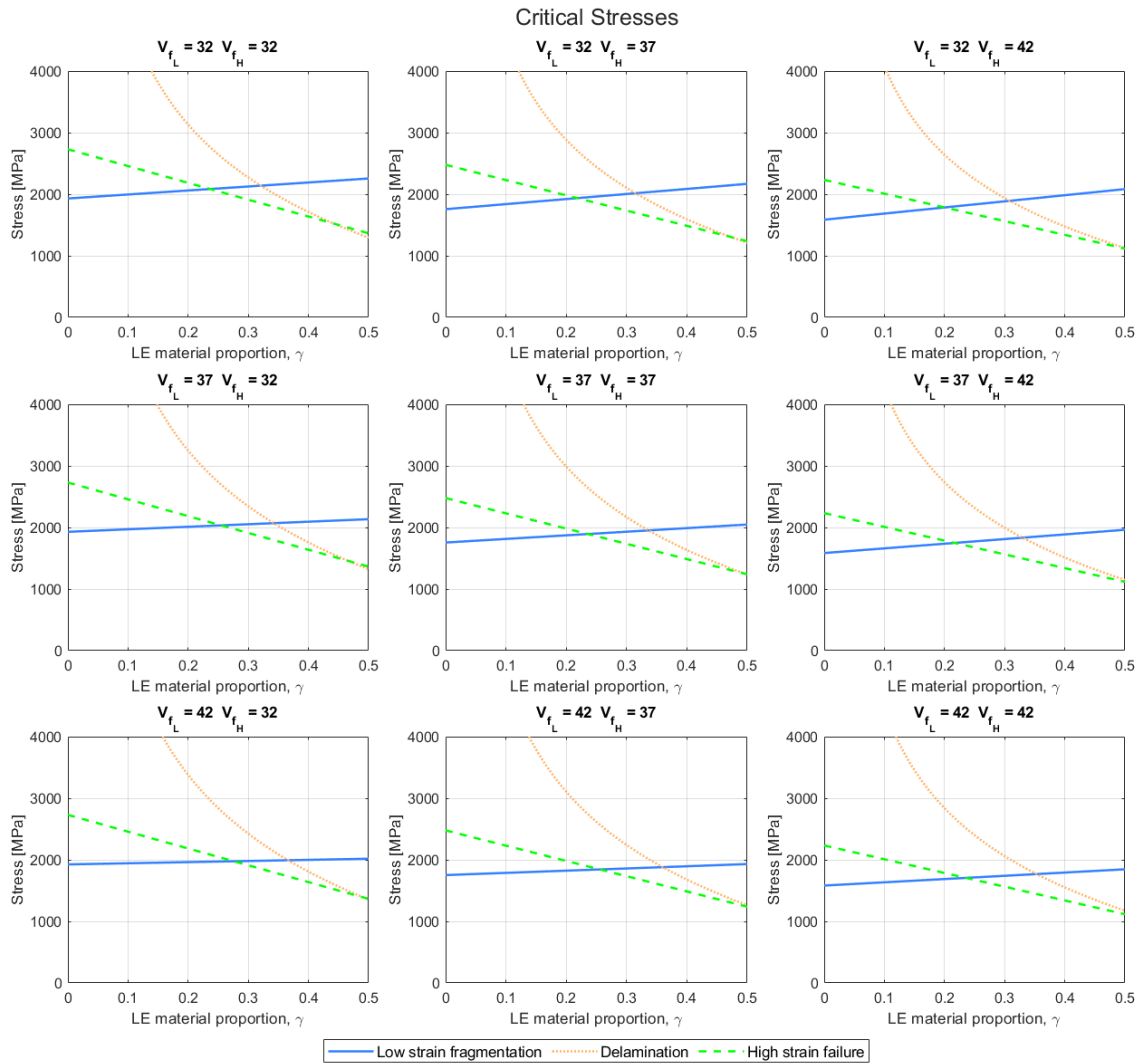


Figure 4.7: Damage model map for T800-HR40 hybrid configuration (Critical Stresses)

Instead of the failure stress the interlayer hybrid can be defined as a function of the low-strain thickness. For this the high-strain thickness must be constant and in this case set to $100 \mu m$. The total ply-block thickness is thus variable and at $\gamma = 0.33$ reaches the limit of thin-ply thickness, at $150 \mu m$. However, for completeness, in the plots that follow the γ value is extended to 0.5.

Given that an approximate 1:1 relation can usually be assumed between the ply thickness in microns and the prepreg FAW in $g \cdot m^{-2}$ (or *gsm*), the plot y-axis is expressed in the latter unit [32].

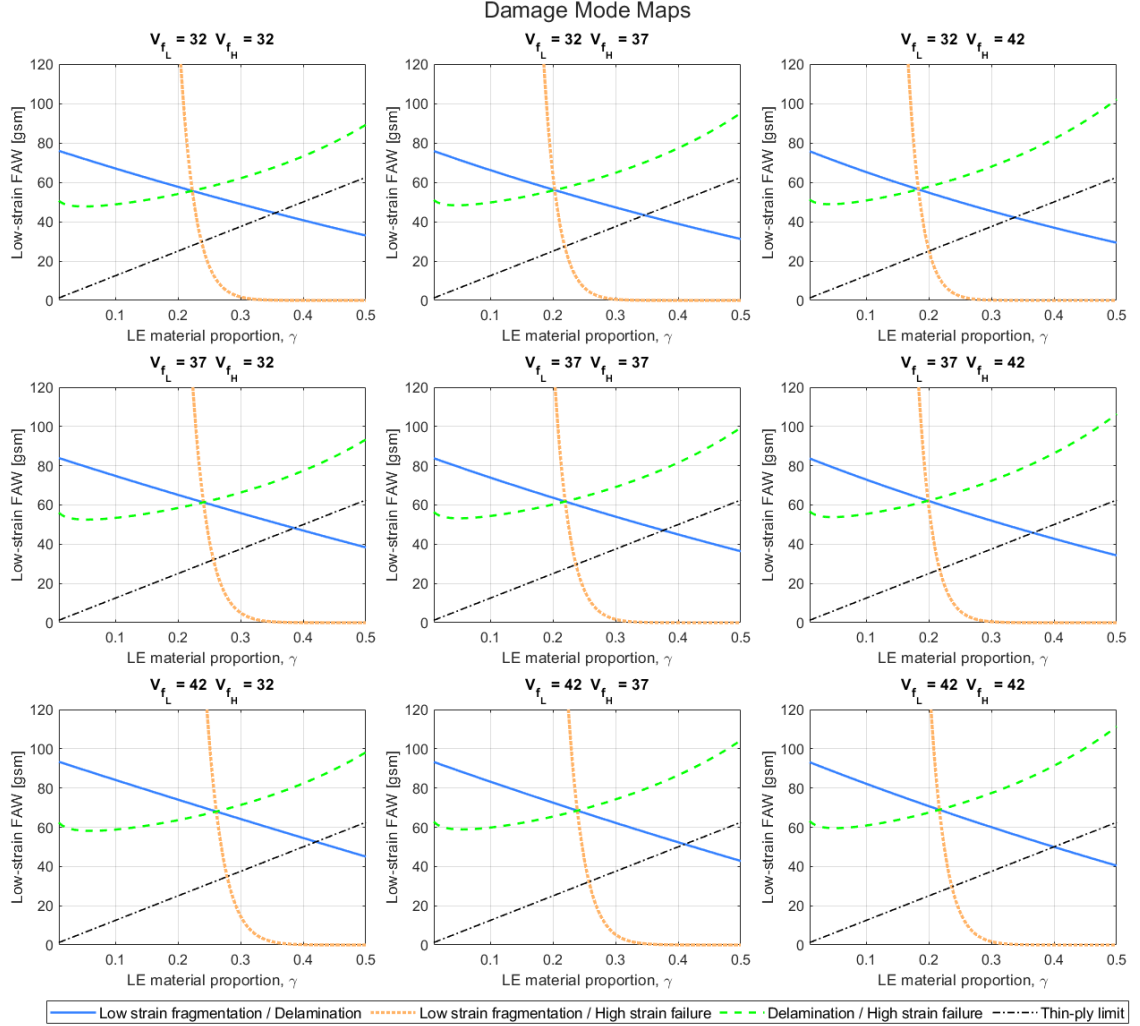


Figure 4.8: Damage model map for T800-HR40 hybrid configuration (Low-strain FAW)

4.1.3 Complementary Design Considerations for Hybrid Composites

While damage mode maps (Figure 4.8) provide information about the expected sequence of failure events, a comprehensive approach to designing hybrid composites, as demonstrated by Broggi [32], also involves evaluating other design aspects. These include the predicted tensile properties, characteristics indicative of translaminar toughness such as fiber pull-out length, and the impact of residual thermal stresses. These analyses provide a more complete picture to help in the selection of optimal hybrid configurations. The tensile and pull-out length calculations were performed using the MATLAB script in Appendix A.3, while the residual stress analysis was conducted with the script detailed in Appendix A.4.

4.1.3.1 Predicted Tensile Properties

The key tensile properties of the hybrid laminate are its longitudinal elastic modulus (E_h) and the ultimate tensile strength (σ_h). These properties can be estimated using the Rule-of-Mixtures and the bilinear Rule-of-Mixtures, respectively, – both are listed in the previous chapter in Section 2.3.4.1 – by considering both the constituent material properties and the LE volume fraction (γ).

The plots of these metrics against γ , for the different resin contents, are illustrated in Figure 4.9,

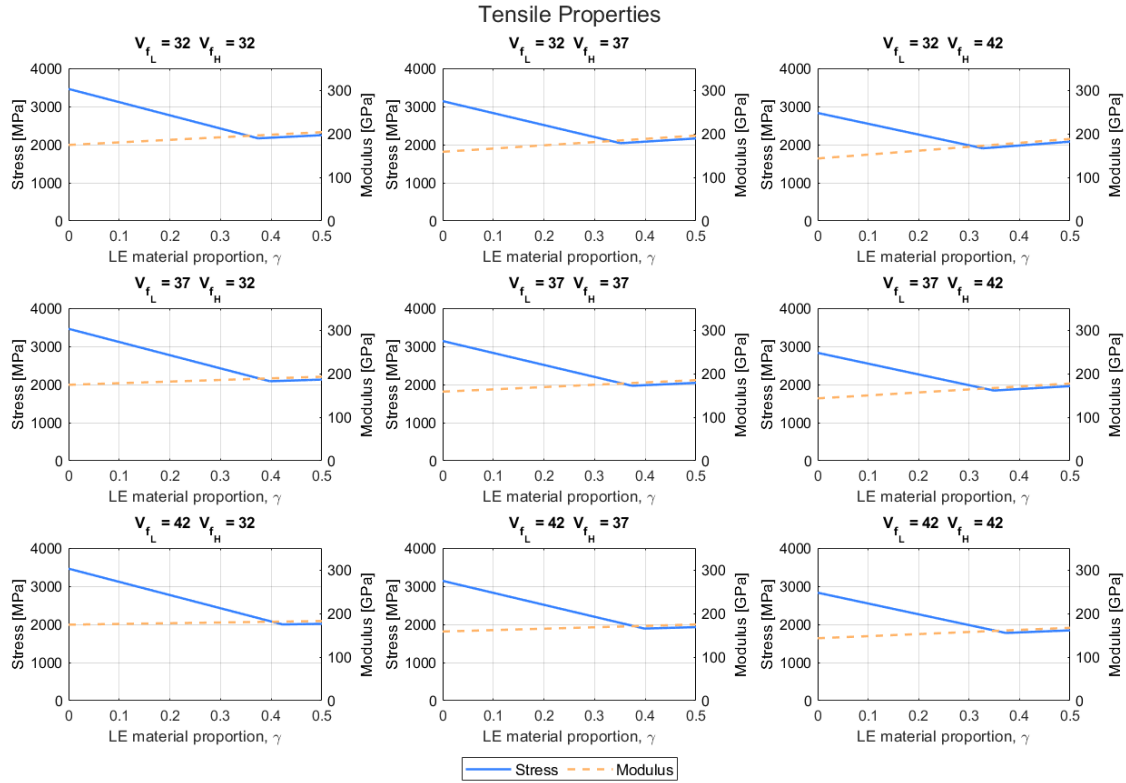


Figure 4.9: Tensile properties for the T800-HR40 hybrid configuration

Such plots allow the identification of configurations that achieve a desirable balance, for example, by maximizing strength while ensuring that the fragmentation of the LE layer (damage onset) is sufficiently delayed or occurs at a manageable stress level relative to the ultimate failure of the HE material. This helps in selecting configurations that not only exhibit a favorable failure sequence but also meet specific stiffness and strength targets.

4.1.3.2 Predicted Pull-Out Length

The energy absorption capacity during fracture, particularly relevant to translaminar toughness, is significantly influenced by the pull-out length of fibers from the LE layer during its fragmentation [32]. An estimation of this characteristic length can be derived from the critical stress recovery length, l_c , often determined using a shear-lag model. The critical length is defined by the equation:

$$l_c = \frac{\sigma_L t_L}{\tau_y} \quad (4.13)$$

where t_L is the LE layer thickness and σ_L is its nominal longitudinal tensile strength. Consistent with the approximation made for the LE fragmentation stress (σ_{LF}) in Equation 4.3, the nominal strength σ_L is used here in place of the average fiber bundle strength (σ'_L).

The term τ_y denotes the interlaminar shear strength (or matrix shear yield strength), which for the purpose of this estimation is considered to be 81 MPa obtained from the suppliers datasheet [159], calculated according to the ISO 14130 with a UD $[0^\circ]_{32}$ layup. This l_c acts as an upper bound for the fiber pull-out length.

The predicted l_c as a function of the LE volume fraction (γ) for the T800/HR40/80EP-402 system is presented in Figure 4.10. Given the similarity in the predicted l_c across the different resin contents investigated, only the plot corresponding to the nominal resin content is presented.

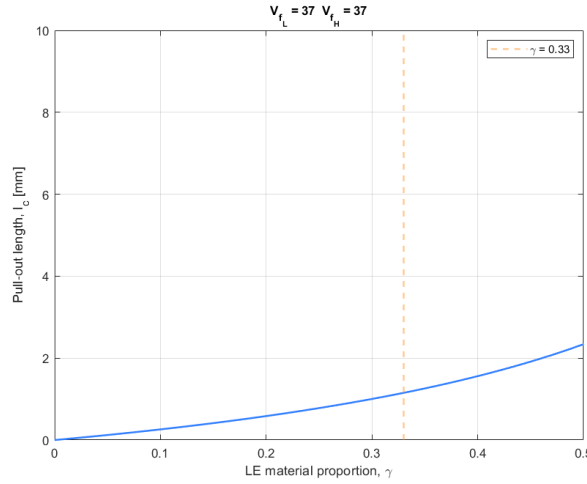


Figure 4.10: Predicted of the maximum pull-out length for the T800-HR40 hybrid configuration

Although maximizing l_c is often a design goal when selecting constituent fibers for new hybrid systems, for the predetermined T800/HR40 combination in this study, the predicted l_c is presented primarily as a characteristic property and not a design variable – the primary scope of this work is not the study of fracture toughness.

Normally, selecting hybrid configurations that maximize the pull-out length can lead to potentially improved toughness. Broggi [32], for instance, employed such an analysis, sometimes normalizing the pull-out length by the associated loss in strength, to identify optimal material systems.

4.1.3.3 Residual Thermal Stresses

As discussed in Section 2.3.2, thermal residual stresses which derive from the mismatch in coefficients of thermal expansion (CTE) between different constituents during cooling from the cure

temperature and are a significant factor in the behavior of composite materials, including hybrid composites [199].

While the output of the residual stress calculations can be a valuable tool for design optimization and configuration selection, within the scope of the present work, these calculations are presented primarily to characterize an additional design variable. Thus, although calculations for some general configurations are presented, the values are not directly incorporated into the subsequent analysis of the experimental campaign detailed in later chapters.

The calculation involves a two step process: First, the effective thermo-mechanical properties of each individual ply are determined from the properties of its fiber and matrix materials and their respective volume fractions. Second, Classical Laminated Plate Theory (CLPT) is applied to the laminate to determine the internal stress distribution resulting from a specified temperature change.

The Rule of Mixtures – which has been introduced earlier to calculate the longitudinal Young's modulus – can be applied to evaluate the major Poisson's ratio:

$$\nu_{12} = \nu_{fLT} V_f + \nu_m V_m \quad (4.14)$$

where ν_{fLT} is the fiber's longitudinal-transverse Poisson's ratio.

As for the transverse and shear properties, the simple Rule of Mixtures often proves inadequate. The Halpin-Tsai equations, on the other hand [85], offer better accuracy over simpler models since they incorporate empirical factors that account for fiber geometry and packing [108].

These equations take the general form:

$$\frac{M}{M_m} = \frac{1 + \xi \eta V_f}{1 - \eta V_f} \quad (4.15)$$

where:

- M is the composite modulus
- M_m is the corresponding matrix modulus
- M_f is the corresponding fiber modulus
- η is given by:

$$\eta = \frac{(M_f/M_m) - 1}{(M_f/M_m) + \xi} \quad (4.16)$$

- ξ acts as a measure of fiber reinforcement, accounting for fiber geometry, packing geometry and loading conditions. The values of ξ are typically determined by comparing the Halpin-Tsai equations with exact elasticity solutions or experimental data through curve-fitting techniques [108].

When calculating the transverse Young's modulus (E_2), the Halpin-Tsai equations are applied with $M = E_2$, $M_f = E_{fT}$ and $M_m = E_m$. The value for the reinforcement factor is $\xi_2 = 2$ since it has shown a good agreement with numerical solutions for fiber volume fractions up to approximately 0.65-0.70 thus within the range required.

$$E_2 = E_m \frac{1 + \xi_2 \eta_2 V_f}{1 - \eta_2 V_f} \quad (4.17)$$

Similarly, for the in-plane shear modulus (G_{12}), the equations are applied with $M = G_{12}$, $M_f = G_{fLT}$, and $M_m = G_m$.

$$G_{12} = G_m \frac{1 + \xi_G \eta_G V_f}{1 - \eta_G V_f} \quad (4.18)$$

where G_m is the standard matrix shear modulus given by

$$G_m = \frac{E_m}{2(1 + \nu_m)} \quad (4.19)$$

For circular fibers, $\xi_G = 1$ is a common value. However, it is noted that for higher fiber volume fractions ($V_f > 0.5$), this can lead to an underestimation of G_{12} for certain fiber arrangements. Some modifications, such as $\xi_G = 1 + 40V_f^{10}$, have been proposed to improve correlation with numerical solutions, but for the purpose of this work, $\xi_G = 1$ is adopted.

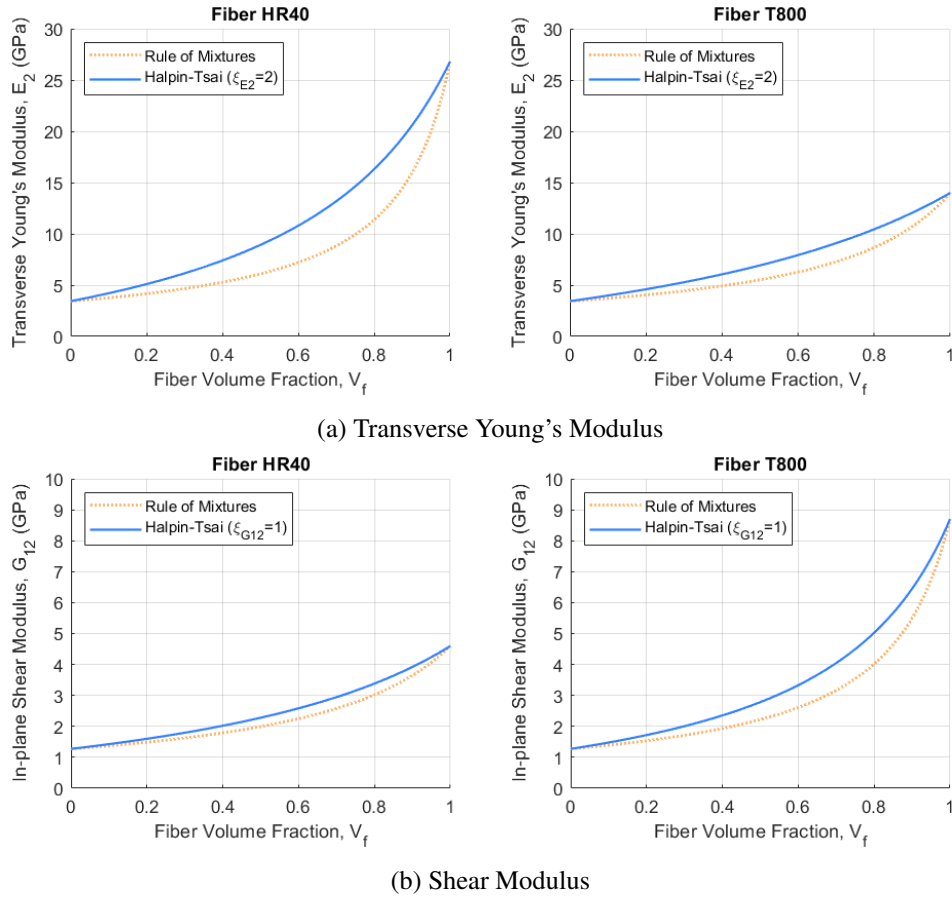


Figure 4.11: Rule of Mixtures and Halpin-Tsai comparison

From this the local reduced stiffness matrix can be constructed:

$$[Q] = \begin{bmatrix} Q_{11} & Q_{12} & 0 \\ Q_{12} & Q_{22} & 0 \\ 0 & 0 & Q_{66} \end{bmatrix}$$

where, for the case of an orthotropic laminate, the components are:

$$Q_{11} = \frac{E_1}{1 - \nu_{12}\nu_{21}}, \quad Q_{22} = \frac{E_2}{1 - \nu_{12}\nu_{21}}, \quad Q_{12} = \frac{\nu_{12}E_2}{1 - \nu_{12}\nu_{21}}, \quad Q_{66} = G_{12}, \quad \nu_{21} = \nu_{12}\frac{E_2}{E_1}$$

and the constitutive equation can be written as:

$$\{\sigma\}_1 = [Q]\{\epsilon\}_1 \quad (4.20)$$

With this micromechanical models (Rule of Mixtures for E_1 and ν_{12} , and Halpin-Tsai equations for E_2 and G_{12}) the engineering constants of the lamina can be determined from the properties of the constituent fiber and matrix materials. The specific material properties for the constituents used are presented in Table 4.9.

Table 4.9: HR40 and T800 material properties.

Property	Symbol	HR40	T800	Unit
Longitudinal Young's modulus	E_{fL}	395	294 [214]	GPa
Transverse Young's modulus	E_{fT}	26.8 [84]	14.0 [61]	GPa
In-plane shear modulus	G_{fLT}	4.6 [84]	8.7 [61]	GPa
Major Poisson's ratio	ν_{fLT}	0.23* [84]	0.226 [61]	-
Longitudinal CTE	α_{fL}	-	-0.4 [214]	$\times 10^{-6}/^{\circ}C$

* Approximated from a similar carbon fiber.

Regarding the Poisson's ratio, their precise values have a relatively minor impact on the determination of other elastic properties. A sensitivity study by Duan et al. [61] demonstrated that for the IMS65 carbon fiber, varying the major Poisson's ratio from 0.2 to 0.1 or 0.3 resulted in changes of only approximately ± 0.1 GPa for the transverse Young's modulus and ± 0.2 GPa for the shear modulus. Given this, the influence of approximations in the assumed Poisson's ratio values is negligible.

As for the longitudinal and transverse Coefficient of Thermal Expansion (CTE), α_1 and α_2 , respectively, these are calculated using Schapery's equations [2, 183] applicable to unidirectional composites.

$$\alpha_1 = \frac{1}{E_1} (E_{fL} V_f \alpha_{fL} + E_m V_m \alpha_m) \quad (4.21)$$

where α_{fL} is the fiber longitudinal CTE.

$$\alpha_2 = (1 + \nu_{fLT})\alpha_{fT}V_f + (1 + \nu_m)\alpha_mV_m - \nu_{12}\alpha_1 \quad (4.22)$$

where α_{fT} is the fiber transverse CTE.

Carbon fibers exhibit significant anisotropy in their thermal expansion:

- **Longitudinal CTE (α_{fL}):** Typically small and often negative (contraction upon heating), particularly in high-modulus fibers due to the alignment of graphitic planes.
- **Transverse CTE (α_{fT}):** Generally positive and much larger than α_L , comparable to polymer matrices.

These values are dependent on several factors like material (PAN, Pitch), heat treatment temperature and the resulting Young's modulus.

A review of literature provides several data points for the longitudinal Young's modulus and the corresponding longitudinal CTE for various carbon fibers. These values, gathered from multiple sources, are presented in Table 4.10.

Table 4.10: Properties of various carbon fibers from literature

Fiber Name	E_{fL} (GPa)	α_{fL} ($10^{-6}/^{\circ}\text{C}$)	α_{fT} ($10^{-6}/^{\circ}\text{C}$)	Modulus Class.	Ref.
T300	233.0	-0.54	10.08	HT	[30]
T650-35	243.0	-0.84	7.8	HT	[174]
IM7	276.0	-0.40	5.60	HT	[128]
T1000	290.0	-1.40	3.80	IM	[128]
M40J	377.0	-0.83	N/A	HM	[174]
P55	379.0	-0.30	11.50	HM	[128]
HMS	379.2	-0.99	6.84	HM	[30]
P75	550.2	-1.35	6.84	HM	[30]
M60J	588.0	-1.10	N/A	HM	[174]
P100	796.3	-1.40	6.84	UHM	[30]
P120 (Al)	787.6	-1.62	N/A	UHM	[228]
P140 (Al)	841.4	-1.593	N/A	UHM	[228]
P120 (Epoxy)	848.3	-1.62	N/A	UHM	[228]
P140 (Epoxy)	965.5	-1.66	N/A	UHM	[228]

N/A indicates data not available in the cited source.

Modulus Classification: HT (High Tensile, 200-280 GPa), IM (Intermediate Modulus, 280-350 GPa), HM (High Modulus, 350-600 GPa), UHM (Ultra-High Modulus, 600-~950 GPa).

A comprehensive study by Wolff [228] investigated the relationship between the axial stiffness and axial CTE for a range of carbon fibers, including PAN-based and pitch-based fibers. Wolff's work demonstrated a clear decreasing trend of α_L with increasing E_L .

Later, Schulgasser [186] proposed a theoretical model to predict the axial CTE of carbon fibers based on the thermoelastic properties. The equation derived is:

$$\alpha_{f_L} = \left(-2.09 + \frac{499}{E_{f_L}} \right) \times 10^{-6} / ^\circ\text{C} \quad (4.23)$$

where E_{f_L} is the longitudinal Young's modulus in GPa.

Figure 4.12 illustrates this relationship, incorporating data points visually extracted from Wolff's study, the literature data from Table 4.10 and the theoretical model.

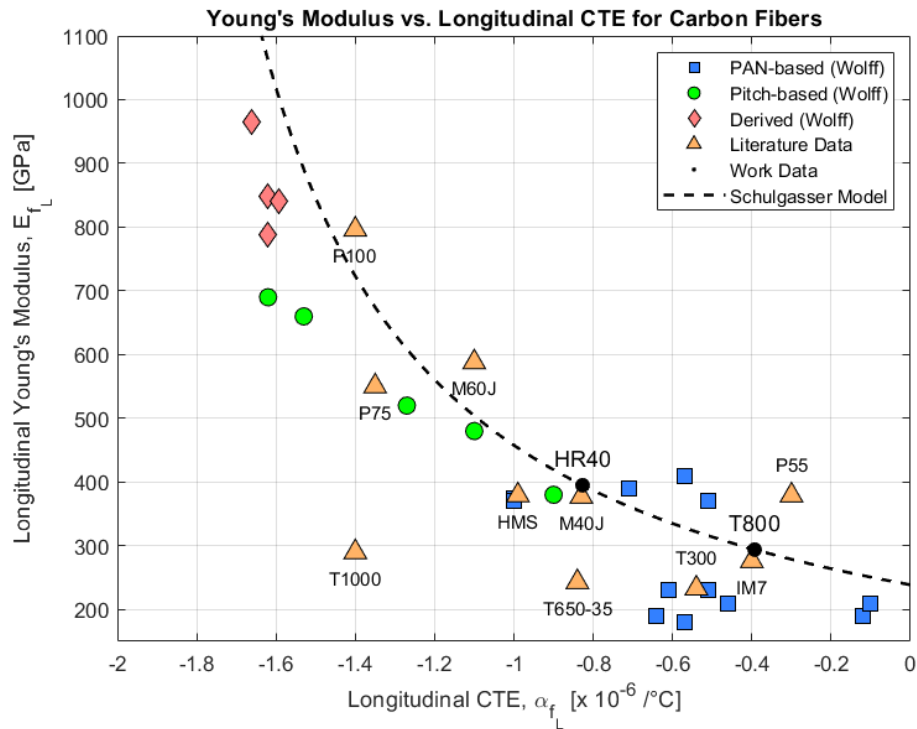


Figure 4.12: Young's Modulus (E_L) vs. Longitudinal CTE (α_{f_L}) for Carbon Fibers.

As it can be seen, Schulgasser's theoretical curve provides a good approximation of the experimental data presented by Wolff. Furthermore, even when accounting for the additional data points the theoretical model continues to show a strong correlation with the general trend. As such, using Equation 4.23, the longitudinal CTE for T800 and HR40 result in -0.827 and -0.393 , respectively. The predicted value of the later fiber was already available in the manufacturers datasheet [214] but as a form of model validation was calculated and the results are equal.

While the longitudinal CTE of carbon fibers is relatively well characterized, the transverse CTE is not as straightforward and no model is available. Instead of being directly correlated with the longitudinal Young's modulus it is more sensitive to other properties such as porosity and the degree of ordering and spacing of graphitic layers in the transverse direction.

Due to these, predicting α_{f_T} is more challenging, and experimental values can show considerable scatter depending on the specific fiber type and measurement technique (as shown in Figures 4.13

and 4.14).

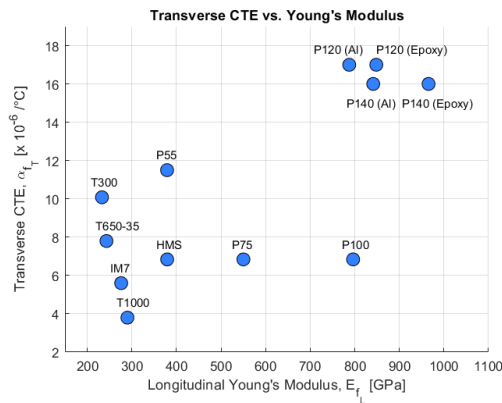


Figure 4.13: Transverse CTE (α_{fT}) versus longitudinal Young's modulus (E_{fL}).

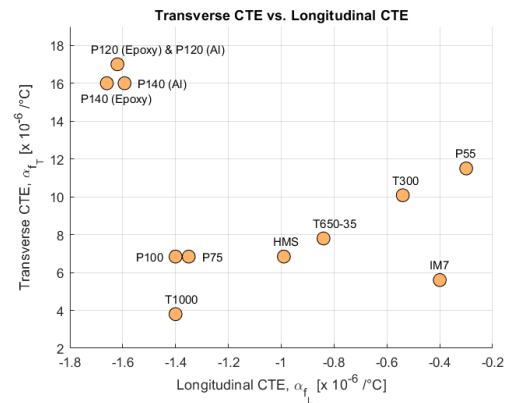


Figure 4.14: Transverse CTE (α_{fT}) versus longitudinal CTE (α_{fL}).

Consequently, estimating α_{fT} for specific dry fibers like HR40 and T800, in the absence of direct manufacturer data, is done by selecting PAN-based carbon fibers with similar fiber properties. In the case of the T800 the analagous fibers are the IM7 and T1000. As for the HR40 one can use the HMS and the M40J. Resulting in the values $7.0 \times 10^{-6}/^{\circ}\text{C}$ and $4.5 \times 10^{-6}/^{\circ}\text{C}$, respectively.

As for the matrix, the properties required are listed in Table 4.11 with the CTE value for the neat resin being an estimate.

Table 4.11: TP402 epoxy thermomechanical properties.

Property	Symbol	Value	Unit	Ref
Tensile Modulus	E_m	3.44	GPa	[158]
Poisson's ratio	ν_m	0.35	-	[76, 77, 163]
CTE	α_m	35-48	$\times 10^{-6}/^{\circ}\text{C}$	[19, 176]

Given the range of CTE values found for TP402 epoxy or similar systems in the literature, an intermediate value of $\alpha_m = 41.5 \times 10^{-6}/^{\circ}\text{C}$ is selected for this analysis. This value is also in line with data for other epoxy systems, such as the $44.9 \times 10^{-6}/^{\circ}\text{C}$ reported by Nawab et al. [152].

The temperature variation is determined from the cure cycle of the TP402 epoxy matrix, as displayed in Figure 4.15, which is based on the manufacturer's recommendations [158]. The material is considered to cool from the final curing temperature to room temperature. With a final cure temperature of $T_{cure} = 135^{\circ}\text{C}$ and a final room temperature of $T_{final} = T_o = 25^{\circ}\text{C}$, the temperature variation is $\Delta T = T_o - T_{cure} = -110^{\circ}\text{C}$.

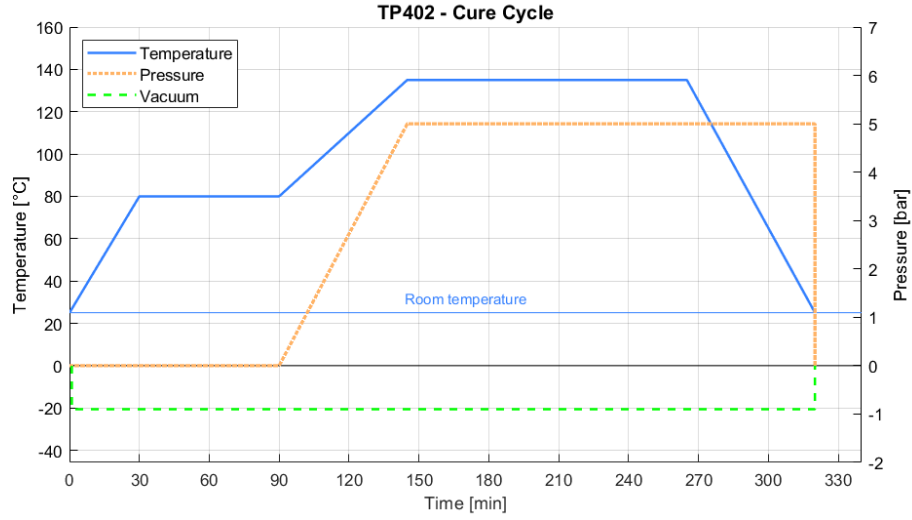


Figure 4.15: Autoclave curing cycle of the TP412 epoxy matrix.

With the effective properties of each lamina type established, CLPT is used to determine the response of the laminate to applied loads, including thermal loads resulting from a temperature change. The standard CLPT formulation according to the NASA Reference Publication 1351 by Nettles [154] is reproduced here for completeness.

In this method, ply 1 corresponds to the top ply of the laminate and the z -coordinate is zero at the laminate mid-plane, with positive z pointing upwards.

- **Coordinate transformation for stresses and strains:**

Stresses and strains in a ply are often determined in the global laminate coordinate system (x, y) but are frequently required in the local ply principal material coordinate system $(1, 2)$. For a ply oriented at an angle θ_k with respect to the global x -axis, the transformation of stresses from global $\{\sigma\}_x = [\sigma_x, \sigma_y, \tau_{xy}]^T$ to local $\{\sigma\}_1 = [\sigma_1, \sigma_2, \tau_{12}]^T$ is given by:

$$\{\sigma\}_1 = [T]_k \{\sigma\}_x \quad (4.24)$$

where $[T]_k$ is the stress transformation matrix for ply k :

$$[T]_k = \begin{bmatrix} c^2 & s^2 & 2cs \\ s^2 & c^2 & -2cs \\ -cs & cs & c^2 - s^2 \end{bmatrix}_k \quad (4.25)$$

with $c = \cos(\theta_k)$ and $s = \sin(\theta_k)$.

To transform stresses from the local $(1, 2)$ system back to the global (x, y) system, the inverse of the stress transformation matrix, $[T]_k^{-1}$, is utilized:

$$\{\sigma\}_x = [T]_k^{-1} \{\sigma\}_1 \quad (4.26)$$

Similarly, engineering strains in the local system, $\{\varepsilon\}_1 = [\varepsilon_1, \varepsilon_2, \gamma_{12}]^T$, can be obtained from the global engineering strains, $\{\varepsilon\}_x = [\varepsilon_x, \varepsilon_y, \gamma_{xy}]^T$ and vice-versa.

- **Transformed reduced stiffness matrix:**

To relate global stresses to global engineering strains, for an individual ply k , the local reduced stiffness matrix $[Q]_k$ is transformed to the global laminate coordinate system based on the ply orientation angle θ_k .

It's useful to first note the transformation for tensorial strains in the following form:

$$[\bar{Q}]_k = [T]_k^{-1} [Q]_k [T]_k \quad (4.27)$$

However, when working with engineering strains, the transformed reduced stiffness matrix is calculated differently:

$$[\bar{Q}]_k = [T]_k^{-1} [Q]_k [T']_k \quad (4.28)$$

where,

$$[T']_k = \begin{bmatrix} c^2 & s^2 & cs \\ s^2 & c^2 & -cs \\ -2cs & 2cs & c^2 - s^2 \end{bmatrix}_k \quad (4.29)$$

The resulting transformed stiffness matrix for ply k is then:

$$[\bar{Q}]_k = \begin{bmatrix} \bar{Q}_{11} & \bar{Q}_{12} & \bar{Q}_{16} \\ \bar{Q}_{12} & \bar{Q}_{22} & \bar{Q}_{26} \\ \bar{Q}_{16} & \bar{Q}_{26} & \bar{Q}_{66} \end{bmatrix}_k \quad (4.30)$$

The explicit formulas for these components are detailed in [154] and correspond to the engineering strains formulation. As a note, while Nettles does not explicitly write Equation 4.28 using a separate symbol $[T']_k$, the component formulas are the direct result of this type of transformation.

- **Laminate stiffness matrices:** These matrices are defined by integrating the transformed ply stiffnesses through the laminate thickness:

$$[A] = \sum_{k=1}^n [\bar{Q}]_k (z_k - z_{k-1}) \quad (4.31)$$

$$[B] = \sum_{k=1}^n \frac{1}{2} [\bar{Q}]_k ((z_k)^2 - (z_{k-1})^2) \quad (4.32)$$

$$[D] = \sum_{k=1}^n \frac{1}{3} [\bar{Q}]_k ((z_k)^3 - (z_{k-1})^3) \quad (4.33)$$

where z_k and z_{k-1} are the z-coordinates of the top and bottom surfaces of ply k , respectively, and n is the total number of plies.

• **Laminate constitutive equations:**

The relationship between the applied mechanical force $\{N\}$, moment $\{M\}$ and the laminate mid-plane strains $\{\epsilon^0\}$ and curvatures $\{\kappa\}$ is given by:

$$\begin{Bmatrix} N_x \\ N_y \\ N_{xy} \\ M_x \\ M_y \\ M_{xy} \end{Bmatrix} = \begin{bmatrix} A_{11} & A_{12} & A_{16} & B_{11} & B_{12} & B_{16} \\ A_{12} & A_{22} & A_{26} & B_{12} & B_{22} & B_{26} \\ A_{16} & A_{26} & A_{66} & B_{16} & B_{26} & B_{66} \\ B_{11} & B_{12} & B_{16} & D_{11} & D_{12} & D_{16} \\ B_{12} & B_{22} & B_{26} & D_{12} & D_{22} & D_{26} \\ B_{16} & B_{26} & B_{66} & D_{16} & D_{26} & D_{66} \end{bmatrix} \begin{Bmatrix} \epsilon_x^0 \\ \epsilon_y^0 \\ \gamma_{xy}^0 \\ \kappa_x \\ \kappa_y \\ \kappa_{xy} \end{Bmatrix} \quad (4.34)$$

Or, written in contracted matrix form:

$$\begin{Bmatrix} \{N\} \\ \{M\} \end{Bmatrix} = \begin{bmatrix} [A] & [B] \\ [B] & [D] \end{bmatrix} \begin{Bmatrix} \{\epsilon^0\} \\ \{\kappa\} \end{Bmatrix} \quad (4.35)$$

The equations above represent the standard formulation of CLPT for determining the mechanical response of a laminate. To apply this theory to the specific case of calculating residual stresses due to thermal effects, it is necessary to consider the strains induced by temperature changes and the resulting internal forces and moments within the laminate. This involves defining the ply thermal expansion coefficients and then calculating the equivalent thermal loads on the laminate.

As previously mentioned, the laminate cools by ΔT from its cure temperature. Note that in the local lamina CTE vector $\{\alpha_{local}\}_k = [\alpha_{1,k}, \alpha_{2,k}, \alpha_{12,k}]^T$, $\alpha_{12,k}(T)$ is the shear CTE component which is typically assumed to be zero for orthotropic materials and the other components are derived from the Schapery's equations.

Several factor affect the coefficient of thermal expansion, including:

- Fiber Volume
- Void Volume
- Layup Angle
- Fabric Skewness
- Stacking Sequence
- Thermal Cycling
- Temperature Dependence
- Moisture Effects
- Viscoelasticity

One of the most relevant among these is the temperature dependence. This means that at different temperatures the CTE itself alters. Johnson, Kural, and Mackey [107] note that the common practice of reporting CTE as a single quantity can lead to significant errors in composite design due to this temperature dependence, which is mainly caused by changes in the resin system. They recommend that CTE values should be obtained from thermal expansion test data for the

specific design temperature range, as the CTE is the slope of the thermal strain-temperature curve between two temperatures.

To account for this, the thermal strain vector for ply k in its local coordinates due to a temperature change from $T_{initial}$ to T_{final} , can be obtained by integrating the local CTE vector:

$$\{\epsilon_{local}^{thermal}\}_k = \int_{T_{initial}}^{T_{final}} \{\alpha(T)_{local}\}_k dT = \begin{bmatrix} \int_{T_{initial}}^{T_{final}} \alpha_{1,k}(T) dT \\ \int_{T_{initial}}^{T_{final}} \alpha_{2,k}(T) dT \\ \int_{T_{initial}}^{T_{final}} \alpha_{12,k}(T) dT \end{bmatrix} \quad (4.36)$$

This integral accounts for the continuous change in the material's expansion or contraction rate as the temperature varies. However, if the temperature dependency of the CTE is negligible for the specific material and temperature interval or if an appropriate average CTE value is used, the thermal strain calculation simplifies to the more common form [107].

$$\{\epsilon_{global}^{thermal}\}_k \approx \{\alpha_{global}\}_k \Delta T \quad (4.37)$$

For the present analysis of carbon-epoxy systems, considering the cool-down from 125°C to room temperature – which is below the typical glass transition temperature where CTE variation is most pronounced for the epoxy matrix – the assumption of using effective constant ply CTEs is employed. Experimental data often shows that for many carbon/epoxy systems, the CTE (slope of thermal strain vs temperature) in temperature ranges below the expected T_g of the resin appears relatively constant or can be represented by a single average value. Therefore, the simplified local thermal strain is used [107, 228, 236].

This local thermal strain vector, in ply k , is then transformed into the global laminate coordinate system using the inverse of the transformation matrix $[T']_k$:

$$\{\epsilon_{global}^{thermal}\}_k = ([T']_k)^{-1} \{\epsilon_{local}^{thermal}\}_k \quad (4.38)$$

From this global thermal strain vector for each ply, $\{\epsilon_{global}^{thermal}\}_k$, the thermal force and moment vectors are calculated as:

$$\{N^{thermal}\} = \sum_{k=1}^n [\bar{Q}]_k \{\epsilon_{global}^{thermal}\}_k (z_k - z_{k-1}) \quad (4.39)$$

$$\{M^{thermal}\} = \sum_{k=1}^n \frac{1}{2} [\bar{Q}]_k \{\epsilon_{global}^{thermal}\}_k ((z_k)^2 - (z_{k-1})^2) \quad (4.40)$$

where $\{N^{thermal}\}$ and $\{M^{thermal}\}$ represent the net forces and moments that would be required to suppress all thermal strains in the laminate.

When considering only thermal effects and no external mechanical loads, hence $\{N\} = \{0\}$ and $\{M\} = \{0\}$, the general constitutive relation is modified to include thermal resultants.

$$\begin{bmatrix} [A] & [B] \\ [B] & [D] \end{bmatrix} \begin{Bmatrix} \{\epsilon^0\} \\ \{\kappa\} \end{Bmatrix} = \begin{Bmatrix} \{N^{thermal}\} \\ \{M^{thermal}\} \end{Bmatrix}$$

Inverting, the mid-plane strains and curvatures are found:

$$\begin{Bmatrix} \{\epsilon^0\} \\ \{\kappa\} \end{Bmatrix} = \begin{bmatrix} [A] & [B] \\ [B] & [D] \end{bmatrix}^{-1} \begin{Bmatrix} \{N^{thermal}\} \\ \{M^{thermal}\} \end{Bmatrix} \quad (4.41)$$

where $\{\epsilon^0\} = [\epsilon_x^0, \epsilon_y^0, \gamma_{xy}^0]^T$ and $\{\kappa\} = [\kappa_x, \kappa_y, \kappa_{xy}]^T$.

The global strain vector at the mid-plane ($z_{mid,k}$) of ply k is:

$$\{\epsilon_{global}\}_k = \{\epsilon^0\} + z_{mid,k} \cdot \{\kappa\} \quad (4.42)$$

Finally, the global and local residual stress vectors in ply k are then obtained from:

$$\{\sigma_{global}\}_k = [\bar{Q}]_k (\{\epsilon_{global}\}_k - \{\epsilon_{global}^{thermal}\}_k) \quad (4.43)$$

$$\{\sigma_{local}\}_k = [T]_k \{\sigma_{global}\}_k \quad (4.44)$$

To validate the presented formulation, a standard laminate configuration can be analyzed. The properties of each ply material are listed in Table 4.12 followed by the layup and the corresponding residual stresses are displayed in Table 4.13.

Table 4.12: Lamina Material Properties.

Material	V_f	E_1 (GPa)	E_2 (GPa)	G_{12} (GPa)	ν_{12}	α_1 ($10^{-6}/^\circ\text{C}$)	α_2 ($10^{-6}/^\circ\text{C}$)
T800	0.53	158.60	7.26	2.96	0.284	-0.399	30.8
HR40	0.53	211.36	9.46	2.37	0.286	-0.0732	29.2

Generally, residual stresses in carbon-epoxy composites are predicted to be in the range of 20 to 60 MPa, which is similar to values reported for other composite systems [4, 236]. These stress levels can represent approximately 5 to 10% of the expected laminate strength. In such laminates, the 0° plies are often found to be loaded in tension in the longitudinal direction, which can negatively impact the overall laminate strength [32].

Table 4.13: Residual stresses prediction for the [T800 0° / HR40 0° / T800 90° / HR40 90°] laminate

Material	Orientation °	FAW g/m^2	Ply #	z	Longitudinal MPa	Transverse MPa
T800	0.0	60	1	Upper	77.62	12.80
				Middle	33.16	14.19
				Bottom	-11.30	15.59
HR40	0.0	60	2	Upper	-8.00	18.76
				Middle	-67.26	20.57
				Bottom	-126.51	22.38
T800	90.0	60	3	Upper	-109.23	18.67
				Middle	-66.62	17.17
				Bottom	-24.01	15.67
HR40	90.0	60	4	Upper	-24.94	18.87
				Middle	31.86	16.93
				Bottom	88.65	14.98

The analysis of the values in Table 4.13 indicates they are within the expected range. It is also important to consider that the mismatch of residual stress between 0° ply-blocks and 90° may promote delamination and secondary damage in these laminates, or in some cases, change the conditions for the onset of damage [32].

This methodology for calculating residual thermal stresses post-cure using CLPT relies on several assumptions. First, this approach primarily considers the laminate to be fully cured prior to the temperature change that induces these stresses. The elastic properties of the individual fully cured plies are taken as constant although, in reality, these can exhibit temperature dependence. Furthermore, no plastic deformation of the constituents as a result of these thermal stresses is considered and also time-dependent effects like viscoelasticity are negligible over this cool-down period.

While this approach, offers a way to assess post-cure thermal stresses, it is important to recognize that the overall development of residual stresses in composite materials is much more complex. More comprehensive models, such as those presented by Gonçalves et al. [78] and Danzi [53], are able to capture the entire manufacturing process, from the resin's initial gel state through to the final cured solid. These advanced models are capable of incorporating additional factors such as chemical shrinkage of the resin during polymerization, progressive evolution of the matrix's mechanical properties as functions of both the degree of cure and the instantaneous temperature and elasto-plastic deformations within the matrix material.

4.1.4 Investigated Material Configurations

With the selection methodology established, it is now possible to make a contextualized choice of valid hybrid configurations for analysis. As introduced in Chapter 2, interply hybridization is a strategy that combines layers of different fiber materials into a single laminate. This approach is generally the easiest to manufacture however it also results in the least amount of dispersion compared to other hybrid architectures.

As mentioned, the materials used in this study are based on the ThinPreg™ 80EP-402 prepreg system. For interply configurations, specimens are produced directly from full-width prepreg rolls. This material is supplied on a silicon paper backing, wound into 300 mm wide rolls. The standard UD preregs, T800/80EP-402 and HR40/80EP-402, are commercially available in a range of fiber areal weights. For this investigation, two FAWs were considered for both the high-strain and low-strain materials: 30 gsm and 60 gsm.

Combining the available materials and layups generates a matrix of 12 potential configurations. The properties of each, including the coordinates required for the damage mode map analysis, are calculated and presented in Table 4.14.

Table 4.14: Calculation of mapping coordinates for all 12 potential hybrid configurations.

Configuration ID	Layup	T800 FAW (gsm)	HR40 FAW (gsm)	Total LE FAW (Y-axis)	Total Laminate FAW (gsm)	LE Proportion (γ) (X-axis)
H1-T ₃₀ -H ₃₀	[T/H/T]	30	30	30	90	0.33
H1-T ₃₀ -H ₆₀	[T/H/T]	30	60	60	120	0.50
H1-T ₆₀ -H ₃₀	[T/H/T]	60	30	30	150	0.20
H1-T ₆₀ -H ₆₀	[T/H/T]	60	60	60	180	0.33
H2-T ₃₀ -H ₃₀	[2T/H/2T]	30	30	30	150	0.20
H2-T ₃₀ -H ₆₀	[2T/H/2T]	30	60	60	180	0.33
H2-T ₆₀ -H ₃₀	[2T/H/2T]	60	30	30	270	0.11
H2-T ₆₀ -H ₆₀	[2T/H/2T]	60	60	60	300	0.20
H3-T ₃₀ -H ₃₀	[3T/2H/3T]	30	30	60	240	0.25
H3-T ₃₀ -H ₆₀	[3T/2H/3T]	30	60	120	300	0.40
H3-T ₆₀ -H ₃₀	[3T/2H/3T]	60	30	60	420	0.14
H3-T ₆₀ -H ₆₀	[3T/2H/3T]	60	60	120	480	0.25

To achieve the desired pseudo-ductile behavior, the fragmentation stress of the LE layers must be lower than the stresses that would cause delamination or premature failure of the HE layers. This corresponds to zone 3 (stable fragmentation) and zone 4 (fragmentation with delamination) in the damage mode map shown in Figure 4.16.

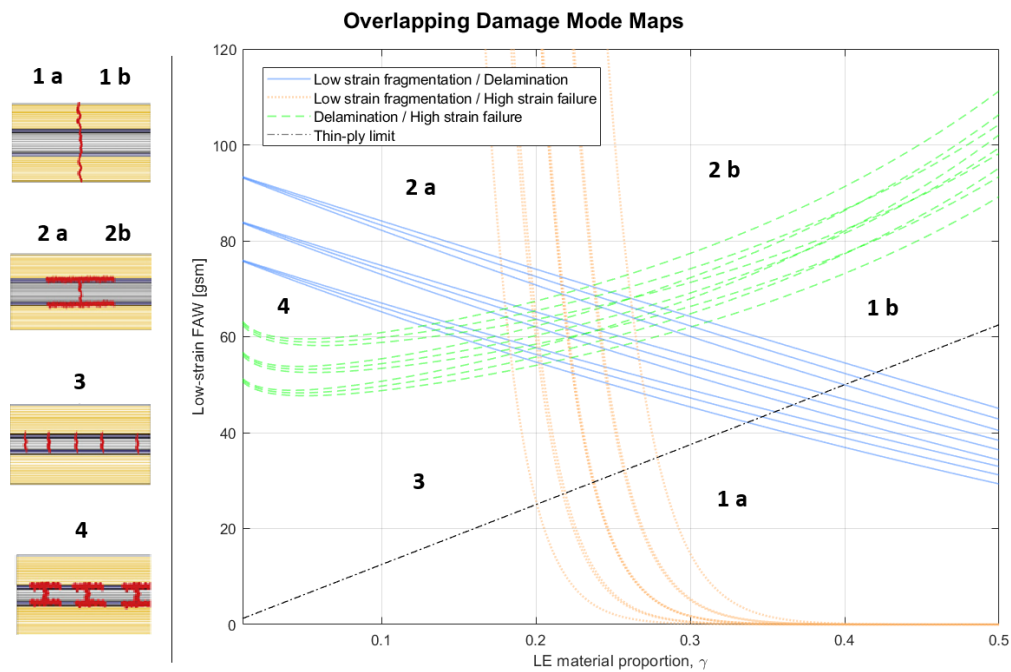


Figure 4.16: Overlapping damage mode map for T800-HR40 hybrid configuration with associated failure scenarios. The bands for each boundary represent the variability due to a range of fiber volume fractions.

It is important to note, however, that the boundaries between these failure zones are not deterministic. A significant source of variability in composite properties is the fiber volume fraction (V_f) of the constituent plies, which can fluctuate within manufacturing tolerances. As these properties are direct inputs to the analytical model, any variation in V_f will shift the calculated position of the failure mode boundaries. Consequently, a single set of lines represents an idealization. A more robust approach is to consider the envelope of possible boundaries. The overlapping damage mode map shown in Figure 4.16 is constructed using this principle and it superimposes the damage mode boundaries calculated for a range of expected V_f values for both the LE and HE materials. This transforms the sharp boundaries into wider transition regions, visually representing the model's sensitivity to material variability.

If the objective were also to improve fracture toughness, then the dissipation of energy should also be maximized, for instance through an increase of the fiber pull-out length. This is related to delaying the onset of fragmentation as much as possible to not jeopardize the ply-thickness advantage with a low-strain onset of damage. While the primary focus here is on achieving pseudo-ductility, this secondary consideration suggests that configurations located deeper within the desired zones, away from the fragmentation boundaries, may offer superior overall performance.

Using this map, each of the 12 potential configurations was screened by plotting its coordinates, as illustrated in Figure 4.17. The outcome of this evaluation, including the predicted failure mode for each case, is presented in detail in Table 4.15.

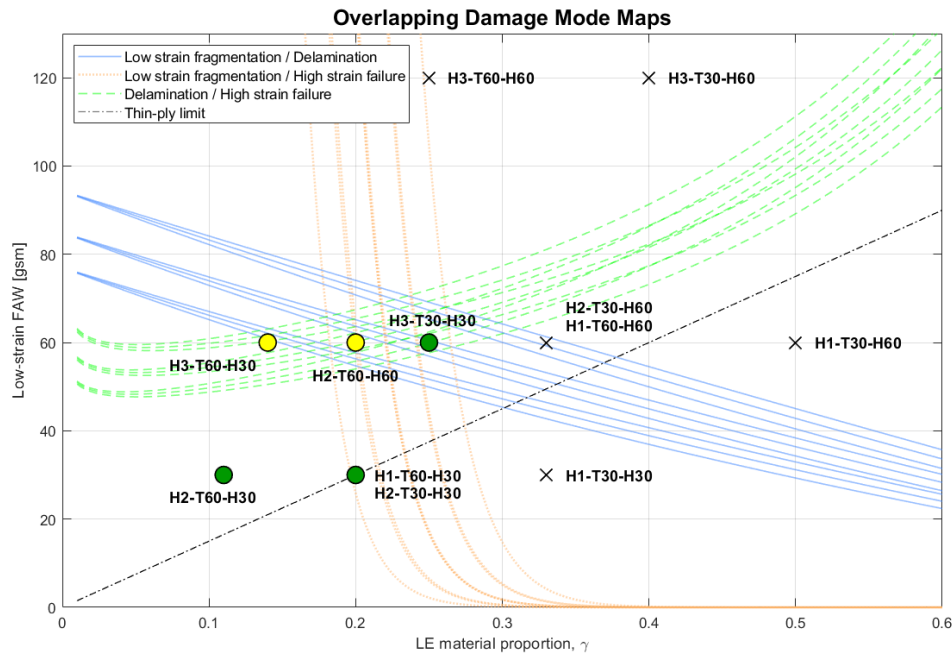


Figure 4.17: Screening of 12 potential hybrid configurations on the damage mode map.

The initial 12 configurations can be filtered using a series of constraints derived from the damage mode map and practical limits.

The first constraint relates to the LE prepreg FAW. As indicated by the damage mode map, using the 60 gsm HR40 material is highly restrictive, as it requires the LE proportion (γ) to be kept below 0.2 to ensure pseudo-ductility. This would require an impractically thick HE layer, leading to a total laminate FAW of over 300 gsm. Therefore, configurations based on the 60 gsm LE prepreg are deemed unsuitable. By extension, configurations with an even higher total LE FAW, namely H3-T₃₀-H₆₀ and H3-T₆₀-H₆₀, are also excluded.

A second, practical constraint is an upper limit of 300 gsm on the total laminate FAW. This filter eliminates the H3-T₆₀-H₃₀ configuration. These filters narrow the focus to the remaining candidates, all of which use the more versatile 30 gsm LE material. For this subset, a final constraint is applied to maintain a safe margin from the brittle failure boundary: the LE proportion must not exceed 0.25. As seen in Figure 4.16, this rule removes the H1-T₃₀-H₃₀ configuration, as its γ of 0.33 is above this threshold. The results of this filtering process are summarized in Table 4.15.

Table 4.15: Final validity assessment of potential hybrid configurations.

Configuration ID	Valid	Zone	Reason
H1-T ₃₀ -H ₃₀	No	1a	LE proportion ($\gamma > 0.25$)
H1-T ₃₀ -H ₆₀	No	1b	Unsuitable LE FAW (60 gsm)
H1-T ₆₀ -H ₃₀	Yes	3	
H1-T ₆₀ -H ₆₀	No	1b	Unsuitable LE FAW (60 gsm)
H2-T ₃₀ -H ₃₀	Yes	3	
H2-T ₃₀ -H ₆₀	No	1b	Unsuitable LE FAW (60 gsm)
H2-T ₆₀ -H ₃₀	Yes	3	
H2-T ₆₀ -H ₆₀	No	3	Unsuitable LE FAW (60 gsm)
H3-T ₃₀ -H ₃₀	Yes	-	
H3-T ₃₀ -H ₆₀	No	1b	Unsuitable LE FAW (120 gsm)
H3-T ₆₀ -H ₃₀	No	3	Total FAW exceeds limit (>300 gsm)
H3-T ₆₀ -H ₆₀	No	1b	Unsuitable LE FAW (120 gsm)

This filtering process therefore yields four distinct configurations that are considered viable:

- H1-T₆₀-H₃₀
- H2-T₃₀-H₃₀
- H2-T₆₀-H₃₀
- H3-T₃₀-H₃₀

The positions of these four viable configurations on the damage mode map also allow for a test of the model's predictions for distinct behaviors:

- **H1-T₆₀-H₃₀ and H2-T₃₀-H₃₀:** These configurations are at the same location on the map, close to the boundary for brittle failure. Their position enables a test of the model's predictions at the edge of the design space, where a pseudo-ductile response is not guaranteed.
- **H2-T₆₀-H₃₀:** This layup lies within the pseudo-ductile region, far from the boundaries. As such, the pseudo-ductile failure mode is predicted, making this configuration a baseline for behavior in this region.
- **H3-T₃₀-H₃₀:** This configuration is located closer to the boundary where delamination is predicted to occur with fragmentation. Its inclusion allows for the investigation of a mixed-mode failure behavior, while also representing a different layup architecture.

4.2 Analysis of a Benchmark Interply Hybrid Study

To contextualize the insights from the selection methodology, this section reviews and analyzes the experimental results from the benchmark interply hybrid study conducted by Danzi [53]. The analysis of this benchmark is particularly relevant as the constituent materials (HR40 and T800 fibers) and ply-level architectures are equivalent to the viable configurations identified in the previous selection. This allows for a direct observation of the mechanical behavior of configurations designed using similar principles. For complete details regarding specimen production, testing setups, and data reduction techniques, the reader is referred to the original work [53].

This research was subsequently extended by Tavares [209], who used the same constituent materials to explore the effects of interply hybridization in QI laminates. That experimental campaign included plain strength tension and compression, as well as open-hole tension and open-hole compression tests. The present analysis, however, focuses on Danzi's original study, which is essential for understanding the fundamental lamina and sublamine properties that govern these more complex behaviors.

The primary purpose of Danzi's experimental campaign was to determine the lamina mechanical properties for hybrid and non-hybrid thin-ply carbon composites, in order to analyze the benefits and drawbacks of the ply-by-ply hybridization technique and to compare thin-ply composites to standard thickness laminates. The study investigated three specific interply hybrid configurations as well as the non-hybrid constituent materials. The layups were fabricated using the HR40 and T800 fiber types, albeit with a different epoxy resin system (ThinPreg™ 80EP-736) and different nominal ply areal weights (20 gsm for HR40 and 50 gsm for T800). The configurations, nominal thicknesses, and the measured effective cured thicknesses are detailed in Table 4.16.

Table 4.16: Plies nominal thicknesses and effective cured thicknesses (data from Danzi [53]).

Configuration	Nominal Thickness (μm)	Cured effective Thickness (μm)
T800	50	54
HR40	20	24
[T800/HR40/T800]	120	130
[2T800/HR40/2T800]	220	241
[3T800/2HR40/3T800]	340	374

The scope of the combined experimental campaigns is summarized in Table 4.17, detailing the test types and specific laminate layups used to determine a range of mechanical properties. The symbols for the measured properties correspond to the longitudinal tensile modulus (E_{1T}), tensile strength (X_T), and major Poisson's ratio (ν_{12}), the transverse tensile modulus (E_{2T}) and strength (Y_T), the in-plane shear modulus (G_{12}) and strength (S_{12}), and the Mode I fracture toughness (G_{Ic}), followed by the un-notched laminate tensile (X_T^L) and compressive (X_C^L) strengths, and the open-hole tensile (X_T^{OHT}) and compressive (X_C^{OHC}) strengths.

Table 4.17: Consolidated experimental plan, showing the lamina-level benchmark campaign by Danzi [53] and the subsequent campaign by Tavares [209].

Test Type	Configuration	Layup [†]	Thickness (mm) [*]	Properties
Benchmark Campaign (Danzi [53])				
Longitudinal Tensile	T800	[0] ₂₀	1.000	E_{1T}, X_T, ν_{12}
	HR40	[0] ₅₀	1.000	
	[T/H/T]	[0] ₈	0.960	
	[2T/H/2T]	[0] ₄	0.880	
	[3T/2H/3T]	[0] ₃	1.020	
Transverse Tensile	T800	[90] ₃₀	1.500	E_{2T}, Y_T
	HR40	[90] ₇₅	1.500	
	[T/H/T]	[90] ₁₂	1.440	
In-plane Shear	T800	[±45] _{25S}	2.700	G_{12}, S_{12}
	HR40	[±45] _{64S}	3.072	
	[T/H/T]	[±45] _{10S}	2.600	
DENT Fracture	T800	[90/0] _{20\$}	1.950	G_{Ic}
	HR40	[90/0] _{50\$}	1.980	
	[T/H/T]	[90/0] _{10\$}	2.280	
	[2T/H/2T]	[90/0] _{6\$}	2.420	
	[3T/2H/3T]	[90/0] _{4\$}	2.380	
Campaign Extension (Tavares [209])				
Plain Strength Tension	T800	[45/−45/0/45/−45/90/0] _S	1.404	X_T^L
	[T/H/T]	[45/−45/0/45/−45/90/0] _S	1.690	
	[2T/H/2T]	[45/−45/0/90] _S	1.687	
Plain Strength Comp.	T800	[45/−45/0/45/−45/90/0] _S	1.404	X_C^L
	[T/H/T]	[45/−45/0/45/−45/90/0] _S	1.690	
	[2T/H/2T]	[45/−45/0/90] _S	1.687	
Open-Hole Tension	T800	[45/−45/0/45/−45/90/0] _S	1.404	X_T^{OHT}
	[T/H/T]	[45/−45/0/45/−45/90/0] _S	1.690	
	[2T/H/2T]	[45/−45/0/90] _S	1.687	
Open-Hole Comp.	T800	[45/−45/0/45/−45/90/0] _S	1.404	X_C^{OHC}
	[T/H/T]	[45/−45/0/45/−45/90/0] _S	1.690	
	[2T/H/2T]	[45/−45/0/90] _S	1.687	

^{*} Total laminate thickness calculated using the nominal ply thicknesses, not the effective cured ones.

[†] The ‘\$’ symbol denotes a center-symmetric laminate where the last ply of the semi-laminate is not repeated.

All the experimental tests were conducted following established standards. The un-notched tensile tests (longitudinal, transverse, and plain strength) followed ASTM D-3039 [25], plain strength compression followed ASTM D-6484 [23], open-hole tension followed ASTM D-5766 [24], and in-plane shear followed ASTM D-3518 [22]. The DENT fracture tests were conducted according to the methodology proposed by Catalanotti et al. [38]. The results from these campaigns are presented and analyzed in the following subsections.

4.2.1 Longitudinal Tensile Properties

The longitudinal tensile response is the most relevant for evaluating the pseudo-ductile behavior of the hybrids thus is the first being presented.

According to Danzi, the tests were performed following the ASTM D-3039 [25] standard on an INSTRON 4208 testing machine equipped with a 200 kN load cell. The specimens consisted of UD laminates with all plies oriented at 0° , stacked to achieve a target thickness of approximately 1 mm. The specimens had a nominal rectangular geometry of $250\text{ mm} \times 15\text{ mm}$, with adhesively bonded aluminum tabs, as shown in Figure 4.18.

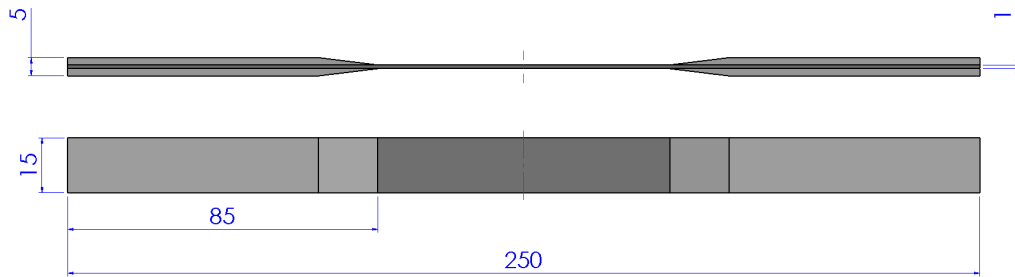


Figure 4.18: Illustration of a longitudinal tensile specimen. All dimensions are in millimeters.

A comparison of the nominal thickness, the effective thickness based on the cured ply thicknesses, and the mean measured thickness of the manufactured specimens is provided in Table 4.18.

Table 4.18: Nominal, effective cured, and mean measured thicknesses of the longitudinal tensile specimens.

Configuration	Layup	Nominal (mm)	Effective Cured (mm)	Mean Measured (mm)
T800	$[0]_{20}$	1.000	1.080	1.116
HR40	$[0]_{50}$	1.000	1.200	1.129
$[T/H/T]$	$[0]_8$	0.960	1.040	1.023
$[2T/H/2T]$	$[0]_4$	0.880	0.964	0.964
$[3T/2H/3T]$	$[0]_3$	1.020	1.122	1.124

The tests were conducted at room temperature under displacement control at a rate of 0.5 mm/min and the strain data was acquired using both rosette-type extensometers and Digital Image Correlation (DIC).

The mechanical properties listed in Table 4.19 were derived from the experimental data using the following standard definitions. The longitudinal tensile strength, X_T , was calculated as the maximum load (P_{MAX}) divided by the initial cross-sectional area (thickness t and width w):

$$X_T = \frac{P_{MAX}}{t \cdot w} \quad (4.45)$$

The longitudinal modulus, E_{1T} , was determined as the slope of the stress-strain curve in the standard strain range of 1000–3000 $\mu\epsilon$:

$$E_{1T} = \frac{\Delta\sigma_{11}}{\Delta\epsilon_{11}} \quad (4.46)$$

The major Poisson's ratio, ν_{12} , was evaluated from the linear portion of the stress-strain curve as the negative ratio of transverse to longitudinal strain:

$$\nu_{12} = -\frac{\epsilon_{22}}{\epsilon_{11}} \quad (4.47)$$

The results from the study showed a mix of behaviors. The thicker hybrid configurations ([2T/H/2T] and [3T/2H/3T]) exhibited the characteristic pseudo-ductile response, marked by an initial linear region up to the fragmentation of the HR40 layers, followed by a stress plateau and subsequent failure of the T800 layers. The thinnest hybrid, however, failed in a brittle manner. The mechanical properties for all configurations are summarized in Table 4.19.

Table 4.19: Longitudinal tensile properties of the benchmark configurations (data from Danzi [53]).

Configuration	Long. Modulus, E_{1T} (GPa)	Long. Strength, X_T (MPa)	Poisson's Ratio, ν_{12}
T800	148.36 $\pm$ 6.37	2923.8 $\pm$ 97.4	0.304 $\pm$ 0.062
HR40	192.60 $\pm$ 8.19	2157.6 $\pm$ 170.3	0.290 $\pm$ 0.040
[T/H/T]	154.72 $\pm$ 6.42	2369.1 $\pm$ 62.7	0.307 $\pm$ 0.059
[2T/H/2T]	144.26 $\pm$ 6.76	2640.4 $\pm$ 199.3	0.310 $\pm$ 0.120
[3T/2H/3T]	144.71 $\pm$ 9.14	2587.1 $\pm$ 151.3	0.337 $\pm$ 0.042

In his analysis of these results, Danzi noted that the longitudinal stiffness and strength of the hybrid systems are intermediate to those of the non-hybrid T800 and HR40 constituents. A key part of the study was to verify that the stiffness of the [T/H/T] configuration, in particular, respected the classical rule of mixtures. Using the lamina stiffness values from the non-hybrid tests and the reported ply thicknesses, the theoretical hybrid modulus, E_{th} , was calculated by the author as follows:

$$E_{th} = \frac{t_{T800} E_{T800} + t_{HR40} E_{HR40}}{t_H} \approx 156 \text{ GPa} \quad (4.48)$$

where t_{T800} and t_{HR40} are the total thicknesses of the respective materials in the hybrid, and t_H is the total hybrid thickness. This theoretical value of 156 GPa fits well within the experimental scatter of the measured average modulus of 154 GPa, thus validating the rule of mixtures for this configuration.

A critical finding of the study was the discrepancy between the prediction and the experimental result for the [T/H/T] configuration. While the damage mode map predicted pseudo-ductile

behavior, Danzi reported that this hybrid system, along with the non-hybrid T800 and HR40 systems, exhibited a brittle failure in longitudinal tension. The author highlights a key mechanistic detail for this behavior: before the final failure, the onset of transverse cracks derived from matrix failure due to the Poisson effect. This initial damage mode, possible in UD specimens, then led to the final catastrophic failure, which was characterized by an sudden release of stored elastic energy, producing an abrupt failure.

This discrepancy is also explained by manufacturing variability. Because the configuration was designed close to a transition boundary on the damage mode map, inherent property variations could shift some specimens from the intended pseudo-ductile region into the brittle failure zone. With a limited number of specimens, it is statistically plausible that this effect masked the desired behavior, a theory supported by Danzi's own observation that some samples in the series did perform as expected.

In addition to these effects, micromechanical mechanisms such as fiber-matrix decohesion – where a fiber break initiates a debond that propagates along the fiber length – could also contribute to a more catastrophic failure. This brittle behavior was characterized by an instantaneous release of stored elastic energy, resulting in an "abrupt explosion" of the sample with massive transverse cracking.

In contrast, the $[2T/H/2T]$ and $[3T/2H/3T]$ systems behaved as predicted, presenting a different failure scenario. The presence of materials with different stiffnesses led to a stable fragmentation process that reduced the speed of the final fracture. Post-mortem analysis of these specimens provided clear evidence of this mechanism, revealing cracks on the delaminated surfaces oriented orthogonally to the applied load. Danzi reported that these cracks showed a regular fragmentation pattern, with a measured distance between cracks of approximately 2 mm, confirming that stable fragmentation had been successfully achieved.

These behaviors observed at the lamina level are consistent with the trends reported in the subsequent study by Tavares [209] on QI laminates. In plain strength tension tests, it was similarly found that the addition of the brittle HR40 layers governed the ultimate failure of the material. The results from these tests are summarized in Table 4.20.

Table 4.20: Plain strength tension results for the QI laminates (data from Tavares [209]).

Configuration	Long. Modulus, E_{1T} (GPa)	Long. Strength, X_T^L (MPa)	Normalized Strength, X_T^L/E_{1T} (MPa/GPa)
T800	53.47	839.6 ± 53.4	15.70
$[T/H/T]$	55.08	640.8 ± 46.8	11.63
$[2T/H/2T]$	58.50	875.8 ± 86.1	14.97

As shown in the table, although the $[2T/H/2T]$ hybrid QI laminate showed a higher absolute strength, which can be attributed to its higher percentage of 0° plies, its strength normalized was lower than that of the T800 baseline.

4.2.2 Transverse Tensile Properties

The transverse tensile tests were conducted to characterize the matrix-dominated properties of the laminates.

The tests were performed following the ASTM D-3039 [25] standard. The specimens were composed of UD laminates with plies oriented at 90° , designed to a target thickness of 1.5 mm. The specimens had a nominal rectangular geometry of $175\text{ mm} \times 25\text{ mm}$ and included adhesively bonded aluminum tabs, as shown in Figure 4.19.

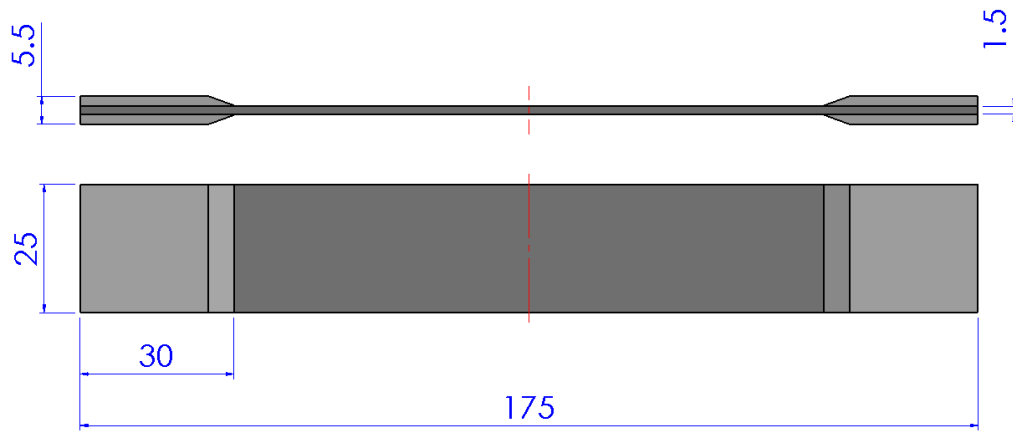


Figure 4.19: Illustration of a transverse tensile specimen. All dimensions are in millimeters.

The relevant thickness details for these specimens are summarized in Table 4.21.

Table 4.21: Nominal, effective cured, and mean measured thicknesses of the transverse tensile specimens.

Configuration	Layup	Nominal (mm)	Effective Cured (mm)	Mean Measured (mm)
T800	$[90]_{30}$	1.500	1.620	1.619
HR40	$[90]_{75}$	1.500	1.800	1.757
[T/H/T]	$[90]_{12}$	1.440	1.560	1.567

The experimental setup and data acquisition methods (load cell, extensometers, and DIC) were analogous to those used for the longitudinal tests. The transverse tensile strength, Y_T , was determined from the maximum load (P_{MAX}) and the cross-sectional area:

$$Y_T = \frac{P_{MAX}}{t \cdot w} \quad (4.49)$$

The transverse modulus, E_{2T} , was evaluated as the slope of the transverse stress-strain curve in the strain range of 1000–3000 $\mu\epsilon$:

$$E_{2T} = \frac{\Delta\sigma_{11}}{\Delta\epsilon_{11}} \quad (4.50)$$

For completeness, it should be noted that for the samples equipped with strain gauges, the longitudinal strain (ϵ_{11}) was calculated by converting the raw electrical signal using manufacturer-supplied conversion factors.

The results for the non-hybrid and the [T/H/T] hybrid configurations are summarized in Table 4.22.

Table 4.22: Transverse tensile properties of the benchmark configurations (data from Danzi [53]).

Configuration	Transverse Modulus, E_{2T} (GPa)	Transverse Strength, Y_T (MPa)
T800	8.01 ± 0.36	63.73 ± 8.24
HR40	6.56 ± 0.43	37.52 ± 1.21
[T/H/T]	8.21 ± 0.24	58.27 ± 3.74

In his discussion, Danzi highlighted the brittle nature of all materials under transverse loading, with abrupt failures and fractures developing orthogonally to the load direction. Two key findings from these tests can be observed.

First, the transverse stiffness of the hybrid configuration was unexpectedly higher than that of both non-hybrid constituents, a result that contradicts the classical rule of mixtures for transverse properties. A potential explanation for this phenomenon is a ply constraint effect. When the laminate is loaded transversely, a mismatch in the Poisson's ratio between the plies would cause them to contract differently in the through-thickness direction. Since they are bonded together, this mismatch may induce a complex triaxial stress state within each ply. This constraint on deformation makes the entire laminate act stiffer than its individual components, leading to the observed increase in the apparent transverse modulus.

Second, a negative trend in transverse strength with hybridization was observed: the highest strength was found for the non-hybrid T800, and this value decreased with the introduction of the HR40 material. While the author suggested this could be due to differences in ply thickness, another hypothesis relates to a mismatch in interfacial bond strength. Transverse strength is largely dictated by the quality of the fiber-matrix interphase. It is plausible that the T800 fiber has better adhesion to the epoxy matrix compared to the HR40 fiber. If this were the case, the weaker HR40/matrix interphase would act as the "weakest link" in the hybrid. Under transverse load, failure would initiate at this weaker interface at a lower stress level and then propagate through the laminate, causing the overall strength of the hybrid to be lower than that of the pure T800 system. It is also worth noting that the reduction in strength may not be linear with the proportion of LE material added since the interfacial bond mismatch could introduce non-linearities. However, verifying this hypothesis would require further experimental studies on a wider range of hybrid configurations.

4.2.3 In-Plane Shear Properties

The in-plane shear properties of the laminates were determined using tensile tests on $[\pm 45]_s$ specimens.

The tests were performed in compliance with the ASTM D-3518 [22] standard. The experimental setup and data acquisition systems were the same as those used for the previous tensile tests. The specimens were designed to a target thickness of approximately 2.5 mm and had a nominal rectangular geometry of $250\text{ mm} \times 25\text{ mm}$ (also with aluminum tabs), as shown in Figure 4.20.

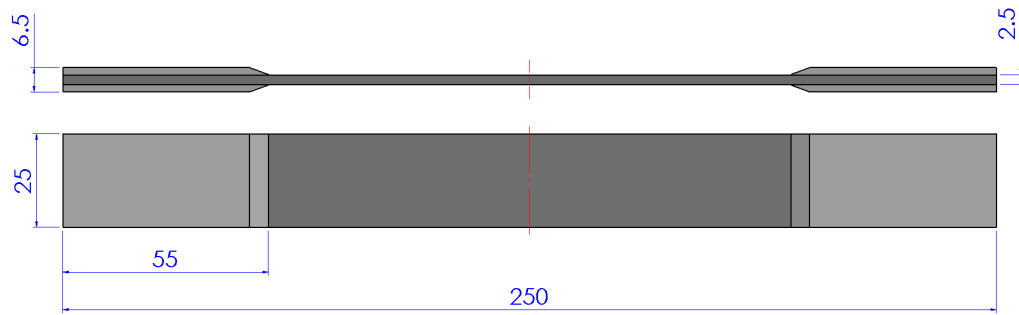


Figure 4.20: Illustration of a shear tensile specimen. All dimensions are in millimeters.

A comparison of the nominal thickness, the effective thickness based on the cured ply thicknesses, and the mean measured thickness of the manufactured specimens is provided in Table 4.23.

Table 4.23: Nominal, effective cured, and mean measured thicknesses of the in-plane shear specimens.

Configuration	Layup	Nominal (mm)	Effective Cured (mm)	Mean Measured (mm)
T800	$[\pm 45]_{25S}$	2.500	2.700	2.845
HR40	$[\pm 45]_{64S}$	2.560	3.072	3.239
[T/H/T]	$[\pm 45]_{10S}$	2.400	2.600	2.623

The shear stress, τ_{12} , and engineering shear strain, γ_{12} , were calculated from the tensile test data. The shear strain was determined from the measured longitudinal and transverse strains:

$$\gamma_{12} = \varepsilon_{11} - \varepsilon_{22} \quad (4.51)$$

The shear stress was then calculated from the applied load (P) and the specimen's cross-sectional area:

$$\tau_{12} = \frac{P}{2 \cdot t \cdot w} \quad (4.52)$$

From these values, the in-plane shear modulus, G_{12} , was evaluated as the slope of the shear stress-strain curve in the strain range of 2000–5000 $\mu\epsilon$:

$$G_{12} = \frac{\Delta \tau_{12}}{\Delta \gamma_{12}} \quad (4.53)$$

Finally, the in-plane shear strength, S_{12} , was obtained from the peak load (P_{MAX}):

$$S_{12} = \frac{P_{MAX}}{2 \cdot t \cdot w} \quad (4.54)$$

The in-plane shear properties reported are summarized in Table 4.24.

Table 4.24: In-plane shear properties of the benchmark configurations (data from Danzi [53]).

Configuration	Shear Modulus, G_{12} (GPa)	Shear Strength, S_{12} (MPa)
T800 (non-hybrid)	4.73 ± 0.38	145.74 ± 2.39
HR40 (non-hybrid)	3.70 ± 0.77	95.21 ± 2.36
[T/H/T]	4.43 ± 0.18	133.39 ± 0.95

In his discussion, Danzi noted that the effect of hybridization is evident in both the stiffness and the strength of the laminate, with the hybrid properties falling between those of the two non-hybrid constituents. This intermediate stiffness behavior can be verified using the classical rule of mixtures. Using the shear moduli from the non-hybrid tests and the ideal constituent ply thicknesses, the theoretical hybrid shear modulus, G_{th} , can be calculated:

$$G_{th} = \frac{t_{T800} G_{T800} + t_{HR40} G_{HR40}}{t_H} \approx 4.54 \text{ GPa} \quad (4.55)$$

This theoretical value, calculated using the ideal constituent ply thicknesses, is in good agreement with the measured average modulus of 4.43 GPa. This again confirms that the rule of mixtures provides a reasonable prediction for the in-plane shear stiffness of this configuration.

One finding was that the HR40 laminate showed a shear modulus lower than the T800 laminate, despite its higher longitudinal stiffness. The author states that this result is in accordance with the trend observed in the transverse tensile tests. However, there appears to be a misstatement in the original text, where it is suggested the T800 transverse modulus is lower than that of HR40. As shown in the previously (Table 4.22), the opposite was reported. Therefore, the trend is actually consistent across both matrix-dominated tests: the T800-based composite shows higher stiffness than the HR40-based composite in both transverse tension and in-plane shear. This is consistent with the author's primary hypothesis that the observed differences can be attributed to a superior fiber-matrix adhesion in the T800 composite.

The failure process was also observed to differ between the materials. The T800-based composites exhibited high plastic deformations with fiber rotation, leading to a concentrated necking area with massive out-of-plane displacements and delaminations. In contrast, the HR40-based laminate showed a much shorter final softening region, with localized delaminations in a wider necking area leading to an abrupt failure. The hybrid configuration exhibited both delaminations through the thickness and along the specimen length, but with a steeper stiffness reduction in the softening region compared to the T800 composite.

4.2.4 Double-Edge Notched Tension (DENT) Properties

The fracture toughness and damage progression of the laminates were assessed using the Double-Edge Notched Tension (DENT) test. These were performed following the procedure proposed by Catalanotti et al. [38] and the experimental setup was consistent with the previous test series although, for this test, strain gauges were not used. One of feature of this campaign was the investigation of size effects where for each material system, specimens of five different widths (10, 20, 30, 40, and 50 mm) were tested. The initial notch length, a_0 , was scaled with the width, w , to maintain a constant ratio of $a_0/w = 0.6$.

The specimen geometry is shown schematically in Figure 4.21.

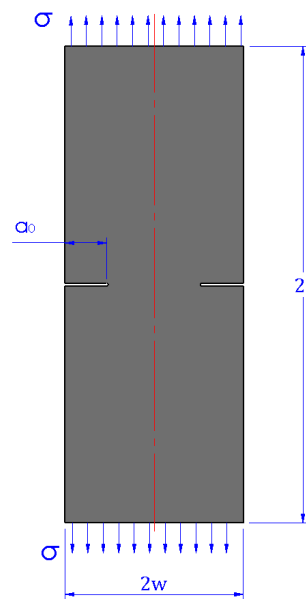


Figure 4.21: Schematic of the DENT test set-up and specimen geometry

The thicknesses for the various cross-ply DENT specimens are provided in Table 4.25.

Table 4.25: Nominal, effective cured, and measured thicknesses of the DENT specimens.

Configuration	Layup [†]	Nominal (mm)	Effective Cured (mm)	Mean Measured Thickness (mm) per Width					
				10 mm	20 mm	30 mm	40 mm	50 mm	All
T800	[90/0] ₂₀ \$	1.950	2.106	2.038	2.156	2.111	2.169	2.085	2.112
HR40	[90/0] ₅₀ \$	1.980	2.376	2.266	2.374	2.397	2.379	2.462	2.376
[T/H/T]	[90/0] ₁₀ \$	2.280	2.470	2.363	2.512	2.429	2.509	2.472	2.457
[2T/H/2T] [*]	[90/0] ₆ \$	2.420	2.651	—	—	—	—	—	—
[3T/2H/3T] [*]	[90/0] ₄ \$	2.380	2.618	—	—	—	—	—	—

^{*} Measured thickness data for these configurations were not reported in the source.

[†] The '\$' symbol denotes a center-symmetric layup where the final ply of the half-laminate is not repeated at the midplane.

The data reduction procedure follows the methodology of Catalanotti et al. [38] to determine the Mode I intralaminar crack resistance curve ($\mathcal{R}$ -curve). The analysis begins with the laminate elastic properties, summarized in Table 4.26. Using classical lamination theory, these properties are used to calculate the laminate's stiffness matrix. This matrix is then inverted to obtain the compliance matrix, $[S]$, which provides the terms required for the fracture mechanics calculations.

Table 4.26: Laminate elastic properties for DENT specimens (data from [53]).

Configuration	Layup	E_x (MPa)	E_y (MPa)	G_{xy} (MPa)	ν_{xy}	ν_{yx}
T800	[90/0] ₂₀	76695	80308	4732	0.030	0.031
HR40	[90/0] ₅₀	103403	105382	4005	0.016	0.016
[T/H/T]	[90/0] ₁₀	77933	85677	4420	0.029	0.032
[2T/H/2T]	[90/0] ₆	73926	87102	4732	0.028	0.033
[3T/2H/3T]	[90/0] ₄	70668	91542	4732	0.026	0.034

The analysis starts with the fundamental relationship between the energy release rate, G_I , the Mode I stress intensity factor, K_I , and an equivalent modulus, $\hat{E}$:

$$G_I = \frac{K_I^2}{\hat{E}} \quad (4.56)$$

For a 2D orthotropic body, the equivalent modulus $\hat{E}$ is defined by the components of the compliance matrix, $[S]$, and two dimensionless elastic parameters, λ and ρ :

$$\hat{E} = \left(s_{11}s_{22} \frac{1+\rho}{2} \right)^{-1/2} \lambda^{-1/4} \quad (4.57)$$

$$\lambda = \frac{s_{11}}{s_{22}}, \quad \rho = \frac{2s_{12} + s_{66}}{2\sqrt{s_{11}s_{22}}} \quad (4.58)$$

The stress intensity factor, K_I , is a function of the remote stress σ , specimen geometry, and a correction factor, κ :

$$K_I = \sigma \sqrt{w} \kappa(\alpha, \rho, \zeta) \quad (4.59)$$

The correction factor κ is a function of the material orthotropy and two dimensionless shape parameters:

- The material orthotropy parameter, ρ .
- The relative crack length, α , defined as the ratio of the crack length to the specimen half-width:

$$\alpha = \frac{a}{w}$$

- The rescaled aspect ratio, ζ , which accounts for material orthotropy through the parameter λ and the geometric aspect ratio, ξ . It is defined as:

$$\zeta = \lambda^{-1/4} \xi, \quad \text{where} \quad \xi = \frac{w}{l}$$

The calculation of the correction factor, κ , is a critical step. For materials with low orthotropy, such as quasi-isotropic laminates where the parameter $0 < \rho < 4$, simpler analytical expressions for κ exist. However, the cross-ply laminates investigated in this study are highly orthotropic, resulting in values of ρ that are significantly greater than 4. Consequently, the simplified expressions are not applicable. The methodology therefore employs a more general and robust polynomial function for κ that is valid for a wide range of orthotropy.

While a general polynomial expression for $\kappa(\alpha, \rho, \zeta)$ exists, for sufficiently long specimens ($\xi \rightarrow 0$), it simplifies to a function of only α and ρ , denoted as $\Phi(\alpha, \rho)$. According to Catalanotti et al., this approximation is acceptable for $\xi < 0.5$. Danzi's experimental campaign was designed to respect this condition, ensuring that the error from using this simplified form remains below 2%. The function ϕ is given by:

$$\kappa|_{\xi \rightarrow 0} = \phi = \sqrt{\tan \frac{\pi\alpha}{2} \sum_{i=1}^5 \sum_{j=1}^5 \Phi_{ij} \alpha^{i-1} \rho^{j-1}} \quad (4.60)$$

where Φ_{ij} are the elements of a numerically derived matrix:

$$\Phi = \begin{bmatrix} 1.7482487564 & -0.053754159533 & 0.0040142704949 & -9.8480085881 \times 10^{-5} \\ -0.76896688866 & -0.0068632911438 & 0.0029984681658 & -0.00010108691939 \\ 0.85633404777 & 0.23922363475 & -0.023289123198 & 0.00062358861997 \\ -0.67470597429 & -0.25334178248 & 0.022297779266 & -0.00056784694513 \\ 0.18495379886 & 0.084067007027 & -0.0068989066533 & 0.00016783852495 \end{bmatrix} \quad (4.61)$$

Replacing the expression for K_I into the energy release rate equation gives G_I in terms of remote stress or applied load, P :

$$G_I = \frac{1}{\hat{E}} w \sigma^2 \kappa^2 = \frac{1}{4w\hat{E}} \left(\frac{P\kappa}{t} \right)^2 \quad (4.62)$$

The $\mathcal{R}$ -curve, which represents the material's intrinsic resistance to crack growth, $\mathcal{R}(\Delta a)$, is determined by finding the envelope of the crack driving force curves, $G_I(\Delta a)$, that correspond to the peak loads, P_u , for each specimen size. Mathematically, this tangency condition is expressed by:

$$\begin{cases} G_I(\Delta a) = \mathcal{R}(\Delta a) \\ \frac{\partial G_I(\Delta a)}{\partial \Delta a} = \frac{\partial \mathcal{R}(\Delta a)}{\partial \Delta a} \end{cases} \quad (4.63)$$

Assuming the size-effect law, $\sigma_u = \sigma_u(w)$, is known, the first part of this condition allows the $\mathcal{R}$ -curve to be calculated directly from the ultimate stress:

$$\mathcal{R}(\Delta a) = \frac{1}{\bar{E}} w \sigma_u^2 \kappa^2 \quad (4.64)$$

Therefore, the central experimental task is to determine the size-effect law by fitting the failure data using one of the regression models shown in Table 4.27.

Table 4.27: Size-effect law fitting models and corresponding formulas for propagation toughness ($\mathcal{R}_{ss}$) and fracture process zone length (l_{fpz}) (adapted from Catalanotti et al. [38]).

Regression Fit	Formula	Parameters	$\mathcal{R}_{ss}$	l_{fpz}
Bilogarithmic	$\ln(\sigma_u) = \ln(M/\sqrt{N+w})$	M, N	$\frac{\kappa_0^2}{\bar{E}} M^2$	$\frac{\kappa_0'}{\kappa_0} N$
Linear Regression I	$1/\sigma_u^2 = A \cdot w + C$	A, C	$\frac{\kappa_0^2}{\bar{E}A}$	$\frac{\kappa_0'}{2\kappa_0} \frac{C}{A}$
Linear Regression II	$w/\sigma_u^2 = A' \cdot w + C'$	A', C'	$\frac{\kappa_0^2}{\bar{E}A'}$	$\frac{\kappa_0'}{2\kappa_0} \frac{C'}{A'}$

The specific fitting parameters of the Danzi's configurations are presented in Table 4.28.

Table 4.28: Size-effect law parameters of the benchmark configurations (data from Danzi [53]).

Configuration	Best Fit Method	Parameter 1	Parameter 2	R^2
T800	Bilogarithmic	$M = 1104$	$N = 7.12$	0.815
HR40	Bilogarithmic	$M = 415.4$	$N = 0.047$	0.900
[T/H/T]	Bilogarithmic	$M = 1232$	$N = 8.54$	0.870
[2T/H/2T]	Bilogarithmic	$M = 1471$	$N = 7.64$	0.857
[3T/2H/3T]	Linear II	$A' = 4.40 \times 10^{-6}$	$C' = 2.47 \times 10^{-7}$	0.877

Once the size-effect law is established, the laminate's $\mathcal{R}$ -curve is calculated. The final step, as noted by Pinho et al. [164], is to isolate the toughness of the load-bearing 0° plies. This is based on an energy balance for self-similar crack propagation, where the energy dissipated by the laminate is the sum of the energy dissipated by the 0° and 90° plies:

$$\mathcal{R} t da = \mathcal{R}_0 t_0 da + \mathcal{R}_{90} t_{90} da \quad (4.65)$$

where $\mathcal{R}$, $\mathcal{R}_0$, and $\mathcal{R}_{90}$ are the $\mathcal{R}$ -curves of the laminate, 0° plies, and 90° plies, and t , t_0 , and t_{90} are their respective total thicknesses. By assuming the contribution of the 90° plies to be negligible ($\mathcal{R}_{90} \ll \mathcal{R}_0$), the equation can be simplified for "balanced" cross-ply laminates, where the number of 0° and 90° plies is equal. In that specific scenario, the total 0° ply thickness equals the total 90° ply thickness, and the relationship simplifies to $\mathcal{R}_0 = 2\mathcal{R}$. However, this simplification is not valid for the laminates in this study. As detailed before, the layups are designed to be symmetric but are not balanced. For instance, a layup denoted as $[90/0]_{2S}$ results in a full laminate of $[90/0/90/0/90/0/90]$, which contains four 90° plies and only three 0° plies. Therefore, the toughness of the 0° plies must be determined using the more general form of the rescaling equation,

where the total laminate thickness, t , and the total 0° ply thickness, t_0 , are calculated from the actual, unequal ply counts:

$$\mathcal{R}_0 = \frac{t}{t_0} \mathcal{R} \quad (4.66)$$

The $\mathcal{R}$ -curves for each of the benchmark configurations were derived from the experimental data using the described methodology. Each plot illustrates the family of crack driving force curves (G_I) calculated for different specimen widths, with the final laminate $\mathcal{R}$ -curve appearing as their lower envelope.

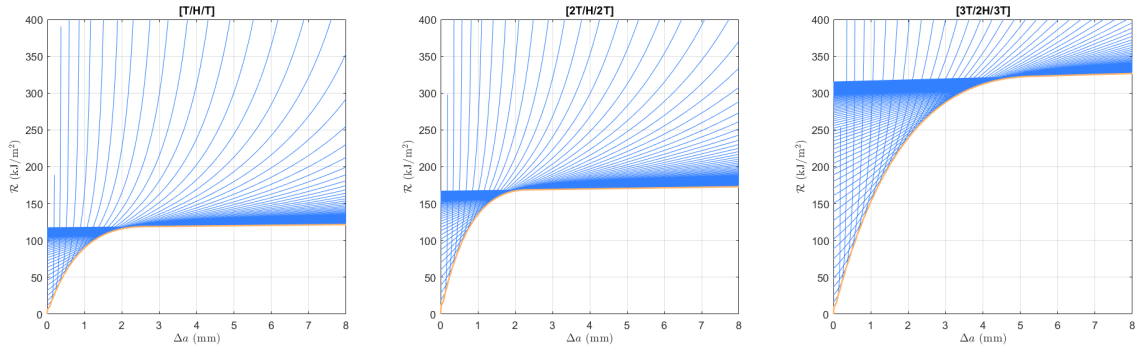


Figure 4.22: Calculated $\mathcal{R}$ -curves for the hybrid configurations: $[T/H/T]$, $[2T/H/2T]$, and $[3T/2H/3T]$

The corresponding $\mathcal{R}$ -curves for the non-hybrid constituent materials are shown in Figure 4.23.

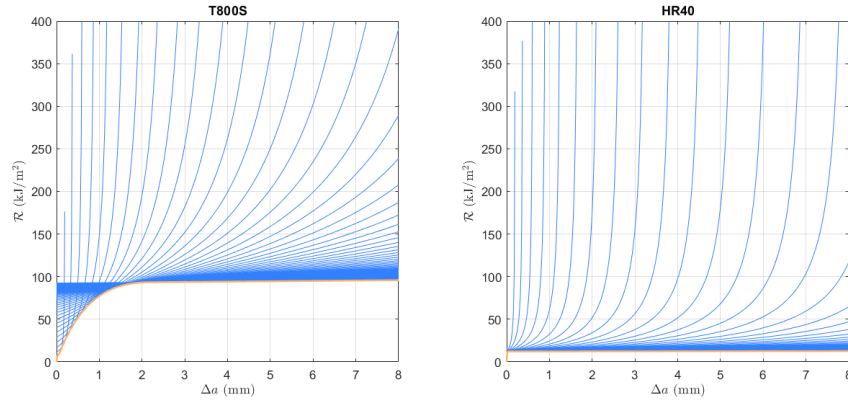


Figure 4.23: Calculated $\mathcal{R}$ -curves for the non-hybrid constituent materials: T800 and HR40

The parameters used in the analysis and the final calculated fracture properties are consolidated for each material system in Table 4.29. This includes the laminate material properties, the width-dependent shape parameters, and the final parameters describing the 0° ply $\mathcal{R}$ -curve. These results, along with the preceding R-curves, were generated using a custom MATLAB script detailed in Appendix A.5.

Table 4.29: Summary of fracture mechanics parameters and final $\mathcal{R}$ -curve results for all benchmark configurations.

Parameter	Symbol	Units	Width (mm)	T800S	HR40	[T/H/T]	[2T/H/2T]	[3T/2H/3T]
Laminate & Material Parameters								
Equivalent Modulus	$\hat{E}$	GPa	–	36.05	39.25	35.31	35.44	34.65
Compliance Ratio	λ	–	–	1.047	1.019	1.100	1.178	1.295
Orthotropy Parameter	ρ	–	–	8.26	13.02	9.21	8.45	8.47
0° Ply Thickness	t_0	mm	–	1.026	1.176	1.170	1.205	1.122
Total Thickness	t	mm	–	2.106	2.376	2.470	2.651	2.618
Shape Parameters								
Initial Crack Ratio	α_0	–	–	0.6	0.6	0.6	0.6	0.6
Aspect Ratio	ξ	–	10	0.0400	0.0400	0.0400	0.0400	0.0400
			20	0.0800	0.0800	0.0800	0.0800	0.0800
			30	0.1200	0.1200	0.1200	0.1200	0.1200
			40	0.1600	0.1600	0.1600	0.1600	0.1600
			50	0.2000	0.2000	0.2000	0.2000	0.2000
Rescaled Aspect Ratio	ζ	–	10	0.0395	0.0398	0.0391	0.0384	0.0375
			20	0.0791	0.0796	0.0781	0.0768	0.0750
			30	0.1186	0.1194	0.1172	0.1152	0.1125
			40	0.1582	0.1592	0.1563	0.1536	0.1500
			50	0.1977	0.1991	0.1953	0.1920	0.1875
Final 0° Ply $\mathcal{R}$ -Curve Results								
Fracture Process Zone	l_{fpz}	mm	–	2.17	0.01	2.61	2.33	5.44
Propagation Toughness	$\mathcal{R}_0$	kJ/m ²	–	191.75	24.70	250.69	371.16	753.26
(Laminate)	$\mathcal{R}$	kJ/m ²	–	93.41	12.22	118.75	168.71	322.83
R-Curve Shape Parameter	γ	mm ^{−1}	–	0.3133	69.0527	0.2634	0.2930	0.1283
R-Curve Shape Parameter	β	–	–	4.7453	3.0000	4.6960	4.7252	4.6024

In his discussion of the results, Danzi reported that hybridization led to a substantial increase in fracture toughness, a trend clearly visible in the propagation toughness ($\mathcal{R}_0$) values in Table 4.29. The non-hybrid materials provided a baseline for this comparison. The HR40 laminate was observed to be brittle, with a low toughness and flat fracture surfaces. The T800 composite showed a slightly higher toughness, with its fracture surfaces exhibiting a minor presence of fiber pull-out.

The increase in toughness in the hybrid systems is attributed to multiple, co-existing energy-dissipating mechanisms. The DIC analysis provided the primary observational evidence for these processes. The first mechanism identified was a progressive and stable fragmentation of the load-bearing 0° plies. This was not observed directly but was inferred from the appearance of high-strain stripes oriented orthogonally to the loading direction on the surface of the 90° plies. Danzi interpreted these stripes as the response of the surface ply to the fragmentation of the underlying 0° plies. This fragmentation initiated at the notch and increased in density as the load progressed towards failure. An ambiguity in this interpretation, however, is whether these surface features are just regions of high transverse strain or if they correspond to the formation of actual transverse cracks within the 90° plies, something that surface analysis cannot clarify.

According to Danzi, the superior performance of the hybrid systems, particularly $[2T/H/2T]$ and $[3T/2H/3T]$, is attributed to the synergistic action of multiple damage mechanisms. The primary mechanism identified was stable fiber fragmentation. This process reduces notch sensitivity by creating a large number of new free surfaces, which significantly increases the amount of energy dissipated during fracture and thus the energy required to propagate the main crack. The author suggested that this widespread fragmentation, along with associated delaminations, in turn promoted a more extensive fiber pull-out mechanism upon final failure of the HE layers.

The second mechanism, observed in the $[2T/H/2T]$ and $[3T/2H/3T]$ configurations, was the formation of split cracks at the notch tip that propagated parallel to the fibers along the loading direction. By blunting the main crack tip, these splits reduced the material's notch sensitivity and contributed to the increase in fracture toughness. Danzi noted that this initial propagation represented a form of stable damage growth that did not compromise the material's load-bearing capability, a process that explains the notch-insensitive behavior later reported by Tavares [209] for these hybrid configurations in open-hole tension tests. The increase in fracture toughness also correlated with the higher pseudo-ductility seen in longitudinal tensile tests, indicating that the same underlying energy-dissipating mechanisms are beneficial for both properties. While these observations provide a qualitative explanation for the increase in toughness, Danzi concluded that the complex interactions between fiber fragmentation, split cracks, and interfacial properties require further investigation.

Therefore, the following section addresses this need directly by conducting a post-mortem analysis of the DENT specimens using high-resolution micro-Computed Tomography (μ CT).

4.2.5 Performance in Compression and Notched Tension

The investigation by Tavares [209] also provided key insights into the performance of these hybrid systems under compression and in the presence of stress concentrations, which were not part of Danzi's original scope. In plain strength compression tests on QI laminates, no synergistic effect was observed. Both hybrid configurations exhibited lower compressive strength than the non-hybrid T800 baseline, with very similar failure mechanisms observed across all materials. The primary benefit of hybridization was revealed in the open-hole tension (OHT) tests. The hybrid configurations demonstrated superior notched performance, with one showing a notched strength significantly higher than that of the T800 baseline. A notch sensitivity analysis confirmed this, showing the T800 to be highly notch-sensitive, while the hybrid laminates exhibited significantly more notch-insensitive behavior. This exceptional damage tolerance was attributed to a complex failure mode involving large split cracks and multiple secondary cracks emanating from the hole, which effectively blunted the stress concentration. This finding is particularly significant, as it shows that the toughening mechanisms of fiber fragmentation and splitting observed at the lamina level by Danzi directly translate into a substantial improvement in the structural performance of notched multidirectional laminates.

4.3 Post-Mortem μ CT Analysis of DENT Specimens

While the mechanical test data and surface DIC analysis from the benchmark study provide a robust, top-down view of the fracture behavior, they cannot fully resolve the internal, three-dimensional damage mechanisms responsible for the observed increase in toughness. Danzi's hypotheses rely on inferences from surface observations and post-mortem visual inspection. To directly test these hypotheses and provide a detailed picture of the internal damage state, a post-mortem analysis of selected DENT specimens was conducted using high-resolution laboratory-based X-ray micro-Computed Tomography (μ CT).

4.3.1 Literature Precedent for CT Analysis of Notched Composites

The use of X-ray computed tomography to investigate the internal fracture of advanced composites provides an established precedent for moving beyond the surface-based deductions of the benchmark study. The existing literature demonstrates that tomography can directly visualize the damage mechanisms before mentioned, offering a means to resolve the ambiguities left from his analysis and to provide a more complete picture of the toughening mechanisms in the hybrid systems.

From the two principal damage mechanisms identified – split cracking and fiber fragmentation – the ability of X-ray CT to visualize the former was first demonstrated by Moffat et al. [148] in a notched $[90/0]_S$ carbon-epoxy laminate, where each ply was approximately 250 μ m thick. While this study provided the first volumetric observations of 0° splits and detailed their micromechanisms, a physical explanation for their initiation was later provided by Yang et al. [231]. In their study of 1.1 mm thick notched $[90/0/90/0]_S$ laminates, they demonstrated through SRCT and simulation

that these splits initiate at offset locations from the notch root, not at the point of maximum tensile stress. This phenomenon was attributed to local stress redistribution caused by matrix shear nonlinearity. This ply-level stress redistribution is a fundamental process driven by the material's response to a stress concentration and is not specific to a particular ply thickness or material system. Therefore, it is reasonable to expect this same mechanism to govern the initiation of splits in the benchmark DENT specimens, regardless of their thin-ply or hybrid nature. The consistent observation of these splits by other researchers, including Scott et al. [187, 188] and Garcea et al. [75], in similar notched cross-ply specimens further highlights the importance of this mechanism.

The second mechanism, fiber fragmentation, was also addressed by this studies. Scott et al. [187, 188] supplied the first direct three-dimensional measurements of fiber fracture accumulation in similar notched $[90/0]_S$ specimens of a commercial T700/M21 carbon-epoxy. Their analysis revealed a progression where fiber breaks are initially isolated but later evolve into clusters at higher loads, validating the progressive nature of the fragmentation process proposed in the benchmark study. An example of this progression is provided in Figure 4.24.

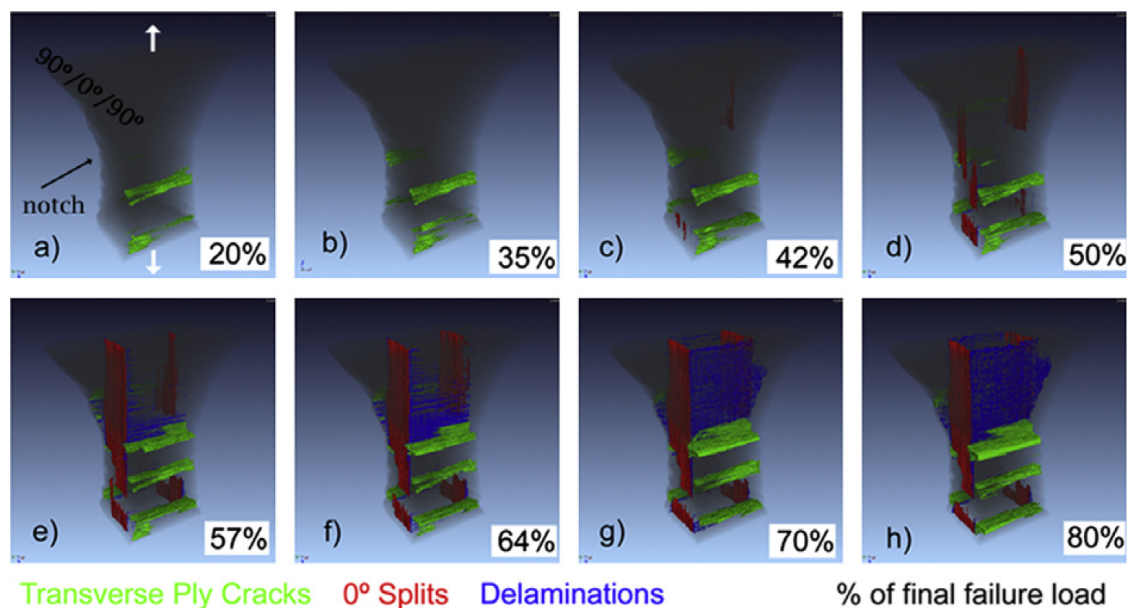


Figure 4.24: Progression of matrix damage visualized via in-situ SRCT at increasing percentages of the final failure load, showing the growth of transverse ply cracks (green), 0° splits (red), and delaminations (blue) (reproduced from [188]).

This work also documented that matrix damage, including splits, occurred well in advance of extensive fiber breaks, allowing them to test whether stress concentrations from this matrix damage dictate the location of subsequent fiber failures. Their three-dimensional analysis found no spatial correlation, indicating that fiber fracture locations were governed by the stochastic distribution of flaws within the fibers themselves. The dynamic aspect of this fragmentation was later investigated with greater temporal resolution by Garcea et al. [75]. Using fast, continuous SRCT on 1 mm thick

toughened $[90/0]_S$ DENT specimens, their work showed that large clusters of breaks form in a sudden burst just before catastrophic failure, captured in the 3D rendering in Figure 4.25. This confirms that fiber fragmentation is an observable process of break accumulation and clustering that can be mapped in 3D, thus providing a procedure to validate the primary hypothesis from the benchmark campaign.

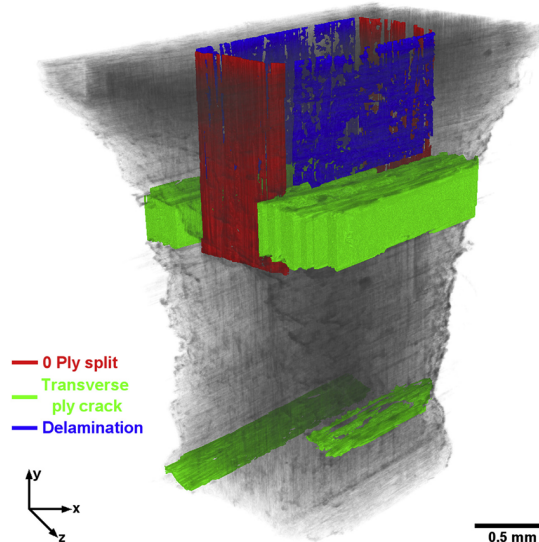


Figure 4.25: 3D rendering of the damage state in the notch region just 1 second prior to catastrophic failure (99.9% UTS). The image shows 0° splits (red), transverse ply cracks (green), and delamination (blue) (reproduced from [75]).

One of the open points from the surface-based analysis was the interpretation of the high-strain stripes observed on the 90° plies. Danzi's conclusion was that the stripes were the surface response to, and therefore observational evidence for, the stable fragmentation of the underlying 0° plies. The ambiguity in this line of reasoning, however, is that surface analysis cannot definitively distinguish between strain in the 90° ply that is a response to underlying damage and the formation of transverse cracks within the 90° ply itself. The tomographic literature offers a potential physical explanation for these observations. The study by Moffat et al. [148] supplied the first volumetric observation of transverse cracks initiating near the notch, and the physical origin of these cracks was later explained by the work of Yang et al. [231]. Their work showed that these cracks form for the same reason as the 0° splits, which is local stress redistribution driven by matrix shear nonlinearity. For the standard-thickness laminates they investigated, the SRCT analysis suggests that the surface strain features are the expression of fully-formed transverse cracks propagating through the thickness of the 90° plies. This would indicate they are a primary damage event and not a secondary effect from fragmentation in the underlying 0° plies, providing a testable hypothesis for the present study.

With the behavior of these damage mechanisms established in standard-thickness laminates, recent research has applied tomographic techniques to the specific thin-ply and hybrid architectures relevant to the benchmark campaign, providing insight into how these established failure processes

are modified. The work presented initially by Kopp et al. [122] and expanded in a subsequent journal article by Ni et al. [157] provides a direct comparison of damage progression in 1 mm thick quasi-isotropic DENT specimens made from standard-ply (AS4/8552) and thin-ply (HTS40/Q-1112) materials. Focusing on their baseline, non-nanostitched configurations, they supplied quantitative evidence for the thin-ply effect, showing that the thin-ply architecture intrinsically suppresses matrix damage and results in 6.5 times less total damage surface area. While the damage mechanisms of matrix cracking and interfacial debonding were the same in both architectures, their behavior differed. The standard-ply laminates developed fewer but larger matrix cracks that blunted the notch stress concentration, whereas damage in the thin-ply laminates was characterized by a high concentration of smaller, localized cracks confined to the notch region (illustrated in Figure 4.26).

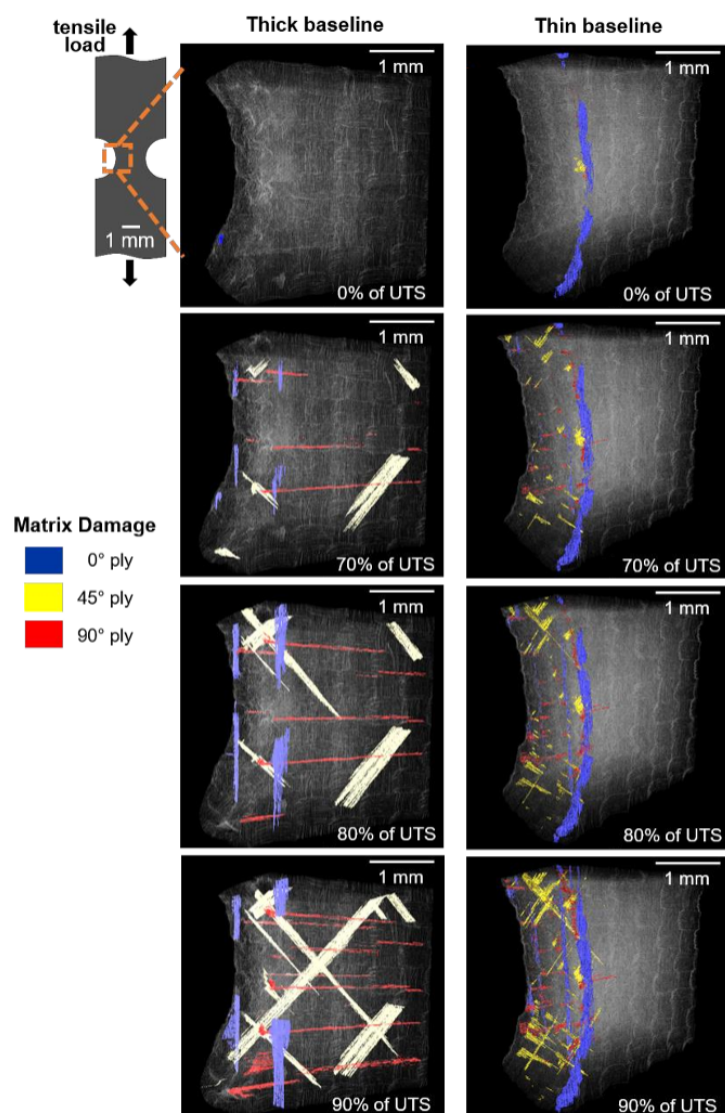


Figure 4.26: Qualitative inspection of 3D matrix damage progression in standard (thick) and thin-ply thickness DENT specimens at increasing load levels (adapted from [122]).

While these works focused on non-hybrid laminates, other studies have examined the specific ply interactions in hybrid systems. The investigation by AhmadvashAghbash et al. [3] on thin-ply carbon-carbon hybrid cross-ply laminates using novel mini-compact tension specimens (Figure 4.27) confirmed that 90° ply-blocks effectively isolate the 0° plies, allowing them to fracture independently. This ply isolation was evidenced by a lack of correlation in the shape and height of pull-out bundles between adjacent 0° plies, as depicted in Figure 4.28. Their volumetric analysis also revealed that the crack fronts in the 90° plies advance further than those in the 0° plies, a three-dimensional effect that surface analysis cannot capture.

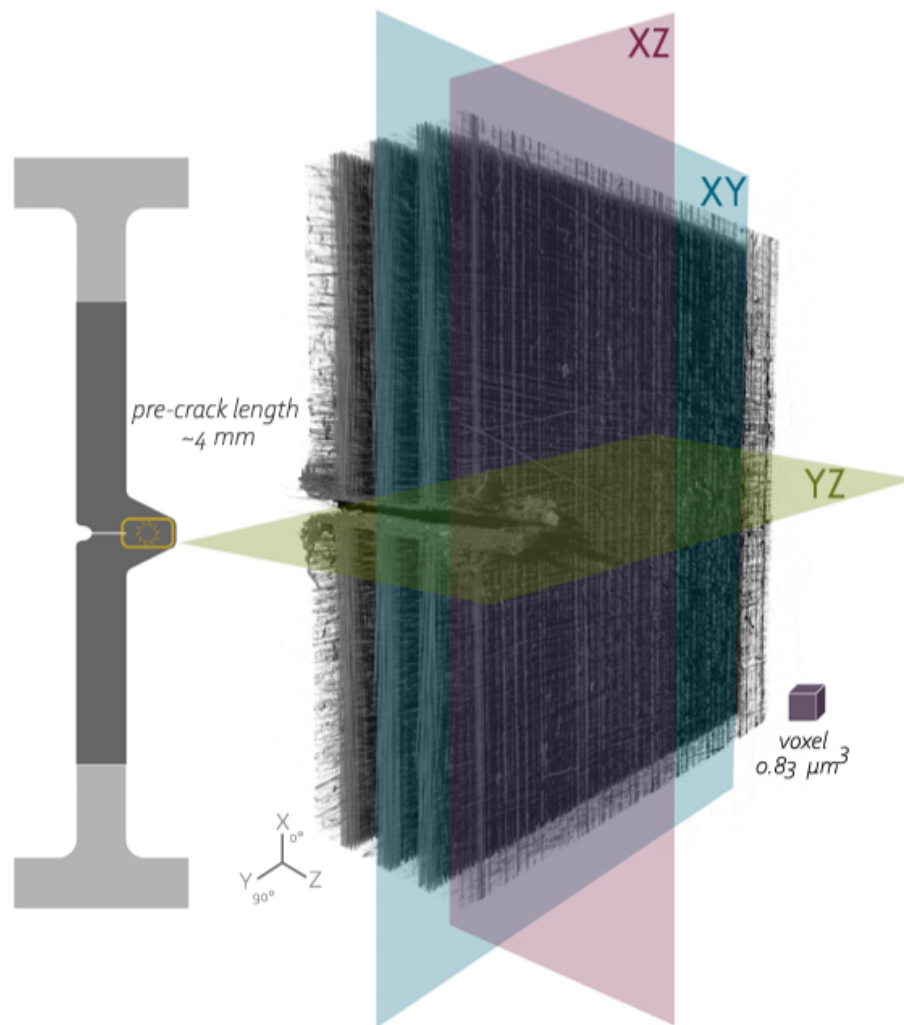


Figure 4.27: Methodology for visualizing 3D SRCT data, showing the specimen with orthogonal slicing planes for (adapted from [3]).

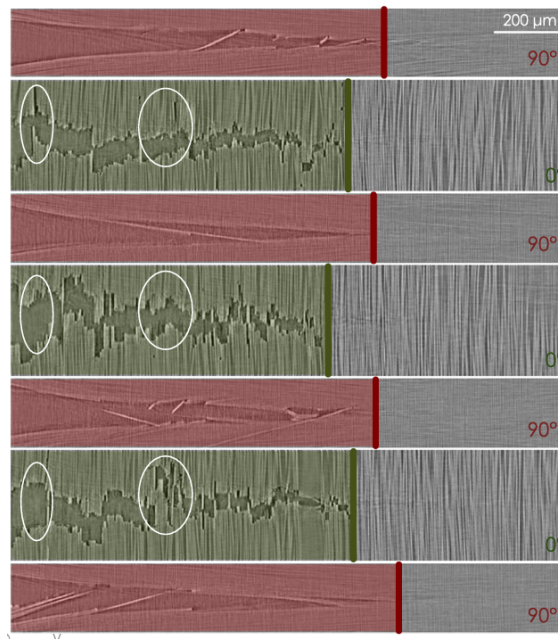


Figure 4.28: SRCT slice showing the lack of correlation between pull-out bundle shapes (highlighted by white ovals) in adjacent 0° plies and different crack fronts (adapted from [3]).

A study by Mesquita et al. [144] provided a micromechanical explanation for the synergistic toughening observed in the benchmark study, using SRCT to analyze notched tensile specimens of thin-ply unidirectional carbon-glass (Grafil 34-700/HYBON 2026) hybrids. Their work confirmed that the hybrid architecture allows the carbon plies to accumulate a significantly higher density of fiber breaks before failure compared to a pure carbon reference composite. Their tomographic data also revealed a previously unreported mechanism: fiber breaks were found to form preferentially near the ply interfaces rather than in the middle of the carbon plies. They attributed this to manufacturing-induced effects but noted that this spatial preference is beneficial, as it forces initial failures to occur adjacent to the high-elongation glass plies, which are most effective at arresting the growth of damage clusters and thereby augmenting the hybrid effect. While their use of different specimen geometries and materials means the global failure modes are not identical to the benchmark study, their findings on these fundamental ply-level interactions are.

Although the preceding literature provides a methodological background, to the best of the authors knowledge, no prior investigation has applied X-ray CT to the internal fracture mechanisms of thin-ply carbon-carbon hybrid DENT specimens. The present post-mortem analysis is therefore intended to provide this initial volumetric characterization, with several objectives. The analysis will map the internal damage state to confirm the presence and density of transverse cracking in the 90° plies and to identify the initiation sites of the 0° splits, which will allow for an assessment of whether the stress redistribution mechanism proposed by Yang et al. [231] is applicable to these thin-ply hybrid systems. Furthermore, this analysis will qualitatively compare the total damage area against the baseline to assess the thin-ply effect observed by Ni et al. [157], investigate the ply isolation effect reported by AhmadvashAghbash et al. [3] by comparing the fracture morphologies

of consecutive 0° plies, and look for evidence of fiber breaks forming preferentially near ply interfaces, a mechanism identified by Mesquita et al. [144] in carbon-glass hybrids. This analysis will be undertaken with the acknowledgment that its post-mortem nature makes extracting some required information more difficult and prevents the analysis of others. For example, the progressive differential crack front advancement between plies reported by AhmadvashAghbash et al. [3], a relevant aspect for a complete understanding of the fracture process, cannot be investigated here and remains a topic for future in-situ studies.

4.3.2 Specimens Analyzed

The original benchmark study by Danzi produced a total of 75 Double-Edge Notched Tension (DENT) specimens. This set comprised three samples for each of five widths (10, 20, 30, 40, and 50 mm) across five distinct material configurations. For completeness and to provide immediate context for the selection process that follows, the layup and nominal thickness for each configuration, though mentioned in a previous section, are summarized again in Table 4.30.

Table 4.30: Nominal thicknesses and layups for the DENT specimen configurations.

Configuration	Layup	Nominal Thick. (mm)
T800	$[90/0]_{20\$}$	2.106
HR40	$[90/0]_{50\$}$	2.376
$[T/H/T]$	$[90/0]_{10\$}$	2.470
$[2T/H/2T]$	$[90/0]_{6\$}$	2.651
$[3T/2H/3T]$	$[90/0]_{4\$}$	2.618

A subset of these specimens was chosen for the present detailed analysis based on several practical considerations. The primary factors for selection were the post-test availability of the specimens, the integrity of their fracture surfaces, and the completeness of their respective pairings. Table 4.31 presents an inventory of the specimens and their condition following the initial mechanical test campaign.

Table 4.31: Inventory of DENT specimens. The cell color indicates the specimen's status: green for present, yellow for incomplete, and pink for missing.

Width	T800	[T/H/T]	[2T/H/2T]	[3T/2H/3T]	HR40
10mm	T800-IV-A1	H-IV-A1	H2-IV-A1	H3-IV-A1	HR40-IV-A1
	T800-IV-A2	H-IV-A2	H2-IV-A2	H3-IV-A2	HR40-IV-A2
	T800-IV-A3	H-IV-A3	H2-IV-A3	H3-IV-A3	HR40-IV-A3
20mm	T800-IV-B1	H-IV-B1	H2-IV-B1	H3-IV-B1	HR40-IV-B1
	T800-IV-B2	H-IV-B2	H2-IV-B2	H3-IV-B2	HR40-IV-B2
	T800-IV-B3	H-IV-B3	H2-IV-B3	H3-IV-B3	HR40-IV-B3
30mm	T800-IV-C1	H-IV-C1	H2-IV-C1	H3-IV-C1	HR40-IV-C1
	T800-IV-C2	H-IV-C2	H2-IV-C2	H3-IV-C2	HR40-IV-C2
	T800-IV-C3	H-IV-C3	H2-IV-C3	H3-IV-C3	HR40-IV-C3
40mm	T800-IV-D1	H-IV-D1	H2-IV-D1	H3-IV-D1	HR40-IV-D1
	T800-IV-D2	H-IV-D2	H2-IV-D2	H3-IV-D2	HR40-IV-D2
	T800-IV-D3	H-IV-D3	H2-IV-D3	H3-IV-D3	HR40-IV-D3
50mm	T800-IV-E1	H-IV-E1	H2-IV-E1	H3-IV-E1	HR40-IV-E1
	T800-IV-E2	H-IV-E2	H2-IV-E2	H3-IV-E2	HR40-IV-E2
	T800-IV-E3	H-IV-E3	H2-IV-E3	H3-IV-E3	HR40-IV-E3
Complete	6	6	2	1	4
26.7 % (20 of 75)					

However, the tomographic process is both time-intensive and requires specialized operation, which required limiting the detailed analysis to a smaller group of specimens. Consequently, from the original set of 75, a single 10 mm width specimen was selected from each of the five material configurations. For each of these five specimens, one half of the fractured sample was prepared for the X-ray CT investigation. This choice was based on the expectation that the ply-level damage phenomena of interest are not primarily governed by specimen size effects, making this selection suitable for an initial characterization. Acknowledging this limitation, a more comprehensive analysis that includes other specimen widths remains a subject for future work.

4.3.3 Experimental Setup

4.3.3.1 Tomography System: ZEISS VersaXRM 615

The tomographic scans for this study were performed using a ZEISS VersaXRM 615, a 3D X-ray microscope located at the Institute of Science and Innovation in Mechanical and Industrial Engineering (INEGI). An exterior view of the complete system is presented in Figure 4.29.

The instrument's interior, shown in Figure 4.30, highlights the primary hardware components: the X-ray source on the left, the central sample rotation stage, and the detector turret with its multiple objectives on the right.



Figure 4.29: Exterior view of the ZEISS VersaXRM 615 instrument used in this study.

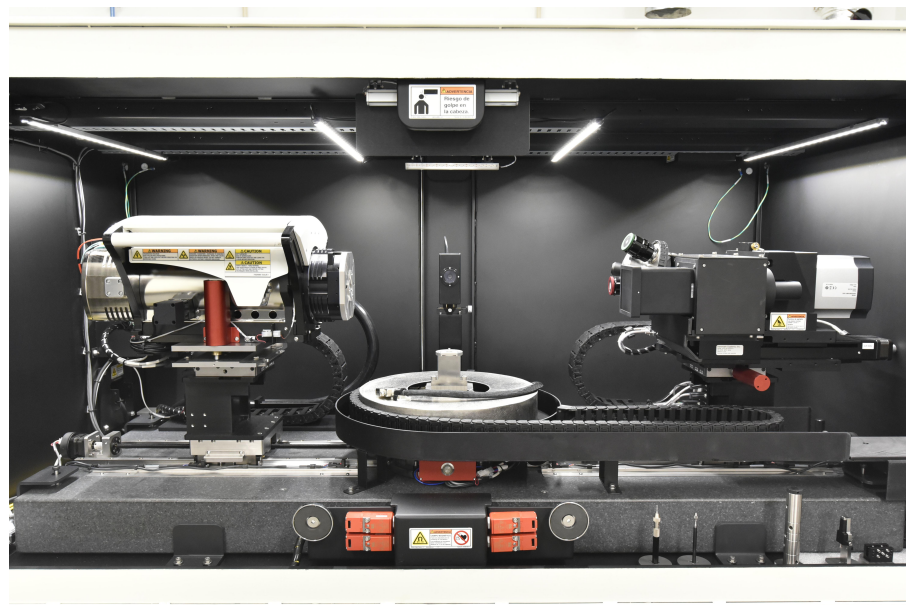


Figure 4.30: Interior view of the ZEISS VersaXRM 615, showing the primary hardware components.

The arrangement of these components distinguishes the system from conventional micro-CT scanners. It employs synchrotron-caliber, two-stage magnification optics, a design which maximizes both absorption and phase contrast. Conventional systems rely on a single stage of geometric magnification, where resolution is directly dependent on the sample's proximity to the X-ray source.

The VersaXRM architecture, illustrated in Figure 4.31, adds a second optical magnification stage. In this configuration, the X-ray projection is first cast onto a scintillator, and the resulting visible light image is then optically magnified before reaching the detector [37].

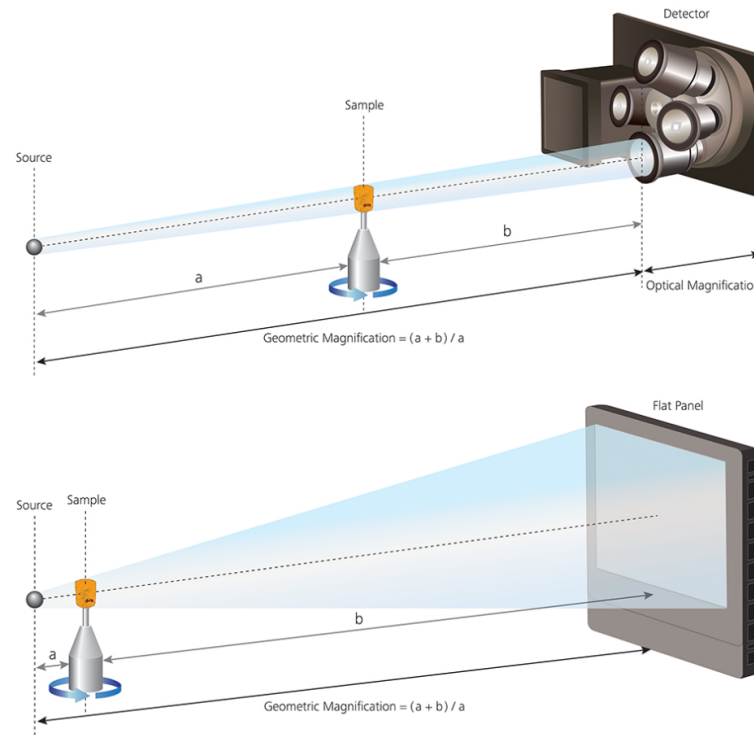


Figure 4.31: Comparison of the ZEISS XRM two-stage magnification architecture (top), featuring an optical magnification stage, with the conventional single-stage geometric magnification architecture of a flat panel micro-CT (bottom) (reproduced from [235]).

This two-stage process is the basis for the instrument's *Resolution at a Distance* (RaaD) capability, a design that makes the final resolution less dependent on the initial geometric magnification. This functionality allows the system to maintain sub-micron resolution even at large working distances and is a prerequisite for *in situ* studies. It permits the non-destructive characterization of a material's 3D microstructure and the observation of its evolution over time [37].

System operation is managed through a guidance and control software that provides a *Scout-and-Zoom* workflow for 3D navigation and multi-scale analysis. The instrument is configured with a high-flux 25 W X-ray source to produce faster submicron-scale data acquisition across a wide range of intact sample sizes and types. For data processing, the system includes a standard AI-based reconstruction algorithm that reduces scan time and improves image quality compared to conventional filtered back-projection methods. The platform's computational functionality is expandable with a full suite of advanced reconstruction tools and other machine learning software for accelerating post-processing tasks, automating analysis pipelines, and performing advanced 3D visualization. The instrument is also designed for future hardware expansion, with options such as a flat panel extension for rapid scanning, an autoloader for batch processing, and an interface kit

for *in situ* experiments, while also integrating with other light and electron microscopes to address multi-scale, correlative challenges [37].

The main technical specifications for the system are summarized in Table 4.32.

Table 4.32: Main technical specifications of the ZEISS VersaXRM 615 (data from [37]).

Parameter	Specification
Imaging Performance	
Spatial Resolution	500 nm
Resolution at a Distance (RaaD)	1.0 μ m (@ 50 mm working distance)
Minimum Achievable Voxel	40 nm
X-Ray Source	
Architecture	Sealed transmission, fast activation
Voltage Range	30 – 160 kV
Maximum Power Output	25 W
Detector System	
Standard Objectives	0.4 $\times$, 4 $\times$, 20 $\times$
Optional Objectives	40 $\times$, Flat Panel Extension (FPX)
Sample Stage Load Capacity	25 kg

4.3.3.2 Scan Parameters and Procedure

The tomographic scans were conducted by Dr. Nathan Klavzer at the INEGI. For each of the five selected specimens, a region of interest around the fracture surface was scanned using the acquisition parameters summarized in Table 4.33. The slight variation in some parameters, such as voxel size and exposure time, reflects minor adjustments made to optimize the scan for each specific specimen. The resulting scan time for each sample was approximately six hours.

Table 4.33: Tomography acquisition parameters used for the experimental scans.

Parameter	Value
Voxel Size	1.45 μ m
Source Voltage	50 kV
Source Power	4.5 W
Number of projections	2401
Exposure Time per Projection	$\approx$ 6.5 s
Number of Projections	2401
Rotation	180° – 360°
Filter	LE1
Pixel Binning	1

4.3.3.3 Data Visualization and Analysis Software

The reconstructed 3D volumes were processed and visualized using the Dragonfly software (Version 2024.1, build 1627, ZEISS edition). Each scan produced a dataset of approximately 16 GB, with a representative volume having dimensions of 2030 x 2030 x 2030 voxels, which corresponds to a physical cube of approximately 2.76 mm per side. The visualization workflow centered on adjusting parameters to distinguish the solid material from lower-density features such as voids, cracks, and delaminations.

A primary step in the procedure was the adjustment of window leveling parameters, which control the mapping of voxel grayscale values to the displayed brightness and color. The histogram of values was viewed in a logarithmic scale (Log Y) to obtain finer control over the selection range. For the generation of 2D cross-sectional images, the selected range was adjusted for each specimen using the software’s area function to optimize the contrast between the material and damage features. Table 4.34 summarizes the typical window leveling ranges used for 2D and 3D visualization:

Table 4.34: Typical window leveling ranges for 2D and 3D visualization.

Parameter	2D Views	3D Views
Lookup Table (LUT)	grayscale	grayscale solid2
Selected Range	11,870 – 15,944	12,618 – 65,535
Plotted Range	-149.99 – 65,684.99	-149.99 – 65,684.99
Data Range	0 – 65,535	0 – 65,535

For three-dimensional renderings, the `grayscale solid2` look-up table (LUT) was selected to provide a suitable visual representation. Opacity was controlled mainly using a `Bi-gated ramp` transfer function, although in certain cases, `Bandpass Gaussian` or `Piecewise functions` were used to isolate specific features.

The 3D rendering scene was configured with a uniform white background, and the `High quality` rendering option was enabled for all exported images. Virtual lighting was centered on the volume of interest, and shadows were selectively used with tone mapping to improve the perception of depth on fracture surfaces. A preliminary step in the analysis was the virtual re-alignment of each dataset to orient the laminate’s principal axes with the software’s orthogonal viewing planes, using the `ply` interfaces as a guide. This process is complicated by inherent manufacturing-induced ply waviness and post-fracture deformation. As a consequence, a single 2D slice through the volume may intersect multiple plies, meaning that a clean view of an individual layer is not always achievable.

4.3.3.4 Nomenclature of Viewing Planes

To maintain consistency throughout the volumetric analysis, a standard nomenclature for the orthogonal viewing planes is adopted, as illustrated in Figure 4.32. The planes are defined relative to the specimen's geometry. The top plane (green) provides a through-thickness view, equivalent to the X-Y plane of the laminate. The front plane (red) and lateral plane (blue) provide cross-sectional views along the specimen's width and thickness, respectively. The lateral plane is the view where the stacking sequence of the plies is most clearly visible. This terminology will be used hereafter to describe the orientation of 2D slices and 3D views.

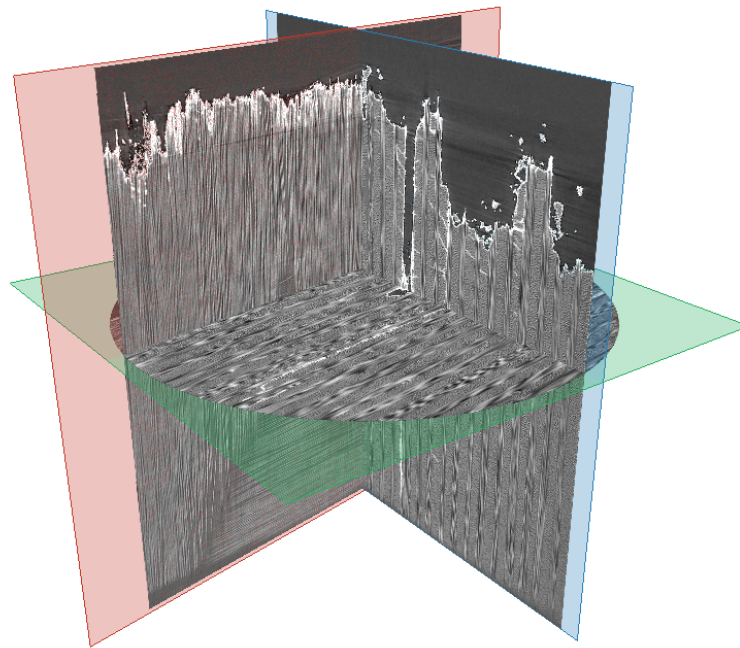


Figure 4.32: Definition of the orthogonal viewing planes used for volumetric analysis: top plane (green), front plane (red), and lateral plane (blue).

4.3.4 Results and Volumetric Damage Analysis

To provide the primary visual context for the volumetric analysis that follows, Figure 4.33 presents the surface strain fields measured by DIC for the 10 mm width specimens from the benchmark study by Danzi. For completeness, and to allow a direct comparison of the size effect on the damage pattern, the corresponding strain fields for the 50 mm width specimens are also shown in Figure 4.34.

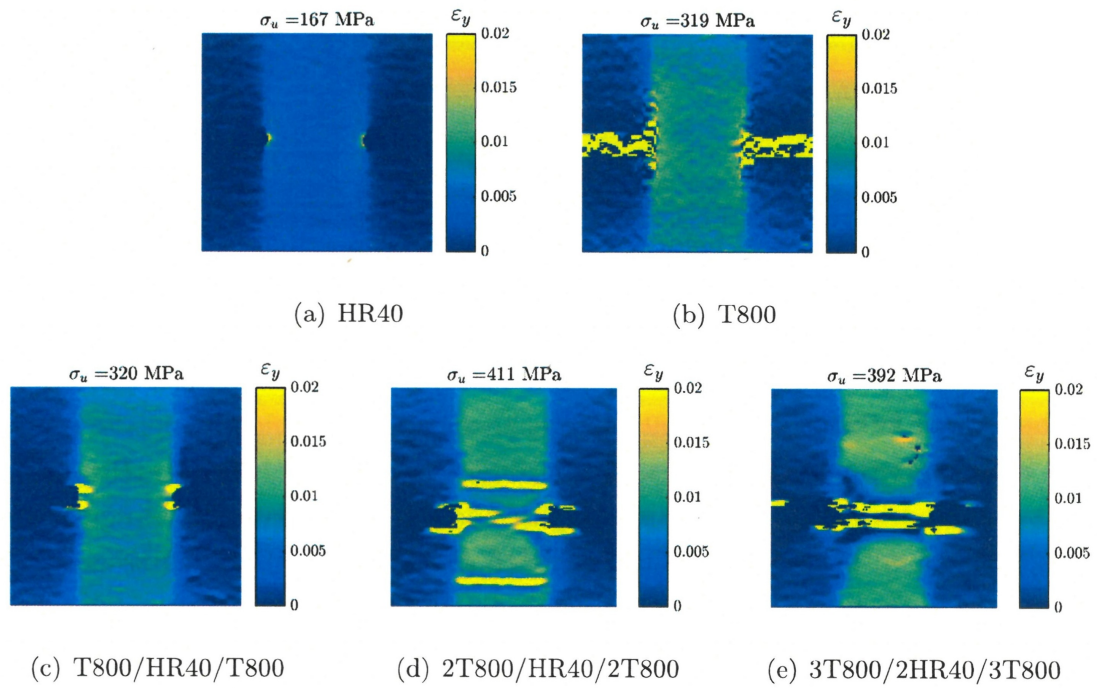


Figure 4.33: Strain field along the loading direction at 95% of the maximum load for the DENT test of 10 mm width specimens (adapted from Danzi [53]).

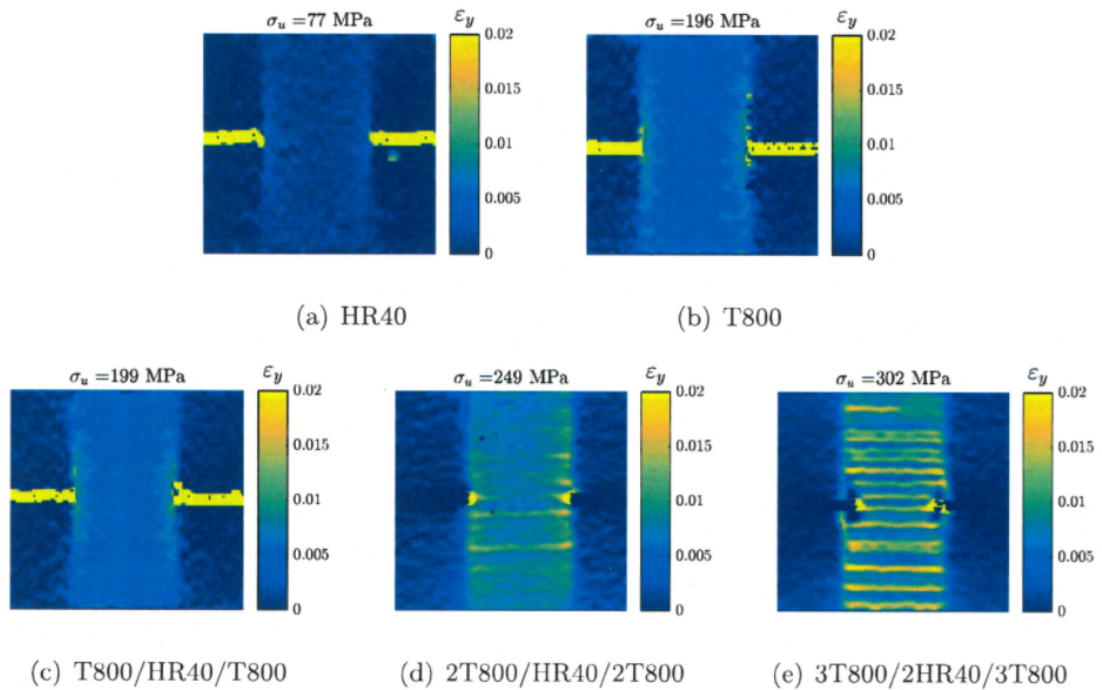


Figure 4.34: Strain field along the loading direction at 95% of the maximum load for the DENT test of 50 mm width specimens (adapted from Danzi [53]).

The following subsections detail the post-mortem μ CT analysis for each of the five material configurations, presenting the findings in order of increasing complexity.

4.3.4.1 HR40 Baseline

The HR40 baseline represents the most brittle behavior in the series. Its DIC strain field (Fig. 4.33a) shows strain highly localized at the notch tips, consistent with a notch-sensitive failure that lacks apparent fragmentation or large split cracks. The volumetric analysis of the fracture surface confirms this, revealing a macroscopically planar fracture with minimal topographical features, as shown in Figure 4.35. This morphology indicates a failure with little energy dissipation from preceding damage accumulation.

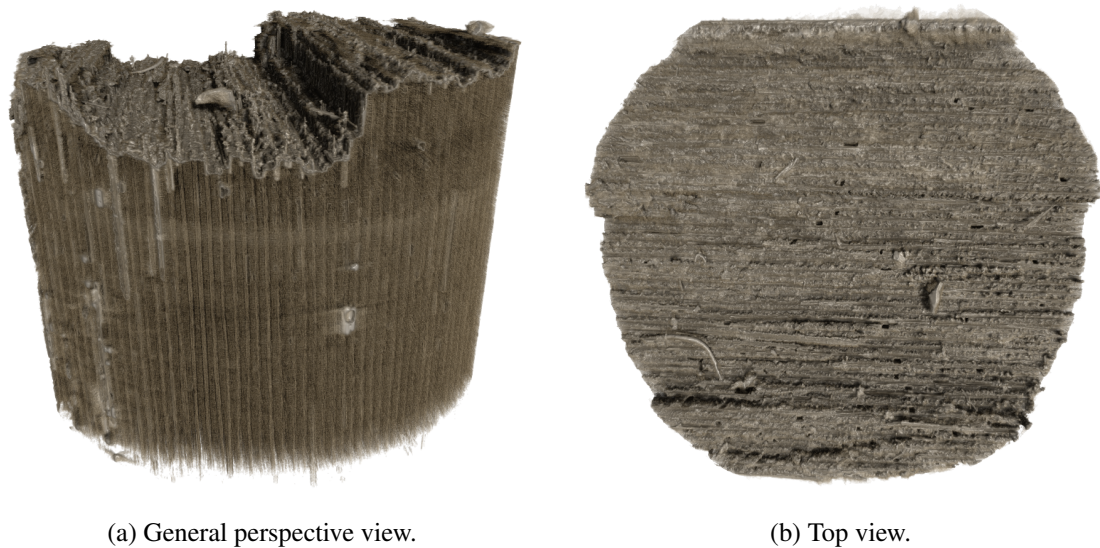


Figure 4.35: Fracture surface of the HR40 baseline specimen, showing increased roughness and topographical complexity from a general perspective (a) and a top view (b).

The volumetric analysis was conducted on a section of the specimen away from the immediate notch region. The internal inspection of this volume reveals the direct consequence of the material's fiber-dominated failure, as detailed in the orthogonal slices in Figure 4.36. The lateral slice (a) and top-down slice (b) show a dense area of fiber fragmentation and failure clusters directly beneath the fracture surface. This damage pattern is a direct result of the low strain-to-failure of the HR40 fibers and the thin-ply architecture itself. The numerous delaminations are therefore the result of this abrupt, fiber-driven event, rather than a preceding toughening mechanism with some of them observed to connect across the interfaces between the 0° and 90° plies.

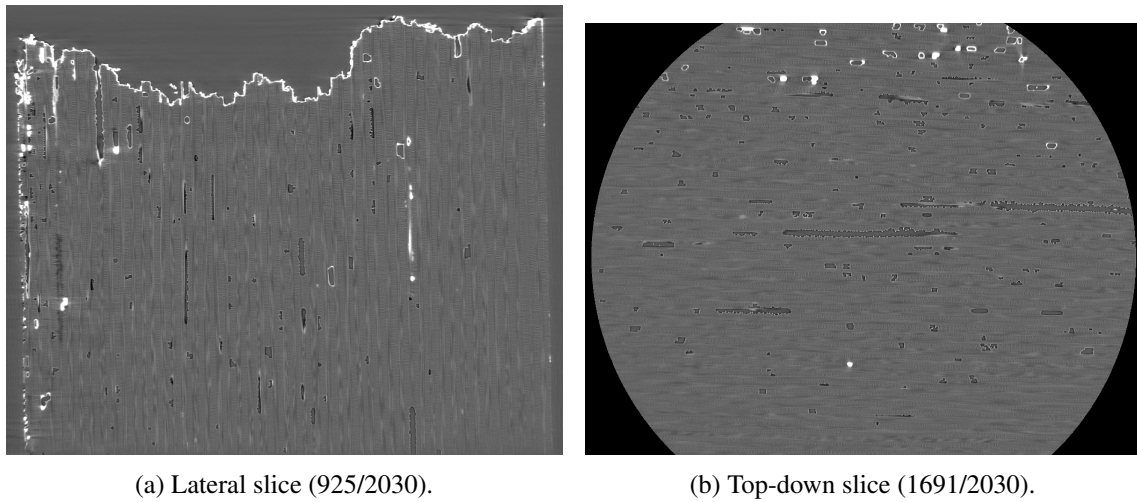


Figure 4.36: CT slices of the HR40 specimen showing clusters of fiber breaks and delaminations.

4.3.4.2 T800 Baseline

The T800 baseline developed a more complex damage zone than the HR40. Its DIC pattern (Fig. 4.33b) is characterized by a primary horizontal damage band between the notches, accompanied by the initiation of short longitudinal splits extending vertically. This indicates the onset of the same toughening mechanisms that are more fully developed in the hybrid systems. The corresponding fracture surface, shown in different perspectives in Figure 4.37, is visibly rougher than that of the HR40, with a more pronounced and irregular topography.

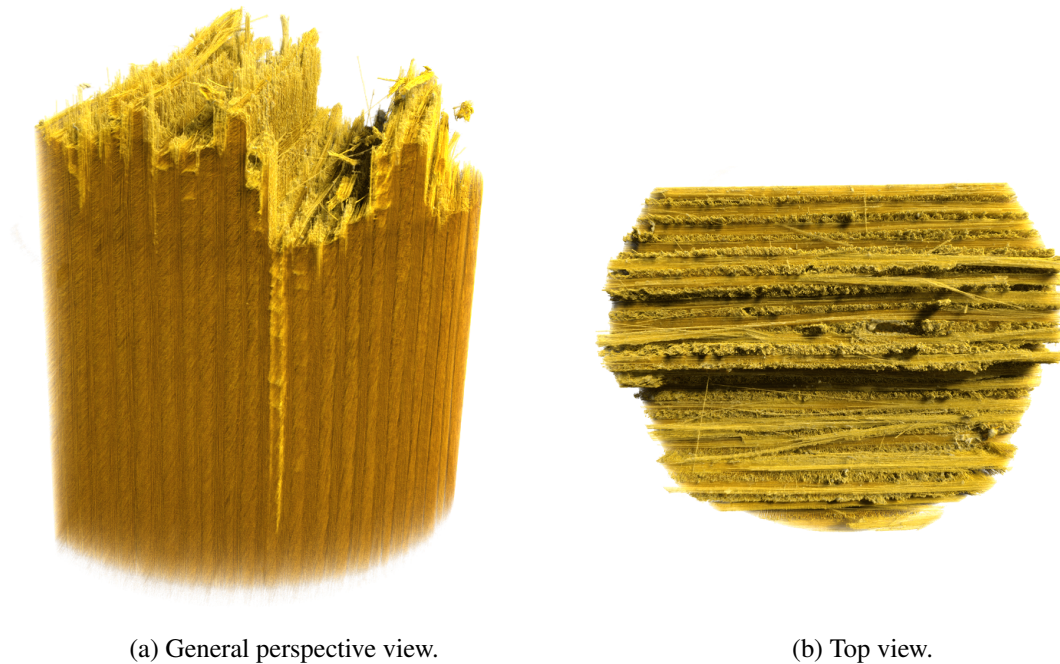


Figure 4.37: Fracture surface of the T800 baseline specimen, showing increased roughness and topographical complexity from a general perspective (a) and a top view (b).

A feature contributing to this surface roughness is the extensive fiber pull-out. The detailed views in Figure 4.38 illustrate the irregular fracture heights of the fiber bundles, which is a departure from the flatter fracture of the HR40 baseline and indicates a greater energy dissipation during failure.

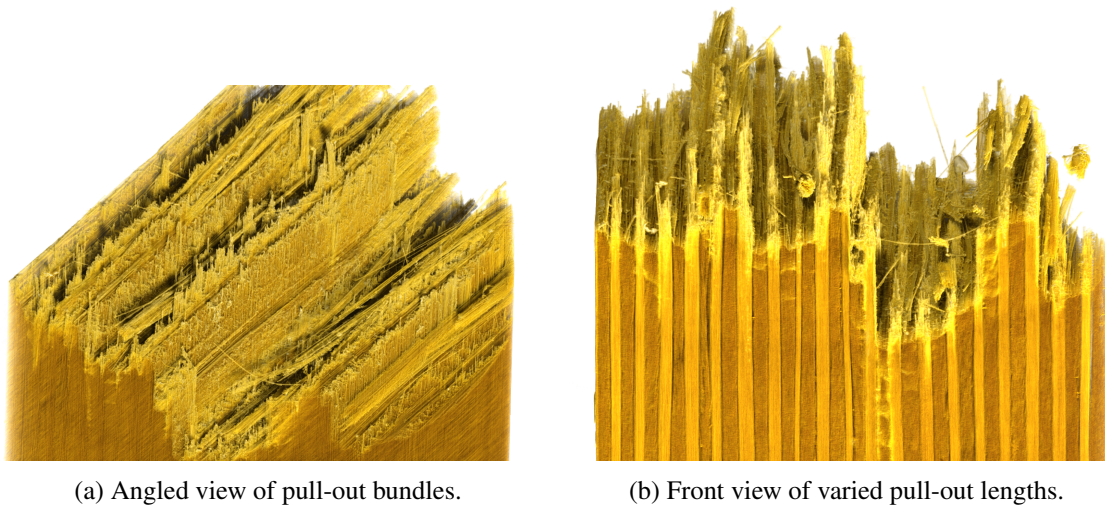


Figure 4.38: Detailed views of fiber pull-out on the T800 fracture surface, showing significant variation in fracture height between fiber bundles.

The internal analysis of the specimen also revealed several manufacturing-induced features, as shown in the top-down slice in Figure 4.39. These include a large subsurface void, resin-rich areas, and local variations in ply thickness. Other features such as transverse cracks that have initiated delamination are also visible.

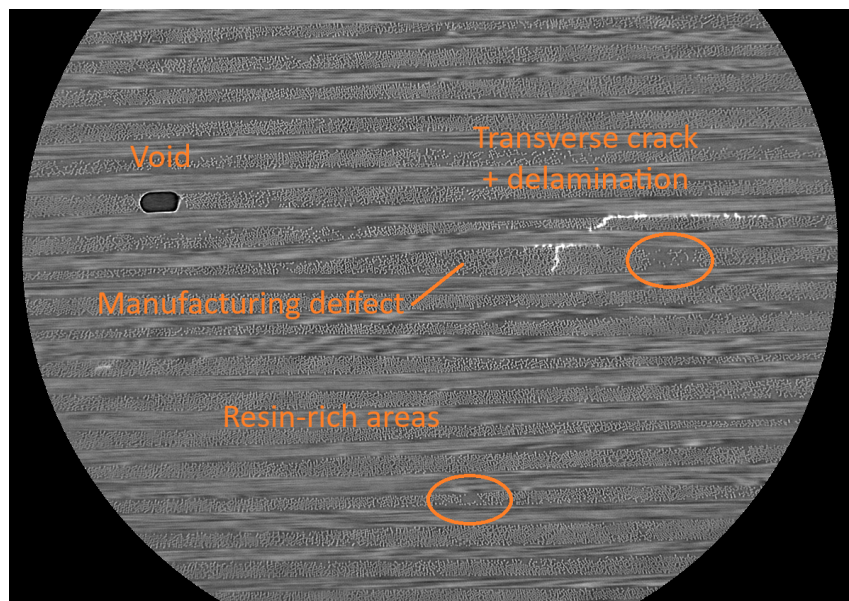
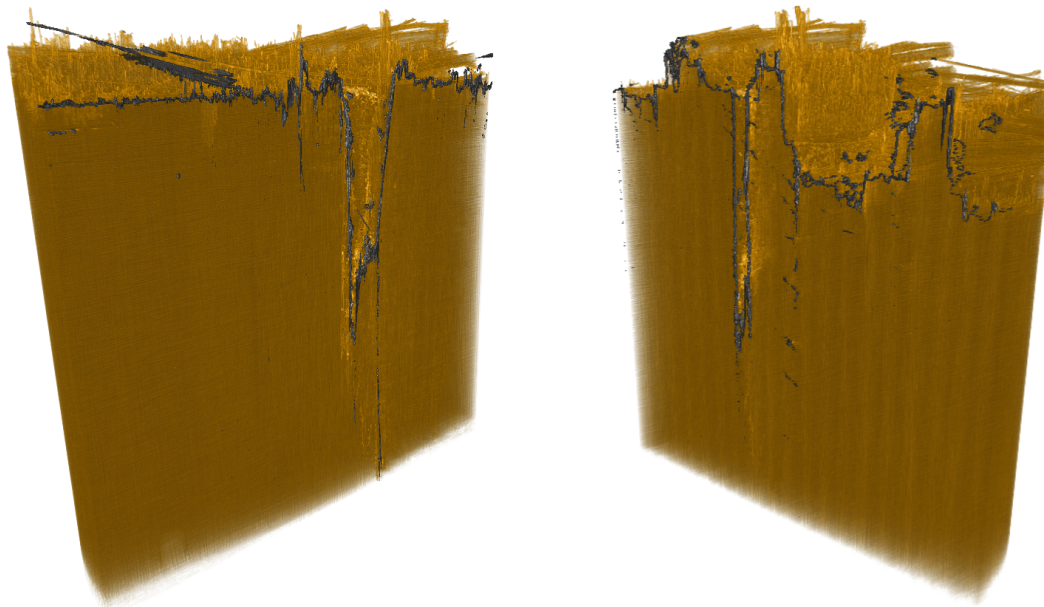


Figure 4.39: Top-down slice of the T800 specimen showing various internal features, including manufacturing defects, voids, and load-induced matrix damage.

Another feature observed on the fracture surface was a deep cavity, shown from two intersecting planes in Figure 4.40. Given its size relative to the surrounding pull-out topography, it is possible this feature resulted from post-test handling or transportation, where a partially fragmented section may have detached from the main body of the specimen.



(a) Front plane intersecting the cavity.

(b) Lateral plane intersecting the cavity.

Figure 4.40: Views of a large cavity on the fracture surface. The gray contour outlines the material's cross-section at the position of the intersecting plane.

The volumetric analysis for this specimen was also conducted away from the immediate notch region. Consequently, the longitudinal splits suggested by the DIC analysis are not present in the scanned volume. The internal inspection does confirm the presence of extensive transverse matrix cracking, as seen in Figure 4.41. This failure mode is fundamentally different from the one observed in the HR40. The higher strain-to-failure of the T800 fibers permits the laminate to reach a state of deformation where the matrix begins to fail progressively through cracking. This is in direct opposition to the HR40, which failed from catastrophic fiber clusters before such widespread matrix damage could occur. The formation of this damage in the T800 is therefore consistent with the high-strain regions observed in the DIC data and provides a baseline for the more extensively developed damage observed in the subsequent hybrid specimens.

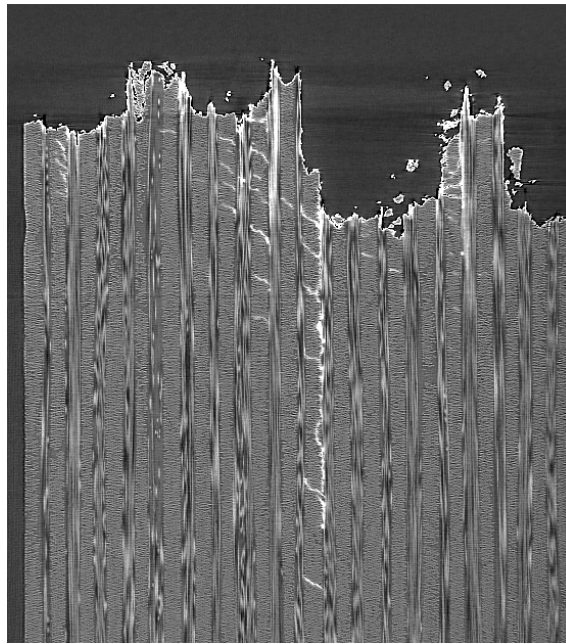


Figure 4.41: Lateral plane view of the T800 specimen showing the formation of transverse matrix cracks, which correlate with the high-strain regions seen in the DIC analysis.

4.3.4.3 [T/H/T] Hybrid

A qualitatively similar failure mode to the T800 baseline is observed in the [T/H/T] hybrid (Fig. 4.33c). The post-mortem analysis reveals a fracture surface, shown in Figure 4.42, that is more complex than that of the T800, with notably longer and more extensive fiber pull-out lengths.

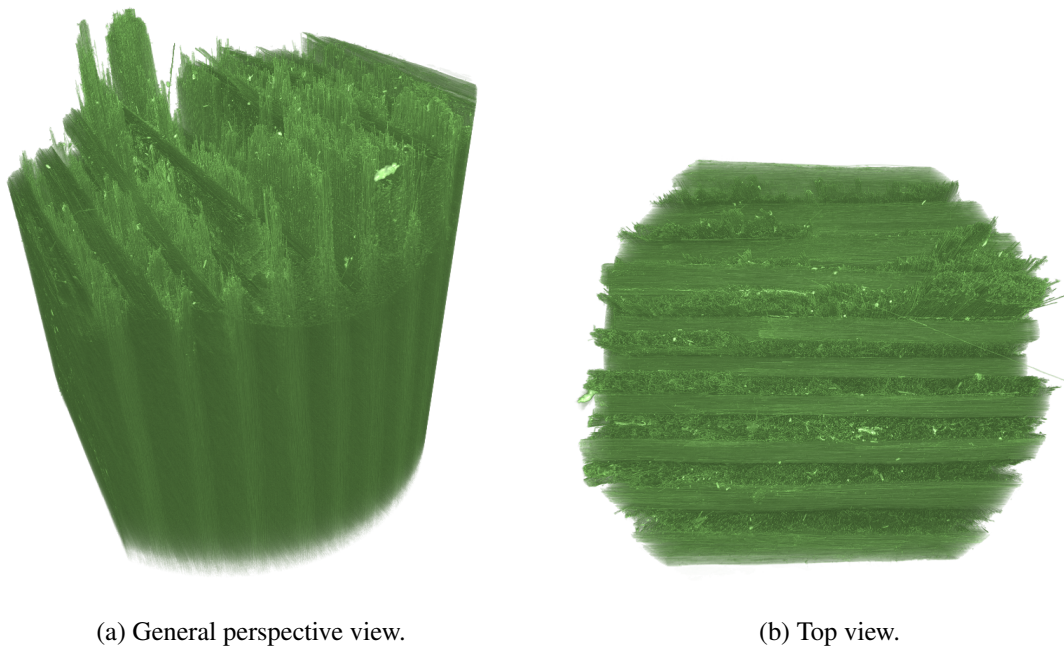
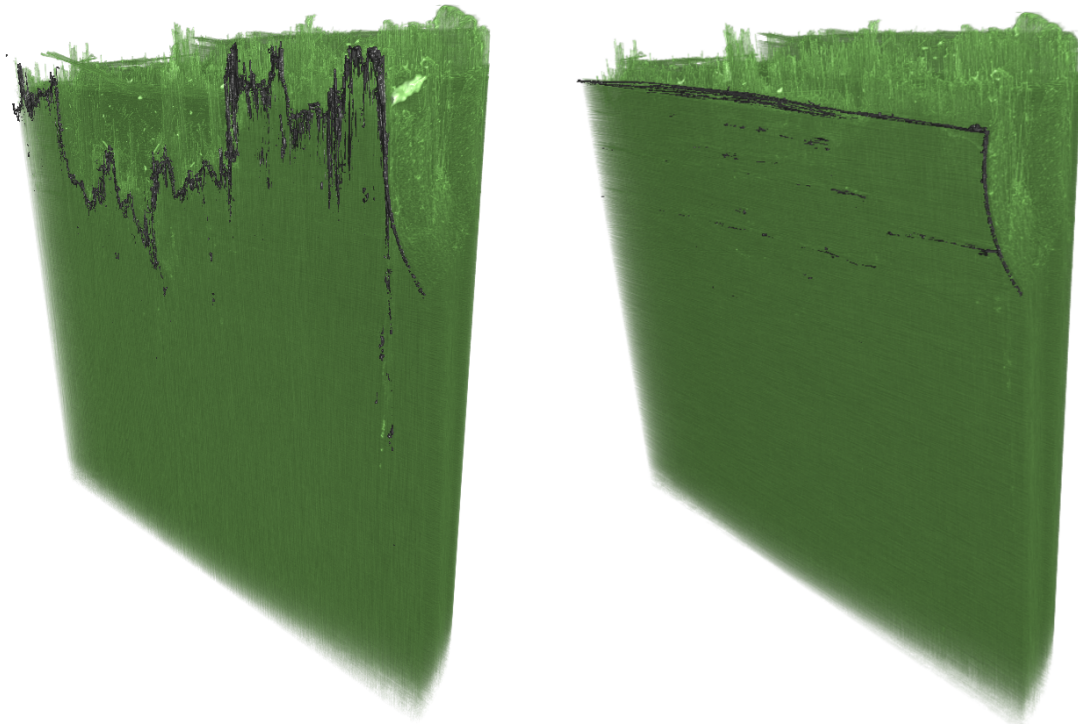


Figure 4.42: Fracture surface of the [T/H/T] hybrid specimen

To investigate the primary toughening mechanisms, a 0° split crack was isolated within the volumetric data, as shown in the 3D renderings in Figure 4.43. These views place the split in the context of the overall fracture volume and also illustrate the difference in texture between the orthogonal ply surfaces. The front view (a) shows the rough surface of the fractured 0° ply, while the lateral view (b) shows the relatively smooth surface of the 90° ply where fibers are oriented parallel to the plane.



(a) Front perspective showing the 0° ply surface. (b) Lateral perspective showing the 90° ply surface.

Figure 4.43: 3D renderings of an isolated 0° split crack from the $[T/H/T]$ specimen, highlighting the crack's geometry in relation to the overall fracture surface.

When this analysis is extended across all valid plies within the volume, a clear and consistent pattern emerges. A sequence of lateral plane slices from the region near the notch, presented in Figure 4.44, illustrates this behavior. The highlighted regions in these slices show that the 0° split cracks consistently form at an offset location from the notch center plane. This observation is in good agreement with the 10° to 20° offset reported by Yang et al. [231] and confirms that their stress redistribution mechanism is applicable to this hybrid system.

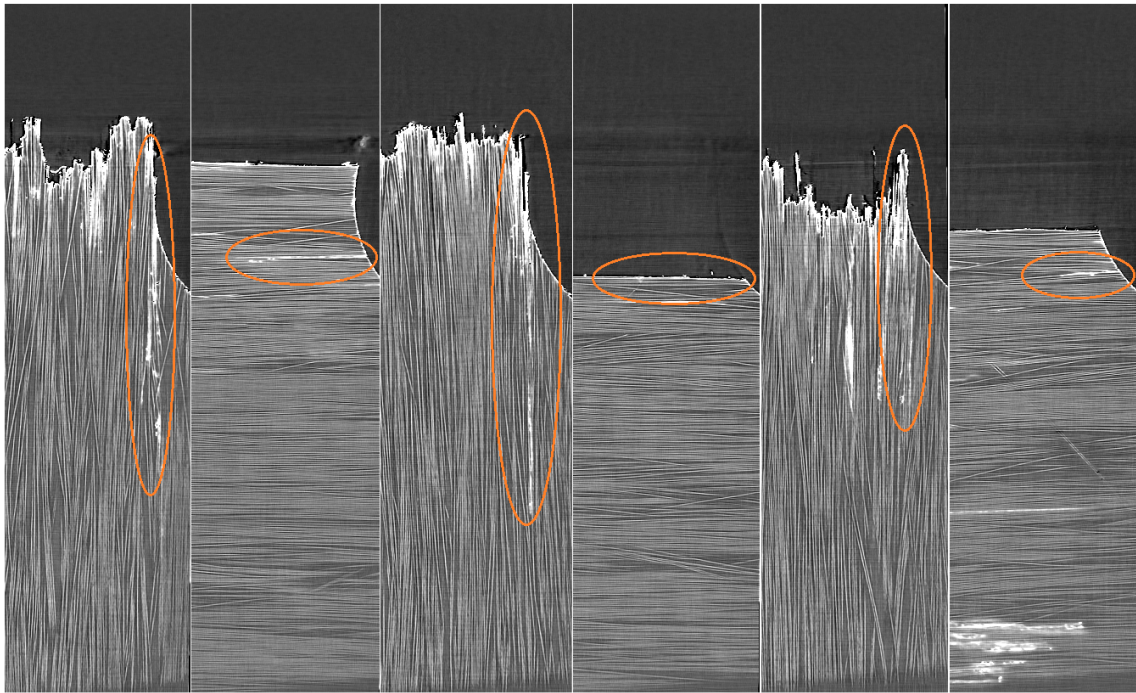
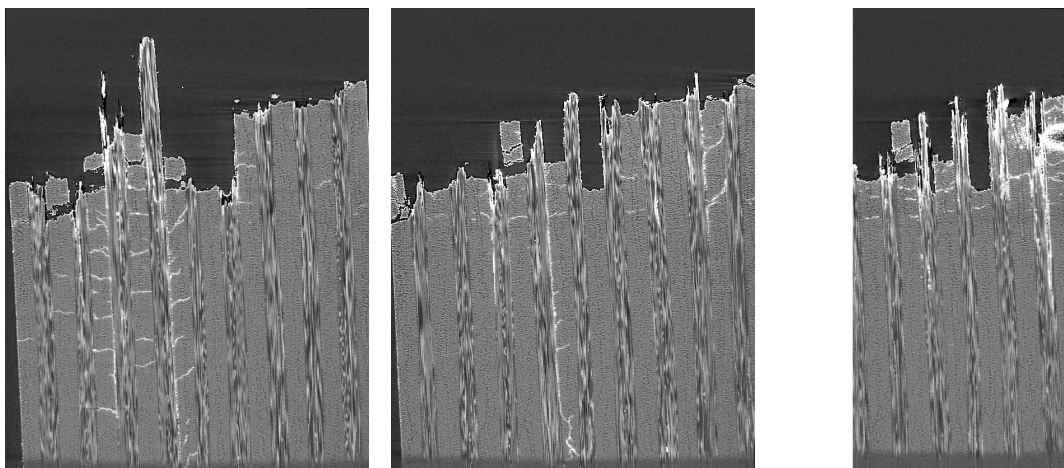


Figure 4.44: Sequence of lateral plane slices from the [T/H/T] specimen showing the initiation of offset 0° split cracks.

In addition to the longitudinal splits, the internal inspection revealed a high prevalence of transverse cracking in the 90° plies, as illustrated in Figure 4.45. The density of these transverse cracks was observed to increase when moving from the lateral plane of the notch towards the center of the specimen, while no visible damage was apparent from the specimen's free edge inward (until reaching the notch). Although less common, some transverse cracks were also found in the 0° plies, typically in proximity to delaminated areas.



(a) Between the center plane and the notch plane (b) Close to the notch plane and respective split crack plane (c) Between the notch plane and the surface of the specimen

Figure 4.45: Tomographic lateral plane cross-sections of the [T/H/T] specimen, selected from a total of 2039 slices. From left to right, the panels show Slice 180, 465, and 1519.

It is hypothesized that the macroscopic similarity in failure mode to the T800 baseline is a result of the laminate's architecture. Fiber failures are thought to initiate in the low-elongation HR40 ply, with the surrounding high-elongation T800 plies acting as a barrier. This contains the fragmentation damage and prevents it from propagating to the primary $0^\circ/90^\circ$ interfaces. By suppressing the formation of large-scale delamination, the laminate behaves like a tougher version of the T800 baseline, where energy is dissipated through these contained mechanisms rather than a shift to a fully delamination-driven failure.

4.3.4.4 [2T/H/2T] Hybrid

A distinct change in failure mode is apparent in the [2T/H/2T] configuration, whose DIC pattern is dominated by long, fully developed vertical high-strain bands (Fig. 4.33d). This shift is attributed to the thicker intermediate layers that promote a notch-insensitive response. Given the similarity of this macroscopic behavior to the [3T/2H/3T] configuration, the post-mortem analysis for this specimen was intentionally performed on a section away from the final fracture. This approach allowed for an investigation of the baseline microstructure of the hybrid material in a relatively undamaged state, with a 3D rendering of the scanned volume shown in Figure 4.46.



Figure 4.46: 3D rendering of the undamaged volume scanned for the [2T/H/2T] specimen, showing the uniform layered microstructure.

The internal scans confirmed a high degree of consolidation with a well-integrated interface between the dissimilar T800 and HR40 materials, as shown in the orthogonal slices in Figure 4.47. However, because both constituent fibers are carbon-based and have similar diameters, there is insufficient material contrast in the μ CT data to distinguish between T800 and HR40 fibers. This finding, while confirming the manufacturing quality, prevents the application of the hybrid

parameter evaluation methods detailed in Section 2.7.1. Specifically, parameters that require the identification and spatial location of individual fiber types, such as Dispersion and the Degree of Hybridization, cannot be calculated. While parameters that describe the overall fiber packing, like Areal Disorder, Inter-Fiber Spacing, and Local Fiber Volume Fraction, could theoretically be evaluated for the laminate as a whole, they cannot be assessed for the individual constituent materials.

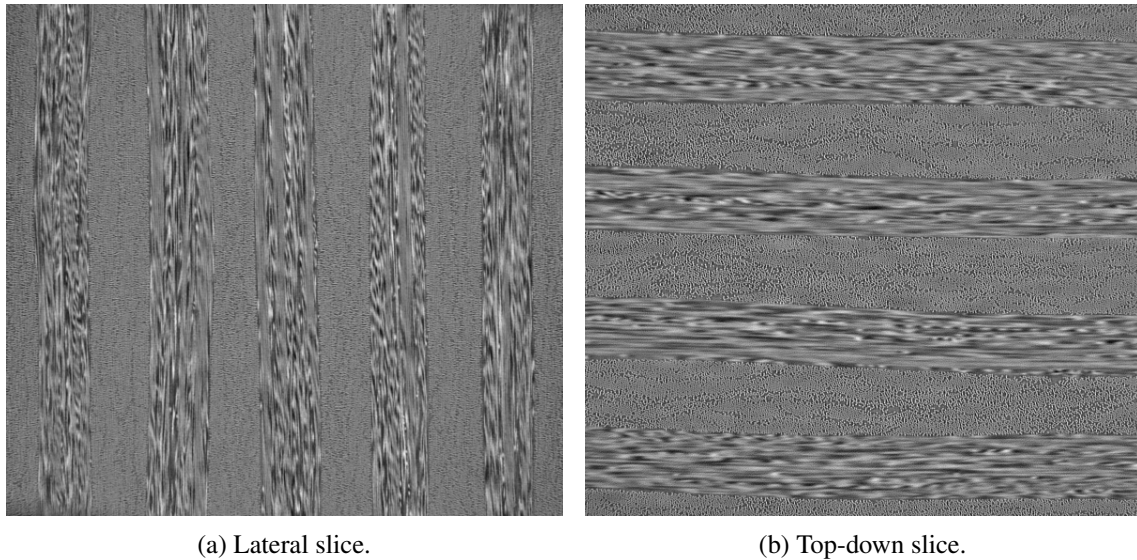


Figure 4.47: Orthogonal slices showing the baseline microstructure of the $[2T/H/2T]$ specimen.

A magnified view of the top-down slice, presented in Figure 4.48, provides a clearer look at the high-density packing and distribution of the individual carbon fibers within a single ply.

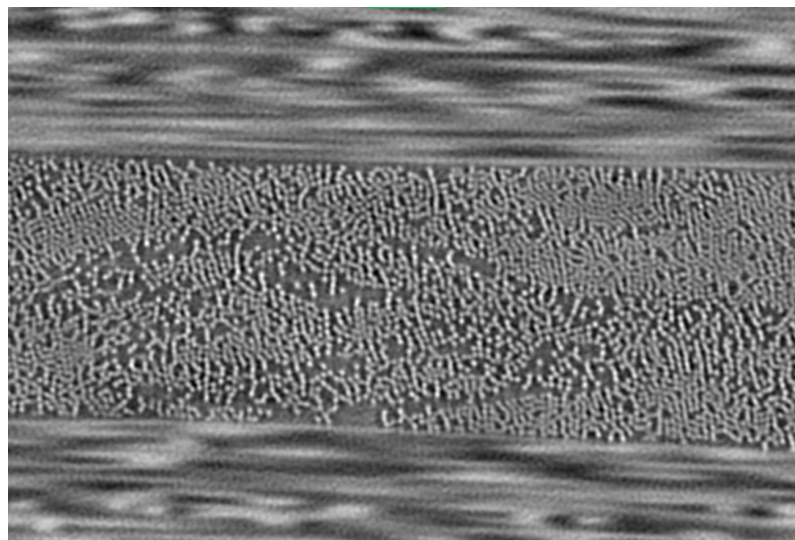


Figure 4.48: Magnified top-down view of a single ply from the $[2T/H/2T]$ specimen.

4.3.4.5 [3T/2H/3T] Hybrid

The [3T/2H/3T] configuration shows the most extensively developed vertical strain bands in the DIC analysis (Fig. 4.33e). The post-mortem analysis for this specimen was focused on a region near the notch, a decision informed by the high density of transverse fracture expected from the DIC data and the very large pull-out lengths characteristic of this layup.

The resulting fracture surface, shown in Figure 4.49, is the most complex of all configurations.

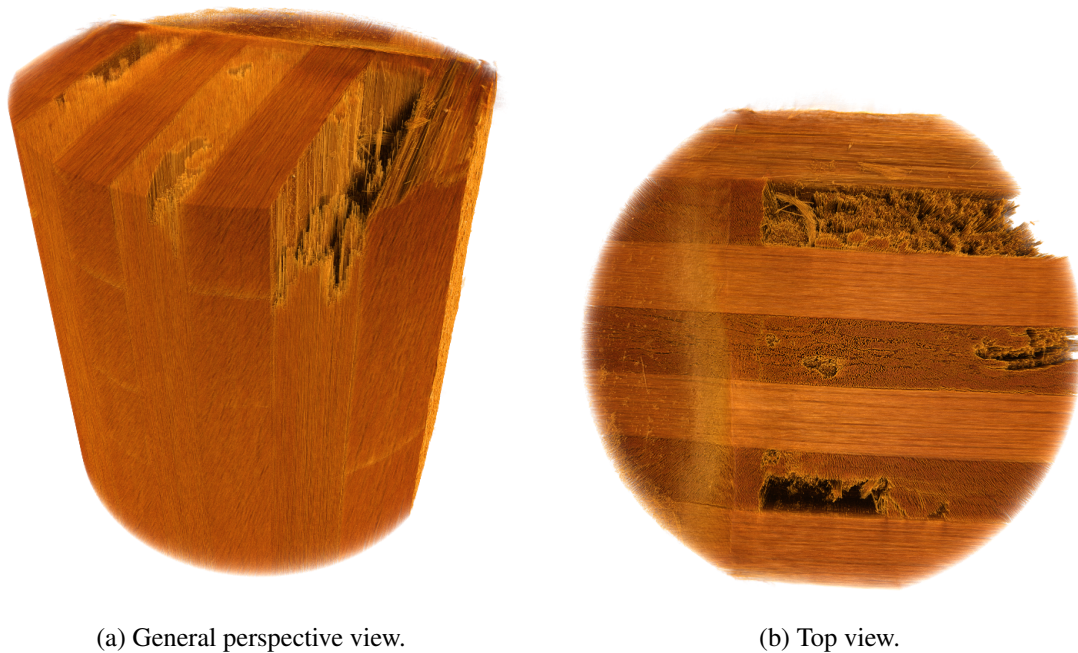


Figure 4.49: Fracture surface of the [3T/2H/3T] hybrid specimen.

The internal analysis revealed a direct interaction between different damage modes. As illustrated in Figure 4.50, the 0° splits originate from the area of the notch and propagate vertically. These vertical splits then lead to the formation of multiple transverse cracks that start in the same region. The lateral view highlights that these transverse cracks form a network and are not confined to a single height, although they are more prevalent near the final fracture plane. While not visible across the full extent of the scanned volume, these transverse cracks are presumed to propagate horizontally to reach the corresponding 0° split originating from the opposing notch of the DENT specimen.

This underlying damage interaction provides a direct explanation for the horizontal high-strain bands observed in the DIC analysis for the 10 mm specimens (Fig. 4.33). While the CT analysis was performed on this specific width, the failure mechanism appears to be a general characteristic of the material. A comparison with the strain fields for the 50 mm width specimens (Fig. 4.34) reinforces this conclusion. The data shows that this pattern of transverse damage becomes even more clear and extensive in wider configurations, suggesting that this is a common aspect of the material's failure process.

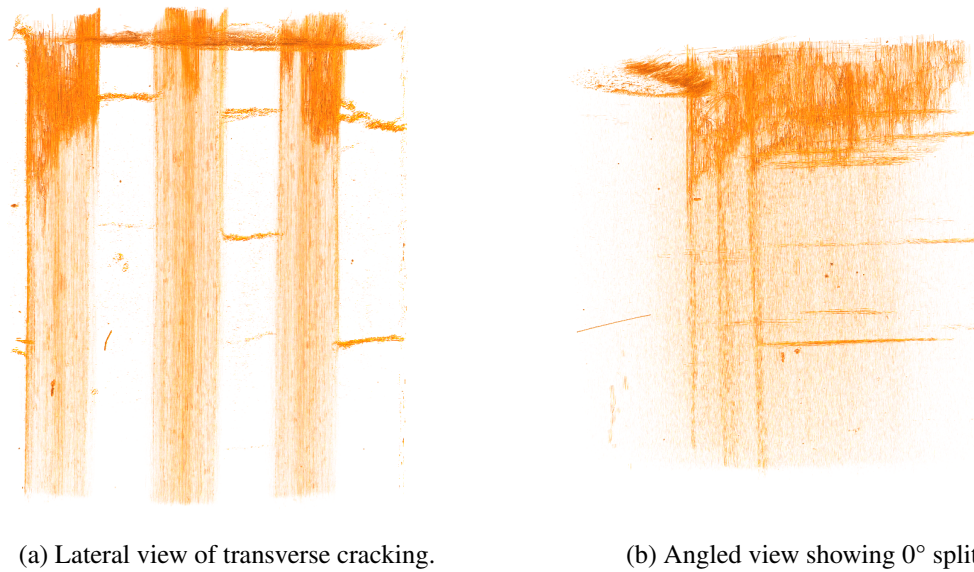


Figure 4.50: Spatial correlation between transverse cracking in a 90° ply and the split crack in the adjacent 0° ply.

Similar to the $[T/H/T]$ hybrid, analysis of the lateral plane slices confirmed that 0° splits initiated at offset locations from the notch root, in agreement with the mechanism proposed by Yang et al. [231]. A sequence of these slices is shown in Figure 4.51.

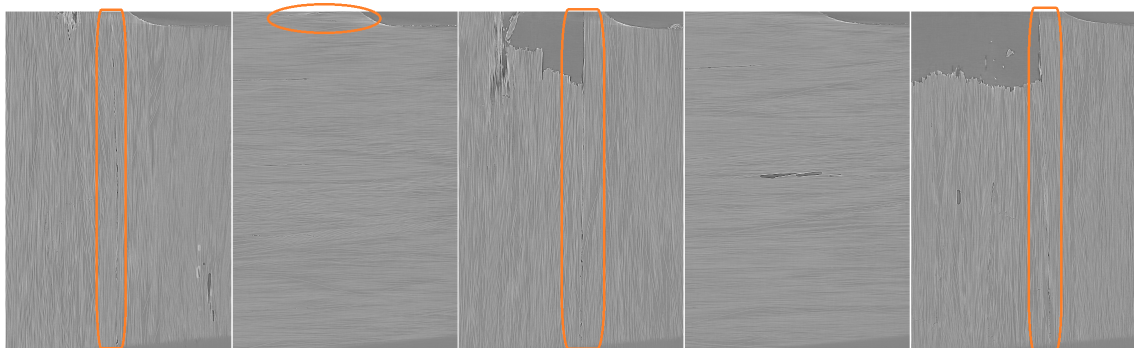


Figure 4.51: Sequence of lateral plane slices from the $[3T/2H/3T]$ specimen. The highlighted regions show the consistent initiation of 0° split cracks at an offset from the notch center, confirming the mechanism proposed by Yang et al. [231]

Furthermore, a spatial correlation was found between the initiation of these 0° split cracks and the location of extensive fiber pull-out. The segmented images in Figure 4.52 show the material volume containing the pull-out bundles. This analysis revealed that the face from which the pull-out originates often coincides with the plane where a split crack initiates.

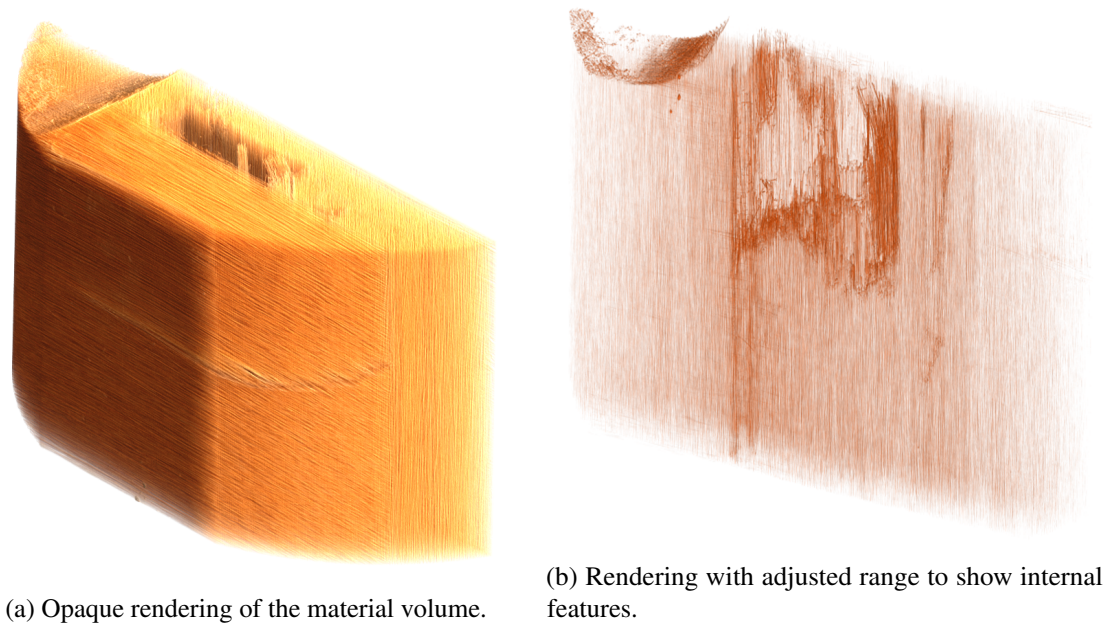


Figure 4.52: Segmented volumetric data isolating the extensive fiber pull-out bundles in the [3T/2H/3T] specimen to show the correlation with split crack initiation planes.

The overall morphology is consistent with extensive fragmentation of the low-elongation HR40 plies in the laminate's core. Because the high-elongation T800 plies were located at the $0^\circ/90^\circ$ interfaces in this layup, the mechanism of preferential fiber fracture at interfaces, as identified by Mesquita et al. [144], could not be validated. However, the analysis of the fracture profiles across the hybrid specimens confirms the ply isolation hypothesis from AhmadvashAghbash et al. [3]. The 3D views show that the fracture surfaces of consecutive 0° plies do not follow one another. If the plies were not isolated, a linear interpolation of the fracture height between two adjacent 0° plies would result in a similar, interpolated fracture behavior in any intermediate ply. The volumetric data does not show this behavior, and as such, the hypothesis that the 90° plies effectively isolate the load-bearing 0° plies is validated.

Chapter 5

Conclusions and Future Work

5.1 Conclusions

This thesis investigated two distinct manufacturing strategies for creating hybrid carbon fiber composites: filament-level hybridization through air-jet co-mingling and ply-level hybridization through the stacking of interply prepregs. The former strategy was pursued through an experimental manufacturing campaign, while the latter combined an in-depth analysis of existing benchmark data with a culminating, high-resolution tomographic investigation to elucidate the governing three-dimensional damage mechanisms.

The experimental campaign focused on producing a co-mingled yarn intended to be a fiber-by-fiber blend of two dissimilar carbon fibers. Due to material constraints, the process was evaluated using a single carbon fiber type. The results demonstrated that the manufacturing process, conducted on machinery not specialized for brittle, high-performance fibers, was unable to achieve the intended filament-level dispersion. Further examination revealed that the constituent tows did not open and intermingle. This outcome is attributed to a combination of factors, including the inherent stiffness of the carbon filaments and the cohesion from fiber sizing, which the machine's configuration and selected process parameters were insufficient to overcome. Furthermore, some observed fiber damage, in the form of loose filaments, was likely attributable not only to the air jet itself but also to the non-standard feeding system and rudimentary guides required for the experiment. Although the process successfully combined the input tows into a single continuous yarn, the resulting internal structure was not a homogeneous blend but rather a parallel, side-by-side arrangement. The necessity for extensive modifications to the experimental setup further demonstrates that co-mingling stiff and cohesive carbon fibers cannot be achieved by exclusively adjusting process parameters on standard textile machinery. This implies that successful dispersion would require a process line where each component, from the fiber feed system to the nozzle geometry, is specifically designed for such materials. While this outcome did not meet the primary objective, the resulting side-by-side yarn could serve as a precursor material for manufacturing intraply hybrid composites using Automated Fiber Placement, possibly eliminating the need for material changes during the layup of a ply, a process which can introduce manufacturing defects

like gaps or overlaps.

The second part of the thesis focused on interply hybrids, beginning with a systematic methodology for selecting pseudo-ductile configurations using analytical damage mode maps. This framework was used to analyze a benchmark study on thin-ply hybrid composites, confirming that while certain hybrid configurations (such as $[2T/H/2T]$ and $[3T/2H/3T]$) exhibited the predicted pseudo-ductile behavior, the predictive model had limitations, as the $[T/H/T]$ hybrid failed in a brittle manner contrary to expectations. To resolve ambiguities from the benchmark's surface-based analysis, a detailed post-mortem analysis of the DENT specimens was conducted using micro-Computed Tomography (μ CT). This volumetric investigation provided direct, three-dimensional evidence of the internal damage state, leading to several valuable conclusions. The analysis confirmed a consistent failure mechanism progression (excluding the brittle HR40 baseline) beginning with the formation of 0° split cracks, followed by transverse cracking that led to delamination and ultimate failure. A central finding was the role of the high-elongation T800 plies, which, being located at the interfaces between the 0° and 90° plies, acted as a barrier that contained the fragmentation of the more brittle HR40 plies located in the sublaminates. This containment mechanism is the primary source of the observed pseudo-ductility and the significant increase in fracture toughness. Further observations included the extensive fiber pull-out in the tougher configurations, particularly the $[3T/2H/3T]$ specimen, and the excellent manufacturing quality confirmed by the clean, well-consolidated microstructure of the $[2T/H/2T]$ specimen, where layers of the same orientation were indistinguishable. Ultimately, the analysis concluded that the specific hybrid layup had less influence on the damage mechanisms than the constituent ply thicknesses, with the differences in performance being justifiably attributed to this factor. In addition to these primary findings, the investigation validated several existing hypotheses from the literature. The analysis confirmed that 0° splits initiate at an offset from the notch root, supporting the stress redistribution mechanism proposed by Yang et al. [231]. It also provided evidence for the ply isolation effect hypothesized by AhmadvashAghbash et al. [3], by showing a lack of correlation in the fracture morphology of consecutive 0° plies. Finally, the μ CT data clarified that the high-strain stripes observed on the surface of the 90° plies correspond to fully formed transverse cracks, confirming they are a primary damage event.

In summary, this work provides a comprehensive evaluation of two pathways for carbon-carbon hybridization. It presents a realistic assessment of the difficulties in achieving filament-level dispersion through co-mingling, while simultaneously offering additional understanding of the toughening phenomena in ply-level hybrids through tomographic characterization.

5.2 Future Work

The findings and limitations of this research naturally lead to several paths for future investigation, which can be broadly categorized into the areas of co-mingling process development and advanced characterization of interply hybrids.

For the co-mingling process, future work should proceed along two parallel tracks. The first is to fully characterize the novel side-by-side yarn produced in this study. This would involve mechanical testing of individual fibers extracted from the yarn to quantify any process-induced damage. Subsequently, composite specimens should be manufactured by impregnating the yarn with a resin and their mechanical properties compared against intraply specimens made by placing individual tows next to each other, in order to validate the proposed application in automated manufacturing. The second track should focus on achieving the original objective of a fiber-by-fiber blend. This requires a systematic optimization of the co-mingling process, centered on the air-jet nozzle. A parametric study using Computational Fluid Dynamics (CFD) could be employed to simulate and optimize different nozzle geometries. Promising designs could then be fabricated using 3D printing and their performance in opening and intermingling the tows could be experimentally compared against the standard nozzle.

For the interply hybrids, the post-mortem μ CT analysis laid the basis for more advanced investigations. A possible next step is to conduct *in situ* testing of DENT specimens using Synchrotron Radiation Computed Tomography (SRCT). This would allow for the real-time visualization of damage initiation and propagation, clarifying phenomena not accessible in post-mortem analysis, such as the differential cracking sequence between 0° and 90° plies and the propagation of splits and transverse cracks. The analysis should also be extended to specimens with widths beyond 10 mm to investigate potential size effects on the observed failure mechanisms. Furthermore, the application of X-ray tomography should be extended from DENT specimens to other mechanical test configurations to build a more complete understanding of damage in thin-ply carbon hybrids.

Across both research areas, significant opportunities exist for the integration of artificial intelligence, particularly deep learning. The vast datasets produced by μ CT (often requiring significant memory and dedicated hardware for analysis) could be processed more efficiently using AI-based segmentation algorithms. Such tools could automate the identification of damage features and, through their ability to recognize patterns and interpolate data, potentially reduce the computational resources required for analysis. For complex 4D datasets from *in situ* tests, pattern recognition could be very useful in understanding the intricate evolution of damage. Finally, it could also be employed to overcome specific characterization challenges, such as developing a method to distinguish between different carbon fiber types in a hybrid cross-section based on subtle textural differences, a task that proved difficult with conventional imaging techniques.

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Appendix A

MATLAB Code Listings

This appendix contains the MATLAB code used for calculations presented in this work.

A.1 Lamina Property Calculation Script

This MATLAB script calculates the fiber volume fraction (V_f) and the longitudinal lamina stiffness (E_L) for composite materials using two different fiber types (T800 and HR40) embedded in a ThinPreg(TM) 80EP-736 matrix. It takes a range of resin weight fractions (W_r) as input. The calculations are performed based on the material properties (Young's modulus E and density ρ) provided for the matrix and fibers.

The script utilizes two helper functions: `fiberVolumeFraction` to convert resin weight fraction to fiber volume fraction, and `laminaStiffness` to calculate the longitudinal modulus based on the Rule of Mixtures ($E_L = E_f V_f + E_m(1 - V_f)$). Both functions include an option to round the results. Finally, the script organizes and displays the calculated V_f and E_L values for both fiber types in a table corresponding to the input resin weight fractions.

Listing A.1: LaminaProperties

```
% -----  
% | SCRIPT: LaminaProperties.m |  
% | AUTHOR: Raul Maciel Oliveira |  
% | DATE: September 2025 |  
% |-----|  
% | CONTEXT: This script was developed for the Master's Thesis: |  
% | "Hybrid composite laminates for a new generation of aerospace |  
% | structures" |  
% | |  
% | Master's Degree in Mechanical Engineering |  
% | Specialization in Aeronautical Structures and Vehicles |  
% | FACULDADE DE ENGENHARIA DA UNIVERSIDADE DO PORTO (FEUP) |  
% -----  
  
%% --- INITIALIZATION ---  
% Clear workspace, command window, and close any figures.  
clear;
```

```

clc;
close all;

%% --- SCRIPT CONFIGURATION ---
rounded = true; % Set to true to round results, false for full precision

%% --- MATERIAL PROPERTIES INPUT ---
% Define the properties of the constituent materials. All properties should be in
% consistent units.

% Matrix Properties: ThinPreg(TM) 80EP-736
Em = 3.44; % Matrix Young's Modulus [GPa]
rho_m = 1.21; % Matrix density [g/cm^3]

% Fiber Properties: T800 Carbon Fiber
Ef_T800 = 294; % Fiber Young's Modulus [GPa]
rho_f_T800 = 1.80; % Fiber density [g/cm^3]

% Fiber Properties: HR40 Carbon Fiber
Ef_HR40 = 395; % Fiber Young's Modulus [GPa]
rho_f_HR40 = 1.82; % Fiber density [g/cm^3]

% Define the range of resin weight fractions to be analyzed
Wr_percent = [32; 37; 42]; % Resin weight content [%]
Wr = Wr_percent / 100; % Convert percentage to fraction for calculations

%% --- CALCULATION ---

% Step 1: Calculate Fiber Volume Fraction (Vf) for both fiber types.
Vf_T800 = fiberVolumeFraction(Wr, rho_m, rho_f_T800, rounded);
Vf_HR40 = fiberVolumeFraction(Wr, rho_m, rho_f_HR40, rounded);

% Step 2: Calculate Lamina Longitudinal Stiffness (EL) using Rule of Mixtures.
EL_T800 = laminaStiffness(Ef_T800, Vf_T800, Em, rounded);
EL_HR40 = laminaStiffness(Ef_HR40, Vf_HR40, Em, rounded);

%% --- DISPLAY RESULTS ---
% Organize the calculated data into a table.
results = table(Wr_percent, Vf_T800, Vf_HR40, EL_T800, EL_HR40, ...
    'VariableNames', {'ResinContent_Percent', 'Vf_T800', 'Vf_HR40', 'EL_T800_GPa', 'EL_HR40_GPa'});

disp('-----');
disp('Calculated Lamina Properties (using Rule of Mixtures):');
disp('-----');
disp(results);

%% --- HELPER FUNCTIONS ---

function Vf = fiberVolumeFraction(Wr, rho_m, rho_f, rounded)
% Computes fiber volume fraction from weight fraction.

% Formula for converting weight fraction to volume fraction
Vf = ((1 - Wr) .* rho_m) ./ ( (1 - Wr) .* rho_m + Wr .* rho_f );

if rounded

```

```

        Vf = round(Vf, 3);
    end
end

function EL = laminaStiffness(Ef, Vf, Em, rounded)
    % Calculates longitudinal modulus using the Rule of Mixtures.

    if rounded
        Vf = round(Vf, 3);
    end

    % Linear Rule of Mixtures formula
    EL = Ef .* Vf + Em .* (1 - Vf);
end

```

A.2 Damage Mode Map Calculation Script

This MATLAB script implements the analytical methodology developed by Jalalvand et al. [103, 106] to calculate and plot the damage mode map boundaries for unidirectional hybrid composites. The script is designed to analyze multiple datasets, including the primary data for this thesis alongside validation cases from the literature, and it accommodates calculations for various combinations of constituent material properties.

The boundaries on the map, which plots the LE layer thickness (t_L) against its relative thickness fraction (γ), are determined by the conditions where two failure modes are predicted to occur simultaneously. The script calculates the absolute LE layer thickness (t_L) for these boundaries over a range of relative LE thickness fractions (γ), while also providing an option to display the vertical axis as the equivalent Fiber Areal Weight (FAW). The specific equations implemented for the boundaries are derived by equating the respective failure-initiating stresses and include an empirical correction factor of 2:

- **LE Fragmentation / Delamination (LD) Boundary ($\sigma_{LF} = \sigma_{del}$):**

$$t_{L,LD} = \frac{4G_{CH}E_H}{S_L^2} \frac{\alpha(1-\gamma)}{\alpha\gamma+1-\gamma}$$

- **LE Fragmentation / HE Failure (LH) Boundary ($\sigma_{LF} = \sigma_{HF}$):**

$$t_{L,LH} = 2 \left(\frac{S_H}{K_t S_L} \frac{\alpha}{\alpha\beta+1} \right)^m \frac{\beta}{2WL}$$

- **Delamination / HE Failure (DH) Boundary ($\sigma_{del} = \sigma_{HF}$):**

$$t_{L,DH} = 2 \left(\frac{K_t(2WL)^{1/m}}{S_H\beta^{1/m}} \sqrt{2G_{CH}E_H \frac{1+\alpha\beta}{\alpha}} \right)^{\frac{2m}{m-2}}$$

After calculating these boundary values, the script generates a tiled layout of damage mode maps, with one map for each material property combination. Additionally, the script can produce a single figure where all boundary curves are overlaid for direct comparison.

Listing A.2: FailureModeMaps.m

```
% -----
% | SCRIPT: FailureModeMaps.m |
% | AUTHOR: Raul Maciel Oliveira |
% | DATE: September 2025 |
% |-----|
% | CONTEXT: This script was developed for the Master's Thesis: |
% | "Hybrid composite laminates for a new generation of aerospace |
% | structures" |
% | |
% | Master's Degree in Mechanical Engineering |
% | Specialization in Aeronautical Structures and Vehicles |
% | FACULDADE DE ENGENHARIA DA UNIVERSIDADE DO PORTO (FEUP) |
% |-----|

%% --- INITIALIZATION ---
% Clear workspace, command window, and close any existing figures.
clear;
clc;
close all;

%% --- SCRIPT CONFIGURATION ---
% Select the dataset to use for the calculations.
% 'OLIVEIRA': Main dataset for this thesis.
% 'BROGGI', 'JALALVAND', 'DANZI': Validation datasets from literature.
modelOption = 'OLIVEIRA';

% Set to true to generate an additional figure showing all curves on a single plot.
showOverlapFigure = true;

%% --- INPUT DATA & PARAMETERS ---
% This section defines the material properties and geometric parameters based on the
% selected 'modelOption'.

switch modelOption
case 'OLIVEIRA'
    % --- Case 1: Main Thesis Data ---
    validation = false;
    ylimUnique = 0.12; % Upper y-limit for thickness [mm]
    fawOption = true; % Plot y-axis as Fiber Areal Weight (gsm) instead of thickness

    % LE Ply Properties (HR40) for different Vf
    EL_options = [232.9, 211.4, 191.0]; % Young's Modulus [GPa]
    SL_options = [2.58, 2.34, 2.11]; % Longitudinal Strength [GPa]
    Vf_L_options = [32, 37, 42]; % Fiber Volume Fraction [%]

    % HE Ply Properties (T800) for different Vf
    EH_options = [174.3, 158.6, 143.2]; % Young's Modulus [GPa]
    SH_options = [3.46, 3.14, 2.83]; % Longitudinal Strength [GPa]
    Vf_H_options = [32, 37, 42]; % Fiber Volume Fraction [%]
```

```

% General Model Parameters
m = 30;           % Weibull Modulus
GIIC = 0.55;      % Mode II Critical Energy Release Rate [N/mm or MPa*mm]
Kt = 1.08;        % Stress Concentration Factor
W = 15;           % Laminate Width [mm]
L = 80;           % Laminate Length [mm]

case 'DANZI'
    % --- Case 2: Validation Data from Danzi et al. ---
    validation = true;
    ylimUnique = 0.12;
    fawOption = false;

    % LE Material (HR40)
    EL_options = [220.6, 207.9, 195.5];
    SL_options = [2.4608, 2.3197, 2.1829];
    Vf_L_options = [35, 38, 41];

    % HE Material (T800)
    EH_options = [165, 155.5, 146.3];
    SH_options = [3.2987, 3.1105, 3.2987];
    Vf_H_options = [35, 38, 41];

    m = 30;
    GIIC = 0.5;
    Kt = 1.08;
    W = 15;
    L = 80;

case 'BROGGI'
    % --- Case 3: Validation Data from Broggi et al. ---
    validation = true;
    ylimUnique = 0.12;
    fawOption = true;

    % LE Material (HR40)
    EL_options = [375.0 * 0.5];
    SL_options = [4.41 * 0.5];

    % HE Material (T300-700)
    EH_options = [294 * 0.5];
    SH_options = [5.880 * 0.5];

    m = 41;
    GIIC = 0.55;
    Kt = 1.08;
    W = 25;
    L = 150;

case 'JALALVAND'
    % --- Case 4: Validation Data from Jalalvand et al. ---
    validation = true;
    ylimUnique = 0.16;
    fawOption = false;

    % LE Material (SkyFlex TR30 Carbon/Epoxy)
    EL_options = [101.7];

```

```

        SL_options = [1.962];

        % HE Material (Hexcel E-Glass/913)
        EH_options = [38.7];
        SH_options = [1.548];

        m = 29.3;
        GIIC = 1.0;
        Kt = 1.08;
        W = 20;
        L = 160;

        otherwise
            error('Invalid modelOption specified.');
```

end

```

%% --- PRE-PROCESSING & FIGURE SETUP ---

% Convert material properties from GPa to MPa for consistency in calculations.
EL_options = EL_options * 1000;
SL_options = SL_options * 1000;
EH_options = EH_options * 1000;
SH_options = SH_options * 1000;

% Define the range for the LE thickness fraction (gamma).
gamma = linspace(0.01, 0.5, 500);

% Initialize the main figure with a tiled layout for subplots.
fig = figure('Position', [100, 100, 1000, 700]);
num_lsm_options = length(EL_options);
num_hsm_options = length(EH_options);
total_subplots = num_lsm_options * num_hsm_options;
t1 = tiledlayout(fig, num_lsm_options, num_hsm_options, 'TileSpacing', 'compact', 'Padding', 'compact');
```

% Conditionally initialize the second figure for overlapping plots.

```

if showOverlapFigure
    fig_overlap = figure('Name', 'Overlapping Damage Mode Maps', 'Position', [150, 150, 900, 650]);
    ax_overlap = axes(fig_overlap);
    hold(ax_overlap, 'on');
    handles_for_overlap_legend = [];
    plotted_overlap_types = struct('LD', false, 'LH', false, 'DH', false, 'tpLimit', false);
end

% Initialize plot handles for creating shared legends.
handles_for_shared_legend = [];
p1 = []; p2 = []; p3 = []; p4 = [];
plot_idx_in_loop = 0;

%% --- MAIN CALCULATION & PLOTTING LOOP ---
% This nested loop iterates through every combination of LE and HE material properties to
% generate a damage mode map for each case.

for i_LSM = 1:num_lsm_options
    for i_HSM = 1:num_hsm_options
```



```

plot_idx_in_loop = plot_idx_in_loop + 1;

% Select current material properties for this iteration
EL = EL_options(i_LSM);
SL = SL_options(i_LSM);
EH = EH_options(i_HSM);
SH = SH_options(i_HSM);

% Set subplot title
if any(strcmpi(modelOption, {'OLIVEIRA', 'DANZI'}))
    Vf_L_current = Vf_L_options(i_LSM);
    Vf_H_current = Vf_H_options(i_HSM);
    titleUnique = sprintf('V_{f_L} = %.0f%%, V_{f_H} = %.0f%%', Vf_L_current,
        Vf_H_current);
else
    titleUnique = '';
end

% Calculate intermediate parameter
alpha = EL / EH; % Stiffness ratio

% Pre-allocate arrays for boundary curves
tL_LD = nan(size(gamma)); % Boundary for LE Fragmentation / Delamination
tL_LH = nan(size(gamma)); % Boundary for LE Fragmentation / HE Failure
tL_DH = nan(size(gamma)); % Boundary for Delamination / HE Failure
tpLimit = nan(size(gamma)); % Thin-ply thickness limit

% --- Calculate boundary locations for the damage mode map ---
for i = 1:length(gamma)
    g = gamma(i);
    beta = g / (1 - g); % Thickness ratio of LE to HE layers

    % Boundary: LE Fragmentation vs. Delamination (LD)
    tL_LD(i) = (2 * GIIC * EH * alpha * (1 - g)) / (SL^2 * (alpha * g + 1 - g));

    % Boundary: LE Fragmentation vs. HE Failure (LH)
    tL_LH(i) = (((SH / (Kt * SL)) * (alpha / (alpha * beta + 1)))^m) * (beta / (2
        * W * L));

    % Boundary: Delamination vs. HE Failure (DH)
    term_DH = (beta^(-1/m)) * (Kt / SH) * nthroot(2*L*W, m) * sqrt(2*GIIC*EH) *
        sqrt((1 + alpha*beta)/alpha);
    exponent_DH = (2 * m) / (m - 2);
    tL_DH(i) = term_DH^exponent_DH;

    % Apply thickness correction factor
    tL_LD(i) = tL_LD(i) * 2;
    tL_LH(i) = tL_LH(i) * 2;
    tL_DH(i) = tL_DH(i) * 2;

    % Calculate thin-ply limit for reference
    tpLimit(i) = 0.125 * g;
end

% Set multiplier for y-axis (thickness vs. FAW)
y_multiplier = 1;
if fawOption, y_multiplier = 1000; end

```

```

% --- Plot on the individual subplot (tiled figure) ---
ax_current = nexttile(tl);
hold(ax_current, 'on');
p1 = plot(ax_current, gamma, tL_LD * y_multiplier, 'LineWidth', 1.5, 'Color',
    [0.2, 0.5, 1], 'LineStyle', '-', 'DisplayName', 'LD Boundary');
p2 = plot(ax_current, gamma, tL_LH * y_multiplier, 'LineWidth', 2, 'Color', [1,
    0.7, 0.4], 'LineStyle', ':', 'DisplayName', 'LH Boundary');
p3 = plot(ax_current, gamma, tL_DH * y_multiplier, 'LineWidth', 1.5, 'Color', [0,
    0.8, 0.2], 'LineStyle', '--', 'DisplayName', 'DH Boundary');
if ~validation
    p4 = plot(ax_current, gamma, tpLimit * y_multiplier, 'LineWidth', 1, 'Color',
        [0 0 0], 'LineStyle', '-.', 'DisplayName', 'Thin-ply Limit');
end

% Store handles from the first subplot for the shared legend
if plot_idx_in_loop == 1
    handles_for_shared_legend = [p1, p2, p3];
    if ~validation, handles_for_shared_legend(end+1) = p4; end
end

% --- Conditionally plot on the second (overlapping) figure ---
if showOverlapFigure
    plot(ax_overlap, gamma, tL_LD*y_multiplier, 'Color', [0.2,0.5,1], '
        HandleVisibility','off');
    plot(ax_overlap, gamma, tL_LH*y_multiplier, 'Color', [1,0.7,0.4], '
        HandleVisibility','off');
    plot(ax_overlap, gamma, tL_DH*y_multiplier, 'Color', [0,0.8,0.2], '
        HandleVisibility','off');
end

% --- Format the individual subplot ---
xlabel(ax_current, 'LE material proportion, \gamma');
ylabel(ax_current, if_else(fawOption, 'Low-strain FAW [gsm]', 'LE thickness, t_L [
    mm]'));
title(ax_current, titleUnique, 'FontSize', 10);
grid(ax_current, 'on'); box(ax_current, 'on');
ylim(ax_current, [0, ylimUnique * y_multiplier]);
xlim(ax_current, [min(gamma), max(gamma)]);
hold(ax_current, 'off');
end
end

%% --- FINAL PLOT FORMATTING ---

% --- Finalize the Tiled Figure ---
title(tl, 'Damage Mode Maps', 'FontSize', 15);
if total_subplots == 1 % Case: Single plot (not tiled)
    legend(ax_current, 'Location', 'best', 'FontSize', 8);
else % Case: Tiled layout with multiple subplots
    lgd = legend(handles_for_shared_legend);
    lgd.Layout.Tile = 'South';
    lgd.Orientation = 'horizontal';
end

% --- Finalize the Overlapping Figure ---
if showOverlapFigure

```

```

% Add dummy plots with correct styles to create a clean legend
h_ld = plot(ax_overlap, NaN, NaN, 'LineWidth', 1.5, 'Color', [0.2, 0.5, 1], 'LineStyle',
    '-', 'DisplayName', 'LD Boundary');
h_lh = plot(ax_overlap, NaN, NaN, 'LineWidth', 2, 'Color', [1, 0.7, 0.4], 'LineStyle',
    ':', 'DisplayName', 'LH Boundary');
h_dh = plot(ax_overlap, NaN, NaN, 'LineWidth', 1.5, 'Color', [0, 0.8, 0.2], 'LineStyle',
    '--', 'DisplayName', 'DH Boundary');
handles_for_overlap_legend = [h_ld, h_lh, h_dh];
if ~validation
    h_tp = plot(ax_overlap, NaN, NaN, 'LineWidth', 1, 'Color', [0 0 0], 'LineStyle', '-.',
        'DisplayName', 'Thin-ply limit');
    handles_for_overlap_legend(end+1) = h_tp;
end

hold(ax_overlap, 'off');
grid(ax_overlap, 'on'); box(ax_overlap, 'on');
title(ax_overlap, ['Overlapping Damage Mode Maps for ' modelOption], 'FontSize', 14);
xlabel(ax_overlap, 'LE material proportion, \gamma');
ylabel(ax_overlap, if_else(fawOption, 'Low-strain FAW [gsm]', 'LE thickness, t_L [mm]'));
ylim(ax_overlap, [0, ylimUnique * y_multiplier]);
xlim(ax_overlap, [min(gamma), max(gamma)]);
legend(ax_overlap, handles_for_overlap_legend, 'Location', 'best', 'FontSize', 10);
end

% Helper function for conditional strings (avoids if/else blocks in plotting calls)
function str = if_else(condition, true_str, false_str)
    if condition, str = true_str; else, str = false_str; end
end

```

A.3 Tensile Properties and Pull-Out Length Calculation Script

This MATLAB script provides a comprehensive evaluation of unidirectional hybrid composite performance by calculating several design-relevant properties as a function of the LE layer thickness fraction (γ). The script is designed to be configurable, allowing the user to select from different input datasets for material properties and geometric parameters, which facilitates parametric studies or validation against benchmark cases.

The primary calculations include the predicted longitudinal tensile strength and Young's modulus of the hybrid laminate. Additionally, the script computes the fiber pull-out length as an indicator of translaminar toughness. A notable function is the re-calculation of the critical stresses for LE layer fragmentation, interlaminar delamination, and HE layer failure. This analysis complements the damage mode maps by illustrating how these failure thresholds change with the hybrid composition for a given thickness. The script then generates a series of plots to visualize the calculated properties against the LE thickness fraction, offering a detailed basis for design and material selection.

Listing A.3: HybridPerformanceAnalysis.m

```

% -----

```

```

% | SCRIPT: HybridPerformanceAnalysis.m
% | AUTHOR: Raul Maciel Oliveira
% | DATE: September 2025
% |-----|
% | CONTEXT: This script was developed for the Master's Thesis:
% | "Hybrid composite laminates for a new generation of aerospace
% | structures"
% |
% | Master's Degree in Mechanical Engineering
% | Specialization in Aeronautical Structures and Vehicles
% | FACULDADE DE ENGENHARIA DA UNIVERSIDADE DO PORTO (FEUP)
% |-----|

%% --- INITIALIZATION ---
% Clear workspace, command window, and close any existing figures.
clear;
clc;
close all;

%% --- SCRIPT CONFIGURATION ---
% Select the dataset to use for the calculations.
% 'OLIVEIRA': Main dataset for this thesis.
% 'BROGGI': Validation dataset from literature.
modelOption = 'OLIVEIRA';

% Select which plots to generate.
% 'STRESS': Plots critical failure stresses vs. gamma.
% 'THICKNESS': Plots tensile strength and modulus vs. gamma.
% 'PULL-OUT': Plots pull-out length vs. gamma.
% 'ALL': Generates all of the above plots.
plotOption = 'ALL';

%% --- INPUT DATA & PARAMETERS ---
% This section defines the material properties and geometric parameters based on the
% selected 'modelOption'.

switch modelOption
case 'OLIVEIRA'
    % --- Case 1: Main Thesis Data ---
    epsilon_L = 1.1 / 100; % Failure strain of the Low-Elongation (LE) material

    % LE Ply Properties (HR40) for different Vf
    EL_options = [232.9, 211.4, 191.0]; % Young's Modulus [GPa]
    SL_options = [2.58, 2.34, 2.11]; % Longitudinal Strength [GPa]
    Vf_L_options = [32, 37, 42]; % Fiber Volume Fraction [%]

    % HE Ply Properties (T800) for different Vf
    EH_options = [174.3, 158.6, 143.2]; % Young's Modulus [GPa]
    SH_options = [3.46, 3.14, 2.83]; % Longitudinal Strength [GPa]
    Vf_H_options = [32, 37, 42]; % Fiber Volume Fraction [%]

    % General Model Parameters
    m = 30; % Weibull Modulus
    GIIC = 0.55; % Mode II Critical Energy Release Rate [N/mm or MPa*mm]
    Kt = 1.08; % Stress Concentration Factor
    W = 15; % Laminate Width [mm]
    L = 81; % Laminate Length [mm]

```

```

        tau = 100;      % Interfacial Shear Strength [MPa]

    case 'BROGGI'
        % --- Case 2: Validation Data from Broggi et al. ---
        epsilon_L = 1.1 / 100; % Failure strain of the LE material

        % LE Ply Properties (HR40)
        EL_options = [375.0 * 0.5]; % Young's Modulus [GPa]
        SL_options = [4.41 * 0.5]; % Longitudinal Strength [GPa]

        % HE Ply Properties (T300-700)
        EH_options = [294 * 0.5]; % Young's Modulus [GPa]
        SH_options = [4.8300 * 0.5]; % Longitudinal Strength [GPa]

        % General Model Parameters
        m = 41; % Weibull Modulus
        GIIC = 0.55; % Mode II Critical Energy Release Rate [N/mm or MPa*mm]
        Kt = 1.08; % Stress Concentration Factor
        W = 25; % Laminate Width [mm]
        L = 150; % Laminate Length [mm]
        tau = 73.6; % Interfacial Shear Strength [MPa]

    otherwise
        error('Invalid modelOption specified. Choose ''OLIVEIRA'' or ''BROGGI''.');
    end

end

%% --- PRE-PROCESSING & SETUP ---

% Convert material properties from GPa to MPa for consistency in calculations.
EL_options = EL_options * 1000;
SL_options = SL_options * 1000;
EH_options = EH_options * 1000;
SH_options = SH_options * 1000;

% Determine the number of material combinations to create subplots.
num_lsm_options = length(EL_options);
num_hsm_options = length(EH_options);
total_subplots = num_lsm_options * num_hsm_options;

% Initialize figures based on the selected plotOption.
if any(strcmpi(plotOption, {'THICKNESS', 'ALL'}))
    fig_tensile = figure('Position', [100, 100, 1000, 700]);
    tl_tensile = tiledlayout(fig_tensile, num_lsm_options, num_hsm_options, 'TileSpacing',
        'compact', 'Padding', 'compact');
end

if any(strcmpi(plotOption, {'STRESS', 'ALL'}))
    fig_stress = figure('Position', [150, 150, 1000, 700]);
    tl_stress = tiledlayout(fig_stress, num_lsm_options, num_hsm_options, 'TileSpacing',
        'compact', 'Padding', 'compact');
end

% Initialize plot handles for creating shared legends.
p1 = []; p2 = []; p3 = []; p4 = []; p5 = [];
handles_tensile_legend = [];
handles_stress_legend = [];
plot_idx_in_loop = 0;

```

```

% Define laminate thickness parameters for the analysis.
tH = 0.1; % Constant thickness of the High-Elongation (HE) layer [mm]
tTotal = linspace(0.1, 0.2, 500); % Range of total laminate thicknesses to evaluate [mm]
pullOut = nan(size(tTotal)); % Pre-allocate array for pull-out length

%% --- MAIN CALCULATION & PLOTTING LOOP ---
% This nested loop iterates through every combination of LE and HE material properties
% defined in the input section.

for i_LSM = 1:num_lsm_options
    for i_HSM = 1:num_hsm_options
        plot_idx_in_loop = plot_idx_in_loop + 1;

        % --- Set subplot titles based on the model ---
        if strcmp(modelOption, 'OLIVEIRA')
            Vf_L_current = Vf_L_options(i_LSM);
            Vf_H_current = Vf_H_options(i_HSM);
            titleUnique = sprintf('V_{f_L} = %.0f%%, V_{f_H} = %.0f%%', Vf_L_current,
                                   Vf_H_current);

            % Set a specific title for the central pull-out plot
            if i_HSM == round(num_hsm_options/2) && i_LSM == round(num_lsm_options/2)
                titlePull = sprintf('Pull-out Length (V_{f_L}=%.0f%%, V_{f_H}=%.0f%%)',
                                     Vf_L_current, Vf_H_current);
            end
        else
            titleUnique = 'Tensile Properties';
            titlePull = 'Pull-out Length';
        end

        % --- Select current material properties for this iteration ---
        EL = EL_options(i_LSM); % LE Modulus [MPa]
        SL = SL_options(i_LSM); % LE Strength [MPa]
        EH = EH_options(i_HSM); % HE Modulus [MPa]
        SH = SH_options(i_HSM); % HE Strength [MPa]

        % --- Calculate intermediate parameters ---
        alpha = EL / EH; % Stiffness ratio
        zeta = (SH - epsilon_L * EH) / (SH + SL - epsilon_L * EH); % Transition thickness
        ratio

        % --- Pre-allocate arrays for this iteration's results ---
        gamma = nan(size(tTotal)); % LE layer thickness fraction
        E_l = nan(size(tTotal)); % Hybrid laminate modulus
        sigma_l = nan(size(tTotal)); % Hybrid laminate strength
        sigma_LF = nan(size(tTotal)); % Critical stress for LE fragmentation
        sigma_DEL = nan(size(tTotal)); % Critical stress for delamination
        sigma_HF = nan(size(tTotal)); % Critical stress for HE failure

        % --- Iterate over thickness range to calculate properties vs. gamma ---
        for i = 1:length(tTotal)
            tL = tTotal(i) - tH; % Thickness of the LE layer [mm]
            g = tL / tTotal(i); % Current LE thickness fraction, gamma

            % Stop if gamma exceeds a practical limit (e.g., 0.5)
            if g > 0.5, break; end
        end
    end
end

```

```

gamma(i) = g; % Store gamma value
beta = tL / tH; % Thickness ratio of LE to HE layers

% Calculate Laminate Modulus (Voigt model)
E_l(i) = (EL * g + (1 - g) * EH) / 1000; % Convert back to GPa for plotting

% Calculate Laminate Tensile Strength
if g < zeta
    sigma_l(i) = SH * (1 - g);
else
    sigma_l(i) = SL * g + epsilon_L * EH * (1 - g);
end

% Calculate Pull-out Length for the central material combination
if i_HSM == round(num_hsm_options/2) && i_LSM == round(num_lsm_options/2)
    pullOut(i) = (SL * tL) / tau;
end

% Calculate Critical Stresses for Different Failure Modes
sigma_LF(i) = SL * ((alpha * beta) + 1) / (alpha * (beta + 1));
sigma_DEL(i) = (1 / (1 + beta)) * sqrt(((1 + alpha * beta) * 4 * GIIC * EH) /
    (alpha * beta * tH));
sigma_HF(i) = (1 / (1 + beta)) * (SH / (Kt * nthroot(L * W * tH, m)));
end

% --- Plotting Section ---

% Plot 1: Tensile Properties (Strength and Modulus)
if any(strcmpi(plotOption, {'THICKNESS', 'ALL'}))
    ax_current = nexttile(tl_tensile);
    hold(ax_current, 'on');

    % Left Y-axis for Stress
    yyaxis(ax_current, 'left');
    p1 = plot(ax_current, gamma, sigma_l, 'LineWidth', 1.5, 'DisplayName', 'Strength', 'Color', [0.2, 0.5, 1], 'LineStyle', '-');
    ylabel(ax_current, 'Strength [MPa]');
    ylim(ax_current, [0, 4000]);

    % Right Y-axis for Modulus
    yyaxis(ax_current, 'right');
    p2 = plot(ax_current, gamma, E_l, 'LineWidth', 1.5, 'DisplayName', 'Modulus', 'Color', [1, 0.7, 0.4], 'LineStyle', '--');
    ylabel(ax_current, 'Modulus [GPa]');
    ylim(ax_current, [0, 350]);

    % General formatting
    xlabel(ax_current, 'LE material proportion, \gamma');
    title(ax_current, titleUnique, 'FontSize', 10);
    grid(ax_current, 'on');
    box(ax_current, 'on');
    ax_current.YAxis(1).Color = 'k'; ax_current.YAxis(2).Color = 'k';
end

% Plot 2: Critical Failure Stresses
if any(strcmpi(plotOption, {'STRESS', 'ALL'}))
    ax_current2 = nexttile(tl_stress);

```



```

        hold(ax_current2, 'on');

        p3 = plot(ax_current2, gamma, sigma_LF, 'LineWidth', 1.5, 'DisplayName', 'LE
            Fragmentation', 'Color', [0.2, 0.5, 1], 'LineStyle', '-');
        p4 = plot(ax_current2, gamma, sigma_DEL, 'LineWidth', 1.5, 'DisplayName', '
            Delamination', 'Color', [1, 0.7, 0.4], 'LineStyle', ':');
        p5 = plot(ax_current2, gamma, sigma_HF, 'LineWidth', 1.5, 'DisplayName', 'HE
            Failure', 'Color', [0, 0.8, 0.2], 'LineStyle', '--');

        % Formatting
        ylabel(ax_current2, 'Critical Stress [MPa]');
        ylim(ax_current2, [0, 4000]);
        xlabel(ax_current2, 'LE material proportion, \gamma');
        title(ax_current2, titleUnique, 'FontSize', 10);
        grid(ax_current2, 'on');
        box(ax_current2, 'on');
    end

    % Store handles from the first subplot for the shared legend
    if plot_idx_in_loop == 1
        handles_tensile_legend = [p1, p2];
        handles_stress_legend = [p3, p4, p5];
    end
end

end

%% --- FINAL PLOT FORMATTING ---

% Add legends and main titles to the figures.
if total_subplots == 1
    % Case: Single plot (not tiled)
    if any(strcmpi(plotOption, {'THICKNESS', 'ALL'}))
        legend(ax_current, 'Location', 'best', 'FontSize', 8);
    end
    if any(strcmpi(plotOption, {'STRESS', 'ALL'}))
        legend(ax_current2, 'Location', 'best', 'FontSize', 8);
    end
else
    % Case: Tiled layout with multiple subplots
    if any(strcmpi(plotOption, {'THICKNESS', 'ALL'}))
        title(tl_tensile, 'Hybrid Laminate Tensile Properties', 'FontSize', 15);
        lgd = legend(handles_tensile_legend);
        lgd.Layout.Tile = 'South';
        lgd.Orientation = 'horizontal';
    end
    if any(strcmpi(plotOption, {'STRESS', 'ALL'}))
        title(tl_stress, 'Critical Stresses for Failure Initiation', 'FontSize', 15);
        lgd2 = legend(handles_stress_legend);
        lgd2.Layout.Tile = 'South';
        lgd2.Orientation = 'horizontal';
    end
end

end

% Plot 3: Pull-out Length
if any(strcmpi(plotOption, {'PULL-OUT', 'ALL'}))
    fig_pullout = figure('Name', 'Pull-out Length Analysis', 'Position', [200, 200, 700,
        500]);
end

```

```

ax_pullout = axes(fig_pullout);
hold(ax_pullout, 'on');

plot(ax_pullout, gamma, pullOut, 'LineWidth', 1.5, 'Color', [0.2, 0.5, 1]);

% Add a vertical line at a specific gamma value for reference
gamma_ref = 0.33;
xline(ax_pullout, gamma_ref, 'LineWidth', 1.5, 'Color', [1, 0.7, 0.4], ...
      'LineStyle', '--', 'DisplayName', sprintf('\gamma = %.2f', gamma_ref));

% Formatting
xlabel(ax_pullout, 'LE material proportion, \gamma');
ylabel(ax_pullout, 'Pull-out length, l_c [mm]');
title(ax_pullout, titlePull, 'FontSize', 12);
xlim(ax_pullout, [min(gamma), max(gamma)]);
ylim(ax_pullout, [0, 10]);
legend(ax_pullout, 'show', 'Location', 'northeast');
grid(ax_pullout, 'on');
box(ax_pullout, 'on');
hold(ax_pullout, 'off');
end

```

A.4 Residual Stress Calculation Script

This MATLAB script implements Classical Laminated Plate Theory (CLPT) to calculate the residual thermal stresses within a user-defined composite laminate. The process involves two main stages. First, the script determines the effective thermo-mechanical properties for each unique ply material by applying micromechanics models, such as Halpin-Tsai for transverse and shear moduli and Schapery equations for coefficients of thermal expansion, based on the input properties of the constituent fibers and matrix. Second, it assembles the laminate's [A], [B], and [D] stiffness matrices and solves for the mid-plane strains and curvatures induced by a specified temperature change. The final output provides the resulting residual stress state (σ_1 , σ_2 , τ_{12}) in the local coordinate system of each individual ply.

Listing A.4: ResidualStressCLPT.m

```

% -----
% | SCRIPT: ResidualStressCLPT.m                               |
% | AUTHOR: Raul Maciel Oliveira                               |
% | DATE:   September 2025                                     |
% |-----|
% | CONTEXT: This script was developed for the Master's Thesis: |
% | "Hybrid composite laminates for a new generation of aerospace |
% | structures"                                                  |
% |                                                              |
% | Master's Degree in Mechanical Engineering                   |
% | Specialization in Aeronautical Structures and Vehicles     |
% | FACULDADE DE ENGENHARIA DA UNIVERSIDADE DO PORTO (FEUP)    |
% |-----|
%% --- INITIALIZATION ---

```

```

% Clear workspace and command window.
clear;
clc;

%% --- INPUTS ---
% This section defines all user-configurable parameters for the analysis.

% --- Constituent Material Properties ---
% Fiber Type 1: HR40
fiber1.name = 'HR40';
fiber1.EfL = 395;           % Longitudinal Young's Modulus [GPa]
fiber1.EfT = 26.8;         % Transverse Young's Modulus (Assumed) [GPa]
fiber1.GfLT = 4.6;         % In-plane Shear Modulus (Assumed) [GPa]
fiber1.nuLT = 0.23;        % Major Poisson's Ratio (Assumed)
fiber1.alphaLfL = -0.393e-6; % Longitudinal CTE [°C]
fiber1.alphaLfT = 4.5e-6;   % Transverse CTE (Assumed) [°C]
fiber1.Vf = 0.531;         % Fiber Volume Fraction

% Fiber Type 2: T800S
fiber2.name = 'T800';
fiber2.EfL = 294;          % Longitudinal Young's Modulus [GPa]
fiber2.EfT = 14;           % Transverse Young's Modulus [GPa]
fiber2.GfLT = 8.7;         % In-plane Shear Modulus [GPa]
fiber2.nuLT = 0.226;       % Major Poisson's Ratio
fiber2.alphaLfL = -0.827e-6; % Longitudinal CTE [°C]
fiber2.alphaLfT = 7e-6;    % Transverse CTE [°C]
fiber2.Vf = 0.534;         % Fiber Volume Fraction

% Matrix Properties: Epoxy Resin (Common to both lamina types)
matrix.name = 'Epoxy Resin';
matrix.Em = 3.44;          % Young's Modulus [GPa]
matrix.num = 0.35;         % Poisson's Ratio
matrix.alpham = 41.5e-6;   % CTE [°C]

% --- Laminate Configuration ---
% Define the stacking sequence by mapping each ply to a material type.
% Use '1' for fiber1 (HR40) and '2' for fiber2 (T800).
ply_material_type_indices = [2, 1, 2, 1];
% Ply orientations in degrees, from top to bottom.
orientations_deg = [0, 0, 90, 90];
% Thickness of each corresponding ply [mm].
ply_thickness_mm = [0.060, 0.060, 0.060, 0.060];
% Temperature change from stress-free (cure) to ambient temperature [°C].
delta_T = -110;

%% --- CALCULATE EFFECTIVE LAMINA PROPERTIES (MICROMECHANICS) ---
% This section calculates the effective thermo-mechanical properties for each unique
% lamina type defined in the inputs.

defined_fibers = {fiber1, fiber2};
defined_laminae = cell(1, length(defined_fibers));
Gm = matrix.Em / (2 * (1 + matrix.num)); % Matrix shear modulus [GPa]

fprintf('--- Calculating Effective Lamina Properties ---\n');

for i = 1:length(defined_fibers)
    % Get constituent properties for the current lamina type

```

```

current_fiber = defined_fibers{i};
Vf = current_fiber.Vf;
Vm = 1 - Vf;

% --- Calculate Lamina Elastic Properties ---
% E1 (Longitudinal Modulus) using Rule of Mixtures
lamina_props.E1 = current_fiber.EfL * Vf + matrix.Em * Vm;

% E2 (Transverse Modulus) using Halpin-Tsai
xi2 = 2; % Halpin-Tsai empirical factor for E2
eta2 = ((current_fiber.EfT / matrix.Em) - 1) / ((current_fiber.EfT / matrix.Em) + xi2);
;
lamina_props.E2 = matrix.Em * (1 + xi2 * eta2 * Vf) / (1 - eta2 * Vf);

% G12 (In-plane Shear Modulus) using Halpin-Tsai
xiG = 1; % Halpin-Tsai empirical factor for G12
etaG = ((current_fiber.GfLT / Gm) - 1) / ((current_fiber.GfLT / Gm) + xiG);
lamina_props.G12 = Gm * (1 + xiG * etaG * Vf) / (1 - etaG * Vf);

% nu12 (Major Poisson's Ratio) using Rule of Mixtures
lamina_props.nu12 = current_fiber.nuLT * Vf + matrix.num * Vm;

% --- Calculate Lamina CTEs using Schapery Equations ---
% alpha1 (Longitudinal CTE)
num_a1 = (current_fiber.EfL * Vf * current_fiber.alphaLT) + (matrix.Em * Vm * matrix.
    alpham);
lamina_props.alpha1 = num_a1 / lamina_props.E1;

% alpha2 (Transverse CTE)
lamina_props.alpha2 = (1 + current_fiber.nuLT) * current_fiber.alphaLT * Vf + ...
    (1 + matrix.num) * matrix.alpham * Vm - ...
    lamina_props.nu12 * lamina_props.alpha1;

% --- Calculate Local Stiffness Matrix [Q] ---
lamina_props.nu21 = lamina_props.nu12 * lamina_props.E2 / lamina_props.E1;
den_Q = 1 - lamina_props.nu12 * lamina_props.nu21;
Q11 = lamina_props.E1 / den_Q;
Q12 = lamina_props.nu12 * lamina_props.E2 / den_Q;
Q22 = lamina_props.E2 / den_Q;
Q66 = lamina_props.G12;
lamina_props.Q_local = [Q11, Q12, 0; Q12, Q22, 0; 0, 0, Q66];

% --- Store and Display Properties ---
lamina_props.name = sprintf('Lamina with %s', current_fiber.name);
defined_laminae{i} = lamina_props;

fprintf('\nLamina Material Type %d: %s (Vf=%.3f)\n', i, lamina_props.name, Vf);
fprintf(' E1=%.2f GPa, E2=%.2f GPa, G12=%.2f GPa, nu12=%.3f\n', ...
    lamina_props.E1, lamina_props.E2, lamina_props.G12, lamina_props.nu12);
fprintf(' alpha1=%.2e /°C, alpha2=%.2e /°C\n', lamina_props.alpha1, lamina_props.
    alpha2);
end

%% --- LAMINATE GEOMETRY SETUP (CLPT) ---
% Define the z-coordinates for the top and bottom of each ply.
% The origin (z=0) is at the laminate's mid-plane.

```

```

num_plyes = length(orientations_deg);
ply_thickness_m = ply_thickness_mm * 1e-3; % Convert mm to m
total_thickness = sum(ply_thickness_m);

% Calculate z-coordinate for each interface, starting from the top surface.
z_coords = zeros(num_plyes + 1, 1);
z_coords(1) = total_thickness / 2; % Top surface of the laminate
for k = 1:num_plyes
    z_coords(k + 1) = z_coords(k) - ply_thickness_m(k);
end

% Calculate z-coordinate for the middle of each ply.
z_mid_ply = (z_coords(1:end-1) + z_coords(2:end)) / 2;

%% --- CALCULATE [A,B,D] MATRICES & THERMAL LOADS ---
% Assemble the laminate stiffness matrices (A, B, D) and the thermal force (N_thermal) and
    moment (M_thermal) resultants.

% Initialize matrices
A_matrix = zeros(3, 3); B_matrix = zeros(3, 3); D_matrix = zeros(3, 3);
N_thermal = zeros(3, 1); M_thermal = zeros(3, 1);
Q_bar_plyes = cell(1, num_plyes); % Store each ply's transformed stiffness matrix

for k = 1:num_plyes
    % Get properties for the current ply's material type
    current_material_idx = ply_material_type_indices(k);
    active_lamina = defined_laminae(current_material_idx);
    Q_local_ply = active_lamina.Q_local * 1e3; % Convert GPa to MPa for stress
        calculations

    % Calculate the transformed reduced stiffness matrix [Q_bar] for the ply
    theta_rad = deg2rad(orientations_deg(k));
    c = cos(theta_rad); s = sin(theta_rad);
    c2 = c^2; s2 = s^2; c4 = c^4; s4 = s^4;

    Q11 = Q_local_ply(1,1); Q12 = Q_local_ply(1,2);
    Q22 = Q_local_ply(2,2); Q66 = Q_local_ply(3,3);

    Q_bar = zeros(3,3);
    Q_bar(1,1) = Q11*c4 + Q22*s4 + 2*(Q12 + 2*Q66)*s2*c2;
    Q_bar(1,2) = (Q11 + Q22 - 4*Q66)*s2*c2 + Q12*(c4 + s4);
    Q_bar(2,1) = Q_bar(1,2);
    Q_bar(2,2) = Q11*s4 + Q22*c4 + 2*(Q12 + 2*Q66)*s2*c2;
    Q_bar(1,3) = (Q11 - Q12 - 2*Q66)*c*s*c2 - (Q22 - Q12 - 2*Q66)*c*s*s2;
    Q_bar(3,1) = Q_bar(1,3);
    Q_bar(2,3) = (Q11 - Q12 - 2*Q66)*c*s*s2 - (Q22 - Q12 - 2*Q66)*c*s*c2;
    Q_bar(3,2) = Q_bar(2,3);
    Q_bar(3,3) = (Q11 + Q22 - 2*Q12 - 2*Q66)*s2*c2 + Q66*(s4 + c4);
    Q_bar_plyes{k} = Q_bar;

    % Integrate through thickness to build A, B, and D matrices
    h_k = z_coords(k) - z_coords(k+1);
    A_matrix = A_matrix + Q_bar * h_k;
    B_matrix = B_matrix + (1/2) * Q_bar * (z_coords(k)^2 - z_coords(k+1)^2);
    D_matrix = D_matrix + (1/3) * Q_bar * (z_coords(k)^3 - z_coords(k+1)^3);

    % Calculate ply's thermal strain vector in global coordinates

```

```

alpha1 = active_lamina.alpha1; alpha2 = active_lamina.alpha2;
alpha_global = [alpha1*c2 + alpha2*s2;
                alpha1*s2 + alpha2*c2;
                2*(alpha1 - alpha2)*c*s];
eps_thermal_global_ply = alpha_global * delta_T;

% Integrate to find thermal resultants
N_thermal = N_thermal + Q_bar * eps_thermal_global_ply * h_k;
M_thermal = M_thermal + (1/2) * Q_bar * eps_thermal_global_ply * (z_coords(k)^2 -
    z_coords(k+1)^2);
end

%% --- SOLVE FOR MID-PLANE STRAINS & CURVATURES ---
% With no external loads, the mechanical strains and curvatures are induced
% solely by the thermal loads.

ABD_matrix = [A_matrix, B_matrix; B_matrix, D_matrix];
NM_thermal = [N_thermal; M_thermal];

% Solve the system [eps0; kappa] = [ABD]^-1 * [N_thermal; M_thermal]
if rcond(ABD_matrix) < eps
    warning('ABD matrix is singular or badly conditioned. Using pseudoinverse.');
```

$$\begin{bmatrix} \text{eps0_kappa} \\ \text{NM_thermal} \end{bmatrix} = \text{pinv}(\text{ABD_matrix}) * \text{NM_thermal};$$

```

else
    eps0_kappa = ABD_matrix \ NM_thermal;
end

eps0 = eps0_kappa(1:3); % Mid-plane strains
kappa = eps0_kappa(4:6); % Laminate curvatures [1/m]

%% --- CALCULATE PLY-BY-PLY RESIDUAL STRESSES ---
% Calculate the stress state in each ply in its local (1-2) coordinate system.

fprintf('\n\n--- Residual Stresses in Plies (Local Coordinates, in MPa) ---\n');

for k = 1:num_plies
    % Get relevant properties for the current ply
    theta_rad = deg2rad(orientations_deg(k));
    current_material_idx = ply_material_type_indices(k);
    active_lamina = defined_laminae(current_material_idx);
    Q_bar_k_current = Q_bar_plies{k};

    % Define the stress transformation matrix from global (x-y) to local (1-2)
    c = cos(theta_rad); s = sin(theta_rad); c2 = c^2; s2 = s^2;
    T_stress = [ c2,  s2,  2*c*s;
                s2,  c2, -2*c*s;
                -c*s, c*s,  c2-s2];

    % Recalculate global thermal strain for this ply
    alpha1 = active_lamina.alpha1; alpha2 = active_lamina.alpha2;
    alpha_global = [alpha1*c2 + alpha2*s2; alpha1*s2 + alpha2*c2; 2*(alpha1-alpha2)*c*s];
    eps_thermal_global = alpha_global * delta_T;

    % Calculate total strain at top, middle, and bottom of the ply
    eps_total_top = eps0 + z_coords(k) * kappa;
    eps_total_mid = eps0 + z_mid_ply(k) * kappa;
    eps_total_bot = eps0 + z_coords(k+1) * kappa;

```

```

% Calculate mechanical strain (total - thermal)
eps_mech_top = eps_total_top - eps_thermal_global;
eps_mech_mid = eps_total_mid - eps_thermal_global;
eps_mech_bot = eps_total_bot - eps_thermal_global;

% Calculate stress in global coordinates (sigma_global = Q_bar * eps_mech)
sigma_global_top = Q_bar_k_current * eps_mech_top;
sigma_global_mid = Q_bar_k_current * eps_mech_mid;
sigma_global_bot = Q_bar_k_current * eps_mech_bot;

% Transform stress to local ply coordinates
sigma_local_top = T_stress * sigma_global_top;
sigma_local_mid = T_stress * sigma_global_mid;
sigma_local_bot = T_stress * sigma_global_bot;

% Display results for the current ply
fprintf('\nPly %d (Angle: %.0f deg, Material: %s):\n', k, orientations_deg(k),
        active_lamina.name);
fprintf(' Top      (z= %.4f m): Sigma_1=%8.2f, Sigma_2=%8.2f, Tau_12=%8.2f\n', ...
        z_coords(k), sigma_local_top(1), sigma_local_top(2), sigma_local_top(3));
fprintf(' Middle (z= %.4f m): Sigma_1=%8.2f, Sigma_2=%8.2f, Tau_12=%8.2f\n', ...
        z_mid_ply(k), sigma_local_mid(1), sigma_local_mid(2), sigma_local_mid(3));
fprintf(' Bottom (z= %.4f m): Sigma_1=%8.2f, Sigma_2=%8.2f, Tau_12=%8.2f\n', ...
        z_coords(k+1), sigma_local_bot(1), sigma_local_bot(2), sigma_local_bot(3));
end

```

A.5 R-Curve Calculation Script

This MATLAB script implements the data reduction methodology for Double-Edge Notched Tension (DENT) tests to determine the crack resistance curve ($\mathcal{R}$ -curve), following the procedure detailed by Catalanotti et al. [38]. The script takes laminate elastic properties, layup details, and pre-fitted size-effect law parameters as inputs for several material configurations. It calculates the laminate's equivalent modulus ($\hat{E}$) and uses a polynomial expansion to determine the orthotropic stress intensity factor correction, κ . By applying the experimentally determined size-effect law, it generates a family of crack driving force curves (G_I) for a range of virtual specimen widths, and the laminate's $\mathcal{R}$ -curve is determined as their lower envelope. Finally, it rescales the laminate toughness to calculate the intrinsic toughness of the 0-degree plies.

Listing A.5: RCurve.m

```

% -----
% | SCRIPT: RCurve.m |
% | AUTHOR: Raul Maciel Oliveira |
% | DATE: September 2025 |
% |-----|
% | CONTEXT: This script was developed for the Master's Thesis: |
% | "Hybrid composite laminates for a new generation of aerospace |
% | structures" |
% | |
% | Master's Degree in Mechanical Engineering |

```



```

% | Specialization in Aeronautical Structures and Vehicles |
% | FACULDADE DE ENGENHARIA DA UNIVERSIDADE DO PORTO (FEUP) |
% -----

%% --- INITIALIZATION ---
% Clear workspace, command window, and close any existing figures.
clear;
clc;
close all;

%% --- SCRIPT CONFIGURATION ---
% Switch to enable or disable the generation of plots.
enable_plotting = true;

% Define custom colors for plots for consistency.
color_orange = [1, 0.7, 0.4]; % For the R-curve envelope
color_blue = [0.2, 0.5, 1]; % For the individual crack driving force curves

%% --- INPUT DATA: MATERIAL PROPERTIES & FIT PARAMETERS ---
% This section defines all material properties and the pre-fitted size-effect
% law parameters for each laminate configuration to be analyzed. The data is
% stored in a structured array named 'materials'.

% --- Define effective cured ply/sublaminate thicknesses [mm] ---
t_T800_ply = 0.054;
t_HR40_ply = 0.024;
t_THT_sub = 0.130;
t_2TH2T_sub = 0.241;
t_3T2H3T_sub = 0.374;

% --- Material 1: T800S Laminate ---
materials(1).name = 'T800S';
materials(1).Ex = 76695; % Laminate modulus in x-dir [MPa]
materials(1).Ey = 80308; % Laminate modulus in y-dir [MPa]
materials(1).Gxy = 4732; % In-plane shear modulus [MPa]
materials(1).vxy = 0.030; % Major Poisson's ratio
materials(1).fit_type = 'Bilogarithmic'; % Size-effect law type
materials(1).M = 1104; % Size-effect parameter M
materials(1).N = 7.12; % Size-effect parameter N
materials(1).n_0 = 19; % Number of 0-degree plies
materials(1).n_90 = 20; % Number of 90-degree plies
materials(1).t_ply_0 = t_T800_ply; % Thickness of a 0-deg ply [mm]
materials(1).t_ply_90 = t_T800_ply; % Thickness of a 90-deg ply [mm]

% --- Material 2: HR40 Laminate ---
materials(2).name = 'HR40';
materials(2).Ex = 103403;
materials(2).Ey = 105382;
materials(2).Gxy = 4005;
materials(2).vxy = 0.016;
materials(2).fit_type = 'Bilogarithmic';
materials(2).M = 415.4;
materials(2).N = 0.047;
materials(2).n_0 = 49;
materials(2).n_90 = 50;
materials(2).t_ply_0 = t_HR40_ply;
materials(2).t_ply_90 = t_HR40_ply;

```

```

% --- Material 3: [T/H/T] Hybrid Laminate ---
materials(3).name = '[T/H/T]';
materials(3).Ex = 77933;
materials(3).Ey = 85677;
materials(3).Gxy = 4420;
materials(3).vxy = 0.029;
materials(3).fit_type = 'Bilogarithmic';
materials(3).M = 1232;
materials(3).N = 8.54;
materials(3).n_0 = 9;
materials(3).n_90 = 10;
materials(3).t_ply_0 = t_THT_sub;
materials(3).t_ply_90 = t_THT_sub;

% --- Material 4: [2T/H/2T] Hybrid Laminate ---
materials(4).name = '[2T/H/2T]';
materials(4).Ex = 73926;
materials(4).Ey = 87102;
materials(4).Gxy = 4732;
materials(4).vxy = 0.028;
materials(4).fit_type = 'Bilogarithmic';
materials(4).M = 1471;
materials(4).N = 7.64;
materials(4).n_0 = 5;
materials(4).n_90 = 6;
materials(4).t_ply_0 = t_2TH2T_sub;
materials(4).t_ply_90 = t_2TH2T_sub;

% --- Material 5: [3T/2H/3T] Hybrid Laminate ---
materials(5).name = '[3T/2H/3T]';
materials(5).Ex = 70668;
materials(5).Ey = 91542;
materials(5).Gxy = 4732;
materials(5).vxy = 0.026;
materials(5).fit_type = 'Linear II'; % Different size-effect law type
materials(5).A_prime = 4.40e-6; % Size-effect parameter A'
materials(5).C_prime = 2.47e-7; % Size-effect parameter C'
materials(5).n_0 = 3;
materials(5).n_90 = 4;
materials(5).t_ply_0 = t_3T2H3T_sub;
materials(5).t_ply_90 = t_3T2H3T_sub;

%% --- INPUT DATA: SIF MODEL & VIRTUAL GEOMETRIES ---

% Polynomial coefficients for the orthotropic stress intensity factor (SIF) correction
function, phi(alpha, rho). These are pre-calculated.
Phi_matrix = [
    1.7482487564, -0.053754159533, 0.0040142704949, -9.8480085881e-05;
    -0.76896688866, -0.0068632911438, 0.0029984681658, -0.00010108691939;
    0.85633404777, 0.23922363475, -0.023289123198, 0.00062358861997;
    -0.67470597429, -0.25334178248, 0.022297779266, -0.00056784694513;
    0.18495379886, 0.084067007027, -0.0068989066533, 0.00016783852495
];
phi_poly = @(alpha, rho) sum(sum(Phi_matrix .* ([1; alpha; alpha^2; alpha^3; alpha^4] *
    [1, rho, rho^2, rho^3])));
phi = @(alpha, rho) sqrt(tan(pi.*alpha/2)) .* phi_poly(alpha, rho);

```

```

% Define a range of virtual specimen widths (2w) to generate the R-curve.
num_curves = 100;
specimen_2w = linspace(sqrt(1), sqrt(1500), num_curves).^2; % Non-linear distribution [mm]
w_vector = specimen_2w / 2; % Half-width [mm]

% Define other geometric parameters.
specimen_2l = 250; % Specimen length [mm]
l = specimen_2l / 2; % Specimen half-length [mm]
alpha0 = 0.6; % Initial crack length to width ratio (a0/w)
a0_vector = alpha0 * w_vector; % Vector of initial crack half-lengths [mm]
delta_a_vector = linspace(0, 8, 500); % Vector of crack extensions [mm]

%% --- PLOTTING SETUP ---
if enable_plotting
    plot_titles = {'T800S', 'HR40', '[T/H/T]', '[2T/H/2T]', '[3T/2H/3T]'};
end

%% --- MAIN PROCESSING LOOP ---
% This loop iterates through each material configuration defined in the 'materials' struct
% , calculates its R-curve, and plots the results.

for i = 1:length(materials)
    fprintf('\n--- Processing Material: %s ---\n', materials(i).name);

    % --- Step 1: Calculate equivalent laminate properties ---
    s11 = 1 / materials(i).Ex;
    s22 = 1 / materials(i).Ey;
    s12 = -materials(i).vxy / materials(i).Ex;
    s66 = 1 / materials(i).Gxy;

    lambda = s11 / s22;
    rho = (2*s12 + s66) / (2*sqrt(s11*s22));
    E_hat = ((1+rho)/2)^(-1/2) * (lambda)^(-1/4) / sqrt(s11*s22); % Equivalent modulus [MPa]

    t0 = materials(i).n_0 * materials(i).t_ply_0; % Total thickness of 0-deg plies [mm]
    t_total = t0 + materials(i).n_90 * materials(i).t_ply_90; % Total laminate thickness [mm]

    % --- Step 2: Apply the size-effect law ---
    phi0 = phi(alpha0, rho); % SIF correction factor at initial crack length
    phi0_prime = (phi(alpha0 + 1e-6, rho) - phi(alpha0 - 1e-6, rho)) / (2e-6); % Derivative

    % Select the correct size-effect law based on 'fit_type'
    if strcmp(materials(i).fit_type, 'Bilogarithmic')
        M = materials(i).M; N = materials(i).N;
        sigma_u_func = @(w) M ./ sqrt(N + w); % Strength prediction function
        lfpz = N * (phi0 / (phi0_prime * 2)); % Fracture process zone length [mm]
        Rss_laminate = (M^2 * phi0^2) / E_hat; % Steady-state laminate toughness [kJ/m^2]

    elseif strcmp(materials(i).fit_type, 'Linear II')
        A_p = materials(i).A_prime; C_p = materials(i).C_prime;
        sigma_u_func = @(w) sqrt(1 ./ (w .* (A_p * (1 ./ w) + C_p)));
        lfpz = (A_p / C_p) * (phi0 / (2*phi0_prime));
    end
end

```

```

    Rss_laminate = (phi0^2) / (E_hat * C_p);
end

% --- Step 3: Generate crack driving force (G) curves ---
G_matrix = zeros(length(w_vector), length(delta_a_vector));
for k = 1:length(w_vector)
    w = w_vector(k);
    a0 = a0_vector(k);
    sigma_u = sigma_u_func(w); % Predicted strength for this width

    alpha_vector = (a0 + delta_a_vector) / w; % a/w ratio during propagation
    phi_vector = arrayfun(@phi(a), phi(a, rho), alpha_vector);
    phi_vector(alpha_vector >= 1.0 | alpha_vector <= 0) = NaN; % Invalidate out-of-
    bounds

    G_matrix(k, :) = (1/E_hat) * w .* (sigma_u^2) .* (phi_vector.^2);
end

% --- Step 4: Determine the R-curve and key parameters ---
R_curve_laminate = min(G_matrix, [], 1, 'omitnan'); % R-curve is the lower envelope
R_curve_laminate(1) = 0; % Set initial toughness to zero
R0_ss = (t_total / t0) * Rss_laminate; % Rescaled steady-state toughness for 0-deg
plies

% --- Step 5: Fit model to the R-curve to find shape parameters ---
r_curve_model = @(p, delta_a) Rss_laminate * (1 - (1 - p(1) * delta_a).^p(2));
fit_indices = delta_a_vector > 0 & delta_a_vector <= lfpz;

options = optimoptions('lsqcurvefit', 'Display', 'off');
fit_params = lsqcurvefit(r_curve_model, [1/lfpz, 3.0], ...
    delta_a_vector(fit_indices), R_curve_laminate(fit_indices), ...
    [1e-6, 1.0], [inf, inf], options);

% --- Step 6: Store all calculated results ---
materials(i).E_hat = E_hat; materials(i).lambda = lambda; materials(i).rho = rho;
materials(i).t0 = t0; materials(i).t_total = t_total;
materials(i).lfpz = lfpz;
materials(i).Rss = Rss_laminate;
materials(i).R0_ss = R0_ss;
materials(i).gamma = fit_params(1); % R-curve shape parameter gamma
materials(i).beta = fit_params(2); % R-curve shape parameter beta

% --- Step 7: Plotting ---
if enable_plotting
    % Create and select the appropriate figure and subplot
    if i <= 2 % Non-Hybrid materials
        if i == 1, figure('Name', 'Non-Hybrid R-curves', 'Position', [100 100 1000
            450]); end
        subplot(1, 2, i);
    else % Hybrid materials
        if i == 3, figure('Name', 'Hybrid R-curves', 'Position', [150 150 1500 450]);
        end
        subplot(1, 3, i - 2);
    end

    hold on;
    plot(delta_a_vector, G_matrix, '--', 'Color', color_blue, 'LineWidth', 0.5);

```

```

        plot(delta_a_vector, R_curve_laminate, '-', 'Color', color_orange, 'LineWidth',
              2.0);

        % Formatting
        title(plot_titles{i});
        xlabel('Crack Extension, $\Delta a$ (mm)', 'Interpreter', 'latex');
        ylabel('Toughness, $\mathcal{R}$ (kJ/m$^2$)', 'Interpreter', 'latex');
        ylim([0 400]); xlim([0 8]);
        grid on; box on; hold off;
    end
end

%% --- FINAL RESULTS OUTPUT ---
% This section prints the key calculated parameters in a formatted table.

fprintf('\n\n--- FINAL RESULTS FOR LATEX TABLE ---\n\n');
fprintf('Parameter\t\tT800S\t\tHR40\t\t[T/H/T]\t\t[2T/H/2T]\t\t[3T/2H/3T]\n');
fprintf([repmat('-', 1, 110) '\n']);
fprintf('E_hat (GPa)\t\t\t%.2f\t\t%.2f\t\t%.2f\t\t%.2f\t\t%.2f\n', [materials.E_hat]/1000
        );
fprintf('lambda\t\t\t\t%.4f\t\t%.4f\t\t%.4f\t\t%.4f\t\t%.4f\n', [materials.lambda]);
fprintf('rho\t\t\t\t\t%.4f\t\t%.4f\t\t%.4f\t\t%.4f\t\t%.4f\n', [materials.rho]);
fprintf('t0 (mm)\t\t\t\t%.3f\t\t%.3f\t\t%.3f\t\t%.3f\t\t%.3f\n', [materials.t0]);
fprintf('t_total (mm)\t\t\t%.3f\t\t%.3f\t\t%.3f\t\t%.3f\t\t%.3f\n', [materials.t_total]);
fprintf('\n--- Final 0-deg Ply R-Curve Results ---\n');
fprintf('l_fpz (mm)\t\t\t%.4f\t\t%.4f\t\t%.4f\t\t%.4f\t\t%.4f\n', [materials.lfpz]);
fprintf('R_ss_laminate (kJ/m^2)\t\t%.2f\t\t%.2f\t\t%.2f\t\t%.2f\t\t%.2f\n', [materials.Rss]);
        ;
fprintf('R_0_ss (kJ/m^2)\t\t\t%.2f\t\t%.2f\t\t%.2f\t\t%.2f\t\t%.2f\n', [materials.R0_ss]);
fprintf('gamma (mm^-1)\t\t\t%.4f\t\t%.4f\t\t%.4f\t\t%.4f\t\t%.4f\n', [materials.gamma]);
fprintf('beta\t\t\t\t\t%.4f\t\t%.4f\t\t%.4f\t\t%.4f\t\t%.4f\n', [materials.beta]);
fprintf('\n');

```