Advanced Placed Ply laminate architectures for improved impact tolerance: experimental characterization and simulation

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Lay Summary

Air travel is one of the largest contributors to greenhouse gas emissions worldwide. If the commercial aviation industry were a country, it would place sixth in the global CO$_2$ emissions rankings, and because these gases are emitted at high altitude, their effect on global warming is even more significant. By reducing the weight of aircraft, it is possible to reduce fuel consumption, which in turn lowers CO$_2$ emissions. This weight reduction can be achieved through the use of composite materials — such as carbon fiber composites — which have a higher strength to weight ratio than the materials conventionally used in aircraft.

While the proportion of composite materials in aircraft has risen steadily over the past 50 years, their further adoption is hampered by their susceptibility to damage from out-of-plane loading. This weakness is a result of their laminated structure. Carbon fiber composites consist of long thin carbon fibers, embedded in a plastic. Layers of these plastic impregnated fibers are stacked on top of one another in different directions to form a composite laminate. Since carbon fibers are extremely strong, composites can resist large loads in the directions of the fibers. However, when forces are applied perpendicular to the fibers, the interfaces between the layers in a laminate fail, because they are only held together by the plastic. This phenomena is known as delamination. Existing methods that address this issue include but are not limited to: weaving (in both 2D and 3D), the insertion of so called z-pins in the out-of-plane orientation, and stitching together the layers in a composite. The disadvantage of many of these methods is that they bend fibers, significantly reducing their strength and stiffness in the direction of the fibers.

In this study, a novel composite manufacturing method is investigated, known as Advanced Placed Ply (AP-PLY), in which robotic fiber placement machines are used to create composite laminates with variable amounts of layer-to-layer connectivity. This approach improves the delamination resistance of a composite, without affecting its strength in the fiber direction.

First, the in-plane properties of different AP-PLY laminates were determined by testing their tensile strength and stiffness, and comparing those values with those obtained for conventional, non AP-PLY composites. The stiffness of the AP-PLY laminates was found to be comparable to their conventional counterparts, while the effect of the AP-PLY structure on laminate strength was found to be dependent on the orientations of the fibers.
in each laminate. Since materials may behave differently when they are loaded dynamically — meaning at high speed — the AP-PLY laminates were subsequently subjected to the same tests at much higher loading velocities. These significantly more complex experiments were conducted at the European Commission’s Joint Research Center in Italy, one of the only facilities worldwide with the apparatus and expertise required to test these types of materials dynamically. The AP-PLY laminates proved to be approximately 10% stiffer than normal, non AP-PLY laminates, while their strengths were comparable with those of the baseline composites. Finally, the impact tolerance of AP-PLY composites was assessed using drop-weight tower tests, in which a steel projectile is dropped from a height onto a composite test specimen. The AP-PLY architecture was found to inhibit the formation of delamination, i.e. the damage caused by the impact loading was minimized by the internal structure of each AP-PLY laminate. After damaging the laminates in the drop weight tower tests, their residual strength was evaluated. This test defines the ability of a composite material to resist loads even when it has been damaged, a property which is critical to ensure the safety of a structure. AP-PLY laminates exhibited excellent residual strengths suggesting the material is a good choice for use in applications where dynamic loads may be encountered, e.g. wingskins, aircraft engines etc.

In addition to experimental studies of AP-PLY composites, their behavior was also modeled numerically. A novel methodology was proposed to efficiently model the geometry of AP-PLY composites, lowering the cost and time required to simulate the response of these materials to different types of loading. The simulations were able to capture the effect of the through thickness fiber connectivity in the AP-PLY laminates, and provide additional insights into the complex deformation of these types of composites. Moreover, their development facilitated the efficient study of different types of AP-PLY laminate structures, which would have been expensive and time consuming to investigate experimentally.
Abstract

The aviation industry accounts for approximately 2.1% of global carbon dioxide emissions. Through the use of increasing quantities of advanced composites it is possible to reduce aircraft weight, improving fuel efficiency and reducing greenhouse gas emissions. Currently, the further adoption of composite materials in structural applications is hampered by their susceptibility to damage from out-of-plane loading. Existing methods to address this shortcoming, e.g. 3D weaving, have undesirable effects on a composite's in-plane strength and stiffness. Recently, an automated fiber placement based preforming method, known as Advanced Placed Ply (AP-PLY), has been developed, which harnesses the flexibility and precision of automated fiber placement machinery to produce laminates with a quasi-woven internal architecture. The through thickness fiber connectivity present in these laminates has been shown to improve their impact tolerance while having little to no effect on their undamaged in-plane strength and stiffness.

This thesis presents a comprehensive study of the mechanical response of AP-PLY laminates. Experimental techniques and numerical models have been employed to determine the influence of the 3D reinforcements present in AP-PLY laminates on their quasi-static and dynamic behavior. Macroscopic mechanical testing (quasi-static and dynamic), optical and scanning electron microscopy, as well as X-ray tomography, were used to characterize the deformation and energy dissipation mechanisms of AP-PLY laminates at different length scales. The AP-PLY laminates exhibit outstanding toughness when subjected to dynamic loads, while their undamaged in-plane response is improved in comparison with conventional angle-ply laminates.

It was found that in the quasi-static regime the tensile moduli of AP-PLY laminates were comparable to those of their conventional non AP-PLY counterparts – there was no reduction in laminate stiffness despite the presence of fiber crimp. The effect of the AP-PLY preforming process on laminate strength was found to be dependent on each laminate’s layup. Strain concentrations at through thickness undulations were observed using digital image correlation. Specimen failure was found to occur along planes aligned with these strain concentrations. The dynamic tensile response of AP-PLY laminates was studied using a split-Hopkinson bar. Due to the large size of the AP-PLY specimens, the experiments were conducted at the European Commission’s Joint Research Center in Italy, who operate the world’s largest
split Hopkinson bar. At strain rates of $30\text{s}^{-1}$, the AP-PLY specimens proved to be approximately 10% stiffer than their conventional equivalents, while strengths were similar for both configurations. The short duration of the dynamic experiments limited the realignment of undulating tows with the loading direction, reducing strain concentrations and mitigating the effect of the AP-PLY architecture on laminate performance. Finally, low velocity impact and compression after impact experiments were conducted to characterize the impact resistance and toughness of AP-PLY composites. Cross-ply, triaxial, and quasi-isotropic AP-PLY laminates were subjected to 30 J and 50 J impacts, and subsequently loaded in compression to evaluate the effect of the impact induced damage on their strength. While the high energy impacts resulted in significantly larger delamination footprints, the corresponding reductions in residual compressive strength were marginal. This was attributed to formation of stable sub laminates in the AP-PLY laminates as a consequence of their through thickness fiber connectivity.

In addition to the experimental characterization of AP-PLY composites, a multiscale numerical modelling framework was developed to efficiently and accurately capture the behavior of AP-PLY laminates. The multiscale framework divided each AP-PLY laminate into simple prismatic regions consisting of varying quantities of fibers and resin. The constituents of each region were rotated in-plane and out-of-plane to account for the orientations of the fibers at through thickness undulations. The use of a multiscale framework to represent through thickness fiber undulations allowed the complex damage mechanisms occurring in AP-PLY composites to be captured using coarse meshes. A continuum damage mechanics approach was used to capture the initiation and propagation of intralaminar damage, and a cohesive zone model was implemented to account for delamination at ply interfaces.

The multiscale framework was found to be in good agreement with the experimental results, correctly predicting laminate stiffness, strength, and strain-to-failure in both the quasi-static and dynamic regime. The strain concentrations at tow undulations were well captured by the numerical modeling approach, and there was good agreement between the experimentally observed and numerically predicted specimen failure modes. Furthermore, the numerical models were able to provide additional insights into the deformation mechanisms occurring within AP-PLY laminates, including the formation and propagation of fiber and matrix cracks, as well as delamination. Lastly, the model facilitated the investigation of AP-PLY design parameters – e.g. the number of tows width gaps between tows placed in a single pass – which would have been prohibitively expensive and time consuming to study experimentally.
Declaration

I declare that the thesis has been composed by myself and that the work has not be submitted for any other degree or professional qualification. I confirm that the work submitted is my own, except where work which has formed part of jointly-authored publications has been included. My contribution and those of the other authors to this work have been explicitly indicated below. I confirm that appropriate credit has been given within this thesis where reference has been made to the work of others.

The work presented in Chapter 6 was previously published in Composites: Part A as *Tensile response of AP-PLY composites: a multiscale experimental and numerical study* by Rutger Kok, Francisca Martinez-Hergueta (supervisor), and Filipe Teixeira-Dias (second supervisor). This study was conceived by all of the authors. The author contributions according to the CRediT framework are: Rutger Kok: Investigation, methodology, formal analysis, software, visualisation, data curation, validation, writing - Original draft preparation. Francisca Martinez Hergueta: supervision, conceptualisation, project administration, writing- review & editing, funding acquisition. Filipe Teixeira Dias: Writing- review & editing, supervision.
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Nomenclature

Abbreviations
ABD Classical laminated plate theory compliance matrix
ADL Allowable Design Limit
AE Acoustic Emission (Monitoring)
AFP Automated Fiber Placement
AITM Airbus Industry Test Method
AP-PLY Advanced Placed Ply
ASTM American Society for the Testing of Materials
ATL Automated Tape Laying
BVID Barely Visible Impact Damage
CAI Compression After Impact
CDM Continuum Damage Mechanics
CFRP Carbon Fiber Reinforced Polymers
CNC Computer Numerical Control
CNC Computerized Numerical Control
COR Coefficient of restitution
cpt Cured ply thickness
CT Computed Tomography
CZM Cohesive Zone Method
DCB Double Cantilever Beam
DDM Discrete Damage Mechanics
DIC Digital Image Correlation
ELSA  European Laboratory for Structural Assessment
FE    Finite Element
FEA   Finite Element Analysis
FRP   Fiber Reinforced Polymer
GEOS  Geometry Engine Open Source
INTA  Instituto Nacional de Técnica Aeroespacial
is    In-situ
ISO   International Standards Organization
MDD   Maximum Design Damage
NCF   Non-crimp Fabric
PEEK  Polyetheretherketone
QI    Quasi-isotropic
RDD   Readily Detectable Damage
RTM   Resin Transfer Molding
RVE   Representative Volume Element
SEM   Scanning Election Microscopy
SHB   Split-Hopkinson Bar
TRI   Triaxial
VARTM Vacuum Assisted Resin Transfer Molding
VCCT  Virtual Crack Closure Technique
XFEM  eXtended Finite Element Method
XP    Cross-ply

Symbols
\( \alpha \)  Cohesive interface stiffness calculation parameter
\( \beta \)  Shear response factor
\( \delta \)  Cohesive surface displacement vector
\( \sigma \)  Stress vector
ε     Strain vector
χ     Undulation ratio
δ_i Normal cohesive surface displacement \((i = n, s, t)\)
δ_m Effective displacement between cohesive surfaces
ℓ_1 Longitudinal characteristic length tow constituent
ℓ_2 Transverse characteristic length tow constituent
ℓ_3 Through-thickness characteristic length tow constituent
ℓ_m Characteristic length pure resin constituent
η Benzeggagh-Kenane mixed mode interaction parameter
γ_ij Shear strain direction \(ij\)
Λ^o_{22} In-situ transverse tensile strength parameter
\mathcal{F} Deformation gradient
\mathcal{G} Fracture energy
\mathcal{G}_{1}^{T/C} Longitudinal tensile/compressive fracture toughness
\mathcal{G}_{2}^{T/C} Transverse tensile/compressive fracture toughness
\mathcal{G}_{3}^{T/C} Through-thickness tensile/compressive fracture toughness
ν_ij Poissons ratio direction \(ij\)
ν_m Poissons ratio pure resin
φ Out of plane tow orientation
ψ In-situ shear strength parameter
σ_{ii} Normal stress direction \(ii\)
τ_ij Shear stress direction \(ij\)
C Stiffness matrix
D Damage matrix
H_m Compliance matrix pur resin regions
H_{tow} Compliance matrix tow regions
K Cohesive surface stiffness vector

XVII
\( T \) Coordinate transformation matrix
\( t \) Cohesive surface traction vector
\( \theta \) In-plane angle of AP-PLY tows
\( \varepsilon_{ii} \) Normal components strain vector
\( \varepsilon_{ii} \) Normal strains direction \( ii \)
\( A_{1}^{T/C} \) Longitudinal tensile/compressive damage dissipation fitting parameter
\( A_{1}^{+} \) Longitudinal tensile compressive damage coupling parameter
\( A_{m}^{T/C} \) Pure resin damage dissipation fitting parameter
\( b \) Longitudinal compressive stiffness retention parameter
\( d_{1}^{C} \) Longitudinal pure compression scalar damage variable
\( d_{1}^{T/C} \) Longitudinal tensile/compressive scalar damage variable
\( d_{2}^{T/C} \) Transverse tensile/compressive scalar damage variable
\( d_{3}^{T/C} \) Through-thickness tensile/compressive scalar damage variable
\( D_{ij} \) Components of \( D \) matrix
\( d_{m} \) Pure resin scalar damage variable
\( E_{ii} \) Youngs moduli direction \( ii \)
\( E_{m} \) Youngs modulus pure resin
\( E_{D} \) Dissipated energy (in low veloticy impact test)
\( E_{i} \) Impact energy
\( F \) Force
\( F_{1}^{T/C} \) Longitudinal failure criterion tension/compression
\( F_{2}^{T/C} \) Transverse failure criterion tension/compression
\( F_{3}^{T/C} \) Through-thickness failure criterion tension/compression
\( F_{m}^{T/C} \) Pure resin failure criterion tension/compression
\( g \) Energy dissipated per unit volume
\( G_{c} \) Mixed mode critical fracture energy
\( G_{ic} \) Mode \( i \) critical fracture energy \( (i = I, II, III) \)
\( G_{ij} \)  
Shear modulus direction \( ij \)

\( h \)  
Cohesive layer thickness

\( I_1 \)  
First invariant stress tensor

\( J_2 \)  
Second invariant deviatoric stress tensor

\( K_{ii} \)  
Cohesive surface stiffness \( (i = n, s, t) \)

\( L \)  
Length of impact specimen

\( L_u \)  
Undulation length

\( r_{1T/C}^T \)  
Longitudinal tensile/compressive elastic domain threshold

\( r_{mT/C}^T \)  
Pure resin tensile/compressive elastic domain threshold

\( s \)  
Time-steps

\( S^L \)  
Shear strength 12-direction

\( S^R \)  
Shear strength 31-direction

\( S^T \)  
Shear strength 23-direction

\( t \)  
Time

\( t_i \)  
Cohesive surface tractions \( (i = n, s, t) \)

\( t_0^i \)  
Cohesive surface traction at damage onset \( (i = n, s, t) \)

\( u, v, w \)  
Displacement in x,y,z directions

\( V^f \)  
Volume fraction

\( V_i \)  
Volume of constituent \( i \)

\( W \)  
Width of impact specimen

\( w_{nom} \)  
Nominal tow width

\( w_r \)  
Width of resin-rich region

\( w_t \)  
Width of central tow region

\( X^C \)  
Longitudinal compressive strength

\( X^T \)  
Longitudinal tensile strength

\( x_0, y_0 \)  
Mid-point of AP-PLY laminate architecture

\( x_F, y_F \)  
Point load coordinates
$Y^C$  Transverse tensile strength

$Y^T$  Transverse tensile strength

$Z^C$  Through-thickness tensile strength

$Z^T$  Through-thickness tensile strength
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Chapter 1

Introduction

1.1 Introduction

There is a growing awareness of humanity’s impact on the environment, and a consensus on the need to take action to reduce our consumption of natural resources and mitigate the effects of climate change. The aviation industry produces approximately 2% of global carbon emissions, more than the vast majority of countries including the UK [20]. Changing attitudes to personal carbon footprints may lessen the demand for leisure flights in the long term. However, air freight forms an integral part of global supply chains and cannot be as easily substituted with other modes of transport. As such, the reduction of aviation industry emissions must be driven by improvements in aircraft efficiency.

To date, efficiency improvements have been achieved in part through the development of lighter and therefore more economical aircraft. The remarkable reductions in aircraft weight over the past few decades can be attributed partly to the increased use of lightweight composite materials. According to Boeing, the extensive use of composites in the construction of the 787 resulted in weight savings of approximately 20 percent compared to more conventional aluminum designs [21]. Over the last 30 years, the percentage of composites content in commercial aircraft (by weight) has risen from under 10% to over 50% (Figure 1.1).

While modern composite materials boast extremely high specific strengths and stiffnesses, their laminated structure makes them susceptible to damage from out-of-plane loading. Arguably, this is one of the primary factors limiting the further adoption of composite materials in structural applications [22]. A wide variety of methods have been proposed to improve the impact tolerance of composites, ranging from methods rooted in chemistry e.g. the development of tougher matrices, to methods which manipulate the fibers in a laminate to improve through thickness properties, e.g. weaving, z-pinning, or stitching. These methods, while effective in improving impact tolerance, generally have a deleterious effect on the undamaged in-plane strength and stiffness of a composite.
In recent years, improvements in reliability and productivity, among other factors, have led to the increasing adoption of automated fiber placement (AFP) and automated tape laying (ATL) technologies in the aerospace industry (Figure 1.2). The increased precision with which these machines can place plies permits the production of laminates with unconventional fiber orientations and novel internal architectures. One such novel composite design strategy, known as Advanced Placed Ply, or AP-PLY for short, involves the use of AFP to produce laminates with quasi-woven internal architectures. This novel preforming method creates through thickness fiber connectivity, improving the impact resistance and tolerance of a laminate while retaining the high in-plane stiffness and strength of more conventional angle-ply laminates.

Previous studies have reported significant improvements in mode I interlaminar fracture toughness and compression after impact (CAI) strength as a result of this novel preforming method [10]. As such, AP-PLY composites are potentially a suitable material choice for aerospace components susceptible to dynamic loads such as engine blades, brackets, interiors, nacelles, propellers/rotors, and wings.

1.2 Motivation

To date, various studies have experimentally characterized the response of AP-PLY composites to different types of loading [10, 19, 23, 24, 25, 26]. However, there are still significant gaps in our understanding of the behaviour of AP-
PLY composites. Continued study of these materials will facilitate their further adoption in structural applications.

In particular, the mesoscale effects of through thickness fiber undulations have yet to be investigated in detail. An in depth understanding of these features of AP-PLY laminates will allow for more informed design decisions. Furthermore, while the mechanical properties of AP-PLY composites at quasi-static strain rates have been previously investigated, the high strain rate behaviour of these materials is relatively unknown. It is essential to understand the effects of strain rate on material performance to ensure the safe operation of structures likely to encounter dynamic loads.

While a number of researchers have developed numerical models of AP-PLY composites they are limited in scope (only modeling delamination) [27, 28], limited in their ability to capture the effect of the through thickness reinforcements [19], or limited in their ability to predict arbitrary 3D crack paths [29]. An automated, fully 3D numerical modeling framework for AP-PLY laminates, which is capable of accounting for the effect of through thickness fiber undulations has yet to be developed. This is a consequence of the complexity of the internal architecture of AP-PLY laminates, which is difficult and time consuming to replicate numerically.

The development of such a numerical modeling framework would have a number of benefits. First and foremost, the ability to track stresses, strains, and damage accumulation in a laminate during loading would provide further insight into the dominant deformation micromechanisms in AP-PLY laminates. Second, accurate and efficient numerical models would facilitate the design of composite structures with AP-PLY features. Lastly, an efficient numerical
tool would facilitate the exploration of the AP-PLY composite design space, ultimately allowing for the optimization of AP-PLY laminate architectures.

1.3 Objectives

In light of the above, the present work has the following aims. First, to develop a numerical modeling framework which can: generate the geometries of AP-PLY laminates with different configurations, accurately and efficiently predict their performance under various types of loading, and provide further insight into their deformation mechanisms. Second, the study intends to establish the effects of through thickness undulations on the in-plane mechanical properties of AP-PLY laminates at different strain rates. Lastly, this thesis investigates the low velocity impact characteristics of different AP-PLY configurations, specifically, the ability of the quasi-woven internal architecture to inhibit the formation and propagation of damage.

This thesis is divided into 8 chapters. Following this brief introduction, Chapter 2, the literature review, will provide an overview of the mechanics of composite materials, discuss existing methods for impact tolerance improvement in composite laminates, and present the results of previous studies on AP-PLY composites. Chapter 3 explains the process of AP-PLY preforming in greater detail, introducing the different parameters that define these types of composites. Chapter 4 presents the numerical modeling framework developed in this study to model AP-PLY composites. Through thickness tow undulations are briefly investigated at the microscale level in Chapter 5. The quasi-static and dynamic tensile characterization of AP-PLY laminates are addressed in Chapters 6 & 7. Finally, chapter 8 presents the results of low velocity impact tests on AP-PLY composites.
Chapter 2

Literature Review

2.1 Introduction to Composite Materials

Composites are materials composed of two or more macroscopically distinct constituents. When combined, the properties of the new material will differ significantly from the properties of its constituents. Examples of composite materials include biological composites such as wood (cellulose and lignin) or bone (collagen and hydroxyapatite), ancient composites such as wattle and daub, and modern materials such as fiber reinforced polymers (FRP).

The term FRPs covers a wide range of materials with very different properties, but all FRPs consist of thin reinforcing fibers, embedded in a polymer "matrix". Commonly used fibers include glass, aramid (e.g. Kevlar), and carbon, which can be continuous or discontinuous, as illustrated in Figure 2.1. In this work the focus will be solely on continuous unidirectional FRPs.

A single unidirectional ply or laminae is extremely strong in the direction of the fibers, but weak in the 90° direction because loads are carried solely by the polymeric matrix. To create a structure that can carry loads in multiple directions, individual unidirectional plies or laminae are stacked in different directions to form a composite laminate (Figure 2.2).
FRPs are exceptionally strong for their weight, a typical unidirectional aerospace grade carbon fiber composite has a specific strength six times higher than high yield steel, and a specific modulus three times greater. Aside from their high specific strength and stiffness, FRPs offer high levels of design freedom, the ability to incorporate additional functionality, good fatigue properties, and, depending on the choice of fiber and matrix, excellent resistance to chemical and environmental degradation [32].

2.2 Composite manufacturing methods

Advanced composites such as carbon fiber reinforced polymers (CFRP) consist of carbon fibers impregnated with a polymer matrix. Depending on the application, the fibers in a preform may be randomly oriented short segments, or long continuous strands oriented in a single or in multiple directions. Matrices are typically either thermosetting resins such as epoxy or polyester, or thermoplastics, for example, poly-carbonate, poly-amide, or poly-ether ether ketone. To produce a composite part, the fibers are combined with the matrix, typically in its liquid (uncured) state, and cured under the application of temperature and pressure. How the fiber and matrix are combined, and the method of application of heat and pressure has a significant impact on the final properties of the composite part.

Over the years, a large number of manufacturing methods have been
developed, ranging from relatively cheap manual processes such as wet layups, to highly automated processes such as AFP. The following section discusses the merits of a variety of manufacturing methods. Since this thesis is concerned primarily with structural composites for the aerospace industry, high volume short fiber composite production methods such as injection moulding are omitted from the discussion.

### 2.2.1 Wet Layup

The most basic, lowest cost, manufacturing method for FRPs is wet layup. Dry fibers, generally in the form of woven fabrics or randomly oriented chopped strand mats, are manually wetted with liquid resin and laid into a mold. The completed part is subsequently cured at room temperature under ambient pressure, in an oven under vacuum, in some form of press at elevated temperature, or in an autoclave [33]. The quality of the resultant product is highly dependent on the skill of the laminator and the curing process. Generally, producing parts using a wet layup is slow, and it is difficult to create high fiber volume fraction parts with low void contents. The advantages of a wet layup are the low cost of tooling, and the ability to choose from a range of fiber and matrix types [34]. Wet layups are typically used for the low-volume production of boat hulls or wind turbine blades [35].

### 2.2.2 Hand (Prepreg) Layup

In contrast to a wet layup, in a hand layup the fibers have been pre-impregnated with resin (generally a thermosetting resin such as epoxy). These so called prepreg sheets are partially cured to facilitate handling, but will not harden until the curing process is completed at elevated temperature and pressure. As in a wet layup, laminates may be cured in a press, or by vacuum bagging (both in or out-of-autoclave). To prevent the prepreg sheets from curing at room temperature prior to use, the material must be stored in a freezer. By pre-impregnating the fibers with resin, the void content of finished products is reduced. In addition, the process is safer and cleaner than a wet layup due to the reduced emission of volatile organic compounds [35].

### 2.2.3 Resin Transfer Molding

In resin transfer molding (RTM), dry fiber preforms are placed into a matched mold and resin is injected into the cavity at high pressure (typically 3.5-7 bar) to infuse the fibers [36]. The advantages of this process are the high dimensional tolerances that can be achieved, the high fiber volume fractions of the resultant parts, and the low quantities of volatile organic compounds that are emitted [35][37].

In vacuum assisted RTM (VARTM) one side of the mold is replaced by a vacuum bag (Figure 2.3). Rather than injecting the resin into the mold at high
pressure, a vacuum is used to pull resin through the mold, impregnating the fibers [35]. Relative to RTM, the open mold nature of the VARTM process allows for greater flexibility in the production of different parts, and is easier to scale up to larger components as the mold does not need to stand up to the high positive pressures used in RTM. On the other hand, the lower compressive pressure on the preform compared with RTM limits the achievable fiber volume fraction to values in the 40%-50% range, compared with values of over 60% for high pressure RTM [37][38].

2.2.4 3D Printing

While 3D printing of pure polymers has existed since the early 90s, it is only recently that printing technology has developed sufficiently to allow for the production of short fiber and continuous fiber polymer composites. In conventional "subtractive" manufacturing, complex geometries are created through the removal of material e.g. through machining. In 3D printing, these complex geometries are created (as the name suggests) additively, eliminating the waste associated with subtractive manufacturing methods. Moreover, the process provides designers with far greater control over the material microstructure.

3D printed polymer composites can be divided into two types, particle reinforced composites, and continuous fiber composites. Particle reinforced composites have significantly improved properties relative to their base materials, but lack the mechanical properties necessary for demanding structural applications. Continuous fiber 3D printed composites are manufactured by simultaneously extruding the fibers and the matrix (Figure 2.4). In terms of mechanical properties, 3D printed continuous fiber composites are not yet competitive with conventional manufacturing methods like ATL or RTM. High porosity, low fiber volume fractions, and a susceptibility
to delamination (as a result of the layered deposition of material) limit the strength of these materials. Moreover, the 3D printing process is slow compared to existing automated manufacturing techniques and limited in terms of part size. Lastly, material choice is constrained by the requirements of the chosen printing machine [40]. Currently, most commercial equipment is limited to the use of thermoplastic resins such as nylon. Polyether ether ketone (PEEK) is the focus of recent academic research but has not yet crossed over into mainstream commercial usage.

2.2.5 Filament Winding

Structures with axial symmetry are most commonly manufactured using filament winding. Examples include golf clubs, fishing rods, pressure vessels, and pipes. In conventional filament winding, dry fibers pass through a resin bath and are wound on to a rotating mandrel (Figure 2.5). The winding head moves back and forth parallel to the axis of rotation of the mandrel, creating a helical pattern of deposited tows. Once the winding process is complete, the structures are cured in an autoclave or an oven [41][42]. In some cases the mandrel becomes a part of the finished product, in others it is removed. Advantages of filament winding include low material costs and ease of automation [43].

In some cases prepreg tows may be used in the place of the dry fibers and the resin bath. It is also possible to mold the complete preform using dry fibers, and subsequently using resin transfer molding (Section 2.2.3) to impregnate the fibers and cure the resin. When winding using thermoplastic resins, the matrix can be "cured" in-situ, eliminating the need for autoclave or oven curing [43].

2.2.6 Automated Tape Laying

In Automated Tape Laying (ATL) a robotic tape placement head places 3-12 inch wide unidirectional tapes onto a flat or slightly contoured mandrel (inclines of up to 15° are permissible), see Figure 2.6. A compaction roller effectively debulks the material as the tape is laid down. On some machines the tapes may be preheated by hot air, infrared lamps, or lasers to improve tape to tape
adhesion. Defects are detected using optical sensors to allow for operator intervention [41].

The advantages of ATL include the high level of productivity, the ability to lay up prepregs with high areal weights, good mechanical properties, the ability to manufacture large parts, and low levels of material waste (compared with manual layups). Furthermore, the computer numerically controlled precision of ATL systems eliminates some of the layup errors which might occur during manual layups, resulting in high repeatability. Limitations of ATL include the high initial capital cost, the limited ability to produce complex parts with large curvatures. Currently, as a result of its high cost, ATL is used primarily in the aerospace industry for the production of large flat parts such as wingskins, tail planes, and wing boxes [45].

2.2.7 Automated Fiber Placement

Automated Fiber Placement (AFP) is similar to ATL but differs in the width of the material laid down with each pass of the placement head. Instead of placing a single wide tape, an AFP head places a collimated band of individual tows (typically 1/8 inch to 1/2 inch wide). Generally, AFP heads can
accommodate 12-32 tows in parallel at one time\cite{41}. A schematic illustration and a photograph of an AFP head are provided in Figure 2.7. Each tow can be individually tensioned, cut, and restarted, facilitating the lay up of complex geometries with concave curvatures. An added benefit of the individual tow control is reduced material waste relative to ATL systems. On the other hand, gaps between tows are larger in an AFP layup than in an ATL layup, and productivity is generally lower than in ATL, one study reported a twofold increase in layup time for a complex fuselage section using AFP rather than ATL \cite{45}.

The narrow width of AFP tows also allows for the production of parts with curved fiber paths. So called fiber or tow steering is a recent development facilitated by the improved heating and tow compaction control of the latest generation of AFP machines \cite{46}. Tow steering expands the design space for composites and can provide greater control over the mechanical properties of a composite than through changes in ply stacking sequences alone \cite{47}. 

\begin{figure}[h]
\centering
\includegraphics[width=\textwidth]{figure2_6.png}
\caption{A gantry mounted ATL system from mTorres \cite{44}}
\end{figure}
To date manufacturing difficulties, such as tow buckling at the inner radius of steered plies, and stress concentrations at so called "tow convergence zones", have limited the adoption of this novel manufacturing technique [46].

2.3 Mechanical Response and Failure Modes

A single unidirectional fiber reinforced composite laminate behaves as a transversely isotropic material. In other words, the material has the same properties in one plane (the transverse/through-thickness plane) and different properties in the direction normal to this plane (the fiber direction). Longitudinal properties are fiber dominated, the stress-strain response is linear-elastic dropping off suddenly when the brittle fibers in the lamina fracture. Orthogonal to the fibers, the properties of a composite are matrix dominated. As such, the transverse and shear stress-strain response is more ductile than in the fiber direction. The non-linearity of the transverse and shear response is dependent on the polymer type.

As damage forms in a composite, the material response is changed. In an undamaged composite, all the external work done to deform a composite is stored as elastic strain energy. However, in a damaged composite, energy is dissipated by the formation and advancement of cracks. As a crack propagates, stored elastic energy in the surrounding material - as well as a proportion of the external work done - is released. The strain energy release rate is also known as the fracture energy, and expresses the energy required to propagate a crack (as energy required per unit of projected area) [48].

Cracks in a composite laminate can form within individual laminae (intralaminar failure) or at the interfaces between plies (interlaminar failure). Following is a discussion of the different failure modes observed in a composite. It is worth noting that energy is also dissipated through the plastic deformation of the matrix, however, the presence of the fibers in a composite restrains the matrix to such an extent that the energy dissipated through this damage mode is negligible [49].

2.3.1 Intralaminar failure modes

Within a ply, the primary failure mechanisms are: matrix cracking, fiber breakage, and fiber-matrix debonding. Depending on the direction of loading, these failure mechanisms result in various failure modes, detailed below.

Tensile matrix failure

Tensile loading transverse to the fibers causes cracks to form perpendicular to the direction of loading. Typically, the cracks form due to fiber-matrix debonding, as illustrated in Figure [2.8a]. The fiber-matrix interface is generally the weakest part of a composite laminate. Transverse matrix cracking
normally results in only minor reductions in the overall stiffness of a structure and dissipates the least amount of energy compared with the other failure modes [50].

Compressive matrix failure

Under compressive transverse loading, the matrix in a composite typically fails through the formation of shear bands. Although the maximum shear stress under pure compressive loading occurs at an angle of 45°, a large number of experimental studies have found that the actual fracture angle in carbon-epoxy composites is closer to 53° ± 2° as depicted in Figure 2.8b [51][52][4]. Since compressive loading results in crack closure, this failure mode is generally not catastrophic and absorbs more energy than the tensile matrix failure mode.

In-plane shear matrix failure

When loaded in shear, matrix damage initiates in the form of cracks oriented at 45° to the fiber direction (Figure 2.9 [53]. As loads rise, the cracks tend to rotate, aligning with the fibers to form axial cracks [54].

Tensile fiber failure

Under axial tension, the fibers in a composite fail by fracturing on a plane perpendicular to the loading direction (Figure 2.10a). Fiber failure is often the
most critical failure mode in a lamina as it has the most significant effect on the residual properties of a laminate. The energy dissipated through tensile fiber fracture significantly exceeds the energy dissipated through other damage modes. At low loads, fibers fail singly in random locations throughout the lamina, while closer to final failure, clusters of four or five broken fibers can be observed [54]. The energy dissipated by tensile fiber fracture comes primarily from friction during fiber pull-out, the fracture of the fiber itself contributes to the released energy only minimally [48].

**Compressive fiber failure**

In axial compression, fiber failure occurs through micro-buckling, a phenomenon known as fiber kinking due to the formation of "kink bands" (Figure 2.10b). Since the matrix restrains the fibers in the directions orthogonal to the fibers, the compressive strength of a lamina is a function of both the fiber and the matrix properties [54]. Furthermore, the waviness of the fibers in a lamina, due to micro-structural defects or processing (e.g. weaving), strongly affects the compressive strength of a lamina [56].

Figure 2.10: (a) X-ray computed tomography image of fiber fractures in an axially loaded carbon fiber epoxy specimen [5]. (b) failure of a carbon epoxy composite under compressive loading through the formation of a kink band, adapted from [6].
2.3.2 In-situ effect

The failure mechanisms and strengths of individual plies are generally measured using unidirectional specimens. However, when a ply in a laminate is constrained by neighboring plies with different orientations, the transverse tensile and shear strengths of the ply are increased. This effect was first observed by Parvizi et al. and is known as the in-situ effect. The magnitude of the in-situ effect is a function of the number of clustered plies in the laminate, and their orientation [57][50]. As illustrated in Figure 2.11, as the number of clustered plies \( n \) increases, the in-situ effect becomes less pronounced, as fewer plies are constrained by neighboring plies with different orientations.

Figure 2.11: In-situ and unidirectional shear strength as a function of ply cluster size. Adapted from [50].

2.3.3 Interlaminar failure

Aside from the failure of its constituent plies, a laminate can also fail due to decohesion of the interfaces between plies, known as delamination. Due to the laminated structure of conventional composites there are no fibers providing reinforcement through the thickness. As a result, a composite’s strength in that direction is dependent primarily on the strength of the weak and often brittle matrix and fiber/matrix interface. Under impact loading, delamination arises primarily from bending stresses. The mismatch of bending stiffnesses between adjacent plies leads to interlaminar stresses that induce interfacial debonding [58]. For this reason, delamination between adjacent plies with the same orientation rarely occurs [59]. In addition, reducing the mismatch angle between adjacent plies in a laminate can reduce a laminate’s delamination footprint, see Section 2.5.2.

Needless to say, delamination affects the through-thickness strength of a composite, but it should be noted that it can also play a role in determining
the in-plane strengths of a composite. When loaded in-plane, composites fail through either the fracture of the fibers, or through the formation of a fracture surface (through delamination and matrix cracking) that doesn’t require fiber failure [60]. For example, Figure 2.12 illustrates a [45°, -45°] laminate loaded in tension which has failed due to pullout, without any of the fibers having fractured. It should be noted that this type of failure occurs predominantly in laminates loaded in shear with large (90°) ply mismatch angles [61][62]. In multi-directional laminates with smaller ply mismatch angles delamination is oftentimes a secondary failure mode, occurring after the initiation of fiber failure and/or matrix damage [61].

Delamination is particularly insidious because it can have a significant impact on the residual strength of a composite, without being readily detectable [60]. Since delamination occurs under the surface of a laminate, it is generally not visible to the human eye, requiring ultrasonic inspection to ascertain the damage footprint.

2.3.4 Effect of strain rate

The mechanical response of composites is dependent on the rate at which they are loaded. The magnitude of the strain rate dependency is a function of the strain rate sensitivity of the composite’s constituents. In the case of carbon fiber epoxy composites, the behavior of the fibers is independent of the loading rate [63]. In contrast, experimental studies on the effect of the loading rate on neat epoxy resin have found that at higher strain rates the modulus and strength of the resin are increased, while the ductility of the material is reduced, leading to more brittle fracture [64][65].

In composites, strain rate dependency can be observed predominantly in the matrix dominated loading directions. Under transverse or shear loading,
the changes in material behavior are similar to those observed in neat epoxy resin, i.e. at higher strain rates, CFRPs exhibit higher moduli, higher strengths, and less ductile failure, see Figure 2.13 [66][67]. Longitudinal compressive loading may also exhibit strain rate dependency. At high strain rates the higher matrix yield strengths result in increased "confinement" of the fibers, complicating the formation of kink-bands [68]. Longitudinal tensile failure is dominated by fiber properties and as such is practically independent of the strain rate [66].

![Tension tests](image)

Figure 2.13: Shear stress-strain response of CFRP at quasi-static and dynamic strain rates. Adapted from [67].

While numerous experimental studies have characterized the strain-rate dependence of conventional angle-ply composites, the effects of high strain rate loading on composites containing 3D reinforcements are less well known. Chen et al. found that the fracture angle of woven composites is reduced at high strain rates, resulting in increased energy dissipation and as corollary higher failure strengths [69]. Similarly, in a study on stitched composites at high strain rates, Tarfaoui et al. found that energy dissipation rose as the strain rate (and the impact energy) were increased [70]. Trochez et al. investigated the high strain rate behavior of AFP manufactured laminates with manufacturing defects (e.g. tow overlaps, tow gaps etc.) finding that the flaws in the laminate had a greater impact on its strength in the high strain rate tests compared with quasi-static experiments [71].

### 2.4 Impact tolerance

Composites are particularly susceptible to delamination damage from low-velocity out-of-plane impact loading. Arguably, this is one of the primary
factors preventing the further use of composites for primary aerospace structures [72]. In metals, the kinetic energy of a projectile is dissipated through both elastic and plastic deformation. Impacts may cause localized permanent deformation, but the load bearing ability of the structure is generally unaffected (for low to medium impact energies) [49]. In contrast, the typically brittle fibers in a composite limit the ability of a laminate to deform plastically. As a result, low-velocity out of plane impacts cause delamination which degrades the load bearing ability of a structure. The effect is most critical in compression, where the reduction in laminate strength may be as high as 50% relative to a pristine laminate [73]. In the aerospace industry, the residual strength of an FRP laminate after impact is one of the primary factors governing the design of a composite structure [10]. The ability of a structure to retain its strength after impact damage is commonly known as impact tolerance.

As previously mentioned, the detection of delamination damage is difficult, and often requires ultrasonic inspection tools and trained operators. Since it is infeasible to constantly monitor all of the composite structures in an aircraft for damage, aircraft design regulations stipulate allowable damage limits (ADLs), which specify the maximum allowable damage size in a structure that will still allow it to meet its regulatory ultimate load requirements [74]. Figure 2.14 illustrates the design philosophy. Large damage sites, which have a significant impact on the strength of a structure, should be easily detectable, and as a result can be repaired prior to an aircraft returning to service. In contrast, barely visible impact damage (BVID), which cannot always be detected during routine maintenance, should not reduce the load carrying ability of a structure below its design limit, and thus should not affect the ability of the aircraft to operate in a safe and effective manner.

![Figure 2.14: Residual strength versus damage size. Adapted from [74].](image-url)
2.4.1 Damage inspection

As delamination damage occurs internal to a laminate, it is often not visible to the human eye. As such, the inspection of delamination damage requires non-destructive testing, the most common of which is ultrasonic inspection.

In ultrasonic inspection high frequency sound waves (typically between 500KHz and 20MHz) are transmitted into a test specimen. When the ultrasonic waves encounter a flaw in the specimen, e.g. a void, or delaminated plies, the acoustic impedance changes, and some of the acoustic energy is reflected. By analyzing the reflected signal it is possible to determine the size of the defects, and their location within the laminate [75]. A typical ultrasound scan of a delaminated composite laminate, in which the depth and extent of delamination can be determined, can be seen in Figure 2.15. Ultrasonic inspection is relatively fast and accurate, but it does require a skilled operator to conduct the testing and interpret the results [76].

Figure 2.15: Interlaminar damage maps obtained using ultrasonic inspection indicating the location of the delamination through the thickness of the specimen. Adapted from [77]

Alternative methods for damage detection in composites include acoustic emission (AE) monitoring or x-ray tomography but these methods are subject to limitations which make their use impractical for damage assessment in impact experiments [75]. In AE monitoring damage in a composite is determined by tracking the emission of ultrasonic waves from a specimen. These ultrasonic waves are generated when strain energy is released through fiber breakage, interface debonding, matrix cracking or delamination [78]. In an impact experiment, the duration of which is typically measured in milliseconds, it is impossible to discriminate between the individual damage events. Moreover, AE does not provide high resolution information on the localization of damage within a specimen [79].

Using x-ray computed tomography (x-ray CT) it is possible to obtain high resolution three-dimensional reconstructions of a damaged laminate. Algorithms can be used to automatically distinguish between fiber fracture,
matrix cracking, and delamination, providing a highly accurate assessment of the damage state within a composite. However, x-ray tomography is expensive, time consuming, and imposes serious limitations on specimen size. On commercially available equipment, the detection of small matrix cracks (5 microns) limits the size of the scanned area to only a few square millimeters, and requires several hours of beam time [61]. State of the art synchrotron x-ray CT allows for much faster image acquisition, but is prohibitively expensive for most applications [80].

2.4.2 Dynamic loading characterization

Drop Weight Tower Testing

Impact testing of composite materials can be broadly divided into three different impact regimes; low velocity impact (speeds in the range $1-10 \, ms^{-1}$), ballistic impact ($>100 \, ms^{-1}$), and hypervelocity impact ($>1000 \, ms^{-1}$). The focus of this study is the low velocity impact regime and as such, methods to characterize the response at higher impact velocities are omitted. The most commonly used method for the characterization of the low velocity impact response of a composite laminate is drop weight tower test according to the American Society for the Testing of Materials (ASTM) standard D7136. A composite specimen measuring 100 mm by 150 mm is placed on a support with a central cutout and restrained by 4 rubber tipped clamps, see Figure 2.16. An impactor is dropped from a predetermined height onto the composite specimen, and the resulting force and displacement of the impactor are recorded.

![Figure 2.16: Support fixture for a drop weight tower test according to ASTM D 7136 [81].](image)

The mass and velocity of the impactor determine the energy imparted to the specimen. The impact velocity is a function of the drop height, but in most modern drop weight towers the impactor can also be accelerated using...
mechanical means to attain higher impact energies without increasing the size of the drop weight tower. The impactor is usually hemispherical in shape and contains an accelerometer and a loadcell which provide measures of load, velocity, acceleration, and displacement. A photocell provides an additional measure of the impact velocity and can be used to trigger data recording equipment during the impact event (for example, high speed cameras). Most drop weight towers are fitted with an anti-rebound system which prevents the impactor from repeatedly striking the specimen. Figure 2.17 provides a schematic of a typical drop weight tower.

![Schematic of a drop weight tower](image)

Upon impact, the kinetic energy of the impactor is transferred to the specimen. In a perfectly elastic impact where there are no losses to dissipation (through friction and damage), all of the impactor's energy is converted into elastic strain energy as the specimen deforms flexurally. When the impactor rebounds, all the energy is returned to the crosshead which returns to the same height it was dropped from. If the elastic limits of the specimen are exceeded, some of the impactor's energy will be "absorbed" during the impact. This absorbed energy is a combination of the energy lost to damage (delamination, fiber failure, matrix cracking etc.) and the energy lost to dissipation (e.g. frictional effects). The energy returned to the crosshead is equal to initial kinetic energy of the impactor minus the absorbed energy. This is best visualized by plotting force-displacement curves, in which the area under the curve represents the energy absorbed by a specimen during the impact event.

The mechanical response of the specimens is dominated by their bending
behavior. Since low velocity impacts do not typically penetrate the specimen, damage consists of matrix cracks, fiber failure, and delamination. As previously mentioned, the damage mode that has the greatest impact on the residual strength of the specimen is delamination [59]. While matrix cracking itself does not affect the residual properties of the laminate significantly, delamination is initiated by matrix cracking. These cracks form as a result of transverse tensile or shear stresses. As discussed in Section 2.3, transverse tensile matrix cracks propagate perpendicular to the tensile stress, while shear stress induced cracks form at an angle. In thick specimens, matrix cracking initiates in the uppermost ply due to high contact stresses. Damage then progresses downwards in a “pine-tree” pattern (Figure 2.18). This pattern is reversed in thin laminates, in which cracking initiates in the bottom ply due to bending stresses [59].

![Figure 2.18: Pine tree damage pattern observed in composite laminates subjected to low velocity impact.](image)

It is worth noting that, in the low velocity regime, although noise and vibrations may cause data analysis issues, the reflection of stress waves during the test is largely insignificant [82]. In fact, several studies have shown that the damage resulting from low velocity impacts can be replicated using a quasi-static indentation test provided the same maximum transverse force is used [83].

**Split-Hopkinson Bar**

While drop weight tower tests provide valuable insights into the formation of damage within a laminate, it is difficult to characterize the strain rate dependent stress-strain response of a material since instantaneous strain rates vary throughout the laminate during the impact (Figure 2.19). The response of a composite to high strain-rate loading (up to 1000/s) can be more easily characterized using a Split-Hopkinson bar (SHB).

A split-Hopkinson bar consists of three bars; the striker bar, the incident bar, and the transmission bar. The specimen is placed between the incident and
Figure 2.19: Contour plots of the strain rate in a CFRP specimen at an impact energy of 94 J. (a) t = 0.5 ms, before onset of damage and (b) t = 1.5 ms, after onset of damage [7]

The transmission bars. The incident and transmission bars are instrumented with strain gauges and a sensor is used to measure the impact velocity of the striker bar. A gas gun is used to accelerate the striker bar. Figure 2.20 provides a schematic of a typical SHB setup.

Figure 2.20: Schematic of a split-Hopkinson bar setup [66].

To conduct an SHB test, the striker bar is accelerated by the gas gun, impacting the incident bar at a velocity $V_0$. This generates an elastic compressive strain pulse that travels along the incident bar until it reaches the interface between the incident bar and the specimen. The magnitude of the strain pulse is measured by the strain gauge adhered to the incident bar. When the compressive stress wave reaches the interface, a portion of the wave is transmitted to the specimen, while the remainder is reflected back into the incident bar as an elastic tensile stress wave. The proportions of the transmitted and reflected waves are dependent on the change in impedance between the incident bar and the specimen. The transmitted wave bounces between the surfaces contacting the incident and transmission bars until the dynamic strength of the material is reached. A portion of the incident wave is transmitted through the specimen to the transmission bar where the pulse is measured by another set of strain gauges. Using the measurements of the incident, reflected, and transmitted waves in the bars, the dynamic material properties of the specimen can be ascertained.

A typical lab based SHB setup is sufficient for the characterization of
unidirectional composites and most textile composites, whose representative volume elements are relatively small. However, for composites with large unit cells, testing can only be conducted at facilities with large diameter bars that can generate extremely high stress pulses. There are only a handful of laboratories worldwide with these capabilities. In the present work, tests on large specimens were conducted at the European Laboratory for Structural Assessment (ELSA) HopLab, part of the European Commission’s Joint Research Center in Italy. Aside from the large and sophisticated equipment required, testing large specimens also places additional demands on the processing of the experimental data. Due to the larger distance between the ends of the incident and transmission bar, the time required for strain waves to travel through the specimen increases, and an equilibrated stress state may not be attained during dynamic loading [84]. If equilibrium cannot be established on the basis of the forces in the incident and transmission bar, an alternative approach is to compare the strains at the interfaces using digital image correlation, which captures the full strain field on the surface of a specimen. In addition to the ability to capture locally varying strains on the specimen surface, the use of DIC data may be preferable to data obtained using strain gauges as the latter has a tendency to overestimate strains when used in SHB experiments on composite materials [85]. For a more in depth discussion of the operation of an SHB, the reader is referred to the work of Körber [66].

2.4.3 Compression After Impact

While drop weight tower testing characterizes the capacity of a composite to absorb energy, it does not reveal anything about the residual structural integrity of the composite post-impact. In other words, drop weight tower testing characterizes the impact resistance, but not the impact tolerance of a composite [86].

Since in-plane compression is the critical loading condition for composites damaged by low velocity impacts, the most common method for determining the impact tolerance of a composite laminate is using a compression after impact (CAI) test. ASTM and Boeing developed a standardized method (ASTM D 7137) in which a coupon damaged in a standard drop weight tower test (or quasi static indentation experiment) is subjected to in-plane compressive loading [87].

In the test, a damaged coupon is installed in a fixture that prevents global buckling of the laminate, see Figure 2.21. Knife-edge supports on the longitudinal edges of the specimen restrain out-of-plane translation but do not restrain rotation. The top and bottom constraints provide a limited amount of rotational restraint as a result of the fixture geometry. The fixture is placed between parallel platens in a universal testing machine and loaded in compression until failure. The residual strength of the specimen can then be compared with the strength of the undamaged material from a standard
compressive test (according to ISO 14126 or one of the ASTM standards such as ASTM D 6641). The residual compressive strength of a composite specimen is particularly sensitive to its delamination footprint. Since delamination is the dominant damage mechanism in low velocity impact, the compression after impact strength of a composite is very much a function of the energy absorbed by the composite in a drop weight tower experiment [88].

2.5 Existing methods for impact tolerance improvement

Given the importance of impact tolerance in the design of composite structures, and the susceptibility of composites to this type of loading, it should come as no surprise that significant effort has been expended in industry and academia alike to try to improve the impact tolerance of FRPs. This section discusses a variety of the existing methods in detail.

2.5.1 Matrix toughening

Although the fibers in a composite carry most of the load, the matrix is crucial in defining the response of the composite to loading. A composite’s mode I and mode II interlaminar fracture toughness are primary determinants of its toughness and resistance to damage from out of plane impacts, and its residual strength after impact [89].

Thermoplastic matrices are inherently tougher than their thermoset counterparts, generally leading to improved performance under impact loading (Figure 2.22). However, the improvement in interlaminar fracture toughness generally comes at the cost of inferior fatigue performance and
increased matrix compliance, which reduce undamaged mechanical properties [91]. Furthermore, thermoplastics have a tendency to creep, are less thermally stable than thermosets, and may suffer from poor interfacial bonding, as indicated by the fact that the points corresponding to thermoplastic matrices lie below the dashed line in Figure 2.22 [92]. Lastly, the high viscosity of thermoplastics in the melt state complicates resin infusion and may necessitate high processing temperatures, increasing costs [91][93]. As a result, as of 2014, thermoplastics constitute only 24% of the total carbon fiber composites market by revenue [94].

The toughness of thermoset matrices can be improved through the addition of plasticizers, including rubber, nanoparticles, or carbon nanotubes, among others. One of the primary difficulties in the effective utilization of particle additives is obtaining a uniform dispersion of reinforcing particles. A uniform dispersion is important as the agglomeration of particles can significantly degrade the mechanical properties of a composite [90]. In addition, the inclusion of plasticizing particles in a matrix generally increases its viscosity, complicating resin infusion and increasing the difficulty of obtaining high volume fractions [95][92].

Aside from the properties of the matrix itself, the interphase region between the fibers and the matrix can also significantly impact a composite’s fracture toughness. Various fiber surface treatments can be used to tailor the bond strength between fiber and matrix [96]. Increasing the energy absorbed through debonding can increase toughness but tends to reduce the mechanical properties [92].
Arguably, the further improvements to a composite’s interlaminar fracture toughness that can be gained through the development of tougher matrices are likely to be marginal [22][97]. Satisfying the demand for improved impact tolerance is likely to necessitate the use of 3D reinforcement techniques.

### 2.5.2 Stacking sequence modification

The stacking sequence of a laminate has a significant effect on its properties. A laminate consisting of plies oriented in a single direction is unlikely to delaminate, but will fail at very low transverse loads due to splitting [49]. Laminates with multiple ply angles will not split, but the discrepancies between the bending stiffnesses of the plies create interlaminar shear stresses that can cause delamination. Liu found that the greater the interface angle (the angle between two neighboring plies) the greater the likelihood of delamination, and the larger its footprint [58].

In traditional laminates, plies of the same orientation are often grouped together. Following Liu’s findings, delamination at the interfaces between grouped plies is unlikely to occur as their mismatch angles are all equal to zero. As a result, high shear stresses develop between groups of plies and delamination at these interfaces is therefore likely to be significant in footprint. Lopes et al. investigated whether altering the stacking sequence and ply angles in a traditional laminate could improve their resistance to delamination [98]. By dispersing the stacking sequence - placing plies at a greater number of different angles - delamination is likely to occur at a larger number of interfaces, leading to more stable damage propagation and a reduced overall delamination footprint. Lopes et al. found no improvement in the performance of the dispersed ply stacking sequence laminates relative to traditional laminates. Computationally, dissipated impact energy was found to be approximately the same for both the baseline and dispersed stacking sequences, although as predicted the delaminations were spread over a larger number of interfaces in the dispersed ply laminate [99]. However, a later study by Sebaey et al. found that by reducing ply mismatch angles, significant reductions in delaminated area could be achieved, and residual compressive strengths could be improved by up to 20%, see Figure 2.23 [100].

The concept of ply-dispersion bears some similarity to the bio-mimetic laminate design strategy employed in laminates with Bouligand structures. Originally discovered in the mantis-shrimp’s dactyl club, Bouligand structures consist of a series of plies in a specifically helicoidal arrangement with very small (1.6° to 6.2°) inter-ply angles, see Figure 2.24. The small mismatch angles between plies result in the accumulation of sub-critical damage, in the form of a diffuse helicoidal delamination pattern [101]. Mencattelli studied the low velocity impact response of Bouligand laminates with ply interface angles ranging from 2.5° to 45° (i.e. a conventional quasi-isotropic laminate). The total projected delaminated area of the laminate with a 2.5° interface angle
was found to be 29.4% lower than a conventional laminate with a 45° interface angle. The compression after impact strength was found to be unaffected by the reduced ply mismatch angle [101]. In a different study by Ginzburg et al. laminates with a Bouligand layup exhibited improved impact tolerance performance relative to conventional laminates at high impact energies (80 J), but were found to be inferior at lower impact energies (40 J) [102]. The adoption of this novel laminate architecture for commercial applications has been limited, due in part to the complexity of the manufacturing process and limitations on ply thickness [102]. Since a large number of plies need to be stacked to produce the helicoidal pattern, this manufacturing method is only suitable for very thin plies, as the use of conventional plies will result in impractically thick laminates.

Figure 2.23: Delaminated areas of laminates with different mismatch angles at various impact energies. Adapted from [100].

Figure 2.24: Schematic illustration of a Bouligand structure [101]
Lastly, the angle of the surface ply in a laminate subjected to an out-of-plane impact also affects the impact tolerance of the laminate. Various studies have shown that laminates with surface plies at an angle of ±45° are more resistant to impacts than laminates with 0° surface plies [103][104]. However, the use of 45° surface plies has undesirable effects on the in-plane and flexural strength and stiffness of a laminate.

### 2.5.3 2D and 3D Weaving

Relative to unidirectional laminates, 2D woven composites are more resistant to impact and exhibit greater CAI strengths [89]. There are several reasons for wovens improved impact resistance. First, the increased compliance of woven laminates relative to conventional angle-ply composites allows for more energy to be absorbed through the structural response of the laminate, rather than damage formation [91]. Second, matrix rich regions are formed in areas where the fibers are crimped, allowing for highly localized plastic deformation which effectively delays the formation and propagation of transverse cracks. Third, when cracks and delaminations do occur, they tend to follow the undulating geometry of the fibers. The resultant more tortuous crack paths are longer than those formed between two non-undulating plies, and thus release more energy for the same delamination footprint [89]. Lastly, woven fabrics eliminate the maximum possible interface angle between two plies (90°), and since delamination extent is a function of the interface angle, this reduces the extent of delamination [97].

However, the improved impact response of woven laminates comes at a cost, namely: poor in-plane properties relative to unidirectional laminates. Due to the crimping of the fibers and the smaller proportion of fibers aligned with the loading direction, the strength of woven composites is less than that of an equivalent unidirectional laminate [91].

3D woven laminates contain binder yarns in the through thickness direction which arrest the growth of delamination cracks resulting from out-of-plane impact loading, improving impact tolerance. Furthermore, 3D weaving allows for the production of near-net shape preforms, reducing the need for machining and joining, and reducing material waste [72].

Compared with z-pinning (section 2.5.7) and stitching (Section 2.5.5), 3D weaving introduces only minimal damage to the fibers as the through thickness reinforcement does not need to be forced through the laminate. Consequently, the longitudinal moduli of 3D wovens are generally within 5-10% of the moduli of conventional 2D laminates (for typical z-binder contents of between 0-5%). In fact, if the z-binders yarns are under tension during the weaving process, they have a consolidating effect on the preform, increasing the in-plane fiber volume fraction, which in turn increases laminate stiffness [105]. In terms of strength, Mouritz et al. reported an approximately linear decrease in tensile strength with increasing z-binder content, see Figure 2.25. Experimental data on the compressive strength of 3D wovens is limited but suggests there is no
relationship between z-binder content and compressive strength [105].

Figure 2.25: Effect of z-binder content on the normalized strength of 3D woven laminates [105].

Another limitation of 3D weaving that is preventing its further adoption of 3D wovens is the fact that conventional looms cannot produce textiles that contain fiber orientations other than 0° and 90°. As a result, the properties of 3D wovens are often highly anisotropic, with poor resistance to torsion and shear [72].

2.5.4 Braiding

An alternative to 3D weaving for creating 3D preforms is braiding. Similar to 3D wovens, the through thickness fibers in braided composites help to arrest the formation and propagation of delamination, improving damage tolerance. In addition braided composites possess higher levels of drapability, and are more resistant to torsional loads than their 3D woven counterparts [72].

One of the disadvantages of braiding is the limited size of the preforms that can be economically manufactured. Most braiding machines are limited to cross sections under 100 mm in width, which is insufficient for many of the applications in the aerospace industry [106]. Furthermore, the braiding process itself is slow, partly because the requirement for the spools to move during the braiding process reduces their maximum size, limiting the length of a production run [72]. Lastly, the in-plane mechanical properties of braided composites cannot compete with those of conventional laminates due to their increased fiber waviness and generally lower volume fraction [72]. In summary, use of braiding is likely to be limited to the production of long
structural elements with relatively small cross-sections, such as drive shafts, I-beams, C-sections, and so forth [72].

2.5.5 Stitching

A stitched laminate is simply a traditionally manufactured laminate whose constituent plies have been joined by through thickness yarns. The effect of through thickness yarns is to improve the interlaminar fracture toughness of the composite. Laminates in either the preform or the prepreg state can be stitched, although the latter is more susceptible to be damaged by the stitching process [22]. Aside from the introduction of an additional production step, stitching does not require significant modifications to the manufacturing process. Moreover, the joining of preforms can ease handling prior to resin infusion and curing [106]. Unidirectional dry fiber plies that are stacked in different orientations and stitched are also known as non-crimp fabrics (NCFs).

The effects of stitching on the mechanical properties of composite laminates have been extensively reviewed by Mouritz et al. and Dransfield et al. [106][22]. Depending on the type of reinforcing yarns, and the state of the laminate (preform or prepreg), stitching can increase a laminate’s mode I fracture toughness by as much as 75-90%. The mechanism which brings about this improvement is primarily the bridging of interfacial cracks by the stitching yarns, which reduces the stress concentrations around a crack tip, ultimately delaying the propagation of delamination. The increase in mode I fracture energy is approximately linearly related to the stitch density [107]. In mode II, Jain et al. reported an increase in fracture toughness of between 80-500% for carbon epoxy laminates stitched with Kevlar threads [108]. Tan et al. compared the low velocity impact response and residual compressive strength of stitched an un-stitched laminates, reporting a reduction in delaminated area of up to 40%, and improvements in CAI strength of up to 20% [109][110][111].

However, the introduction of through thickness yarns negatively affects the undamaged in-plane properties of a laminate. The stitching process causes localized fiber breakage, increases fiber waviness, and creates resin rich zones, reducing both the stiffness and strength of a laminate. Reductions in tensile and compressive strength vary by fiber type and the laminate configuration, but may be as high as 20-30% and 5-55% respectively [106].

2.5.6 Knitting

Knitting can be used in a similar fashion to 3D weaving to produce near net-shape preforms, or as an alternative to stitching to produce NCFs. Knitted NCFs are similar in structure to stitched NCFs, and possess many of the same qualities; improved impact tolerance and improved processability. Predictably,
they are also hampered by the same shortcoming: poor in-plane mechanical properties resulting from increased fiber crimp [72].

The reported fracture toughness properties of knitted near net shape preforms are impressive - ranging from 2-10x relative to conventional laminates [112][113]. It should be noted that the aforementioned studies all used glass fiber composites, and the improvement in fracture toughness for CFRPs is likely to be more limited [114]. Subjected to low-velocity impacts, knitted laminates exhibit smaller delaminated areas and superior residual compressive strengths (Figure 2.26)[115].

Despite their excellent response to out-of-plane loading, the in-plane stiffness and strength of knitted composites are inferior to those of woven, braided, and unidirectional composites. The tight radii that the fibers must conform to result in strengths and stiffnesses closer to those of random fiber mats than conventional angle ply laminates [114]. As such these composites are best suited for applications in which toughness is more critical than in-plane strength and stiffness.

2.5.7 Z-pinning

Z-pinning involves the insertion of reinforcement pins (carbon fiber, steel or titanium) into a laminate using an ultrasonic gun. In contrast to weaving and braiding, z-pinning can be used to reinforce prepreg laminates and allows laminates to be cured using conventional methods e.g. vacuum bagging or autoclaving [116]. The effects of z-pinning have been extensively reviewed by Mouritz et al. [116][117].

Various authors have investigated the effect of z-pinning on the interlaminar fracture toughness of a laminate [118][119][120]. Cartie et al.
Figure 2.27: Effect of z-pin content on the residual compressive strength and delaminated area of composites subjected to impact loading [117].

reported a doubling of the mode I fracture toughness for every 0.5% increase in z-pin content [118]. The improvement was attributed to the bridging of interfacial cracks by the z-pins. The additional work required to debond and pull-out the z-pins increased the energy required for crack propagation. In mode II, shear induced z-pin pullout, and increased friction in delamination zones from deflected z-pins (an effect known as snubbing) lead to increased fracture toughness [119]. Experimental studies on the response of z-pinned laminates to low-velocity impacts report reductions in delaminated area of 19-64%, and improvements in residual compressive strength see Figure 2.27 [121][117].

However, the insertion of pins into a prepreg lamina does have deleterious effects on the lamina’s in-plane properties. Fiber breakage and increased fiber waviness near the pins, as illustrated in Figure 2.28, may result in lower stiffnesses and strengths relative to unpinned laminates. The reductions in strength and stiffness are functions of both the z-pin content, and the diameter of the pins themselves [116]. However, z-pinning is an expensive procedure, and despite numerous demonstration projects it has seen limited use in commercial environments [117].

2.5.8 Interleaves/Interlayers

Interleaving refers to the inclusion of thin thermoplastic films, or short fiber “veils”, at the interfaces between composite plies. Interleaves allow for increased plastic deformation between plies, and the short fibers improve fracture toughness through fiber bridging [95][122]. Disadvantages of interleaving are the potential reductions in fiber volume fraction, and reduction
in specific mechanical properties resulting from the additional weight but low
stiffness and strength of the interleaves [90].

Beylergil et al. conducted an experimental study on the effect of
non-woven poly-amide veils on the mechanical performance of carbon-epoxy
laminates, the laminate configuration is illustrated in Figure 2.29. The
polyamide veils were found to increase the mode I initiation and propagation
fracture toughness by 350% and 720% respectively [8]. However, the
reduction in fiber volume fraction resulting from interleaving negatively
impacted the tensile, compressive, and flexural strength of the composite.
Research on extremely thin nanofiber interleaves (thermoplastic or carbon
nanotubes) indicate significant improvements in mode I and mode II fracture
toughness with minimal reduction in mechanical properties or areal density
[123][124].

In spite of the efficacy of these inter layers in improving interlaminar fracture
toughness values, the use of these interleaves and veils is not easily combined
with the high rate automated manufacturing processes used in the aerospace
industries [125].

2.5.9 Hierarchical composites

Bio-mimetic composites emulate the hierarchical structures observed in
biological composites, such as nacre, to improve fracture toughness.
Testament to the importance of intelligent structural design over base material
properties, nacre is up to 20-30 times stronger and three orders of magnitude
tougher than its primary constituent, argonite [126][127].

The structure of nacre can be replicated in composites by cutting plies to
form a tiled or "brick-and-mortar" laminate (Figure 2.30). Cutting the plies
results in the creation of resin-rich regions, which can be used to "guide" the
Figure 2.29: Schematic illustration of unidirectional carbon fiber epoxy laminate interleaved with non-woven polyamide veils. Adapted from [8]

path of a crack through a laminate. In deflecting the cracks in a tortuous path around the composite "bricks" brittle failure of the composite is avoided [127]. Bullegas et al. used laser micro-milling to create a pattern of microcuts in a cross-ply laminate to encourage fiber bundle pullout. As illustrated in Figure [2.31] the introduction of these fiber discontinuities alters the fracture surface completely, and results in an increase in the intralaminar fracture toughness by up to 214% [128].

Naturally, cutting through the continuous fibers in a composite will necessarily reduce its mechanical properties. For CFRPs, the introduction of ply-cuts results in a loss in tensile strength of approximately 12-15% [126][128]. As such, the use of these composites is best suited to applications where ductility is of greater importance than strength and stiffness. Furthermore, the cost and time required to create these hierarchical structures is a barrier to their adoption in the aerospace industry.
2.6 Numerical Modeling of Composite Materials

Historically, the assessment of the structural integrity of composite structures has relied heavily on experimental testing. Up to 10,000 tests, at length scales ranging from the coupon level to entire fuselages, are required for the certification of an airframe structure \[129\]. As such, accreditation of an airframe is an extremely costly and time consuming process. More recently, the development of sophisticated material models, and the dramatic increase in computational power available (at relatively low cost) to academic and industrial engineers, have made it possible to accurately predict the behavior of composite components using analytical and numerical models. By replacing some of the experimental tests with virtual tests using validated numerical models it is possible to reduce the cost of certification \[130\] [131] [132]. In addition, sophisticated numerical models can provide engineers with a deeper understanding of the initiation, propagation, and interaction of the different damage modes occurring in the material, helping to guide the design process.

In the numerical simulation of composites at the coupon level, laminates are typically divided into individual laminae. Intralaminar and interlaminar damage can then be treated separately. The following sections briefly discuss the methods employed in this study to capture these two damage mechanisms.
2.6.1 Unidirectional Lamina Failure

The accurate prediction of the strength of a composite laminae is highly dependent on the accurate prediction of damage initiation in its constituent laminae. A large number of failure theories have been put forward over the years. The Tsai-Hill theory, one of the first criteria developed specifically for composite materials, is a quadratic interaction criterion in which failure of the laminae occurs when the value of Equation 2.1 exceeds unity.

\[
\left( \frac{\sigma_{11}}{X^T} \right)^2 - \left( \frac{\sigma_{11} \sigma_{22}}{X^T Y^T} \right)^2 + \left( \frac{\sigma_{22}}{Y^T} \right)^2 + \left( \frac{\sigma_{12}}{S_L} \right)^2 \geq 1
\]  

(2.1)

Where \( X^T, Y^T \), and \( S_L \) represent the fiber, transverse, and shear strengths of the lamina respectively. Although reasonably accurate at predicting tensile failure of unidirectional laminae at various off axis angles, the coupling of failure in the fiber and matrix directions leads to inaccurate predictions of laminae strength in more complex loadcases [54]. Moreover, the criteria does not provide an indication of the mode of failure. Tsai and Wu later developed their eponymous failure criteria, proposing a polynomial function of the stress components where the coefficients are related to the strengths of the material. Similar to the Tsai-Hill criteria, the Tsai-Wu model gives no indication of the failure mode, and couples fiber and matrix direction damage initiation. Hashin was the first to propose a criteria with independent fiber and matrix direction failure modes. Fiber failure is assumed to be independent of the transverse normal stresses, and vice versa for matrix failure. While originally developed to predict failure in two-dimensions, the criteria can be easily extended to predict through thickness failure [133]. The three dimensional Hashin failure criteria are given by Equations 2.2 - 2.7. \( X^T, X^C, Y^T, Y^C, Z^T, \) and \( Z^C \) denote the tensile and compressive strengths in the longitudinal, transverse, and through thickness directions respectively, and \( S_L, S^T, \) and \( S^R \) represent the shear strength in the 12, 23, and 13 directions.
\[ F_{1T} = \left( \frac{\sigma_{11}}{X^T} \right)^2 + \alpha \left( \frac{\sigma_{12}}{S^L} \right)^2 \geq 1 \]  
(2.2)

\[ F_{1C} = \left( \frac{\sigma_{11}}{X^C} \right)^2 \geq 1 \]  
(2.3)

\[ F_{2T} = \left( \frac{\sigma_{22}}{Y^T} \right)^2 + \alpha \left( \frac{\sigma_{12}}{S^L} \right)^2 \geq 1 \]  
(2.4)

\[ F_{2C} = \left( \frac{\sigma_{22}}{2S^T} \right)^2 + \left( \frac{Y^C}{2S^T} \right)^2 - 1 \left[ \frac{\sigma_{22}}{Y^C} + \left( \frac{\sigma_{12}}{S^L} \right)^2 \right] \geq 1 \]  
(2.5)

\[ F_{3T} = \left( \frac{\sigma_{33}}{Z^T} \right)^2 + \alpha \left( \frac{\sigma_{12}}{S^R} \right)^2 \geq 1 \]  
(2.6)

\[ F_{3C} = \left( \frac{\sigma_{33}}{2S^T} \right)^2 + \left( \frac{Z^C}{2S^T} \right)^2 - 1 \left[ \frac{\sigma_{33}}{Z^C} + \left( \frac{\sigma_{13}}{S^R} \right)^2 \right] \geq 1 \]  
(2.7)

More recent developments include the LaRC failure criteria developed by Pinho et al. which builds on the work of Puck and Schürmann, accounting not only for the different failure modes, but also the inclination of the cracking plane under transverse loading [134][52]. A further development of the LaRC criteria by Catalanotti et al. accounts for the in-situ effect further discussed in Section 2.3.2[51]. While the LaRC criteria achieved the most promising results in the World Wide Failure Exercise [135], the required material parameters, e.g. the shear friction angle and the fracture angle under pure transverse compression, can be difficult to obtain for non-unidirectional laminae. For example, the effect of the out-of-plane undulations occurring in AP-PLY laminates on the shear friction angle cannot be easily determined experimentally, and is difficult to estimate numerically. In addition, the determination of the failure plane inclination can be computationally expensive if determined using brute force algorithms [136]. More sophisticated root finding algorithms to determine the failure plane inclination have been developed by various researchers, but are not straightforward to implement. Moreover, the use of the LaRC failure criteria in explicit finite element analyses can lead to numerical instabilities [137][138]. As such, in this study, an adaptation of the 3D Hashin failure criteria are used. A more in-depth discussion on the implementation of these criteria is provided in Section 4.4.

### 2.6.2 Composite Laminate Failure

The failure of a single lamina rarely results in the catastrophic failure of a multi-directional laminate. Intralay cracks in a lamina are constrained by neighboring plies, stabilizing their growth and prolonging ultimate failure of the laminate. To predict the ultimate strength of a laminate, models must account for the formation and propagation of these intralaminar cracks, as well as the cracks forming along the interfaces between plies.
Typically, intralaminar and interlaminar failure are treated separately. This is because the locations of the cracking planes for interlaminar fracture are known \textit{a priori} (the interfaces between plies), while intralaminar cracks could occur at any location within a ply. As such, computationally efficient discrete damage mechanics (DDM) frameworks e.g. cohesive damage laws, can be used to simulate delamination, while continuum damage mechanics (CDM) or embedded discontinuity methods, e.g. the eXtended Finite Element Method (XFEM), are used for intraply failure.

It should be noted that by treating intra and inter-ply damage mechanisms separately, the interaction between intraply cracks and delamination cannot be captured. In other words, matrix cracks do not lead to delamination and delamination does not promote matrix cracking. Various authors have sought to address this issue. Mami \textit{et al.} developed a fully three-dimensional CDM model capable of capturing both delamination and intra-ply failure \cite{139}. While the results of the study were promising, the approach necessitates the use of a very fine mesh if the ply interfaces are to be captured effectively, increasing computational cost. Alternatively, Wisnom \textit{et al.} made use of cohesive zone models to capture both intra and interply failure, but this approach requires cohesive zones or elements to be placed in every possible crack location, an intractable task if the dominant crack paths are unknown \textit{a priori}. The fact is that dynamic crack growth cannot be replicated by the conventional cohesive elements/interfaces implemented in Abaqus. While cohesives might be sufficient for quasi-static analyses, under dynamic loading, extrinsic finite elements with an initial rigid response are needed. Considering this, it is better to use CDM for dynamic analysis and minimise the number of non-rigid cohesive interfaces in your model.

While it is important to acknowledge that the separate treatment of delamination and intraply failure has inherent limitations, the approach has nonetheless been used to good effect in a wide variety of studies\cite{99} \cite{111} \cite{140} \cite{141}. A brief introduction to these topics is provided below.

\section*{Continuum Damage Mechanics for Intralaminar Failure}

In CDM, damage is assumed to take the form of micro-cracks, which reduce the load bearing cross-sectional area of a volume of material from $A$, to $\tilde{A}$, see Figure \ref{fig:2.32}. The extent of damage is quantified using a scalar variable $d$, which can be expressed as a function of the undamaged and damaged load bearing areas (Equation \ref{eq:2.8}). The damage variable is equal to zero in the undamaged state, and one at complete failure.

\begin{equation}
(1 - d) = \frac{\tilde{A}}{A}
\end{equation}

Various laws have been proposed in the literature governing the evolution of damage. These laws are generally defined as a function of the fracture
energy of the material, a topic which is discussed in more detail in Section 4.4. Regardless of type of evolution law used, the "effective" stress in the material $\sigma$ (i.e. the stress in the damaged state) can consequently be related to the stress in an equivalent undamaged volume, $\tilde{\sigma}$.

$$\sigma = (1 - d) \tilde{\sigma} \quad (2.9)$$

Through strain equivalence, the stress in the undamaged and damaged states is equal, $\varepsilon = \tilde{\varepsilon}$. Assuming a linear relationship between stress and strain, Equation 2.9 can be rewritten as:

$$\sigma = (1 - d) E \varepsilon \quad (2.10)$$

The principle can be easily extended from one to three dimensions. Assuming a transversely isotropic material we can define a compliance tensor $H$ as a function of six damage variables ($d_N, N = 1, ..., 6$), which define the extent of damage in the three normal and three shear directions.

$$\begin{bmatrix}
\varepsilon_{11} \\
\varepsilon_{22} \\
\varepsilon_{33} \\
\gamma_{13} \\
\gamma_{23} \\
\gamma_{12}
\end{bmatrix} = \begin{bmatrix}
\frac{1}{(1-d_1)E_1} & -\frac{\nu_{12}}{E_1} & -\frac{\nu_{13}}{E_1} & 0 & 0 & 0 \\
-\frac{\nu_{21}}{E_2} & \frac{1}{(1-d_2)E_2} & -\frac{\nu_{23}}{E_2} & 0 & 0 & 0 \\
-\frac{\nu_{31}}{E_3} & -\frac{\nu_{32}}{E_3} & \frac{1}{(1-d_3)E_3} & 0 & 0 & 0 \\
0 & 0 & 0 & \frac{1}{(1-d_4)G_{13}} & 0 & 0 \\
0 & 0 & 0 & 0 & \frac{1}{(1-d_5)G_{23}} & 0 \\
0 & 0 & 0 & 0 & 0 & \frac{1}{(1-d_6)G_{12}}
\end{bmatrix} \begin{bmatrix}
\sigma_{11} \\
\sigma_{22} \\
\sigma_{33} \\
\tau_{13} \\
\tau_{23} \\
\tau_{12}
\end{bmatrix} \quad (2.11)$$

Where $E_1$, $E_2$, $E_3$, $G_{12}$, $G_{13}$, and $G_{23}$ are the elastic constants of the material, and $\nu_{12}$, $\nu_{13}$, $\nu_{23}$ are the Poisson’s ratios.

The principles of CDM were first proposed by Kachanov et al., then extended to anisotropic materials (reinforced concrete) by Bazant & Planas,
and finally adapted for composite materials by Matzenmiller et al. \[142\][143][53]. Since then, the CDM approach has been further developed in a wide variety of studies including, the work of Chang & Chang, the LaRC04 based model presented by Maimi et al., and the model developed by Tan et al. which also accounts for the transverse matrix cracking plane inclination \[144\][145][146][111]. For the full details of the CDM implementation used in this study, see Section 4.4.

2.6.3 Cohesive Zone Models for Interlaminar Failure

There are two main approaches to the simulation of delamination; the virtual crack closure technique (VCCT), and the cohesive zone method (CZM). Although both approaches produce similar results, the CZM is easier to implement and more computationally efficient \[147\]. In the present work the CZM has been used to model interlaminar failure.

In a cohesive damage model the interfacial constitutive laws are defined as functions of damage variables, analogous to those in a continuum damage mechanics framework (see Section 2.6.2). The stiffness of the interface between the two separating surfaces is progressively degraded as damage develops. In Abaqus, cohesive interfaces can be implemented using finite thickness cohesive elements, or using cohesive surfaces with zero thickness. Cohesive elements are typically used where the adhesive interface has a non-negligible thickness, and where the bulk properties of the adhesive are known \[148\]. Surface based cohesive zones operate in a similar fashion to cohesive elements but do not require element definitions and are therefore somewhat easier to implement. Furthermore, cohesive surfaces allow the post-debonding behavior of the cohesive interface to be defined, e.g. frictional contact between the delaminated surfaces. Since the thickness of the interface between adjacent plies in a composite laminate is very small, interlaminar failure was captured using surface based cohesive damage modeling.

The contact stresses on the crack plane can be defined as \( t_N, t_S, t_T \), representing the normal traction and the traction in the first and second shear directions. The corresponding displacements or contact separations are \( \delta_N, \delta_S, \delta_T \). In the elastic regime, i.e. prior to damage initiation, the traction vector on the cohesive surface is related to the separation vector through the cohesive stiffness matrix (Equation 2.12).

\[
\begin{bmatrix}
  t_N \\
  t_S \\
  t_T 
\end{bmatrix} =
\begin{bmatrix}
  K_{nn} & K_{ns} & K_{nt} \\
  K_{ns} & K_{ss} & K_{st} \\
  K_{nt} & K_{st} & K_{tt}
\end{bmatrix}
\begin{bmatrix}
  \delta_n \\
  \delta_s \\
  \delta_t
\end{bmatrix} = K\delta
\]

(2.12)

Once a damage initiation criteria is met, the stiffness of the interface is degraded according to either a linear or exponential damage evolution law. The area under the traction-separation curve is equal to the energy dissipated by
the damage process i.e. the interfacial fracture energy of the material (Figure 2.33).

Damage initiation can be predicted by a maximum stress or separation criterion, or using either a stress or separation based quadratic interaction criterion. The most common approach is to use a stress (i.e. traction) based quadratic interaction criterion, see Equation 2.13. Note that compressive normal tractions close the crack and therefore do not contribute to the initiation criteria.

\[
\left\{ \frac{t_n}{t_n^0} \right\}^2 + \left\{ \frac{t_s}{t_s^0} \right\}^2 + \left\{ \frac{t_t}{t_t^0} \right\}^2 = 1
\]  

(2.13)

Once damage at the contact point is triggered, a damage variable \(d\) defines the degradation of the contact stresses (Equation 2.14). The value of \(d\) ranges from zero to one with the latter representing a completely debonded interface.

\[
t_n = \begin{cases} (1 - d) \tilde{t}_n, & \tilde{t}_n \geq 0 \\ \tilde{t}_n, & \text{otherwise} \end{cases}
\]

\[
t_s = (1 - d) \tilde{t}_s \\
t_t = (1 - d) \tilde{t}_t
\]  

(2.14)

Where \(\tilde{t}_n, \tilde{t}_s, \) and \(\tilde{t}_t\) represent the tractions calculated assuming no damage, i.e. using a linear elastic traction-separation law. Note that damage does not affect the transfer of compressive normal tractions because the cracks close under the application of compressive forces. When an interface is subjected to a combination of normal and shear stresses, the debonding process is said to be "mixed-mode". In this case damage evolution must be defined as a function of the mode-mixity. It is useful to define the effective separation between the two surfaces according to Equation 2.15.

\[
\delta_m = \sqrt{\langle \delta_n \rangle^2 + \delta_s^2 + \delta_t^2}
\]  

(2.15)
Figure 2.34 illustrates the traction-separation response under mixed mode loading. Pure normal and pure shear separation are represented by the unshaded triangles. The fracture energy under mixed-mode conditions, $G_c$, can be defined in a tabular fashion or using analytical expressions. For composite materials, in which the critical fracture energies in the two shear directions are generally equal, the Benzeggagh and Kenane (BK) fracture criterion is mostly commonly used (Equation 2.16).

\[
G_c = G_{Ic} + (G_{IIc} - G_{Ic}) \left( \frac{G_{III} + G_{I}}{G_{II} + G_{I} + G_{III}} \right) \eta
\]

(2.16)

Where $G_{Ic}$ and $G_{IIc}$ represent the mode I and mode II critical fracture energies, $G_I$, $G_{II}$, and $G_{III}$ are the current fracture energies in modes I, II, and III as calculated by the FEA code, and $\eta$ is the mixed mode interaction parameter.

2.6.4 Multiscale

Multiscale simulation is a modeling framework in which the stress-strain response of a structure is predicted by modeling material behavior at multiple length scales. Historically, the methodology has been applied using a global-to-local approach. Critical regions are first identified using a model of an entire structure, and refined analyses of these local features are used to capture the initiation and propagation of damage [9].

More recently, multiscale frameworks have been developed that operate from the bottom up [9]. The process is illustrated in Figure 2.35.
up approach relies on the exploitation of periodicity and symmetry in the topology of the structure to be modeled. First, a structure is broken down into unit cells, also known as representative volume elements. A unit cell is the smallest volume whose properties are representative of the larger structure it is a feature of. Next, the stress-strain response of a unit cell is simulated. Since a unit cell is relatively small in size this can be done using high fidelity models which are able to capture local damage mechanisms. Finally, the results are homogenized to define a constitutive model, and this information is passed to the next length scale. The process can be repeated to bridge multiple length scales. The bottom up multiscale approach has been used to good effect in a large number of studies on composite materials, including conventional angle ply laminates [9, 151, 152, 153], and textile composites [133, 154, 155, 156].

A subset of multiscale models are FE² models, in which simulations at multiple length scales run concurrently rather than sequentially [157]. This approach can provide improved accuracy but requires significantly more computation time since the simulations at each length scale are interdependent [158, 159].

### 2.7 Advanced Ply Placement

The preceding sections have introduced advanced composite materials, the methods by which they are manufactured, their susceptibility to damage from low velocity out-of-plane impacts, and the existing methods that have been developed to improve their impact tolerance. Undoubtedly, existing methods such as z-pinning and weaving improve the interlaminar strength and impact
tolerance of FRPs. However, as the previous sections have discussed, these reinforcement techniques negatively impact the undamaged mechanical properties of a laminate, and are difficult to combine with automated manufacturing techniques such as AFP and ATL.

An alternative manufacturing process, which better harnesses the fiber placement flexibility and precision of AFP machines, is known as advanced ply placement (AP-PLY). In an AP-PLY laminate, tapes are placed in a predefined pattern to create interlayer fiber connectivity, as illustrated in Figure 2.36. Instead of placing tapes to form complete plies, as in conventional laminates, the AFP head is modified to place tapes in sets. A tape set is a group of tapes with the same fiber orientation, in which each tape is separated from the next by a predefined gap. The size of the gap can be varied but must always be a multiple of the tape width. The gaps left between tapes in a set are later “filled in” by another tape set, but only after tape sets have been placed in all the other desired fiber orientations first.

Figure 2.36: Schematic illustration of the layup process for one type of quasi-isotropic AP-PLY composite laminate.
The first investigation into the AP-PLY preforming method was conducted by Nagelsmit who experimentally characterized the mechanical properties and impact response of various types of what they called Advanced Placed Ply (AP-PLY) laminates [10]. AP-PLY laminates and their baseline counterparts were manufactured using an AFP machine. The effect of tow width (1/4 inch and 1/2 inch), matrix type (thermoset and thermoplastic), and the AP-PLY configuration on the properties of the laminates was studied. All laminates were either cross-ply or quasi-isotropic by nature but the tow placement sequence (defined as "series", "alternating", or "totally interwoven"), and the tow spacing parameter were varied.

AP-PLY preforming was found to improve the mode I interlaminar fracture toughness by a significant margin of 89.2% relative to a non-AP-PLY baseline. The improvement in mode II fracture toughness was less impressive but nonetheless still significant at 20.0%. Unfortunately, Nagelsmit does not specify the configurations of the AP-PLY laminates that were tested. In addition to characterizing the interlaminar shear performance of the AP-PLY laminates, Nagelsmit also conducted short block and open hole compression tests on AP-PLY and baseline laminates. In the short block compression tests, the modulus of the AP-PLY laminates was found to be equal to or greater than the modulus of the baseline laminates, in spite of the presence of fiber undulations. In terms of strength, the average failure stress of the AP-PLY laminates was somewhat lower (-4.16%) than the average baseline laminate strength, likely due to the increased misalignment of the fibers in the AP-PLY specimens. However, in the open hole compression tests, the AP-PLY architecture was found to positively impact laminate strength, resulting in an increase in failure stress of 0.6% relative to the baseline.

The impact response of the AP-PLY laminates was studied using drop-weight tower tests. In general, AP-PLY laminates exhibited smaller delamination footprints than their baseline counterparts. Nagelsmit suggests that the reason for this improvement is the increased dissipation of impact energy through fiber failure and matrix cracking rather than delamination. In subsequent CAI testing, the reduced delamination footprint was found to positively affect residual strength. The AP-PLY laminates possessed residual strengths up to 15% higher than those of the baseline laminates. Specimens with small tow widths, thermoplastic matrices, and a "totally interwoven" AP-PLY configuration exhibited the greatest residual strength.

After the publication of Nagelsmit’s thesis, a number of research groups experimentally characterized AP-PLY laminates. Hoang et al. characterized the in-plane tensile and compressive behavior of two different quasi-isotropic AP-PLY laminates. In the first configuration, type 1, 45°/-45° tows and 0°/90° tows were AP-PLY separately, in sets of two. In the type 2 laminates, all the tow orientations (45°/-45°/0°/90°) were AP-PLY. Hoang et al. found that the modulus and strength of laminates of type 1 were within 1.5% of the properties of a conventional laminate, in both tension and compression. Type 2 laminates, due to their increased fiber waviness, exhibited marginally lower
tensile stiffness (5%), tensile strength (7%), compressive strength (4%), and compressive stiffness (4%), compared with the baseline laminate [24].

In the work of Waddington et al. cross-ply AP-PLY preforms were produced using robotic dry fiber placement and subjected to low-velocity impact and compression after impact testing [25]. While there are differences between AFP and robotic dry fiber placement, the internal architecture of the resulting laminates is the same. Waddington et al. found that the extent of delamination resulting from low velocity impacts was significantly reduced as a result of the AP-PLY architecture. The observed circularity of the delamination footprint suggested to the authors that the AP-PLY architecture transfers loads in a similar manner to woven preforms. In contrast to the results presented by Nagelsmit, Waddington et al. found that AP-PLY preforming did not result in an increase in residual strength. The authors suggested that the reduction in delaminated area was offset by the increased fiber waviness.

Zivkovic et al. studied the mechanical properties of a 28 ply quasi-isotropic AP-PLY laminate and compared them with a baseline non-AP-PLY laminate. They evaluated the tensile, compressive, flexural, and interlaminar behaviour of the two laminates according to ASTM standards. The AP-PLY architecture was found to have a negative effect on the tensile and flexural strength, and mode I fracture toughness (reductions of 3.1%, 15.1% and 14.9% relative to the baseline laminate, respectively). The compressive strength and mode II fracture toughness were improved by 3.2% and 9.9% respectively [26].

Rad et al. investigated the thermal warpage and tensile response of a number of different AP-PLY preforms [160]. Their results show that the curing induced thermal warpage of asymmetric AP-PLY laminates is significantly less than comparable asymmetric conventional laminates. In addition, the AP-PLY laminates were found to exhibit tensile properties comparable to those of traditional non-AP-PLY laminates.

In subsequent studies by Rad et al. the impact performance of hybrid AP-PLY/conventional angle-ply laminates was investigated [160] [23]. Two hybrid AP-PLY configurations and one non-AP-PLY control panel, all quasi-isotropic and containing 24 plies, were subjected to both low and high velocity impacts. In one type of hybrid laminate the outer 8 plies were AP-PLY, and in the other the inner 8 plies were AP-PLY. In the low velocity impacts, the configuration in which the AP-PLY plies were on the outside of the laminate absorbed 15-20% more energy than its non-AP-PLY equivalent, and exhibited a smaller damaged area. In the high velocity experiments, the configuration with external AP-PLY plies absorbed the most energy, and exhibited the least rear surface damage. The configuration with internal AP-PLY layers was notable for its increased penetration resistance.

While much of the existing literature has been primarily experimental in focus, a number of studies have endeavored to numerically or analytically predict AP-PLY laminate behavior. Nagelsmit developed an analytical model based on laminated plate theory and fracture mechanics to determine the
interfacial stresses between plies. By determining which plies had the greatest tendency to delaminate, Nagelsmit was able to optimize the application of the AP-PLY process, reinforcing only those interfaces most susceptible to interfacial failure. The resulting hybrid AP-PLY layup configuration was found to exhibit a higher residual strength than a fully AP-PLY laminate while requiring less time to manufacture. Zheng et al. developed a more sophisticated method for the prediction of delamination in AP-PLY composites using 2D finite element analysis [27]. An idealized AP-PLY laminate cross-section, illustrated in Figure 2.37 was subjected to 3-point bending, and the onset and propagation of delamination were captured using the cohesive zone method. Zheng et al. state that the numerical results were in reasonable agreement with experimental observations, but acknowledge that the quantity of experimental data available was limited. Moreover, the high computational cost of the approach (in excess of 60 hours per simulation) limits the utility of the approach.

A later study by the same author used an approach based on the strain energy release rate of a crack to determine its propagation through an AP-PLY composite [28]. In doing so, the authors were able to account for the effect of varying stiffness in AP-PLY composites on delamination growth. A genetic algorithm was subsequently used to determine an optimal laminate configuration that met predefined stiffness requirements while delaying the onset of delamination.

Rad et al. were the first to develop a model to predict the response of AP-PLY composites to in-plane loading [160]. They simulated tensile tests on various AP-PLY specimens using 3D shell elements and Tsai-Hill failure criteria. A script was used to assign ply orientations to each region, as illustrated in Figure 2.38. The numerical model exhibited a tendency to overestimate laminate stiffness by approximately 9-23 %. This is likely

Figure 2.37: Schematic of the 2D geometry for the delamination model developed by Zheng et al.. Adapted from [27].
because the model does not account for the undulation of the fibers in the AP-PLY laminates. It is worth noting that the study did not numerically model laminate strength. The Tsai-Hill criteria determine failure initiation at the lamina level, however, without specifying damage evolution the model is not capable of estimating laminate failure. Rad et al. also used the model to determine warping due to residual stresses, results which were in good agreement with experimental data.

Figure 2.38: Illustration of regional stacking sequences in a quasi-isotropic AP-PLY laminate. Possible stacking sequences for each region of the laminate are depicted on the right [19].

More recently, Li et al. have developed a program to generate 3D geometries for AP-PLY laminates [29]. The approach, dubbed tow wise modeling (TWM) models each tow in the laminate individually. The program takes the start and end point of a tow path (outputs of the software used to operate the AFP head) and discretizes the tow, i.e. creates a mesh of the tow geometry. Next, each tow is added to the global mesh, and an algorithm is used to detect where the tow intersects with previously placed tows. If tow intersections are detected, the relevant nodes are shifted in the z-direction, creating the through thickness tow undulations. The process is illustrated in Figure 2.39.

The resultant geometries are in good agreement with microscopy images of a number of laminate cross-sections. However, there are a few limitations to the program that the authors acknowledge. First, the algorithm sometimes generates geometries with small tow intersections, an issue common in textile geometry creation software. Mesh refinement will eliminate these issues, but the added computational cost may be significant. Marginal reductions in element size may lead to relatively large increases in computation time, as the relationship between mesh size and elapsed processing time is on the order of $O(n^{1.5})$. Aside from the interpenetration issue, the current methodology assumes the fiber path between layer is linear, where in reality it is more accurately represented using a quarter sine wave. The authors state that future work will focus on using a Bezier spline to define the curved fiber path of tow undulations.
Figure 2.39: The AP-PLY laminate geometry creation process using the tow-wise modeling approach developed by Li et al. [29].
Chapter 3

AP-PLY Composites

3.1 Material Design Parameters

Since the AP-PLY process does not require any modifications be made to an AFP machine, any fiber-matrix combination that can be used to produce conventional AFP laminates can also be used to manufacture an AP-PLY laminate. Following is a discussion of the material specific parameters that can be varied in the design of an AP-PLY laminate.

**Tow Width**

Tow width is typically limited by the capabilities of an AFP machine which tend to be designed to operate using a specific tow width. Most commercial AFP tows are available in either 1/8", 1/4", or 1/2" widths [161].

**Tow Thickness**

Manufacturers of AFP tows produce different thicknesses to meet a variety of requirements. The cured ply thickness is also affected by the processing methods used to produce a laminate. Higher curing pressures generally result in lower cured ply thicknesses. Generally, cured ply thicknesses for AFP tows range from 0.1 mm to 0.35 mm.

**Tow Constituents**

The most common materials used in AFP manufacturing are thermosetting prepregs. Due to their low gel/melt point relatively little heating power is required from the AFP head. This, in combination with their low viscosity, makes for easier processing [162]. The disadvantages of thermoset resins — for AFP specifically — are the need for a secondary processing step (generally autoclave curing) and the need to store the prepreg tows in a freezer to inhibit premature curing of the material.
Thermoplastics have a much higher viscosity (in the melt state) and melting temperature compared with thermosetting resins. Advantages of thermoplastics for AFP manufacturing include their recyclability and rework-ability, their suitability for in-situ consolidation, and their long shelf life at room temperature [162]. However, much higher temperatures are required (typically around 400° C) during the layup process, and the processing window is significantly smaller. Attaining autoclave level properties and low porosity solely through in-situ consolidation remains a challenge [163].

Dry fiber tows do not come pre-impregnated with a thermoplastic or thermostetting resin. Introducing the resin after the preforming process reduces costs, although it does necessitate an additional manufacturing step [164] [165]. Furthermore, dry fibers can be stored for extended periods at room temperature, and they are well suited for tow steering since there is no resin constricting the movement of the fibers. The lack of resin also reduces the build-up of resin in the AFP head, reducing maintenance requirements. Dry fibers are often supplied with a thermoplastic veil, which provides the tackiness required for AFP preforming [162].

3.2 AP-PLY Design Parameters

In addition to choosing a fiber and matrix combination that best aligns with their requirements, engineers can tailor the properties of an AP-PLY laminate through the modification of the AP-PLY pattern. The following section introduces the parameters that can be varied and briefly discusses their potential impact on a laminate’s properties.

Tow Skipping

Tow skipping refers to the number of tow width gaps left between tows placed in the same set. A tape skipping parameter value of zero will result in a conventional angle-ply composite. Figure 3.1 illustrates the layup process for two cross-ply AP-PLY laminates with different tow skipping values.

Alternating and Series Patterns

The flexibility of AFP preforming allows tows to be placed in either "series" or "alternating" patterns [10]. In the latter, the tow orientation is changed each time a single tow is placed, resulting in a laminate resembling those manufactured using helical filament winding, see Figure 3.2a. Alternatively, in a series pattern the placement orientation is changed after every set of tows is placed, rather than after each individual tow (Figure 3.2b). This has the advantage of requiring fewer movements of the placement head, reducing manufacturing time. Nagelsmit found that the scatter in material properties of laminates laid up in an alternating pattern was higher than in series pattern
Figure 3.1: Schematic illustrating the layup process for two cross-ply AP-PLY laminates with different tow skipping parameters.
laminates [10]. As such, no alternating pattern laminates were investigated in the present work.

![Alternating Pattern](image1.png)

![Series Pattern](image2.png)

Figure 3.2: Alternating and series AP-PLY tow placement patterns. Adapted from [10].

**Hybrids**

The AP-PLY process can be used to reinforce only select locations within a laminate, resulting in a hybrid laminate containing both conventional unidirectional plies and AP-PLY sub-laminates. In doing so, it is possible to improve the interlaminar strength of a laminate at critical interfaces, while reducing the impact of the AP-PLY process on manufacturing time [23, 160]. Hybridization can be applied on a layer by layer basis, as in the work of Rad et al., see Figure 3.3, or by targeting specific locations within a laminate as in Figure 3.4.

**Totally Interwoven Laminates or Directional Skipping**

In a conventional AP-PLY laminate, there is no through thickness fiber connectivity at the interface between a completed "package" of AP-PLY tows. A package in this case refers to a collection of tows that forms a laminate with a uniform thickness, for example, steps 1-16 in Figure 2.36 form a single package. Nagelsmit et al. proposed and patented an adaptation of the regular AP-PLY preforming procedure in which the last tow set from one package is interlaced with the first layer of the next package, see Figure 3.5 [10, 166]. Laminates with this architecture are known as "totally interwoven".

A similar concept was developed by Rad et al. called directional shifting [19]. After a set of tows has been placed in each of the desired fiber
Figure 3.3: Schematic view of a hybrid AP-PLY laminate [23].

Figure 3.4: Schematic view of a hybrid AP-PLY laminate in which specific regions in a laminate are reinforced through the thickness.

Figure 3.5: Schematic illustration of the layup process for a quasi-isotropic totally interwoven AP-PLY laminate.
orientations, the tow placement order changes depending on the value of the directional shift parameter. For example, if the value of the directional shift parameter is 1, then the first fiber orientation placed in the previous set will be the final orientation placed in the next set, see Table 3.1 and Figure 3.6. The directional shift parameter was not investigated in the present work.

<table>
<thead>
<tr>
<th>Tow set</th>
<th>Stacking Sequence</th>
</tr>
</thead>
<tbody>
<tr>
<td>1</td>
<td>$[\theta_1, \theta_2, \theta_3, \theta_4]$</td>
</tr>
<tr>
<td>2</td>
<td>$[\theta_2, \theta_3, \theta_4, \theta_1]$</td>
</tr>
<tr>
<td>3</td>
<td>$[\theta_3, \theta_4, \theta_1, \theta_2]$</td>
</tr>
<tr>
<td>4</td>
<td>$[\theta_4, \theta_1, \theta_2, \theta_3]$</td>
</tr>
</tbody>
</table>

Table 3.1: Layup sequence for a directionally shifted AP-PLY laminate with 4 ply orientations. Adapted from [19].

Figure 3.6: Schematic illustration of the layup process for a quasi-isotropic AP-PLY laminate with a directional shift parameter of 1. After placing tows in the $[0^\circ, 90^\circ, 45^\circ, -45^\circ]$ directions, the next set of tows is placed in a different order, namely: $[90^\circ, 45^\circ, -45^\circ, 0^\circ]$.

### 3.3 Undulation Ratio

To predict the mechanical response of AP-PLY laminates it is essential to accurately represent the through thickness undulations introduced by the preforming process in the numerical models. The undulation ratio variable was defined to quantify the gradient of the through thickness fiber connections. The undulation ratio is simply the ratio of the amplitude of a single tow undulation over its length, i.e. the distance between the point at
which a tow departs from its plane to the point it reaches the next layer (Equation 3.1). Figure 3.7 illustrates the undulation ratio schematically, and provides an image of a real AP-PLY tow undulation obtained using scanning electron microscopy.

\[
\chi = \frac{\text{cpt}}{L_u}
\]

Figure 3.7: Annotated SEM image of a tow undulation in an AP-PLY laminate illustrating the undulation ratio.

\[
\chi = \frac{\text{cpt}}{L_u}
\]  

(3.1)

Where \( \text{cpt} \) represents the cured ply thickness and \( L_u \) represents the length of an undulation. Due to the variability of the manufacturing process, the tow undulations present in an AP-PLY laminate may possess marginally different undulation ratios. In practice, it is very difficult to replicate this variability numerically without reconstructing the geometry of an AP-PLY laminate from tomographs. As such, in this study, an average undulation ratio was used for each AP-PLY laminate.

### 3.4 Representative Volume Element

A representative volume element (RVE) is the smallest part of a material whose properties are representative of the whole. When characterizing the properties of a material, it is essential that tests are conducted on specimens which contain at the very least one RVE. This ensures that the results of the test represent the behavior of the material as a whole, not just a subset. The concept of an RVE is also useful in the field of numerical modeling, because it allows the properties of a material to be determined from a (generally high fidelity) study of only a small volume, reducing computational cost significantly.

While AP-PLY laminates are manufactured through the patterned deposition of tows, this does not guarantee that an RVE can be identified for
all AP-PLY laminates. This is most effectively illustrated by plotting the $B_{11}$ component of an AP-PLY laminates ABD matrix, an approach first developed by Rad et al. [19]. From Figure 3.8 it is straightforward to identify the RVEs of the laminates with stacking sequence [0°, 90°] and [0°, 60°, -60°]. However, while many sub-regions of the quasi-isotropic laminate appear to be similar, no two are identical. For AP-PLY laminates which contain no strictly defined RVE, Rad proposed the use of an "approximate" RVE, whose properties are roughly equivalent to the properties of the entire laminate. The size of a conventional or approximate AP-PLY RVE is dependent on the orientation of a laminate’s constituent tows, the tow width, and the tow skipping parameter. For example, the RVE for a cross-ply laminate with a tow width of 10mm and a tow skipping parameter of 1 is 20 mm × 20 mm. Increasing the tow skipping parameter to 3 will increase the RVE size to 40 mm × 40 mm.

Figure 3.8: $B_{11}$ component of the ABD matrix of (a) cross-ply (b) [0°, 60°, -60°] and (c) quasi-isotropic AP-PLY laminates. The presence and size of laminate RVEs can be identified using this figure. Laminates (a) and (b) contain strictly defined RVEs, bordered in red. Laminate (c) does not contain an RVE, only an "approximate" RVE.

### 3.5 Manufacturing and Laminate Quality

Manufacturing AP-PLY laminates is possible using most commercially available AFP machinery. Generally, AFP heads place a collimated band of 8 to 12 tows with each pass. To manufacture an AP-PLY laminate, a number of these channels — corresponding to the tow skipping parameter — are left empty, see Figure 3.9.

When access to AFP machinery is limited, it is possible to replicate the AP-PLY preforming process manually. In this case, the tows are simply placed by hand on to the tooling surface, and compacted using squeegees and/or intermittent debulking. Printed guides can be used to ensure the tows are
placed in the correct order and are properly aligned with the desired tow orientations. The manual preforming process is illustrated in Figure 3.10.

Manufacturing an AP-PLY laminate is likely to take significantly longer than manufacturing a conventional laminate of the same thickness because more passes of the AFP head are required. Various AP-PLY design parameters affect the time required to produce a laminate. Increasing tow skipping or using an alternating pattern will both increase the number of tow passes required and therefore increase manufacturing time. In contrast, manufacturing a hybrid laminate in which only a select number of plies are interlaced will reduce the impact of the AP-PLY process on the manufacturing time. Nagelsmit reported that their optimal AP-PLY laminate required 42% more time to manufacture than a comparable reference non AP-PLY laminate. In commercial applications the impact of the AP-PLY process on manufacturing time should be considered during the design process.
In terms of laminate quality, AP-PLY laminates are susceptible to the same defects occurring in more conventional AFP manufactured laminates, namely; gaps, overlaps, twisting, bridging, wrinkling, and so forth. These defects and their effects on laminate properties have been extensively studied [167, 168, 169]. Given the greater complexity of AP-PLY laminate’s internal geometry, it is not unreasonable to expect that AP-PLY laminates may contain more defects than conventional angle-ply composites. However, in practice, previous studies have found AP-PLY laminates to possess equivalent laminate quality — in terms of fiber volume fraction and void content — compared to their conventional angle-ply counterparts [26, 10, 19]. Rad et al. noted that the voids occurring at tow undulations are filled by the matrix during curing/consolidation [19]. Nagelsmit found that compared to woven laminates (manufactured using the same fiber and matrix combination), AP-PLY laminates contained fewer voids and tow undulations, see Figure 3.11. The improved quality of the AP-PLY composites was attributed to their more stable fiber architecture and the compaction pressure provided by the AFP machine [10].

Due to the anisotropic thermal expansion coefficients of unidirectional laminae, composite laminates are susceptible to warping. This issue is typically circumvented by creating a "balanced" layup, which contains an equal number of plies in each direction. However, since the additional plies needed to balance a laminate may not be required from a loading perspective, they add unnecessary weight.

Rad et al. investigated whether the AP-PLY laminate architecture could reduce warping in non-balanced laminates [19]. They found that AP-PLY laminates exhibited up to 58% lower out-of-plane displacements compared with non AP-PLY asymmetric laminates, see Figure 3.12. Based on the results of numerical model, Rad et al. suggested that the variation in the local stacking sequence causes the directionality of warpage to change from one region to the next, resulting in no net deformation at the global scale. Laminates built up with smaller tow widths were found to exhibit the smallest out-of-plane deformation, likely due to the small size of each stacking.
sequence region. The effect of the AP-PLY preforming process was not investigated in this study.

Figure 3.12: Comparison of the post curing warpage of (a) a conventional angle ply quasi-isotropic laminate, and (b) an AP-PLY quasi-isotropic laminate [19].
Chapter 4

Multiscale Numerical Modeling Framework

4.1 Introduction

The previous chapter introduced the concept of AP-PLY composites and provided an overview of the parameters that define the internal architecture of a laminate. These parameters, in combination with the myriad of fibers and matrices commercially available, mean the design space for AP-PLY composites is immense. Exploring the design space solely through experimental testing would be prohibitively expensive and time consuming, which is why experimental studies need to be supported by high fidelity numerical models that provide insights into material behavior and guide the design process. If these numerical models are to provide accurate predictions of the stress-strain response of AP-PLY composites it is imperative that they capture the effects of the through thickness reinforcements [170]. Accomplishing this is dependent on the ability of the numerical model to accurately represent the complex internal architecture of AP-PLY composites, an extremely challenging geometric modeling problem [29].

Through thickness fiber connectivity is not unique to AP-PLY composites, and a variety of methods have been developed to account for their presence in woven, braided, or knitted composites (among others). One of the more common approaches is to generate mesoscale specimen geometries using a software package such as TexGen or WiseTex [171][172]. However, while this methodology has proven to be effective in capturing the behaviour of textile composites [173][174][175][176][177], the bezier spline based geometry generation approach employed by these software packages is not easily adaptable to the generation of AP-PLY composite tow orientations.

Alternatively, various researchers have obtained good correlations between their numerical models and experimental results through the reconstruction of specimen geometries from x-ray imagery [178][179][180][181][182]. This approach, while undoubtedly effective in
capturing the internal architecture of an AP-PLY laminate, is time consuming, expensive, and limits the geometries that can be simulated to laminates that have been manufactured, and is therefore not suitable for design space exploration.

As discussed in Section 2.7, Li et al. recently developed a geometry creation program specifically for AP-PLY laminates. The generated geometries show good agreement with micrographs of manufactured laminates, but the computational cost of the model is significant, and it is still subject to interpenetration issues when the mesh is not sufficiently refined [29].

### 4.2 Multiscale numerical model

In this study, a new approach is proposed in which the macroscale variations in strength and stiffness resulting from the presence of tow undulations are captured through the use of multiscale modeling. This low computational cost modeling strategy facilitates the exploration of the AP-PLY composite design space by balancing the need for numerical accuracy with the desire for rapid pre-processing and analysis times. The methodology is adapted from the work of Shah et al. on 3D woven composites [133]. The complete modeling framework is readily available on GitHub.

![Illustration of the idealized geometry used in the numerical models.](image)

First, AP-PLY laminates are divided into regions of three different types: straight fiber regions, undulation regions, and resin-rich regions. Figure 4.1 illustrates schematically the idealized geometry of a cross-ply AP-PLY laminate divided in such a manner.
Three different microscale unit cells are defined corresponding to the three different region types. Depending on the region they represent, the unit cells contain differing proportions of fiber tow and pure resin constituents. By dividing AP-PLY laminates in this manner, through thickness tow undulations are split into four geometrically congruent subregions — illustrated in Figure 4.2 — which contain either two tow constituents with different in-plane and out-of-plane orientations, or a tow constituent and a pure resin constituent.

![SEM image of tow undulation](image1)

![Mesoscale representation of tow undulation](image2)

![Micro-constituent Types](image3)

![Undulation Region Unit Cells](image4)

![Straight Fiber Tow Region Unit Cell](image5)

![Resin Rich Region Unit Cell](image6)

**Figure 4.2:** Schematic of the unit cells for the different region types.

Every element in a finite element model of an AP-PLY laminate is assigned one of the three aforementioned region types. An example of a laminate mesh divided in such a manner is provided in Figure 4.3. Regardless of the region type assigned to an element, a single stress vector is used to represent its stress state, even if the element contains more than one material constituent. The element stress vector is determined by volumetrically averaging the stresses in each constituent. The complete solution process is as follows:

1. Abaqus Explicit provides a six component input strain vector for each element.

2. The strains on each constituent are calculated by assuming iso-strain conditions. In other words, the strains on each constituent are assumed to be equal to the global element strains.

3. Strains are rotated from the global coordinate system to the material coordinate system for each constituent.
4. Stresses in each element constituent are calculated using a continuum damage mechanics approach. Different models are implemented for composite tows and pure resin materials (see Sections 4.4.1 and 4.4.2).

5. The homogenized stress in an element is calculated by taking the volumetric average of the stresses in each constituent.

6. The homogenized stress in an element is returned to Abaqus which uses this value to determine strains in the next simulation timestep.

Figure 4.3: Screen capture from an FEA simulation of an AP-PLY laminate indicating the assignment of region types to each element.

**Limitations**

The global strain state in each element is provided as input to the material subroutine by the Abaqus Explicit solver. Within each unit cell, the strains on each constituent are assumed to be equal to the element-level strains through an isostrain assumption. It is worth acknowledging that there are limitations to the validity of the isostrain assumption.
For straight tow regions, the isostrain assumption is a good approximation of the strain state in the unit cell. While there may be slight local variations due to the random distribution of fibers, the discrepancies are likely to be negligible. However, in regions consisting of two constituents, such as the undulation regions, the isostrain assumption may not accurately represent the strain state in each constituent. To illustrate the limitation, consider the situation illustrated in Figure 4.4 in which an undulation unit cell, consisting of a tow constituent and a pure resin constituent, is stressed in the $x$ direction. Under iso-strain conditions, the deformation of both constituents will be the same. However, since the stiffness of the tow constituent is much greater than that of the pure resin region, a more realistic representation of the deformation of the two constituents is given by Figure 4.4c. Similarly, under transverse loading, the assumption of isostrain conditions does not accurately reflect the strain state in a unit cell containing two distinct constituents.

![Diagram](image)

Figure 4.4: Illustration of deformation in an undulation unit cell subjected to loading in the $x$ direction (a) assuming isostrain conditions, and (b) considering constituent stiffness.

However, in AP-PLY laminates where the volume fraction of the second constituent is relatively small (~18%), and in which the unit cells are typically no larger than 1.5 mm, the assumption of isostrain conditions is reasonable and results in good approximations of laminate behaviour. The model could be improved through the implementation of a more accurate method to capture
the strain state in each constituent at the microscale, but this is a non-trivial task and would likely increase computational cost, and is beyond the scope of the present work.

Aside from the isostrain assumption, an additional limitation is the use of a simple volumetric average to determine the homogenized stress state in an element. An alternative approach is to determine the homogenized stiffness of each unit cell using an analytical model, such as the one proposed by Henry et al. [183]. However, if the undulating tows are not orthogonal, the homogenized stiffness matrix is fully anisotropic and an iterative approach is required to compute it. To keep the computational cost of the modelling approach low, the simpler volumetric stress-averaging method was chosen for this study. This approach has been previously employed by Shah et al. to good effect in their simulations of 3D woven composites [133].

Lastly, because elements in FEA models may contain more than one constituent, delamination between these constituents cannot be captured. As such, debonding of tows can be modeled between straight tow regions, but the propagation of cracks into undulation regions cannot be numerically replicated. Accurately capturing delamination in these regions requires a microscale model, such as the one proposed by Li et al. [29, 184]. However, a microscale model has a much higher computational cost and may be subject to mesh interpenetration issues.

4.3 AP-PLY Laminate Geometry Creation

The generation of multiscale models of AP-PLY laminates is completely automated using Abaqus’ Python API. The model generation process can be divided into three steps: the creation of the idealized AP-PLY laminate geometries, material property assignment, and meshing. The geometry creation framework is implemented in Python using the Shapely package. Shapely is a Python package for the manipulation and analysis of planar geometric objects based on the widely deployed GEOS and JTS libraries [185]. Shapely uses the GEOS library for all its operations, and since GEOS is written in C++, it is highly optimized and efficient.

The user provides as inputs the properties of the tows - elastic constants, strengths, fracture toughnesses, cured ply thickness, and tow width - and provides the AP-PLY process parameters - the orientations of the tows (in the order they are to be placed), the tow spacing (see Section 3.2), the undulation ratio (see Section 3.3), the total number of plies in the laminate (must be divisible by the number of tow orientations), and the planar dimensions of the specimen.

The program first creates a virtual tooling surface, by generating a Shapely Polygon which defines the exterior boundaries of the specimen, as illustrated in Figure 4.5a. Consequently, the program generates a series of lines that define the boundaries of the straight tow and undulation/resin-rich regions (Figure
4.5b). By default, the tow pattern origin is at the midpoint of the specified coupon, \((x_0, y_0)\). However, if a different laminate geometry is desired, the user can shift the origin along the x and y axes.

![Figure 4.5: Creation of virtual tooling surface boundary and division into polygonal regions.](image)

Figure 4.5: Creation of virtual tooling surface boundary and division into polygonal regions.

Where \(\theta\) represents the orientation of a tow, \(w_t\) the width of the central tow region, and \(w_r\) the width of the resin rich regions. The sum of the width of the central tow region and the two adjacent resin rich regions is equal to the nominal tow width, \(w_{nom}\). The width of the aforementioned regions is determined by the value of the undulation ratio, according to Equations 4.1 and 4.2

\[
\begin{align*}
\theta &= L_u/2.0 \\
\theta &= w_{nom} - 2w_t
\end{align*}
\]

The boundary lines are used to partition the specimen into a "grid" of smaller polygons (Figure 4.5c). Each polygon is stored in a Python dictionary along with a unique numeric ID. The next step in the process is to emulate the AFP process by virtually placing tows onto the tooling surface. This is accomplished by assigning tow orientations to the polygons that make up the specimen. In essence, the polygons operate as containers, storing information about the ply stacking sequence in each region of the laminate. To accomplish this, each polygon in the grid is initialized with additional attributes (relative to a pure Shapely Polygon); \(\text{region.type}\), and \(\text{angle}\).

The \(\text{region.type}\) attribute is a list containing the types of each layer in a polygonal area. There are three possible region types: "Tape" representing
straight tow regions, "Undulation" representing regions in which there is fiber curvature, or "Resin" representing the resin rich straight tow regions occurring at the tapering edges of each tow. The \textit{angle} attribute works in the same manner as the \textit{region\_type} attribute but stores the orientation of the tow. Both attributes are initialized as None.

A function iterates over the tow orientations and calculates the boundaries of the tows to be placed with each pass of the AFP head. The boundaries define the central area of each tow, excluding the resin-rich regions, and are a function of the tow width, the tow orientation, and the tow spacing parameter. For each of the tows to be placed, the program checks which of the polygons in the grid are within the bounds of the tow. If a region is within the tow bounds, the angle of the tow and the "Tape" region type are appended to the \textit{angle} and \textit{region\_type} attributes of the corresponding polygon. Since the tow properties are appended to each attribute, the program automatically offsets tows in the through thickness direction if another tow has already been placed in the same region. Figure 4.6 illustrates the placement process schematically.

![Figure 4.6: Illustration of the virtual tow placement process.](image)

The same process is repeated for the resin rich regions which border the central portion of the tow on either side. Shapely’s buffer operation, which dilates a polygon by a specified distance (a distance defined by the undulation ratio in this case), is used to determine the bounds of the resin rich regions. For any polygons in the grid within the specified bounds, the tow angle and the region type "Resin" are appended to the respective attributes (Figure 4.7).
Once all of the tow passes have been placed, the program determines the polygons in which the tows undulate between layers. For every tow in the laminate, the program first merges all of the straight tow regions in the same layer. Starting from the bottom, the program dilates the regions in a layer and in the layer above it by a distance defined by the undulation ratio. The intersection of these two areas identifies the undulation regions. The region_type attributes of the identified polygons are updated accordingly. This process is then repeated for the next pair of layers in the laminate until the top ply is reached. A new polygon attribute, und_pairs is defined which specifies the angles of the two micro-constituents. Figure 4.8 diagrammatically illustrates the undulation region identification process. Note that undulations are assumed to occur only in the central straight tow regions, not in the resin rich regions that bound a tow. This decision was made due to the complexity and variability of the distribution of microconstituents at the edge of a tow which is difficult to capture without direct reconstruction of the geometry from x-ray CT imagery. The omission of fiber curvature in the resin rich regions has a minimal impact on the numerical predictions since they constitute a very small volume fraction of the laminate (approximately 5% in a quasi-isotropic AP-PLY composite).

Having populated the grid with each tow pass and having identified the undulation regions, the geometric information is passed in to Abaqus.
Partitioning a part in Abaqus along the boundaries used to create the polygon grid will result in a complex and inefficient mesh. To avoid meshing issues, the program first cleans up the geometry by removing superfluous cells, edges and vertices. First, areas of the same type are merged to form a single polygon, see Figure 4.9. Next, any vertices which do not define the essential shape of the polygon are removed. This is accomplished by calculating the angles between the edges at a vertex, and removing the vertex if the angle is smaller than a specified tolerance. The coordinates of the resulting polygon are used to create a planar sketch of a part which is subsequently extruded to a depth defined by the cured ply thickness. The part is assigned material properties and a material orientation based on the region_type, angle, and und_pairs attributes of the corresponding polygon in the grid.

Figure 4.9: Geometry clean-up for Abaqus part sketching. The removal of superfluous vertices improves meshing.

All the parts in a single layer are merged to form one instance. Tie constraints or cohesive zones can then be used to define the interlaminar behavior. Examples of the resulting model geometries are provided in Figure 4.10.
Figure 4.10: Examples of AP-PLY laminate geometries created using the proposed framework.
The advantages of this approach are numerous. Firstly, AP-PLY laminate geometries with an arbitrary number of tow orientations can be generated automatically without the need to manually fix interpenetration issues. This is true regardless of the amplitude of the undulations, even very shallow undulation angles where other methods may struggle, can be captured without issues. Secondly, laminates can be meshed coarsely without any loss in the accuracy of the numerical predictions (a coarse mesh may be preferable to ensure correct macro-to-meso strain transformation). Lastly, the methodology can be extended to investigate the effect of AFP manufacturing defects e.g. tow drops or tow gaps.

4.3.1 Elastic AP-PLY Laminate Geometry Creation

In certain cases it may be desirable to create computationally efficient 2D and 3D elastic models of AP-PLY laminates. These elastic models capture the effects of locally varying stacking sequences, but do not model the through thickness tow undulations present in AP-PLY laminates. The advantages of these elastic models are: significantly lower computational cost — especially in the case of shell models — and the ability to run models without needing to define a material subroutine. These lower cost models can be run standalone to provide first-pass estimates of AP-PLY laminate behaviour, or applied regionally in larger models to reduce their computational cost. For example, in drop weight tower simulations, the regions far away from the impactor tip are not (or very minimally) damaged during the course of a simulation, so an elastic model will suffice to capture the behaviour of these regions.

The geometry creation process for an elastic model is identical to the process described in the previous section, except the resin rich region boundaries are not used in the division of the laminate into polygonal regions. In elastic models, each region is of the region_type Tape, and has an angle attribute to define the material orientation in the region (relative to the global coordinate system). The geometry creation process is summarized in Figure 4.11.
4.4 Intralaminar Material Behaviour

As previously discussed, the numerical modeling methodology splits AP-PLY laminates into three different regions, straight tow, resin-rich, and undulation. Each unit cell contains varying proportions of fiber tow and resin material constituents. The simplest is the straight fiber region unit cell, which consists of only a single constituent; a non-undulating tow. Damage initiation and evolution are defined according to the constitutive model for a unidirectional composite tow detailed in Section 4.4.1.

The straight fiber tow regions are bounded by resin rich unit cells, representing the edges of AFP tows where the fiber volume fraction is relatively low. Each straight fiber unit cell consists of resin pockets and a non-undulating tow. The stress-strain response of the neat resin is governed by the material model described in Section 4.4.2 while the behaviour of the non-undulating tow is governed by the aforementioned constitutive model for a unidirectional composite, see Section 4.4.1.

Lastly, undulation regions consist of unit cells containing either a tow and a resin pocket, or two tows with differing in-plane orientations. The stress-strain response of each constituent is modeled using the relevant constitutive model. Depending on the in-plane (\( \alpha \)) and out-of-plane orientation (\( \phi \)) of each constituent, it may be necessary to transform the strains from the element's "global" coordinate system to the material coordinate system by rotating about the y and z axes. Since the out-of-plane orientation of the undulating fibers varies along the length of the unit cell, the mean out-of-plane angle illustrated in Figure 4.12, \( \phi_{\text{avg}} \), is used for the strain transformation (Equation 4.3). The strain transformation matrix is given by Equation 4.4 where \( l_i, m_i, n_i, i = 1, 2, 3 \).
4.4.1 Constitutive behaviour: fiber tows

Damage in the impregnated fiber tows is defined by a continuum damage mechanics framework that degrades the stiffness of the material as damage accumulates based on the models developed by Maimi et al. and Shah et al. [145, 146, 133]. This model was found to provide an effective balance between stability, accuracy, and computational efficiency. Material behavior prior to failure is linear-elastic. After the onset of damage, the gradual unloading of a ply is simulated according to damage...
Figure 4.13: Flowchart illustrating the multiscale algorithm for 3D damage modeling in AP-PLY composites. The $t$ superscript denotes the time step (0 indicates the initial time step). The $D$ variable represents the damage matrix.

... evolution laws expressed as function of three damage variables: $d_1$, representing longitudinal fiber damage, $d_2$, representing transverse matrix damage in the plane of the ply, and $d_3$ representing out-of-plane matrix...
damage. All damage variables are equal to zero prior to damage initiation, and increase to unity at strains corresponding to failure. The compliance tensor of the material can be expressed as a function of the damage variables and the elastic constants of the material as:

$$H_{\text{low}} = \begin{bmatrix} \frac{1}{(1-d_1)E_{11}} & -\frac{v_{12}}{E_{11}} & -\frac{v_{12}}{E_{11}} & 0 & 0 & 0 \\ -\frac{v_{12}}{E_{11}} & \frac{1}{(1-d_2)E_{22}} & -\frac{v_{23}}{E_{22}} & 0 & 0 & 0 \\ -\frac{v_{12}}{E_{11}} & -\frac{v_{23}}{E_{22}} & \frac{1}{(1-d_3)E_{22}} & 0 & 0 & 0 \\ 0 & 0 & 0 & \frac{1}{(1-d_1)(1-d_2)G_{12}} & 0 & 0 \\ 0 & 0 & 0 & 0 & \frac{1}{(1-d_2)(1-d_3)G_{23}} & 0 \\ 0 & 0 & 0 & 0 & 0 & \frac{1}{(1-d_1)(1-d_3)G_{31}} \end{bmatrix}$$  \hspace{1cm} (4.5)

To ensure mesh objectivity, the constitutive model employs the crack band model proposed by Bazant and Oh [186]. The energy dissipated through damage accumulation in each constituent in an element, are regularized using the characteristic lengths of those constituents. In the present work, the characteristic lengths of each micro-constituent, \(i\), were defined as the cubic root of their volume \(V_i\), which means the most accurate results are obtained using elements with an aspect ratio close to one [149, 187].

$$\ell = \sqrt[3]{V_i} \hspace{1cm} (4.6)$$

$$g^k_M = \frac{G^k_M}{\ell}; \hspace{0.5cm} M = 1, 2, 3; \hspace{0.5cm} k = T, C \hspace{1cm} (4.7)$$

where \(G^k_M\) is the fracture toughness of the material along the loading direction \(M\), \(g^k_M\) is the energy dissipated per unit volume (of the constituent), \(T\) and \(C\) are the tensile and compressive loads, respectively, and \(\ell\) is the characteristic length of the constituent. The strain-softening relationships for fiber and matrix damage modes are illustrated in Figure 4.14.

The initiation of damage under longitudinal loading is governed by simple non-interactive maximum strain criteria \(F^T_1\) and \(F^C_1\): While fiber kinking under compressive loading is better predicted using failure criteria (such as LaRC04) that account for fiber misalignment, studies have shown this does not necessarily result in a significant improvement in accuracy [137]. Moreover, the LaRC04 criteria require iteration to determine the angle of the fracture plane for matrix damage. While efficient algorithms have been developed to determine the fracture plane angle, they are non-trivial to implement [136]. To align with the aim of the project — to deliver a computationally efficient numerical modeling framework — the simpler non-interactive criteria were used in this study.

$$F^k_1 = \frac{\varepsilon_{11}}{\varepsilon_{0k}}; \hspace{0.5cm} k = T, C \hspace{1cm} (4.8)$$
where $\varepsilon_{11}^{0k}$ represents the strain corresponding to the strength of the material, i.e. $\varepsilon_{11}^{OC} = X_{11}^C/E_{11}$. After the onset of damage, the stiffness of the material is degraded according to the scalar damage variable $d_k$, defined by an exponential law. In tension, the exponential law is given by Equation (4.9).

$$d_T^1 = 1 - \frac{1}{r_T^1} \exp \left[ A_T^1 \left( 1 - r_T^1 \right) \right]$$ (4.9)

where $r_T^1$ is the longitudinal tensile elastic domain threshold, initially equal to one and increasing monotonically with damage evolution. This is because cracks that form under compressive loading open on load reversal [145].

$$r_T^1 = \max \left\{ 1, \max_{s=0,t} \left\{ F_T^s \right\}, \max_{s=0,t} \left\{ F_C^s \right\} \right\}$$ (4.10)

where $s$ denotes a single time step, in the range from 0 to $t$, and $t$ is the current time step. $A_T^1$ is a parameter that ensures the correct dissipation of fracture energy and is a function of the characteristic length in the fiber direction $\ell_1$. $E_{11}$ is the Young’s modulus in the fiber direction, $G_T^1$ is the longitudinal tensile fracture energy, and $X_T^1$ is the longitudinal tensile strength.

$$A_T^1 = \frac{2\ell_1 (X_T^1)^2}{2E_{11}G_T^1 - \ell_1 (X_T^1)^2}$$ (4.11)

It is worth noting that under longitudinal tensile loading a linear-exponential softening curve would better capture the experimentally observed failure sequence of fiber matrix failure followed by fiber bridging and fiber pull-out [146, 187]. However, determining the pull-out stress and the energy dissipated under each mode is non trivial, and the discontinuity of the stress-strain response can cause stability issues. For these reasons, a simpler exponential damage evolution law is used in this study.
In compression, the damage variable $d^C_1$ must be expressed as a function of both the longitudinal damage variable $d^T_1$ and the compressive elastic domain threshold $r^C_1$. While cracks formed under tensile loading will close under compressive loads, the broken and misaligned fibers cannot carry any additional load [146].

\[
d^C_1 = 1.0 - (1.0 - d^{C*}_1)(1.0 - A^{±}_1 d^T_1)
\]  

(4.12)

where $d^{C*}_1$ is the exponential damage evolution function for purely compressive damage, given by Equation 4.13.

\[
d^{C*}_1 = 1 - \frac{1}{r^C_1} \exp \left[ A^C_1 \left(1 - r^C_1\right)\right]
\]  

(4.13)

Note that since the tensile cracks close under load reversal the compressive elastic domain threshold is not affected by tensile damage, see Equation 4.14. The $A^C_1$ is defined in the same fashion as in the tensile mode.

\[
r^C_1 = \max \left\{ 1, \max_{s=0,t} \left\{ F^{C}_1 \right\} \right\}
\]  

(4.14)

\[
A^C_1 = \frac{2\ell_1 \left(X^C\right)^2}{2E_{11}G^C_1 - \ell_1 \left(X^C\right)^2}
\]  

(4.15)

The $A^{±}_1$ parameter defines the extent to which damage accumulated in tension affects the compressive response

\[
A^{±}_1 = b \frac{E_{11} - E_{22}}{E_{11}}
\]  

(4.16)

where $E_{11}$ and $E_{22}$ are the longitudinal and transverse moduli of a lamina. The $b$ parameter is used to control the extent of stiffness retention. When $b = 1$, the loads are assumed to be carried solely by the matrix. When $b = 0$ fibers are assumed not to have lost alignment and there is no loss in stiffness. In the present work, an intermediate value of 0.5 has been used.

Finally, the longitudinal damage variable $d_1$ can be expressed as a function of the tensile and compressive damage variables and the sign of the longitudinal normal stress. This accounts for the closure of cracks occurring under load reversal.

\[
d_1 = d^T_1 \frac{\langle \sigma_{11} \rangle}{|\sigma_{11}|} + d^C_1 \frac{\langle -\sigma_{11} \rangle}{|\sigma_{11}|}
\]  

(4.17)

Under loading transverse to the fibers, a composite will fail through matrix cracking and/or fiber matrix decohesion. Damage initiation is predicted by a
three-dimensional adaptation of the Hashin failure criteria [133]:

\[ F^{2T} = \left( \frac{\langle \hat{\sigma}_{22} \rangle}{Y_{isT}} \right)^2 + \left( \frac{\hat{\tau}_{12}}{S_{ls}^L} \right)^2 + \left( \frac{\hat{\tau}_{23}}{S_{ls}^T} \right)^2 \]  
(4.18)

\[ F^{3C} = \left( \frac{\langle -\hat{\sigma}_{33} \rangle}{Z_{is}^T} \right)^2 + \left( \frac{\hat{\tau}_{31}}{S_{ls}^R} \right)^2 + \left( \frac{\hat{\tau}_{31}}{S_{ls}^R} \right)^2 \]  
(4.19)

\[ F^{3T} = \left( \frac{\langle \hat{\sigma}_{33} \rangle}{Z_{is}^T} \right)^2 + \left( \frac{\hat{\tau}_{31}}{S_{ls}^R} \right)^2 + \left( \frac{\hat{\tau}_{31}}{S_{ls}^R} \right)^2 \]  
(4.20)

\[ F^{3C} = \left( \frac{\langle -\hat{\sigma}_{33} \rangle}{Z_{is}^T} \right)^2 + \left( \frac{\hat{\tau}_{31}}{S_{ls}^R} \right)^2 + \left( \frac{\hat{\tau}_{31}}{S_{ls}^R} \right)^2 \]  
(4.21)

where \( Y_{isT}, Y^C, Z_{is}^T, \) and \( Z^C \) are the tensile and compressive strengths in the transverse and through-thickness directions, respectively, and \( S_{ls}^L, S^T, S_{ls}^R \) are the shear strengths in the 12, 23, and 31 directions, respectively. The \( \hat{\cdot} \) indicates a trial stress component. The \( is \) subscript indicates in-situ strengths. As discussed in the work of Camanho et al., the matrix dominated strengths of an individual ply are affected by the constraining effect of neighboring plies of a different orientation [50, 51]. This effect is known as the in-situ effect and can be accounted for by modifying the strengths of a laminae using analytical models.

The transverse tensile strength \( Y_{isT} \) and longitudinal in-situ shear strength \( S_{ls}^L \) were calculated according to the analytical model proposed by Camanho et al. [50].

\[ Y_{isT} = \sqrt{\frac{8G_{Ic}}{\pi t \Lambda_{o22}^L}} \]  
(4.22)

\[ S_{ls}^L = \sqrt{\frac{(1 + \beta \psi G_{12}^2)^{1/2} - 1}{3 \beta G_{12}}} \]  
(4.23)

Where the parameter \( \psi \) is dependent on the location of a ply within a laminate; for an embedded (thin) ply: \( \psi = \frac{48G_{Ic} t}{\pi \Lambda_{o22}^L} \), for an outer ply: \( \psi = \frac{24G_{Ic} t}{\pi \Lambda_{o22}^L} \). The \( \beta \) parameter represents the non-linearity of the shear stress–shear strain relation, a value of \( 4.72 \times 10^{-8} \) was used in the present work, see Appendix A.1. \( t \) represents the ply thickness, \( G_{Ic} \) and \( G_{11c} \) represent the normal and shear interlaminar fracture energies, and \( \Lambda_{o22}^L \) is given by Equation [4.24]

\[ \Lambda_{o22}^L = 2 \left( \frac{1}{E_{22}} - \frac{\nu_{21}^3}{E_{11}} \right) \]  
(4.24)

Four damage variables \( (d_{2T}, d_{3C}, d_{3T}, d_{3C}) \) are defined that correspond to the four failure criteria. When the value of a failure criterion exceeds unity, the corresponding damage variable is updated to induce softening of the material.
in the relevant direction. For matrix damage, stiffness degradation is linear, and is defined by a damage evolution law of the form:

\[ \bar{d}_k^i = \frac{\varepsilon_{f k}^i}{\varepsilon_i^k} \left( \varepsilon_i^k - \varepsilon_0^k \right) \quad i = 2, 3, k = T, C \tag{4.25} \]

where \( \varepsilon^0 \) is the strain at damage onset, \( \varepsilon \) is the current strain, and \( \varepsilon^f \) represents the ultimate failure strain, given by:

\[ \varepsilon_i^f = \frac{2G_i}{\sigma_0^0 i} \quad i = 2, 3 \tag{4.26} \]

where \( G_i \) is the fracture energy of the material in the relevant direction, \( \ell_i \) is the characteristic length and \( \sigma_0^0 \) is the stress at damage initiation. Consequently, the damage variables \( d_2 \) and \( d_3 \) can be calculated as:

\[ d_i = 1.0 - (1.0 - d_T^i) \ast (1.0 - d_C^i) \quad i = 2, 3 \tag{4.27} \]

Note that because the shear moduli are degraded by a combination of the \( d_1, d_2, \) and \( d_3 \) variables, the greater toughness of unidirectional composites in the shear direction is not accounted for. In addition, the shear response is assumed to be linear elastic up to the point of failure, while in reality matrix plasticity results in non-linearities prior to the formation of damage. These aspects of the material response will be the focus of future studies.

### 4.4.2 Constitutive behavior: pure resin

Pure resin regions are assumed to be isotropic with initial stiffness \( E_m \). As such their stiffness is degraded using a single scalar damage variable \( d_m \). The compliance matrix, which is a function of the damage state, is:

\[
\mathbf{H}_m = \frac{1}{E_m} \begin{bmatrix}
\frac{1}{(1-d_m)} & -\nu_m & -\nu_m & 0 & 0 & 0 \\
-\nu_m & \frac{1}{(1-d_m)} & -\nu_m & 0 & 0 & 0 \\
-\nu_m & -\nu_m & \frac{1}{(1-d_m)} & 0 & 0 & 0 \\
0 & 0 & 0 & \frac{1+\nu_m}{(1-d_m)} & 0 & 0 \\
0 & 0 & 0 & 0 & \frac{1+\nu_m}{(1-d_m)} & 0 \\
0 & 0 & 0 & 0 & 0 & \frac{1+\nu_m}{(1-d_m)}
\end{bmatrix} \tag{4.28}
\]

Damage onset is predicted using the following pressure dependent loading functions adapted from the work of Liu et al. [188]:

\[
F_m^T = \frac{3J^2 + I_1 (Y^C - Y^T)}{Y^C Y^T} \quad \text{if } I_1 \geq 0 \tag{4.29}
\]

\[
F_m^C = -\frac{3J^2 + I_1 (Y^C - Y^T)}{Y^C Y^T} \quad \text{if } I_1 < 0 \tag{4.30}
\]
where $I_1$ is the first invariant of the stress tensor, $J^2$ is the second invariant of the deviatoric stress tensor (Equation 4.31 & 4.32) and $Y^T$ and $Y^C$ are the tensile and compressive strength of the pure resin region, respectively, assumed to be equal to the transverse strengths of the unidirectional tows.

\[
I_1 = \sigma_{11} + \sigma_{22} + \sigma_{33} \quad (4.31)
\]

\[
J_2 = \frac{1}{6} \left[ (\sigma_{11} - \sigma_{22})^2 + (\sigma_{22} - \sigma_{33})^2 + (\sigma_{33} - \sigma_{11})^2 \right] \quad (4.32)
\]

After failure initiation, damage is dissipated according to the following exponential damage evolution law following the same methodology described in Section 4.4.1:

\[
d^k_m = 1 - \frac{1}{r^k_m} \exp \left[ A^k_m \left( 1 - r^k_m \right) \right] \quad k = T, C \quad (4.33)
\]

where $A^T_m$ and $A^C_m$ are the tensile and compressive fitting parameters used to ensure correct dissipation of fracture energy, and $r^T_m$ and $r^C_m$ represent the elastic domain thresholds under tensile and compressive loading, respectively, defined as

\[
A^k_m = \frac{2 \ell_m Y^k}{2 E_m G^k_m - 2 \ell_m Y^k} \quad k = T, C \quad (4.34)
\]

\[
r^T_m = \max \left\{ 1, \max_{s=0,t} \left\{ F^T_M \right\}, \max_{s=0,t} \left\{ F^C_M \right\} \right\} \quad (4.35)
\]

\[
r^C_m = \max \left\{ 1, \max_{s=0,t} \left\{ F^C_M \right\} \right\} \quad (4.36)
\]

where $\ell$ is the constituent’s characteristic length, $G^T_m$ and $G^C_m$ are the tensile and compressive fracture energies, and $E_m$ is the bulk resin modulus. Finally the damage variable $d_m$ is:

\[
d_m = 1.0 - (1.0 - d^T_m) \ast (1.0 - d^C_m) \quad (4.37)
\]

### 4.4.3 Stability Controls

To avoid unrealistic element distortion resulting from the strain-softening constitutive behavior, elements were deleted from the mesh if the determinant of the deformation gradient $\mathcal{F}$ (i.e. the ratio of the deformed to the undeformed element volume) exceeded predefined limits [189]. The implementation of these deletion criteria improved stability and prevented simulations from aborting prematurely. Note that care must be taken when using the deformation gradient for deletion criteria as this approach may result in premature element deletion when the mesh is very fine. As the size of an element is decreased, its ultimate failure strain rises to ensure the correct amount of energy is dissipated, in line with the work of Bazant and Oh [186].
If the deletion criteria are not adjusted to account for the higher failure strain, elements may be deleted from the mesh before they have dissipated all their energy. Careful consideration of the deformation gradient bounds for each element size are therefore required when applying this methodology.

\[
\text{Delete element if } \begin{cases} 
0 < \det \mathbf{F} < 0.5 \\
\text{or } \det \mathbf{F} > 2.5
\end{cases} \tag{4.38}
\]

The use of hourglass and distortion controls in finite element analyses results in the generation of "artificial" strain energy. To ensure the validity of the simulation results, it is important that this artificial energy does not exceed 2% of the total energy in the system [149]. Figure 4.15 illustrates the energies during a dynamic simulation of split-Hopkinson bar experiment. The artificial strain energy reaches a maximum of 0.65% of the strain energy in the system, remaining well below the aforementioned 2% threshold.

![Figure 4.15: Plot of strain energy, kinetic energy, and artificial strain energy over the course of an SHB tensile test simulation.](image)

### 4.4.4 Validation

The fiber tow constitutive model was validated through the simulation of tensile experiment on a baseline (i.e. non AP-PLY) specimen with a \([0^\circ, 90^\circ]_{2S}\) layup and dimensions conforming to the ISO 527 standard, 150 mm × 25 mm. Figure 4.16(a) compares the experimental stress-strain response of a baseline composite with its numerically predicted behavior at different mesh densities and Figure 4.16(b) shows the response for one single element. Numerical and experimental results were in very good agreement in terms of

83
stiffness and strength. Failure is predicted accurately even at relatively low mesh densities. At high loads matrix damage accumulation in the numerical model softens the stress-strain response, however, this does not reduce the accuracy of the strain-to-failure prediction. The main failure mode, fiber fracture in the plies oriented with the loading direction, was effectively captured by the numerical model. The drop off in the load is the same regardless of the mesh density, indicating that the energy dissipated in the formation of a crack is independent of the element size. This is further evidenced by the plots of the stress-strain response of a single element, which show that as the element size is increased, the area under the stress-strain curve, which represents the volumetric fracture energy density, is decreased. The same principle applies for the pure resin regions, as illustrated in Figure 4.17.

Figure 4.16: Mesh convergence plots showing (a) experimental and numerical stress-strain curves for a conventional cross-ply laminate using different mesh densities and (b) stress strain response in single elements with different dimensions.
4.5 Interlaminar Material Behaviour

Where applicable, the behavior of the interfaces between laminae was defined using the cohesive zone method (CZM). A bilinear traction separation law was implemented using the built-in cohesive interaction functionality in Abaqus [148]. Cohesive interactions were preferred to cohesive elements for several reasons. Firstly, the thickness of the interlaminar region in a laminated composite is very small, and therefore best approximated using a zero thickness interaction which does not add mass to the system [148]. Secondly, upon delamination of the interface, cohesive interactions can model frictional effects between contacting plies. Thirdly, cohesive interactions have a minimal effect on the stable time increment, reducing computational cost. Finally, cohesive interactions were easier to implement in the multiscale numerical framework developed in this study.

The stiffness of the cohesive regions was determined according to the methodology developed by Turon et al. in which the penalty stiffness is expressed as a function of the transverse stiffness of the laminate, $E_M$, the thickness of the cohesive layer, $h$, and a constant $\alpha$ (Equation 4.39) [190]. The value of $\alpha$ must be significantly larger than 1, in this study, as in Turon et al. a value of 50 was used.

$$K_M = \frac{\alpha E_M}{h}$$ (4.39)
Damage initiation was governed by a quadratic stress criterion:

\[
\left(\frac{\tau_1}{\tau_{01}}\right)^2 + \left(\frac{\tau_2}{\tau_{02}}\right)^2 + \left(\frac{\langle \tau_3 \rangle}{\tau_{03}}\right)^2 \leq 1
\]  

(4.40)

Where \(\tau_N(N = 1, 2, 3)\) and \(\tau_{0N}(N = 1, 2, 3)\) are the interface stresses and maximum interface stresses in the in-plane (1,2) and normal (3) directions, respectively. Mixed mode behavior was defined by the energy based criterion of Benzeggagh and Kenane [150].

\[
G_c = G_{Ic} + (G_{IIc} - G_{Ic}) \left( \frac{G_{II} + G_{III}}{G_{I} + G_{II} + G_{III}} \right) \eta
\]

(4.41)

where \(\eta\) represents the Benzeggagh-Kenane parameter. After debonding of an interface, a penalty contact algorithm governs the interaction between plies. A friction coefficient of 0.15 was defined for all tangential contact between debonded plies.

To model the through thickness reinforcements present in AP-PLY laminates, the cohesive interaction properties were modified for connected undulation regions in adjacent layers. The modification of the cohesive interface properties was preferred over the use of tie constraints due to constraint conflict issues when applying the latter. The cohesive interface strength between these undulations regions was set to equal the longitudinal tensile strength of the CFRP tows. The stiffness was estimated by substituting the longitudinal stiffness into Equation 4.39 in the place of the transverse laminate stiffness.
Chapter 5

Microscale Characterization

5.1 Introduction

The through thickness tow undulations present in AP-PLY laminates can have a detrimental impact on their in-plane mechanical response. It is critical to the safe operation of structures containing AP-PLY laminates, that the effect of these undulations on the properties and failure modes of AP-PLY laminates is well understood. To further understand the formation and propagation of damage at tow undulations, an X-ray tomography study was conducted.

Using computed X-ray tomography (X-ray CT) it is possible to obtain 3D images of composite laminae at extremely high resolutions, facilitating the non-destructive inspection of fiber architectures, manufacturing defects, and/or in-service damage accumulation [80]. To study the initiation and propagation of damage under increasing loads, a sequence of 3D images at different load intervals is required. This can be achieved through either ex-situ or in-situ tomography. In the former, specimens are loaded by an external system, from which they are removed at certain intervals for imaging. In the latter, the loading rig is situated inside of the X-ray tomography system, allowing the specimens to be imaged under the application of a load. In-situ loading is often preferable because it simplifies post processing of the images, and because cracks in a specimen may close when unloaded [80]. In the present work, in-situ X-ray tomography was used to investigate the fiber architecture of tows containing fiber undulations, and to better understand the initiation and propagation of damage in these regions.

5.2 Materials

Specimens were manufactured using SHD VTC401 carbon fiber epoxy prepreg. To create the tow undulations, plies were stacked such that a symmetrical tow undulation was formed at the center of each laminate. The process is illustrated in Figure 5.1. The laminates were cured in a hot press at 4 bars of pressure and a curing temperature of 110° C. The thickness of the
cured laminate was $1.08 \text{ mm} \pm 0.05 \text{ mm}$. Glass fiber end tabs were adhered to the specimen ends using SHD Composites VTFA400 adhesive film.

Specimens were designed considering X-ray attenuation, and the specifications of the in-situ rig. The dimensions of the specimens are illustrated in Figure 5.2. The width of the specimens was specified to be the lowest value attainable using CNC machining. This was done to ensure the ratio of the specimen width to the specimen thickness was as close to unity as possible, to ensure approximately equal X-ray attenuation regardless of the orientation of the specimen relative to the source.

5.3 Quasi-static Characterization

In light of the costs and time consuming nature of in-situ tomography experiments, only a single specimen was tested and imaged at the NXCT lab. To establish the loading intervals at which to image the specimen, an
additional 6 specimens, all identical to the specimen tested in the X-ray CT scanner, were tested to failure on a 50 kN Instron universal testing machine at the University of Edinburgh. Custom grips were manufactured to allow the specimens to be clamped in the same manner as in the in-situ rig used in the X-ray CT experiments, see Figure 5.3. Specimens were aligned in the clamps using four 3 mm diameter steel pins. The loads were transferred through friction between the clamps and the specimen. The surfaces of the clamps were knurled to increase the friction coefficient. Once in place the specimens were loaded to failure at a rate of 0.1 mm/min. Displacement was tracked using a 2D digital image correlation (DIC) system. A 1.4 megapixel Allied Vision Manta G-146 camera was used to capture specimen deformation at 17.6 frames per second. The images were processed using Correlated Solutions’ VIC-2D software package to determine the strains in the specimen.

Figure 5.3: Experimental setup for the quasi-static testing of the undulating tow specimens.

5.4 In-situ X-ray Computed Tomography

The in-situ tomography experiments were carried out at the National X-ray Computed Tomography (NXCT) laboratory in Manchester (formerly known as the Henry Moseley X-ray Imaging Facility). Access to the equipment at the NXCT lab, and the funding for the experiments were provided through a Henry Royce PhD Equipment Access grant (EPSRC grant numbers
EP/R00661X/1 and EP/T02593X/1). The tomography experiments were conducted using a Zeiss Xradia 520 system (Figure 5.4). A 3 mm × 3 mm field of view allowed entire tow undulations to be captured at a resolution of 1.5 \( \mu \text{m} \) per pixel. Taking aliasing into consideration, this resulted in a minimum detectable damage size of approximately 4.5 \( \mu \text{m} \) (3 times the pixel size), sufficient to capture small matrix cracks [80]. For reference, carbon fibers measure approximately 7 \( \mu \text{m} \) in diameter.

A 5 kN Deben CT5000 in-situ rig was used to apply tensile loads to the specimen, see Figure 5.5. The rig uses a 15 mm thick polyetherimide support tube to load the specimens, which minimizes X-ray attenuation. The experiments were conducted using an interrupted in-situ acquisition strategy. In other words, specimen extension was paused every 250 N to allow for image acquisition. Note that no images were taken at 250 N to reduce total beamtime, no damage was expected to occur at this low load. Specimens were installed in the in-situ rig in the same manner as previously described in Section 5.3. An exploded view of the in-situ rig and the specimen is provided in Figure 5.6. The Fiji software package was used to post-process the 3D images [191].

5.5 Results and Discussion

Aside from analyzing the initiation and propagation of damage, the X-ray CT scans also facilitate the inspection of the laminate architecture. Three different areas were identified, as highlighted in Figure 5.7, the transverse tows, the longitudinal undulated tow, and an additional resin region. In the resin rich regions, the fiber volume fraction is low compared to the rest of the laminate. The lack of fiber reinforcement in these areas may allow for increased plastic deformation, possibly facilitating the alignment of the tows with the loading direction as loads increase.

Figure 5.8 illustrates the evolution of damage with increasing load. Perpendicular cracks were first observed at a load of 500 N, corresponding to a nominal strain of approximately 0.7%. Similar values for the onset of transverse matrix cracking in carbon epoxy composites have been previously
Figure 5.5: Photograph of the Deben CT5000 in-situ rig installed in the Zeiss Xradia 520 system, from [12].

Figure 5.6: Schematic of specimen attachment within the in-situ rig. Adapted from [13]

Figure 5.7: X-ray CT image illustrating the resin-rich and tow regions.
reported in [158]. The cracks originated at the free surfaces of the specimen, in the 90° tows, and propagated inwards towards the tows aligned with the loading direction. The cracks widened as loads were increased. Furthermore, as illustrated in Figure 5.9, the density of the matrix cracks increased exponentially as the load rose. As a result, the transverse tows gradually softened, increasing the load borne by the undulating tows. No damage was observed in the central transverse tows. The constraining effect of the neighboring undulating plies arrests crack formation, similar to the in-situ effect discussed in Section 4.4.1. It is likely that the greater capacity for plastic deformation of the resin-rich region also plays a role in the inhibition of crack initiation. Failure of the specimen due to fiber fracture was attained at 1600N.

Figure 5.8: X-ray CT images illustrating damage evolution in a specimen with through thickness tow undulations. Matrix cracks are highlighted in red.

The stress-displacement response of a single specimen, tested in the Instron machine at the University of Edinburgh, is depicted in Figure 5.10a. The force-displacement responses of the specimen tested in-situ and the specimens tested in the Instron, are illustrated in Figure 5.10b. Since the in-situ rig is displacement controlled, when the extension is paused to allow for image acquisition, there is some stress relaxation in the specimen, resulting in intermittent load drops. Given that the failure strength of the specimen was in line with those tested previously on the Instron machine, these small load fluctuations probably did not significantly affect the specimen behaviour. The stress-strain and force-displacement responses of the specimens are non-linear. The initial non-linearity in the force-displacement response, from 0 mm to 0.1 mm, can be attributed to flexure of the clamping system. The fact that the stress-strain response — based on DIC data — is linear in this regions lends further credence to this statement. After taking up the slack in the clamping system, the stiffness of the specimens is gradually reduced as strain (or displacement) is increased. This behaviour is likely the result of damage accumulation in the transverse tows.

In terms of strength, calculating the theoretical failure load of the composite
using classical laminated plate theory yields a value of 2.048 kN (or 948 MPa). This suggests that the tow undulations have a negative impact on laminate strength. The through thickness fiber curvature induces stress concentrations which precipitate failure at lower loads than in conventional laminates. It should be noted that in addition to the tow undulations the dogbone geometry of the specimens is likely to have further increased stress triaxiality.

Figure 5.10: Stress-strain response (left) and force displacement response (right) of laminates containing tow undulations subjected to tensile loading.
5.6 Conclusions

In summary, cross-ply carbon epoxy composite laminates containing through thickness tow undulations were imaged using X-ray computed tomography while subjected to tensile loading using an in-situ rig. The resulting images revealed that matrix cracks develop in the tows orthogonal to the loading direction. The density of these cracks increases exponentially as the load rises, a result of both crack widening and the formation of new cracks. As matrix damage accumulates the stress-strain response softens. Finally, the curved fiber path of the tow undulations creates stress concentrations, ultimately reducing laminate strength.
Chapter 6

Quasi Static Mechanical Properties

6.1 Introduction

The in-plane quasi-static tensile behaviour of AP-PLY laminates was investigated first, both experimentally and numerically, to ascertain the effect of the AP-PLY tow undulations on macroscale laminate properties. Rad et al. and Nagelsmit previously investigated the undamaged in-plane mechanical properties of AP-PLY composites, concluding that the quasi-woven internal architecture had a non-negligible influence on stiffness and strength, however, neither study used DIC or 3D numerical models to better understand material behaviour at these critical points and determine the mechanisms that could promote the catastrophic failure of structures manufactured using AP-PLY composites. This chapter describes, in detail, the results of the experimental and numerical characterization of cross-ply and quasi-isotropic AP-PLY laminates.

6.2 Manufacturing

Two AP-PLY laminates with different internal architectures were manufactured: (i) a cross-ply laminate \([0/90]_{2s}\) (XP\textsubscript{AP-PLY}) and (ii) a quasi-isotropic laminate with stacking sequence \([0/45/90/-45]_s\) (QI\textsubscript{AP-PLY}). The latter represents the state of the art in terms of the complexity of its internal architecture \([160, 26, 24]\). The AP-PLY panels were laid up by hand in a process emulating automated fiber placement. Tows were cut out of a roll of prepreg (SHD Composites VTC401) to a width of 10 mm, then placed into a mold in a predefined sequence. Guides were used to ensure correct alignment. Figure \ref{fig:layup} illustrates the layup process for the quasi-isotropic AP-PLY laminate. In both laminates a gap of three tow widths was left between tows placed in the same pass, as in \([19, 23]\). The \(300 \times 300 \text{ mm}^2\) panels were cured in a hot press under 4 bars of pressure at 120°C for 120
minutes. The curing cycle is illustrated in Figure 5.2. In addition, two reference — non AP-PLY — laminates were manufactured for comparison with the AP-PLY panels, (XP_{ref} and QI_{ref}). The discrepancies between the thicknesses of the AP-PLY and baseline specimens were negligible. The average thicknesses of the cross-ply AP-PLY and baseline specimens were 1.68 mm and 1.63 mm respectively. The quasi-isotropic AP-PLY and baseline thicknesses were 1.61 mm and 1.63 mm. A fiber volume fraction of approximately 53.2% was obtained for all the laminates. Glass fiber end tabs were adhered to all specimens using an epoxy adhesive film (SHD Composites VTFA400).

![Layup Process](image)

Figure 6.1: Layup process for a quasi-isotropic AP-PLY laminate. Note layup steps 5-15 are omitted for brevity.

### 6.3 Quasi-static Tensile Characterization

#### 6.3.1 Experimental Methodology

Specimens were extracted from the panels using a water cooled diamond saw. The dimensions of the reference non-AP-PLY specimens conformed to the ISO 527 standard (25 mm x 250 mm with 50 mm end tabs). However, the AP-PLY specimen dimensions were governed by the size of the RVE of each interlacing configuration. As discussed in the Section 3.4, certain AP-PLY laminate configurations, including the quasi-isotropic AP-PLY laminate in this study - do not contain a well defined RVE. Where an RVE was not readily identifiable, an "approximate" RVE was determined whose properties were approximately representative of the behavior of its parent laminate see Figure
Figure 6.2: Curing cycle for specimens manufactured using SHD Composites VTC401 prepreg.

6.3 [19]. Glass fiber end tabs were adhered to all specimens using an epoxy adhesive film (SHD Composites VTFA400).

Figure 6.3: Non-standard design for AP-PLY laminate specimens.

Aside from the specimen dimensions, testing was carried out in accordance with the ISO 527 standard. Six specimens from each panel were tested to failure. Testing was conducted using an MTS 300 kN universal testing machine, at 2 mm/min crosshead displacement. Full-field displacements were recorded using a 2D digital image correlation system. A 1.4 megapixel Allied Vision Manta G-146 camera was used to capture specimen deformation at 17.6 frames per second. The images were processed using Correlated Solutions’ VIC-2D software package. The experimental setup is illustrated in Figure 6.4.
6.3.2 Material Characterization

Introduction

For most composites in use today, the individual lamina (i.e., the individual layer or ply) is the basic unit or building block, whether it is in the design, the analysis, or the fabrication process stage. This lamina may be a unidirectional prepreg, a fabric, a chopped strand mat, or another fiber form, with or without the matrix present prior to laminate fabrication. Whatever the material form or fabrication process, the properties of the individual laminae (regions, layers, or whatever form the composite takes) must be known for design and analysis purposes [192].

Lamina Tensile Response

The longitudinal and transverse tensile strength of the composite laminae were determined according to ISO standard 527-5 [14]. Unidirectional laminates 300 mm in length and 300 mm in width were manufactured by hand by placing plies of SHD VTC401 onto a mould (all plies oriented in the same direction). The resulting preforms were cured in a hot press at 110°C, and at a pressure of 4 bar, see Figure 6.2. Glass fiber end tabs were adhered to the specimens using SHD VTFA400 epoxy adhesive film. Type A specimens, as defined in the ISO standard, were extracted for the longitudinal tensile specimens, while...
type B specimens were extracted for the transverse tensile experiments. The dimensions are presented schematically in Figure 6.5. Finally, specimens were speckled with white paint to provide sufficient contrast for effective digital image correlation.

Figure 6.5: Dimensions of specimens for characterization of tensile response of composite lamina according to ISO 527-5 [14]

Once aligned in the MTS Criterion 300 kN universal testing machine, specimens were clamped using hydraulic wedge grips, then loaded to failure at a rate of 2 mm per minute. A 2D digital image correlation system was used to track strains in the specimens. The images were post processed using the VIC-2D software package. The longitudinal and transverse moduli were calculated from the slope of the stress-strain curve over the strain interval from 0.05% to 0.25%. Poisson’s ratios were determined through linear curve fitting of the longitudinal and transverse strains over the recommended strain interval \((0.3 < \varepsilon < \varepsilon_y)\) where \(\varepsilon_y\) is the strain at yield. Longitudinal and transverse lamina strengths, \(X_{1T}^T\) and \(X_{2T}^T\), were simply defined as the ultimate values of \(\sigma_{11}\) and \(\sigma_{22}\) for the 0° and 90° tensile tests, respectively [192]. The results of the tensile tests in both orientations are illustrated in Figure 6.6.

Due to their high strength and high levels of anisotropy, testing unidirectional CFRPs is very challenging. During the longitudinal tensile tests, a large number of specimens failed near the clamps, rather than in the gauge section. As a result, the longitudinal tensile strength values were lower than expected, and lower than the strengths quoted by the manufacturer. As such, the experimental data was used to determine the longitudinal modulus of the material, but the longitudinal strength value used in the numerical models was taken from the material’s technical data sheet [193].

**Lamina Compressive Response**

The compressive strength of the composite laminae was determined using a combined loading compression (CLC) fixture as outlined in the ASTM D 6641 standard [194]. Laminates were manufactured as in Section 6.3.2. Specimens
were extracted from the cured panels using a wet saw, and end tabbed using glass fiber tabs adhered using the aforementioned epoxy adhesive film. A rosette strain gauge was superglued to each side of the specimen to provide measurements of the strains, and to detect buckling. Specimens were placed within a Wyoming Test Fixtures CLC fixture, which was placed between two self aligning parallel platens attached to an MTS Criterion 300 kN universal testing machine. The dimensions of the specimens, and the test fixture, are illustrated in Figure 6.7. Specimens were loaded to failure at a rate of 1.3 mm per minute. The advantage of using combined end and shear loading is that it reduces the likelihood of buckling failure in the specimens. Five specimens were tested and all exhibited acceptable failure modes. The results of the compressive testing campaign in the longitudinal and transverse orientations are illustrated in Figure 6.8.

As in the tensile experiments, characterization of the material strength under compression is difficult, due to challenges related to buckling and clamping. The strengths obtained during the experimental campaign were significantly lower than expected for a CFRP epoxy laminate containing Toray T700 fibers. For example, the transverse compressive strength obtained in the experiments was 84 MPa, while values in the range 186-290 MPa are more common (see [195, 111, 196, 197]), and are closer to the manufacturer provided value of 186 MPa. Similarly, longitudinal compressive strengths for this type of composite typically range from 1012 MPa to 1422 MPa, close to the manufacturer provided value of 1102 MPa. As such, for both the transverse and longitudinal compressive strength, the manufacturer provided values have been used in the numerical models, rather than the values obtained experimentally.
Laminate Off-Axis Response

To determine the in-plane shear strength and modulus of the composite laminae, ±45° tension tests were conducted. Following ISO 14129, laminates containing exclusively ±45° plies were manufactured according to the same methodology as in Section 6.3.2. Five specimens were extracted from the panels using a wet saw, according to the dimensions specified in the ISO 14129 standard, see Figure 6.9. Since expected failure loads were relatively low, end tabs were not required. Specimens were aligned in an MTS Criterion 300 kN universal testing machine, and then clamped using hydraulic wedge grips. The speed of testing was set to 2 mm/min. Strains parallel and perpendicular to the loading direction were tracked using a 2D digital image correlation system. The shear shear strength and shear moduli were determined according to Equations 6.1 and 6.2.

\[
\tau_{12M} = \frac{F_M}{2bh} \quad (6.1)
\]

where \( F_M \) is the load at failure, \( b \) is the width of the specimens, and \( h \) is the specimen thickness.

\[
G_{12} = \frac{\tau_{12}'' - \tau_{12}'}{\gamma_{12}'' - \gamma_{12}'} \quad (6.2)
\]

where \( \tau_{12}' \) is the shear stress at shear strain \( \gamma_{12}' = 0.001 \) and \( \tau_{12}'' \) is the shear stress at shear strain \( \gamma_{12}'' = 0.005 \). The results of the off-axis testing are illustrated in Figure 6.10.
Characterization of Interlaminar Properties

The material’s longitudinal, transverse, and shear fracture energies, as well as its interlaminar shear strength, were taken from the literature. Nonetheless, for completeness, the experimental methodologies used to obtain these values are very briefly described in this section.

**Longitudinal Fracture Energies** To determine the fracture energy associated with longitudinal failure compact tension and compression tests are used. There is no standardized methodology for these types of tests, but the principles are explained in detail in [134]. Cross-ply specimens are manufactured and an initial crack and two holes for loading pins are machined, see Figure 6.11. The geometry of the specimens differs depending on the loading mode - tension or compression. To test the specimens, they are placed in a universal testing machine such that equal and opposite forces are applied through the two holes in the specimen, propagating the initial sharp crack. Loading continues until the specimen fractures into two pieces. Subsequently, the critical stress intensity factor can be calculated based on
the maximum force attained, and the strain energy release rate can subsequently be determined from the value of the critical stress intensity factor [198].

Figure 6.11: Schematic of (a) a compact tension and (b) a compact compression test specimen. The arrows indicate the direction of loading.

Transverse Fracture Energies The mode I fracture energy of a fiber reinforced polymer composite is most commonly determined using a double cantilever beam (DCB) test (detailed in the ASTM D5528 standard [199]). DCB specimens typically measure 25 mm wide and 356 mm long, and contain an insert which creates an initial crack (i.e. debond) at the specimen mid-plane, see Figure 6.12. End blocks are adhered to the ends of the specimen, through which a tensile load is applied, acting normal to the crack plane. As the specimen is pulled apart, the crack length grows. Using the
force measured by the testing machine and the crack length measurement, the fracture energy required for the initiation and propagation of a mode I crack can be determined.

Figure 6.12: Schematic of the test setup for a standard double cantilever beam test to determine the mode I fracture toughness of an FRP composite.

To determine the mode II fracture energy, an end-notched flexure (ENF) test must be conducted. In similar fashion to a DCB test, specimens are manufactured which contain an initial crack, typically measuring approximately 60 mm in length, see Figure 6.13. These specimens are subjected to what is essentially a 3 point bending test, in which a cylinder loads the specimen flexurally, creating shear forces at the crack tip. The force and crack length are monitored throughout the duration of the experiment to calculate the mode II fracture toughness.

Figure 6.13: Schematic of the test setup for an end-notched flexure test to determine the mode II fracture toughness of an FRP composite.

**Interlaminar Shear Strength**

To determine the shear strength of the interface between plies in a composite laminate, an interlaminar shear test is typically conducted. The complete standardized testing procedure is outlined in ISO 14130 [200]. In brief, a short but thick specimen, the dimensions of which are given in Figure 6.14, is subjected to three point bending. Due to the large thickness of the specimens,
this results in high shear loads at the beam ends. Typically, the laminate will fail at the interfaces between plies at the ends of the specimen. The interlaminar shear stress can be determined on the basis of the peak load and the dimensions of the specimen.

![Schematic of the test setup for an end-notched flexure test to determine the mode II fracture toughness of an FRP composite.](image)

**Figure 6.14:** Schematic of the test setup for an end-notched flexure test to determine the mode II fracture toughness of an FRP composite.

### 6.3.3 FEA Implementation

The tensile tests were simulated using the multiscale modeling framework described in Chapter 4. Material properties were determined experimentally according to the relevant standards and are summarized in Table 6.1 with the exception of the longitudinal fracture toughnesses which were taken from the literature. \( G_{II} \) was determined based on the mode II component of the fracture toughness and the fracture angle under pure transverse compression (53°) [146]. The undulation ratio and volume fractions of the unit cell constituents were estimated from SEM micrographs. For the press manufactured carbon epoxy laminates studied in this chapter the undulation ratio was 0.0683.

Coupons were discretized with 8 node reduced integration linear solid C3D8R elements. Mesh seeds were defined such that the element sizes were approximately equal to the size of the mesoscale unit cells and the length of the undulation (≈ 1.5mm). This is the optimal element size to ensure a realistic macro-to-meso strain transformation [133]. Note that due to the automatic partitioning of the complex geometry some elements may be smaller than the mesoscale unit cell. Mesh topology was dependent on the laminate stacking sequence. Specimens were automatically meshed using swept meshes and the advancing front algorithm. As a result of this process, the quasi-isotropic specimen mesh was largely unstructured, while the cross-ply specimens exhibited a much more regular mesh aligned with the
Table 6.1: Mechanical properties of the SHD Composites VTC401.

<table>
<thead>
<tr>
<th>Property</th>
<th>Value</th>
<th>Source</th>
</tr>
</thead>
<tbody>
<tr>
<td><strong>Elastic properties</strong></td>
<td></td>
<td></td>
</tr>
<tr>
<td>$E_{11}$ (GPa)</td>
<td>124.35</td>
<td>ISO 527-5 [14]</td>
</tr>
<tr>
<td>$E_{22} = E_{33}$ (GPa)</td>
<td>7.231</td>
<td>ISO 527-4 [201]</td>
</tr>
<tr>
<td>$G_{12} = G_{31}$ (GPa)</td>
<td>3.268</td>
<td>ISO 14129 [16]</td>
</tr>
<tr>
<td>$G_{23}$ (GPa)</td>
<td>2.638</td>
<td>estimated as in [202]</td>
</tr>
<tr>
<td>$\nu_{12} = \nu_{31}$ (-)</td>
<td>0.339</td>
<td>ISO 527-4 [201]</td>
</tr>
<tr>
<td>$\nu_{23}$ (-)</td>
<td>0.374</td>
<td>estimated as in [203]</td>
</tr>
<tr>
<td><strong>Strengths</strong></td>
<td></td>
<td></td>
</tr>
<tr>
<td>$X_T$ (MPa)</td>
<td>2550</td>
<td>[193]</td>
</tr>
<tr>
<td>$X_C$ (MPa)</td>
<td>-1102</td>
<td>[204]</td>
</tr>
<tr>
<td>$Y_T = Z_T$ (MPa)</td>
<td>44</td>
<td>ISO 527-4 [201]</td>
</tr>
<tr>
<td>$Y_C = Z_C$ (MPa)</td>
<td>-184</td>
<td>[204]</td>
</tr>
<tr>
<td>$S_{12} = S_{31}$ (MPa)</td>
<td>55</td>
<td>ISO 14129 [16]</td>
</tr>
<tr>
<td>$S_{23}$ (MPa)</td>
<td>83</td>
<td>[204]</td>
</tr>
<tr>
<td><strong>Fracture energies</strong></td>
<td></td>
<td></td>
</tr>
<tr>
<td>$G_1$ (N/mm)</td>
<td>134.0</td>
<td>[196]</td>
</tr>
<tr>
<td>$G_1^C$ (N/mm)</td>
<td>95.0</td>
<td>[196]</td>
</tr>
<tr>
<td>$G_2$ (N/mm)</td>
<td>0.38</td>
<td>[205]</td>
</tr>
<tr>
<td>$G_6$ (N/mm)</td>
<td>1.62</td>
<td>[205]</td>
</tr>
</tbody>
</table>

geometry of the tows, see Figure 6.15. Simulation runtimes for the 100 mm × 40 mm AP-PLY specimens varied from 33 mins to 600 mins running in parallel on 4 cores in a Intel Xeon E3-1230 Windows machine depending on the AP-PLY configuration.

It is important to acknowledge that damage localization and mesh dependency are the result of deficiencies in classical local continuum damage mechanics models [206]. Since cracks are smeared over the length of an entire element, the size and shape of the elements in a mesh can affect the formation and propagation of cracks within it. Claudio Lopes first exploited this mechanism by aligning and biasing lamina meshes in the fiber direction, improving the accuracy crack path predictions [140]. The approach has since been used to good effect by a number of other researchers [207, 141, 208]. However, a systematic review of the effect of mesh alignment in the simulation of composite materials, by Millen et al., found that mesh alignment was unnecessary in cases where matrix crack paths were not established a priori, e.g. un-notched tension or low velocity impact [209]. As such, in the present work, no effort has been made to purposefully align meshes with the fiber
direction. In some cases, Abaqus’ automated meshing algorithms spontaneously generated meshes aligned with the fiber direction. As a result of this process, the quasi-isotropic specimen mesh was largely unstructured, while the cross-ply specimens exhibited a much more regular mesh aligned with the geometry of the tows. Midplane symmetry was used to reduce computational cost. Specimens were fully clamped at one end and a 0.5 mm/s velocity was imposed at the opposite boundary to simulate the quasi-static experiment. The internal and kinetic energy in the model were evaluated to ensure inertial forces were negligible.

6.4 Results and Discussion

6.4.1 Experimental results

Figure 6.16 shows representative stress-strain curves of the baseline and AP-PLY laminates and the results are summarized in Table 6.2. Laminate moduli were evaluated over a strain range from 0.002 to 0.008, prior to damage initiation. No significant difference was found between the initial
The result is consistent with previous studies of AP-PLY laminates which have reported minor changes in undamaged in-plane stiffness in spite of the presence of fiber crimp [10, 24, 26].

In terms of strength, the AP-PLY process was found to reduce the strength of the cross-ply laminates by as much as 16.7%. The discrepancy can be attributed to stress concentrations induced by the through-thickness fiber undulations (see Figure 6.19). Post-mortem examinations of the specimens indicated that ultimate failure of the specimens occurred along tow boundaries, never splitting a tow in the direction parallel to the fibers. Additional stress concentrations were also detected near the clamped ends of the specimens due to the high gripping pressures used to prevent slippage of the large non-standard width specimens, in spite of the use of larger end tabs. This was not an issue for the baseline specimens whose dimensions conformed to the ISO standard.

In contrast to the cross-ply specimens, the averaged quasi-isotropic AP-PLY specimen strength was 7.6% higher than the baseline configuration. Notably, there was a distinct kink in the stress-strain response of the baseline specimens at a load of approximately 500 MPa. This softening behavior was not observed in the AP-PLY specimens, which exhibited linear elastic behavior up to final failure. The non-linear behavior of the quasi-isotropic baseline specimens is attributed to more extensive matrix cracking in the specimens prior to final failure, see Figure 6.18. In the AP-PLY specimens the $\pm 45$ and 90 degree tows do not form a continuous ply from one (clamped) end of the specimens to the other. As a result of the discontinuity of these tows, they tend not to form matrix cracks parallel to the local fiber direction within the tows themselves. Instead, these tows debond from the rest of the laminate, i.e. matrix cracks only form between tows. While the baseline cross-ply specimens also exhibit matrix cracking, they do not exhibit the same softening behavior as the baseline quasi-isotropic composite because they contain a greater proportion of load oriented plies. Similar behavior is
observed in woven composites, in which extensive matrix cracking does not result in a non-linear stress strain response [210].

6.4.2 Numerical response: AP-PLY composites

Table 6.2: Experimental and numerical moduli and strengths for baseline and AP-PLY laminates.

<table>
<thead>
<tr>
<th>Configuration</th>
<th>Modulus (GPa)</th>
<th>Strength (MPa)</th>
</tr>
</thead>
<tbody>
<tr>
<td></td>
<td>Exp. FEA</td>
<td>Exp. FEA</td>
</tr>
<tr>
<td>XP&lt;sub&gt;AP-PLY&lt;/sub&gt;</td>
<td>65.27 ± 3.53</td>
<td>61.71 ± 47.55</td>
</tr>
<tr>
<td>XP&lt;sub&gt;base&lt;/sub&gt;</td>
<td>63.59 ± 1.23</td>
<td>63.26 ± 55.61</td>
</tr>
<tr>
<td>QI&lt;sub&gt;AP-PLY&lt;/sub&gt;</td>
<td>44.96 ± 0.57</td>
<td>42.60 ± 28.85</td>
</tr>
<tr>
<td>QI&lt;sub&gt;base&lt;/sub&gt;</td>
<td>44.56 ± 0.95</td>
<td>44.25 ± 29.79</td>
</tr>
</tbody>
</table>

The numerical framework described above was used to simulate the tensile response of the AP-PLY and baseline panels. Figure 6.17 compares the experimental and numerical stress-strain curves and results are summarized in Table 6.2. Laminate moduli were evaluated over a strain range from 0.2% to 0.8%, prior to damage initiation. The tow region constitutive model was able to accurately predict the stiffness and strength of the baseline laminates, and their failure modes, see Fig. 6.18. The cross-ply specimens failed simultaneously at different points, both in the center and near the clamps. This phenomena was well captured by the numerical model. The progressive ply failure in the quasi-isotropic specimens was also well predicted. Matrix cracking occurred at low strains in the 90° plies, followed by the ±45° laminae, spreading through the entirety of each ply. Final failure of the the specimens was caused by fiber fracture in the 0° layers, with simultaneous perpendicular and diagonal cracks.

The prediction of the mechanical response of the cross-ply AP-PLY panels was in very good agreement with the experimental results. The discrepancies between the experimental and numerical stiffness and strength amounted to 5.5% and 5.7% respectively. Reasonable agreement was also obtained for the response of the quasi-isotropic panel. Stiffness was estimated by the numerical model to within 5.2% of the experimental modulus. However, as the complexity of the internal architecture increased, the numerical model tended to underestimate the strength, by approximately 7.4%.

The models presented a linear-elastic response until the onset of matrix cracking. As loads were increased, strain concentrations developed at the through-thickness tow undulations due to the differences in stiffness between adjacent tows with different out-of-plane orientations. Figure 6.19 compares the strain field on the surface of a cross-ply specimen (obtained using DIC)
with the numerical model predictions at 1.1% nominal strain. The size and location of the strain concentrations were captured relatively accurately by the numerical model, even using a coarse mesh. Additional strain maps acquired from other cross-ply specimens present similar strain concentrations at tow undulations, while there are no significant strain concentrations observed in quasi-isotropic specimens, see Appendix A.3.

The numerical models predicted the location and angles of the planes along which the specimens fractured, which were always aligned with the undulation regions along transverse tow boundaries. Figures 6.20 and 6.21 compare the experimentally observed fracture mechanisms with those predicted by the numerical model. Despite the relatively coarse mesh, the model was able to predict the crack paths accurately.

In the case of the quasi-isotropic laminate, failure occurred at a $\pm 45^\circ$ angle. Fiber failure also happened on this inclined plane even in tows oriented in the loading direction, where failure would normally be expected to occur on a plane normal to the tow. In the cross-ply specimens failure occurred on a plane orthogonal to the loading direction aligned with one or more of the
Figure 6.18: Numerical predictions and experimental observations of damage in cross-ply (a) and quasi-isotropic (b) baseline laminates. Note the matrix cracking in the transverse and ±45 degree tows.

Figure 6.19: (a) Experimental measurements and (b) numerical predictions of the strain field on the surface of a cross-ply laminate at 1.1% nominal strain. (c) Finite element discretization divided in different unit cell regions.

undulation regions. Matrix cracking was only predicted in the vicinity of the tow boundaries, instead of spread over the entirety of each transverse ply, as in the case of the baseline laminates. For example, the transverse crack which initiated on the failure plane and runs along the tow boundary was well captured by the matrix cracking criteria in the numerical model.

The accuracy of the numerical models decreased as the complexity of the internal architecture rose due to the limitations of the homogenization approach. As a result of the assumption of isostrain conditions in the unit cells, the numerically predicted damage evolution within each element may not accurately represent the actual damage accumulation within a laminate. For example, in undulation regions containing tow and pure resin constituents, the strains on the tows may be artificially high, as the homogenized stiffness of these elements is reduced by the presence of the pure resin constituent.
Figure 6.20: Numerical predictions and experimental observations of damage in cross-ply AP-PLY laminates. The finite element mesh used to discretize the top ply of the specimen is illustrated in (a). Subfigures (b) and (c) exhibit the damage envelopes corresponding to fiber damage and transverse/through-thickness damage, respectively. Deleted elements are not shown. Experimentally observed failure mechanisms are exhibited in (d).

As a result, fiber damage may be triggered in the tow constituent marginally earlier than in reality, resulting in lower strengths than expected. In addition, the use of volumetric stress averaging means that damage occurring in one constituent has the effect of reducing the stiffness of the entire element, when in reality damage in one constituent does not necessarily affect the mechanical properties of the other. This effect can be observed in the numerical response of the quasi-isotropic laminates, which diverges from the experimental response at high loads because of matrix damage accumulation. It should be noted that the non-linear response of the resin and the shear response of the tows were not implemented. Incorporating these phenomena into the constitutive models could potentially improve strength predictions, particularly in the case of the quasi-isotropic laminates (or other composites with high resin content). However, while it is important to acknowledge the potential for reduced accuracy in the predicted strength of AP-PLY laminates with a large number of different tow orientations, the quasi-isotropic laminates studied in the present work represent the current state-of-the-art in terms of geometric complexity [10, 160, 24, 26].

Despite the aforementioned limitations, the multiscale homogenization/CDM framework presented in this study predicts the mechanical response of the AP-PLY composites with reasonable accuracy and at a reduced computational cost compared to microscale or FE² approaches [211, 158, 212, 157, 159]. The automated pre-processing (comprised of specimen partitioning, material property assignment, and meshing) is 6 times faster than the approach developed by Li et al. [29] when performed on a 4 core (Intel Xeon E3-1230) Windows machine with 16 GB of
Figure 6.21: Numerical predictions and experimental observations of damage in quasi-isotropic AP-PLY laminates. The finite element mesh used to discretize the top ply of the specimen is illustrated in (a). Subfigures (b) and (c) exhibit the damage envelopes corresponding to fiber damage and transverse/through-thickness damage, respectively. Deleted elements are not shown. Experimentally observed failure mechanisms are exhibited in (d).

RAM. Furthermore, the use of a coarse mesh ($\approx 1.5$ mm characteristic length) to reproduce the response of the undulations drastically reduces the computational cost of the models when compared against microscale approaches that require meshes of the order of 0.07 to 0.35 mm to discretize the fiber curvature, as in studies of 3D woven and braided composites [213, 214, 212, 156, 215, 155]. While these microscale approaches might be able to replicate the stress-state in an AP-PLY composite with greater accuracy, the high number of degrees of freedom required preclude their use in the analysis of large structural components [216]. The methodology presented in this paper strikes a balance between accuracy in the prediction of the stress-strain response and computational efficiency.

6.4.3 Effect of specimen size

As discussed previously, the AP-PLY quasi-isotropic specimens characterized in this study contained only an “approximate RVE” as a true RVE could not be identified for laminates with this configuration. In spite of this, the experimental results exhibited low levels of variability: the coefficients of variation of the modulus and strength amounted to 1.7% and 4.9% of their mean values, respectively. To evaluate whether the dimensions of the specimen impact the numerical model results, various virtual specimens with dimensions ranging from $30 \times 30$ [mm$^2$] to $80 \times 80$ [mm$^2$] were simulated.

Figure 6.22 illustrates the stress-strain response of the virtual specimens. As in the experimental results, the use of an approximate RVE can be
observed to have a minimal impact on laminate performance. Even the specimen size that is smaller than the approximate RVE, $30 \times 30$ [mm$^2$], produces results in line with the larger specimens. The mean failure stress was found to be 673 MPa with a coefficient of variation of only 2.5%. The averaged laminate stiffness is 44.21 GPa with a coefficient of variation of 1.8%. These results suggest the mechanical properties of an AP-PLY component with no strictly identifiable RVE can be determined experimentally or numerically within a reasonable scattering compatible with the requirements of primary structural components.

Figure 6.22: On the left, the numerical stress-strain curves for various sizes of QI$_{AP−PLY}$ specimens. On the right, the corresponding specimen bounds overlaid on the mesoscale geometric idealization of a quasi-isotropic AP-PLY laminate.

6.4.4 Effect of tow-skipping parameter

To investigate the effect of the tow skipping parameter on laminate performance, numerical models of cross-ply and quasi-isotopic laminates were generated in which either 1 or 5 tows were skipped (versus the 3 tow gap used for the experimental characterization). This parameter determines the density of the undulation regions in a laminate, hence a low tow-skipping value implies a higher number of undulations.

Laminate stiffness was unaffected by the tow skipping parameter for both the quasi-isotropic and cross-ply configuration (in agreement with the experimental results). Furthermore, the tow skipping parameter had a negligible impact on the strength of the cross-ply laminate: the maximum stresses were almost identical for all three laminate configurations (Figure 6.23a). In the quasi-isotropic configuration however, increasing the number of skipped tows led to an increase in laminate strength (Figure 6.23b). In AP-PLY laminates, the magnitude of the stress intensity factor resulting from an undulation region is dependent on the mismatch angle between the regions’ micro-constituents. In the cross-ply laminate the stress intensity...
factor at all tow undulations was the same, and the undulations were sufficiently spread out so they did not interact. In the quasi-isotropic laminate, however, reducing the spacing of the tows resulted in interactions between the different tow undulation regions, increasing the stress intensity factor and thereby negatively affecting laminate strength.

These results show the potential of the numerical framework to analyze the influence of preforming parameters in the laminate’s mechanical performance. In particular, it could be used to optimize the structural response of components manufactured by automated fiber placement subject to complex loading states, such as low-velocity impact, a potential application for the aerospace sector.

![Figure 6.23](image)

**Figure 6.23:** Predicted stress-strain response of (a) cross-ply and (b) quasi-isotropic AP-PLY laminates with different numbers of tows skipped between tows placed in the same pass.

### 6.4.5 Effect of the undulation ratio

The undulation ratio used in this study was obtained using SEM micrographs. For the given material and processing method, the undulation ratio was found to be relatively constant, varying by $\pm 10\%$ from the mean value of 0.0683. A sensitivity study was conducted to determine the effect of the undulation ratio on the numerical model predictions. In the model this was implemented by changing the length of the undulation and the adjacent resin rich regions. This methodology resulted in a variation in the total fiber volume fraction of the laminate of $\pm 0.3\%$ from the initial 53.2\%, which was within the bounds of the experimental scattering. The results are illustrated in Figure 6.24.

Laminate moduli were found to be relatively insensitive to changes in the undulation ratio. The most likely explanation for this result is that changes to the undulation ratio have two competing effects. First, as previously mentioned, increasing the undulation ratio increases the laminate FVF marginally. However, this change also increases the out-of-plane inclination of
the fibers in the undulation regions, reducing the stiffness of these regions and in turn the stiffness of the laminate as whole.

In terms of strength, increasing the undulation ratio was found to have a negative impact on the strength of the laminate for both the cross-ply and quasi-isotropic laminates. As mentioned previously, increasing the undulation ratio results in larger out-of-plane fiber inclinations in the undulation regions, leading to higher stress concentrations. As a result, the longitudinal fiber failure criteria are triggered at lower nominal stresses in laminates with high undulation ratios. This effect is more significant in the cross-ply laminates where stress-concentrations are higher due to the greater ply mismatch angles. These results suggest that the AP-PLY process is best suited to thin ply composites in which the amplitude of the fiber undulation, and therefore the undulation ratio, is very small. High consolidation pressures during curing are likely to have a beneficial effect on laminate strength, for the same reason.

Figure 6.24: Predicted stress-strain response of (a) cross-ply and (b) quasi-isotropic AP-PLY laminates with different undulation ratio value.

6.5 Conclusions

In summary, for a given undulation ratio, the AP-PLY process was found to have a negligible impact on laminate stiffness, regardless of the AP-PLY configuration. In spite of the fiber crimp present in the AP-PLY laminates, they retained the excellent in-plane stiffness characteristic of conventional angle-ply laminates. The effect of AP-PLY preforming on laminate strength was found to depend on the layup: cross-ply laminates were found to be sensitive to the stress concentrations introduced by AP-PLY preforming, resulting in a lower strength compared with their non-AP-PLY counterparts. However, the quasi-isotropic AP-PLY laminates exhibited higher strengths than the baseline laminates, possibly due to the capacity of the through-thickness reinforcements to arrest the propagation of matrix cracking and constrain it to the tow boundaries.
The multiscale continuum damage mechanics framework was able to accurately predict laminate stiffness and strength, to within 5.5% and 5.7% of the experimental results for the cross-ply laminate, and 5.2% and 7.4% for the quasi-isotropic laminate, respectively. Failure mechanisms were similarly well captured by the modeling framework, which correctly predicted the location and orientation of the crack planes.

An investigation on the effect of specimen size on laminate performance demonstrated that coherent results can be attained using "approximate" RVEs. Mechanical properties were consistent for all specimen sizes and can be used for future damage tolerant design purposes, independent of the dimensions of the component. Secondly, from the parametric study on the effect of the tow-skipping parameter, it is evident that increasing the number of gaps left between tows placed in the same pass may increase the strength of quasi-isotropic laminates. Cross-ply laminate strength and stiffness were unaffected. Lastly, lower undulation ratios, can increase laminate strength in both quasi-isotropic and cross-ply configurations. Stiffness was unaffected by changes to the undulation ratio within the analysed range.

To encourage future research and the reuse of the experimental results presented in this chapter, the complete set of raw data has been uploaded to a public data repository.
Chapter 7
Dynamic Mechanical Properties

7.1 Introduction

Having established the quasi-static in-plane tensile response of AP-PLY composites, this chapter investigates the response of this family of composite materials to high strain rate tensile loading using split-Hopkinson bar experiments. Reliable data on the dynamic properties of AP-PLY composites are sparse mainly due to experimental difficulties [85]. Conventional SHBs are limited to specimens with relatively small dimensions, however, the large representative volume elements (RVE) of AP-PLY composites, in the present study measuring approximately 40 mm x 40 mm, require the use of more advanced non standard equipment. First, exceptionally large bar diameters and pulse magnitudes are required to generate forces sufficient to break the specimens. Second, a large pulse duration with a progressive rising time is required to avoid the premature failure of the brittle composite [217]. An additional consequence of the brittle nature of carbon fiber composites is that they are more sensitive to mechanical gripping, which may result in measurements of strength that underestimate the actual material strength [218]. Since the gauge length and distance between bar ends are much greater in large SHBs, it is more difficult to achieve force equilibrium, requiring more detailed analysis to ascertain the validity of the experimental data [219]. In summary, characterizing the tensile high strain-rate behavior of brittle materials with large RVEs, such as AP-PLY composites, using specimen sizes representative of global material behavior, is highly demanding of both equipment and analytical expertise.

In this context, numerical simulations can be used in tandem with experiments to determine dynamic material properties. This methodology mitigates the uncertainties arising from boundary-conditions [220, 221] and can provide valuable insights into the triggering sequence of failure mechanisms that can not be captured experimentally due to the time-resolution limitation of high-speed imaging techniques and post-mortem fractography [67]. In the present work, the high strain rate mechanical properties of AP-PLY composites were determined using a hybrid
experimental and numerical approach. The response of the AP-PLY laminates were compared with those of conventional angle ply composites to quantify the effect of the through thickness fiber undulations.

### 7.2 Materials And Methods

#### 7.2.1 Materials

The AP-PLY configurations of the specimens for the dynamic material characterization were the same as those used in the quasi-static experiments. The tow stacking sequence and tow skipping parameter were left unchanged, but the specimen dimensions were altered to match the requirements of the testing apparatus. As described in Chapter 6, The cross-ply and quasi-isotropic specimens, both conventional and AP-PLY, were laid up onto moulds by hand using SHD Composites VTC401 unidirectional prepregs slit into 10 mm wide tows, with a gap of three tow widths left between tows placed in the same pass. The 300 mm x 300 mm panels were cured in a hot press under 4 bars of pressure at 120° C for 120 minutes. Cross-ply and quasi-isotropic specimens contained 20 and 32 plies (equivalent) respectively, resulting in average thicknesses of 4.14±0.05 mm and 4.20±0.16 mm for the AP-PLY and baseline cross ply specimens, respectively, and 6.95±0.15 mm and 6.88±0.04 mm for the baseline and AP-PLY quasi-isotropic laminates. The AP-PLY process did not impact laminate thickness or fiber volume fraction.

Specimens were extracted from the laminates using computerized numerical control (CNC) machining according to the design reported in Figure 7.1. The dogbone specimen geometry was chosen to ensure the size of the clamping area was sufficient to prevent slippage of the specimen during the tests. A parametric study was conducted to determine the radius that would minimize stress concentrations at the shoulder, although it was not possible to eliminate these concentrations entirely. Initial attempts at using a waterjet to cut out the specimens resulted in delamination and debonding of the end tabs. Switching to CNC machining for specimen extraction was found to produce high quality specimens with no internal damage.

The width of the gauge section varied between 30 mm and 40 mm to ensure the response was representative of the mechanical behavior of the parent laminates [222]. Aluminium end tabs were adhered to the specimens to improve stress transfer between the clamps and the specimen. The end tabs were extracted using a water jet and adhered to the specimens using a two part epoxy adhesive (Araldite Standard/2021 or Permabond ET5428). Clamps were used to provide a consolidating pressure while the adhesives cured in an oven according to the manufacturer’s recommended cure cycle (60°C for 1 hour). Excess adhesive was removed by machining to achieve the desired dimensional tolerances for testing. The tabbing procedure is
Figure 7.1: Dynamic tensile testing specimen dimensions (in mm).

illustrated in Figure 7.2.

Figure 7.2: End tabbing procedure for the dynamic tensile characterization specimens.

7.3 Experimental Setup

The experimental campaign was carried out at the facilities of the European Laboratory for Structural Assessment: Large Hopkinson Bar Facility (ELSA-HopLab) [223]. The Split Hopkinson Bar (SHB) at the HopLab facility consists of steel incident and transmitter bars 72 mm in diameter and 12 m and 90 m in length respectively, see Fig. 7.3. It is the world's largest Hopkinson Bar, specifically designed to test components, sub-assemblies and large material samples. In the present work, loading pulses were generated by first pretensioning a steel cable, then suddenly releasing it using explosive bolts. This generated a stress pulse with an input velocity of $5 \text{ ms}^{-1}$ and a measured strain rate of approximately $30 \text{ s}^{-1}$ over the specimen's 40 mm gauge length.
The specimens were threaded into the bars with custom clamps, see Fig 7.4. Four M10 bolts provided sufficient clamping force to prevent slippage of the specimens. The friction coefficient between the specimen and the clamps was further increased by knurling the surface of the grips. Four pins were used to arrest movement between the upper and lower clamp plates. A minimum of six specimens per configuration were tested.

The incident and transmission bars were equipped with semiconductor strain-gauges connected in a full-bridge configuration. Signals were post-processed in MATLAB [224]. Note that the SHB at the HopLab facility was designed to test specimens with a failure load on the order of 500 kN, but since manufacturing AP-PLY composite specimens of sufficient thickness to reach such a high failure load was not feasible, the final specimen design resulted in a failure load of approximately 150 kN. As a result, the analysis
utilized exclusively the data from the semiconductor strain gauges, owing to their significantly higher sensitivity. While semiconductor strain gauges exhibit some non-linearity over their measurement range, this could be mitigated by linearizing over a small application range.

In addition to measuring strain using gauges, a 2D digital image correlation system was used to capture the strain field in the specimens under loading [225]. The use of DIC is especially important because, as a result of the brittle nature of the specimens and the potential for fixture flexure, the global displacements measured by the gauges may not be representative of the strain field in the specimen. A high-speed camera (Photon SA1.1) was used to capture the deformation of the specimens at 50,000 frames per second and at a resolution of 512 x 208 pixels. Cold lights (Veritas Constellation 120) were used to illuminate the specimen surface during the test. Figure 7.5 illustrates the DIC setup for the SHB tests. The specimens were speckled by hand using a white marker, resulting in an average speckle diameter of 1 mm. Given the resolution of the camera, this resulted in approximately 5 pixels per speckle. At least 3 pixels are required per speckle to avoid anti-aliasing issues [226]. Post-processing of the DIC images was conducted using the MatchID software package [227]. Subset and step size were set to 15 and 17, respectively, and interpolation was conducted using a bi-cubic spline algorithm. Strains were determined using a filter size of 15. The noise floor was estimated to be 50 microstrain, based on the standard deviation of the axial strain in an image of an unstressed specimen [227]. The spatial resolution of the DIC system — estimated on the basis of the subset, step, and strain filter size — was 3 mm.

Figure 7.5: DIC setup for the SHB experiments, illustrating (a) camera positioning orthogonal to the specimen surface, (b) the lighting setup for the experiments, and (c) the DIC speckle.

### 7.4 Data Processing

In a quasi-static tensile test, the crosshead velocity is low enough to ensure that the effect of stress waves is negligible. As such, the stress-strain can be
easily deduced from the force displacement data provided by the testing apparatus. The analysis of data from a split-Hopkinson bar test is significantly more complex. The stress-strain response of the material must be evaluated based on the incident, reflected, and transmitted stress waves measured in the incident and transmission bar. Figure 7.6a illustrates the raw signals from the 6 strain gauges adhered to the input and output bars (3 measurement points for each bar, ohmic and semiconductor strain-gauges). Note there is a slight offset due to the static preload of the bar. The resulting incident, transmitted and reflected waves, illustrated in Figure 7.6b were calculated using a deconvolution algorithm capable of compensating for wave dispersion distortions [228]. There are some oscillations in the input force history which can be attributed to small propagation errors from the deconvolution algorithm and the additional masses of the specimen clamps.

Figure 7.6: Raw strain gauge signals (a) and the resulting waves (b) calculated using a deconvolution algorithm.

The force-time data obtained using the strain gauges on the bars were combined with the displacement-time data obtained using DIC to generate the equivalent force-displacement curves. Linear interpolation was used to account for the higher data recording frequency of the strain sensors relative to the high speed cameras. Since the force in the transmission bar is typically measured with greater accuracy, this force was assumed to act over the entirety of the specimen. A potential limitation of this approach is that the forces are calculated in the bars at the ends of the specimen, while the strains are measured over the gauge section of the specimens, resulting in a small delay. This issue can be resolved by synchronizing the two measurements using the normalized stress and strain curves from the DIC system and the SHB gauges, see Figure 7.7a. The resulting stress-strain relationship, illustrated in Figure 7.7b, further emphasizes the importance of using DIC for strain measurement. Due to the flexure of the clamping system, the material response obtained using data from the SHB strain gauges is overly soft, and does not represent the true behaviour of the specimens.
Figure 7.7: On the left, the synchronization of SHB and DIC data using normalized stress and strain data. On the right, the stress-strain response of a quasi-isotropic AP-PLY specimen using strain data obtained using strain gauges and DIC.

7.4.1 Validation

The dynamic force equilibrium between bars was analyzed to validate the stress-strain curves. Figure 7.8a plots the force in the input and output bars against time for one of the cross-ply specimens. Despite the noise of the input bar signal, force equilibrium is reached in the 3.2 ms to 3.8 ms time interval. Analysis of the strains in the specimen at the input and output ends using DIC confirms the state of equilibrium in the specimen during the test, see Figure 7.8b.

Figure 7.8: Validation of state of equilibrium: (a) input and output bar forces as measured using SHB strain gauges, and (b) strain at the input and output ends of a specimen determined using DIC. Adapted from [17]
7.5 FEA Implementation

As in the previous chapter, the experiments were simulated using the multiscale AP-PLY modeling framework detailed in Chapter 4. The material properties used were the same as in the quasi-static simulations, see Table 6.1. At the maximum strain rates achieved in this study (\(30 \text{s}^{-1}\)) neither the longitudinal nor the transverse properties of the tows are expected to differ significantly from the quasi-static properties [229, 230].

Specimens were discretized using 8 node reduced integration linear solid C3D8R elements. Since the stacking sequence in each specimen is repeated and symmetric, an 8-ply sub-laminate was sufficient to represent the response of each 20-32 ply thick laminate. To further reduce computational cost, only the central (non-clamped) portion of the specimens was modeled. As before, element dimensions were chosen to match the size of the mesoscale unit cells, 1.5 mm. This optimal element size ensured a realistic macro-to-meso strain transformation [133].

![Figure 7.9](image)

(a) (b)

Figure 7.9: On the left, the boundary conditions applied to the specimens in the models of the dynamic tensile experiments. On the right, the experimental and numerical input and output bar velocities used to define the boundary conditions in the numerical simulations.

A zero displacement boundary condition was applied to the left hand side of each of the specimens. A resultant velocity boundary condition was applied to the other end of the specimens based on the input and output bar velocities measured experimentally by the strain gauges, see Figure 7.9. This approach, previously validated in studies of titanium alloys [231] and woven composites [7], is computationally efficient as it does not require the simulation of the entire SHB system. To avoid numerical instability issues at the onset of damage, Abaqus’ default bulk viscosity option was used, and enhanced hourglass and distortion control were enabled. The energies in the simulation were monitored to ensure the artificial strain energy resulting from the section controls did not exceed 2% of the total energy in the system [149].
The kinetic energy represented, on average, 0.6% of the total energy in the model.

7.6 Results and Discussion

7.6.1 Dynamic response of AP-PLY composites

This section analyzes the performance of AP-PLY composites at high strain rates by considering the results of the experimental and numerical studies together. Figure 7.10 presents the experimental and numerical stress-strain response of the AP-PLY specimens subjected to tensile loading at an average strain rate of $30 \text{s}^{-1}$. Despite the interaction of secondary waves and the initial noise of the pulse, the experiments show good repeatability, see Table 7.1. The moduli of the specimens were estimated over the strain interval from 0.4% to 0.8%, after dynamic force equilibrium was established and avoiding the non-linear response outside of this strain range.

Numerical models exhibit good agreement with the experimental results. Predictions of the strengths of the cross-ply and quasi-isotropic specimens are within the experimental error bounds, differing by -5.2% and 7.0% from the experimental values, respectively. The discrepancy between the numerical prediction and experimentally determined quasi-isotropic stiffness is similarly small at only 1.3%. In all specimens, baseline and AP-PLY, the gradient of the stress-strain response increases incrementally at strains of around 0.2 - 0.4%, a phenomenon not registered at material level by the numerical model. Rather than reflecting a change in material behavior, the change in slope, which occurs before dynamic equilibrium is established, is thought to result from localized stress concentrations in the clamped area. These concentrations resulted in the premature failure of 4 cross-ply specimens in the bolted joints of the grips instead of in the gauge section.

It is worth noting that as a result of the assumptions made in the modeling methodology, the simulations cannot capture relative motion between constituents within undulation regions. As a result of the isostrain assumption, each constituent within an undulation region undergoes the same deformation, which limits the ability of the model to accurately capture strain gradients at the microscale. Additionally, delamination between tows in the undulation regions cannot be captured, because the interface is contained within a unit cell and there is no numerical modeling mechanism to support crack initiation and propagation within an element. These deformation mechanisms could be more accurately modeled using a fully microscale model, or through further development of the present model using an approach such as XFEM to permit discontinuities within elements. However, these approaches would increase the computational cost of the model and are beyond the scope of the present work. In the case of tensile loads, delamination is observed as a secondary failure mechanism, and the existing
Table 7.1: Experimental and numerical results of AP-PLY composites at high strain rates.

<table>
<thead>
<tr>
<th>Configuration</th>
<th>Modulus (GPa)</th>
<th>Strength (MPa)</th>
</tr>
</thead>
<tbody>
<tr>
<td></td>
<td>Experimental</td>
<td>Numerical</td>
</tr>
<tr>
<td>XP&lt;sub&gt;AP-PLY&lt;/sub&gt;</td>
<td>69.38 ± 3.29</td>
<td>64.91</td>
</tr>
<tr>
<td>QI&lt;sub&gt;AP-PLY&lt;/sub&gt;</td>
<td>47.41 ± 2.63</td>
<td>46.78</td>
</tr>
</tbody>
</table>

methodology captures the deformation mechanisms with a reasonable level of accuracy.

Fig. 7.10: Comparison of experimental and numerically predicted stress-strain response of AP-PLY (a) cross-ply and (b) quasi-isotropic laminates.

Fig. 7.11 and 7.12 compare the experimental and numerical evolution of the strain fields for AP-PLY laminates. The symmetry of the strain field measured using DIC, in which similar strain values are measured at both clamped edges, lends further credence to the assumption of force equilibrium in the specimen. The strain fields measured on the surface of the specimens depend on each laminate’s stacking sequence. The cross-ply laminate’s strain field is homogeneous over the central region, but strain gradients develop at the shoulders of the specimen. In contrast, the quasi-isotropic strain field is more heterogeneous, with higher strains occurring in an hourglass shaped region in the center of the specimen. A complete set of strain maps for every specimen is included in Appendix A.4.

The simulations are able to capture the overall macro-deformation of the specimens, as well as the location and size of the strain micro-gradients at the tow undulations. These microgradients have been previously registered experimentally during quasi-static testing [222]. In the present work, the resolution of the DIC system was insufficient to capture the strain gradients at the tow undulations. As mentioned in section 7.3, the spatial resolution of the DIC system was approximately 3 mm, in excess of the size of the undulation regions, of approximately 1.5 mm.
The damage mechanisms are effectively captured by the numerical models. Matrix cracking initiates in undulation regions near the specimen shoulders at approximately 0.6% nominal strain. Localized matrix softening places additional stress on the fiber tows in these areas, inducing fiber fracture. In the cross-ply specimens, catastrophic failure through fiber fracture occurs on a plane orthogonal to the loading direction, near the specimen shoulder, see Figure 7.13. In the quasi-isotropic specimens, the failure surface is v-shaped (i.e. at 45° angles to the loading direction), a phenomenon which is reflected in the experimental results, see Figure 7.14. Equally well represented is the extensive matrix damage occurring in the vicinity of the fracture plane.

### 7.6.2 Influence of undulations at high strain rates

Table 7.2 summarizes the experimental results at quasi-static and dynamic strain rates. It includes both stacking sequences (cross-ply and quasi-isotropic) and also compares AP-PLY and baseline laminates. Given the overlapping error bounds, we conclude there is no significant difference between the moduli of the AP-PLY specimens at high and low strain rates. Similarly, there is no appreciable strain rate dependency in the baseline laminates. Figure 7.15 presents the quasi-static and dynamic response of the
AP-PLY and baseline laminates. In the dynamic regime, there are marginal improvements in laminate stiffness resulting from the AP-PLY preforming process. However, given the complexity of the dynamic experiments and the intrinsic oscillations in the stress-strain response, the discrepancy may be the result of experimental scatter alone. The same conclusion was drawn at quasi-static strain rates, where the preforming process also had a negligible impact on the stiffness of the laminates [222].

The strength of all the specimens, both AP-PLY and baseline, is significantly reduced at high strain rates. The proximity of the failure plane to the specimen shoulders, and the strain concentrations measured using DIC, suggest that this lower strength may be a consequence of the geometry of the specimens, rather than strain rate dependency at the material level. As mentioned previously, brittle carbon fiber composites are sensitive to mechanical gripping at high strain rates, potentially reducing their measured strength [218]. Moreover, while stress concentrations arising from the through-thickness reinforcements in the cross-ply AP-PLY laminates could trigger failure in the specimens at relatively low loads, the fact that the strength of the baseline laminates was reduced to a similar or greater extent suggests that the fiber undulations are not the primary cause of the low dynamic tensile strength. The reduction in the quasi-isotropic specimen strength at high strain rates is less substantial compared with the cross-ply
Figure 7.13: Experimentally observed and simulated failure mechanisms in a cross-ply AP-PLY laminate.

Table 7.2: Quasi-static vs dynamic laminate moduli and strengths.

<table>
<thead>
<tr>
<th>Configuration</th>
<th>Modulus [GPa]</th>
<th>Strength [MPa]</th>
</tr>
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<tbody>
<tr>
<td></td>
<td>25 × 10⁻⁴s⁻¹</td>
<td>30 s⁻¹</td>
</tr>
<tr>
<td>XP_{AP-PLY}</td>
<td>65.27 ± 3.53</td>
<td>69.38 ± 3.29</td>
</tr>
<tr>
<td>XP_{base}</td>
<td>63.59 ± 1.23</td>
<td>62.01 ± 3.63</td>
</tr>
<tr>
<td>QI_{AP-PLY}</td>
<td>44.96 ± 0.57</td>
<td>47.41 ± 2.63</td>
</tr>
<tr>
<td>QI_{base}</td>
<td>44.56 ± 0.95</td>
<td>43.88 ± 1.51</td>
</tr>
</tbody>
</table>

specimens. The high strain rate strengths of the cross-ply baseline and AP-PLY specimens fell by 46% and 32% respectively, relative to their quasi-static strength, while the reduction in quasi-isotropic baseline and AP-PLY strengths were only 8% and 14% respectively. This is because, as illustrated in Figure 7.12, the strain concentrations at the shoulders of the quasi-isotropic specimen are less pronounced than in the cross-ply specimens. A possible explanation for the smaller magnitude of these strain concentrations is the greater distribution of tow orientations, and as a result the smaller mismatch angles between tows.

At high strain rates, the strengths of the cross ply AP-PLY laminates were found to be insensitive to the internal architecture. This contrasts with data from the quasi-static regime, in which the strength of cross-ply laminates was negatively affected (~17% loss) by the AP-PLY preforming process [222]. Due to the low failure stresses of the dynamic specimens compared with their quasi-static counterparts — a consequence of the stress triaxiality caused by
Figure 7.14: Experimentally observed and simulated failure mechanisms in a quasi-isotropic AP-PLY laminate.

the specimen geometry — matrix cracking in the dynamic regime is likely to be less significant. As reported in Chapter 5, the density of matrix cracks at tow undulations increases exponentially with increasing load, which means the crack density of the dynamic specimens is expected to be lower than in the quasi-static specimens. A consequence of the lower crack density in the dynamic regime is that the softening of the tow undulation regions is minimized. This, in turn, reduces the magnitude of the stress concentrations in these regions, thereby decreasing the sensitivity of the specimens to the AP-PLY tow undulations. In short, in the dynamic regime, the strain concentration factors at through thickness undulations are smaller, while the reinforcement that these features provide is unaffected.

In contrast to the cross-ply results, the trends in quasi-isotropic laminate strength were similar for both high and low strain rate loading. In both the quasi-static and dynamic regimes, the discrepancy between the AP-PLY and baseline strengths was within the experimental error bounds, with the former generally exhibiting marginally higher strengths. The matrix cracking effects occurring in cross-ply laminates are likely to be less significant in the quasi-isotropic composites, because the former contain a larger proportion of tows oriented orthogonal to the loading direction, in which matrix cracking will be more significant, and because the mismatch angles between tows in the quasi-isotropic laminates are generally smaller, reducing the magnitude of the strain concentrations. Furthermore, in the AP-PLY laminates, matrix cracks formed predominantly at the boundaries between tows, rarely splitting a tow in the transverse direction. This contrasts with the more widespread matrix cracking observed in the baseline laminates, and may explain the slightly higher average
Figure 7.15: Stress-strain response of (baseline and AP-PLY) cross-ply and quasi-isotropic laminates under quasi-static (a) & (c) and dynamic (b) & (d) loading.

strengths of the AP-PLY laminates in both loading regimes.

The numerical model predictions of stiffness and strength at both low and high strain rates are in good agreement with the experimental results, despite using strain rate independent material properties [222]. This suggests that the effect of the strain rate at the material level is minor, and that the realignment of the fibers may play a larger role. Carbon fiber epoxy composites are strain rate insensitive in the fiber direction [229, 232, 230], and at the relatively low dynamic strain rates attained in this study (30 s⁻¹), the transverse properties are likely to be only minimally affected [229, 230]. While microscale numerical modeling and/or in-situ high-speed synchrotron tomography might shed further light on the underlying causes of the discrepancy between the baseline and AP-PLY specimens at high strain rates, they are beyond the scope of this study.

In summary, these findings prove that the through thickness tow undulations present in an AP-PLY composite do not have a detrimental effect on its dynamic mechanical properties. Furthermore, through thickness undulation have the capacity to arrest cracks, delaying the propagation of delamination, and, depending on the laminate configuration, improving the
lamine strength. This contrasts with AFP defects e.g. tow overlaps or tow gaps, which have been observed to reduce a composite’s strength [71]. While overlaps and gaps are unpredictable features that occur randomly throughout a laminate, the through thickness undulations introduced by the AP-PLY process occur in a structured pattern and, more importantly, can be strategically distributed to improve the local damage tolerance of composite structures subjected to dynamic loads.

7.7 Conclusions

The high strain rate tensile behavior of AP-PLY composites was investigated. Due to the large representative volume elements and brittle nature of the laminates, experiments were conducted at the European Laboratory for Structural Assessment, one of the only laboratories worldwide capable of accurately analyzing this family of composite materials. The bespoke SHB equipment in combination with DIC and high-speed imaging ensured dynamic force equilibrium was rapidly established, providing accurate measurements even at low strains. In addition, the SHB experiments were simulated using a multiscale numerical modeling framework to gain further insight into the high strain rate deformation mechanisms of the novel pseudo woven composites.

Quasi-isotropic and cross-ply AP-PLY specimens were subjected to tensile loading at a strain rate of $30 \ s^{-1}$. The moduli of the AP-PLY composites were found to be independent of the strain rate. While the dynamic experimental AP-PLY laminate strengths were significantly less than the quasi-static strengths, this discrepancy was attributed to the geometry of the specimens and the resulting strain concentrations, rather than strain rate dependency in the constituent materials. Numerical model results provided further evidence that the material strength of the AP-PLY composites was not affected by the higher strain rate.

The response of the AP-PLY laminates was compared with baseline — non AP-PLY — laminates to quantify the effect of the three-dimensional reinforcement. It was possible to establish that the through thickness undulations introduced by the AP-PLY preforming process did not result in a reduction in laminate stiffness in the dynamic regime. While the dynamic AP-PLY laminate moduli were found to be greater than those of the baseline laminates, the discrepancies were small, and could be attributed to the resolution of the SHB equipment.

In terms of strength, different trends were obtained depending on the predominant failure micromechanism. The failure of the quasi-isotropic AP-PLY laminates was dominated by localized matrix cracking at the tow boundaries regardless of the applied strain rate. This mechanism promoted tow debonding rather than the extensive matrix cracking observed in the baseline laminates, showing the capacity of the AP-PLY architecture to influence laminate deformation. As a result, the quasi-isotropic AP-PLY
laminates exhibited slightly higher strengths than their baseline non AP-PLY counterparts at both high and low strain rates.

In the cross-ply AP-PLY laminates, strain concentrations at through thickness undulations were found to be the catalysts for failure at low strain rates. In the dynamic regime however, these concentrations had a much smaller impact on the laminate response. This change in behavior was attributed to reduced matrix cracking, resulting in smaller strain gradients at tow undulations. As a result, the reduction in cross-ply AP-PLY laminate strength, relative to the baseline, registered in the quasi-static loading regime, was not observed at high strain rates.

These results prove AP-PLY laminates are effective substitutes for conventional angle-ply laminates in dynamic loading scenarios. They possess the same exceptional specific stiffness as angle-ply laminates, while their improved damage tolerance at high strain rates can result in higher strengths — depending on the stacking sequences.

To encourage future research and the reuse of the experimental results presented in this chapter, the complete set of raw data has been uploaded to a public data repository.
Chapter 8

Impact Response

8.1 Introduction

Low velocity impact tolerance is a major concern in the design of critical aerospace structures. As previously mentioned, conventional composites consisting of laminated unidirectional plies possess poor impact tolerance, due to the propensity of their constituent plies to delaminate when subjected to dynamic out-of-plane loading. The resultant damage can significantly affect the residual strength of a composite. Alternative laminate architectures, e.g. 3D-wovens, non-crimp fabrics, and z-pinned composites, contain through thickness reinforcements which improve their out-of-plane response [105, 233, 49]. However, the fiber crimp and fiber breakage introduced during the manufacture of these composites negatively impact their undamaged in-plane mechanical properties. In the previous chapters, it has been established that the AP-PLY preforming process does not significantly reduce a laminates in-plane stiffness. The effect of the through thickness undulations on laminate strength was found to depend on the laminate layup. In this chapter, the impact response of three different AP-PLY laminates is investigated, to ascertain whether the through thickness connectivity present in these types of laminates affects has an effect on the initiation and propagation of delamination, and the residual strength.

Cross-ply, triaxial, and quasi-isotropic AP-PLY laminates were first subjected to drop-weight tower testing at impact energies of 30 J and 50 J. The resulting damage was inspected using non-destructive ultrasound, before the residual compressive strength of the specimens was determined using compression after impact tests. Low velocity impact tests were then simulated using the previously discussed multiscale numerical framework, showing good agreement with the experimental results for most AP-PLY configurations. The numerical model was subsequently used to simulate the response of baseline non-AP-PLY laminates, facilitating a comparison of the impact response of the conventional and AP-PLY composites.
Figure 8.1: Schematic illustration of the layup process for a quasi-isotropic AP-PLY laminate.

8.2 Materials and manufacturing

AP-PLY panels were manufactured by vacuum assisted resin transfer moulding and the fiber preforms were produced using an mTorres AFP machine at FIDAMC in Madrid. Three different panels with dimensions 500 mm x 500 mm were manufactured with a cross-ply (denoted XP) \( [0^\circ, 90^\circ]_{3S} \), triaxial (TRI) \( [0^\circ, 60^\circ, -60^\circ]_{4S} \), or quasi-isotropic (QI) \( [0^\circ, 45^\circ, -45^\circ, 90^\circ]_{3S} \) layup. Photos and schematic representations of the three panels are provided in Figure 8.2. The quasi-woven AP-PLY architecture was created by leaving a single tow width gap — measuring 12.7 mm — between tows placed in the same pass.

All laminates were manufactured using Hexcel HiTape UD210 dry tapes (containing HexTow IMA-GP 12 K fibres) impregnated with Hexcel RTM6 resin using vacuum assisted resin transfer molding. A 4 gsm thermoplastic veil provided the tackiness required for AFP preforming. Excess material was trimmed using an ultrasonic cutter before the panels were vacuum bagged for resin transfer moulding. A distribution mesh and breather fabric were used to facilitate the homogeneous distribution of resin throughout the laminate \([234][235]\). Thermocouples were placed on the top surface of each panel and on the tooling surface to monitor the temperature during the resin infusion and curing process. The RTM6 resin was initially heated to 60°C for 30 minutes before being poured into a boiler. Once in the boiler, the resin was degassed at 5 mBar and 80°C for 30 minutes. After degassing, the resin was transferred to a pre-heated (120°C) mold containing the fiber preform. Finally, after impregnation, the system was cured at 180°C for 90 minutes in an oven \([236]\).

Specimens were extracted at the National Institute for Aerospace Technology in Madrid (INTA), using a diamond saw, according to the
dimensions and tolerances specified in ASTM D 7137, specifically, 150 mm x 100 mm. The dimensions of the specimens were measured at two points for the length and width, and four points for the thickness. The average thickness of each specimen type was determined to be 5.16±0.02 mm, 5.04±0.07 mm, and 5.15±0.05 mm, for the XP, TRI, and QI configurations, respectively. The resulting average cured ply thickness was approximately 0.213 mm.

It is worth noting that due to the different number of tow orientations in each laminate, the number of interfaces containing through thickness reinforcements differs. The XP laminates, for example, consist of 12 sets of AP-PLY plies, meaning there are non-reinforced interfaces between every 2nd layer. In contrast, the TRI laminates are split into 8 sets of 3 AP-PLY plies, and the QI laminates consist of 6 sets of 4 AP-PLY plies. Plotting the $B_{11}$ component of each laminate’s ABD matrix, see Figure 8.3, reveals the local variation in the properties of each laminate, and allows for the visual identification of each laminate’s representative volume element (RVE). Note the QI laminate does not contain a strictly defined RVE. However, following the methodology proposed by Rad et al. it is possible to determine an "approximate RVE" whose properties differ only minimally from those of its parent laminate [160]. Planar RVE dimensions for the XP, TRI, and QI laminates measure 25.4 mm × 25.4 mm, 25.4 mm × 29.3 mm, and 25.4 mm × 25.4 mm respectively. To improve the repeatability of the experiments, the extraction of the specimens was conducted such that the internal architecture and the impact location did not vary by more than ±4 mm.
8.3 Experimental techniques

8.3.1 Ultrasound inspection

Laminates were inspected before and after impact through ultrasound non-destructive inspection using an Olympus Omniscan X3 phased array ultrasound device. Amplitude and time-of-flight data were recorded using a 5.0 MHz transducer. Ultrasound data was post-processed using Olympus’ OmniPC software and plotted using Python. The edges of the plates, where defect content was higher, were discarded during the extraction of the specimens.

8.3.2 Low velocity impact

Low velocity impact testing was conducted in accordance with the ASTM D 7136 standard using an Instron-CEAST Fractovis 6875 drop tower, instrumented with acceleration and displacement sensors [81]. Two different energy levels were selected; 30 J and 50 J. In the tests, a 5.85 kg impactor with a 16 mm diameter hemispherical tip was dropped from a height onto each coupon, resulting in 3.2 ms$^{-1}$ and 4.13 ms$^{-1}$ for each impact energy. An anti rebound system prevented specimens from being impacted more than once. The bottom surface of the specimens was constrained by a rigid plate with a central cutout measuring 125 mm by 75 mm. Four clamps, placed near the corners of the specimens, were used to arrest the in-plane movement of the specimen during the impact.

Photron SA-Z 2100K cameras were used to record the back face of the specimens at 20,000 frames per second at a resolution of 1 megapixel. Space constraints in the impact tower precluded recording the back face of the specimens directly, so a mirror, inclined at 45 °, was used [18]. An additional benefit of this system is that it allows the movement of the impactor
8.3.3 Compression After Impact

The residual compressive strength of the AP-PLY laminates was evaluated through compression after impact (CAI) tests conducted according to the Airbus AITM 1-0010 standard, which is similar to the more commonly used ASTM D 7137 standard except for the fixture which is modified to apply a positive clamping pressure on the specimen edges [87, 237]. The specimens were placed in the CAI fixture, which imposes simply supported boundary conditions on the side edges and clamped boundary conditions on the top and bottom edges. The fixture was placed in an Instron universal testing machine with hinged compression platens, minimizing eccentricity in the compressive loading. The loading rate was set at 1.25 mm/min. Strain was measured using the crosshead displacement of the universal testing machine, corrected to account for the frame stiffness. The CAI testing setup is illustrated in Figure 8.5.

It is worth noting that the compressive strengths obtained from CAI experiments, for both damaged and undamaged specimens, reflect the resistance to buckling of a laminate rather than its in-plane strength [99]. As such, to accurately assess the undamaged in-plane strength of each laminate, short block compression tests were conducted, according to an adaptation of the the AITM 1-0008 standard. Specimens containing a single RVE (approximate in the case of the quasi-isotropic laminate) were extracted from the laminates using a wet saw. The resulting specimen dimensions measured 25.4 mm \times 25.4 mm for the XP and QI configurations, and 29.3 mm \times 25.4 mm for the TRI configuration. The specimens were placed between the parallel platens of a 250kN Instron universal testing machine,
and loaded to failure in compression at a rate of 1.25 mm/min (same as in the CAI experiments). Digital image correlation was used to record the strains in the specimens and to detect buckling. A minimum of six specimens were tested for each configuration. Specimens which failed through buckling or brooming of the specimens ends were excluded from the final data set. The short block compression testing setup is illustrated in Figure 8.6.

### 8.4 Numerical modeling

Low velocity impact tests on the AP-PLY laminates, were simulated using the multiscale modeling framework described in Chapter 4. In addition to the AP-PLY laminates, conventional angle-ply laminates with the same tow orientations were numerically modeled to facilitate a comparison of their impact response. The material properties of the prepreg fiber tows are shown in Table 8.1. Due to time and material constraints brought on by the COVID-19 pandemic, it was not possible to characterize the material properties of the HiTape UD210 prepreg. As such, the elastic constants and strengths of the lamina were calculated using a computational micromechanics approach in Autodesk Helius, based on the properties of the IMA fibers and RTM-6 resin provided by Hexcel [238, 236]. The fracture toughnesses of the HiTape UD210 material were estimated by considering...
Figure 8.6: Experimental test setup for the short block compression tests (left) and image from the DIC system of a quasi-isotropic short block compression specimen (right).
the fracture toughnesses of IMA/M21 prepregs, and effects of thermoplastic veils on fracture toughness reported in [8, 239, 240]. It should be noted that in-situ properties were not used in the simulation of the impact experiments, as there was insufficient material data available to determine their values. Moreover, previous studies have found that in-situ properties did not improve numerical predictions on impact response [111]. At the interfaces between connected undulation regions the interlaminar strengths and fracture toughnesses were set to equal the longitudinal tensile lamina properties, and the interfacial stiffnesses were scaled based on the unidirectional longitudinal laminate stiffness, see Section 4.5.

Table 8.1: Quasi-static Mechanical properties of Hexcel HiTape UD210

<table>
<thead>
<tr>
<th>Elastic Properties</th>
<th>$E_{11} = 156.0$ GPa, $E_{22} = E_{22} = 7.24$ GPa, $G_{12} = G_{31} = 4.65$ GPa, $G_{23} = 3.10$ GPa, $\nu_{12} = \nu_{31} = 0.309, \nu_{23} = 0.401$</th>
</tr>
</thead>
<tbody>
<tr>
<td>Strengths</td>
<td>$X^T = 3135$ MPa, $X^C = -1760$ MPa, $Y^T = Z^T = 76$ MPa, $Y^C = Z^C = -180$ MPa, $S_{12}^T = S_{31} = 101$ MPa, $S_{23} = 101$ MPa</td>
</tr>
<tr>
<td>Fracture Energies</td>
<td>$G_{1}^T = 134$ N/mm, $G_{1}^C = 95.0$ N/mm, $G_{2}^C = 0.78$ N/mm, $G_{6}^C = 2.73$ N/mm</td>
</tr>
<tr>
<td>Interlaminar Properties</td>
<td>$G_{1c} = 0.78$ N/mm, $G_{1c}^C = 2.73$ N/mm, $\tau_0^0 = 76$ MPa, $\tau_0^0 = 101$ MPa, $\eta = 1.75$</td>
</tr>
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The impact specimens were divided into two different regions, a linear elastic region which is not damaged during the impact experiment, and a highly non-linear region in which intra and interlaminar damage can occur. The size of each region was established based on post-mortem non-destructive inspections of experimental specimens [241]. The virtual impact test setup is illustrated in Figure 8.7.

The primary role of the linear elastic region is to transfer the loads and displacements from the non-linear damageable region to the specimen supports. Since shear displacements and through thickness stresses dissipate with distance from the impact site, the aforementioned load and displacement transfer can be accomplished using a single layer of shell elements [241]. The shell part was partitioned along the tow boundaries, and assigned locally varying stacking sequences to reflect the AP-PLY architecture. Each ply was defined using a single integration point, resulting in 24 integration points through the thickness of the shell part. AP-PLY tow undulations were not modeled in the shell part. The rotational and translational degrees of freedom of the solid element and shell regions were coupled using Abaqus’ built-in shell-to-solid coupling constraint [149].

To mitigate the effects of spurious wave oscillations occurring as a result of the solid to shell coupling, a linear elastic solid element transition region
was defined between the shell and damageable regions. As in the shell part, the locally varying stacking sequences resulting from AP-PLY preforming were modeled but through thickness tow undulations were not. A tie constraint was used to couple the displacements at the interface between the damageable and linear elastic solid element regions.

In the damageable region, 3D solid elements — specifically C3D8R elements — were used to model the effects of matrix cracking and fiber fracture. The mesh size was set to approximately equal the mesoscale unit cell size (approximately 1 mm). This ensured realistic strain transformation from the element to the micro-constituent level, while minimizing the computational cost \[133\]. It is worth noting that some variation in element size was unavoidable given the automated geometry generation methodology.

A general contact definition was created specifying hard contact between all surfaces, with the exception of the cohesive surfaces between tows in the damageable region. Using a general contact interaction has the additional advantage of allowing for frictional contact between surfaces when the cohesive interface is completely damaged. The initial velocity of the impactor — calculated based on the desired impact energy — was specified using a predefined field, its movement in every other direction was constrained. The clamps exercised a 1.1 kN load on the top specimen face, the minimum force specified in the ASTM D 7136 standard \[81\].

Validation

The elastic bending response under impact loading was compared with analytical predictions of specimen deflection under a point load. It should be noted that the bending response under low velocity impact and quasi-static
point load are not identical, however, the elastic impact response should be in agreement with the theoretical bending stiffness of the different laminates to ensure the numerical approach is able to represent the stiffness of the laminate. A full description of the analytical model can be found in Appendix A.2 [242, 243]. The elastic force-displacement curves for the three different laminates, and the analytical model predictions are plotted on Figure 8.8. Note the analytical model predicts larger deflections than the elastic finite element model, these differences are expected given the differing boundary conditions. Zero deflection is assumed along the specimen boundaries for the analytical model, which does not replicate exactly the boundary conditions in a drop weight tower experiment. Nonetheless the the differences in bending stiffnesses are well captured, with XP panels exhibiting greater deflection even at lower loads, and QI and TRI panels showing similar bending responses.

![Figure 8.8: Fully elastic (no damage) force displacement behavior of the three different AP-PLY laminates at an impact energy of 30 J.](image)

**8.5 Results and Discussion**

In light of the novelty of the AP-PLY manufacturing process, and the absence of existing best practices, it is worth discussing the quality of the undamaged AP-PLY specimens. First, since the top faces of the laminates were constrained only by a vacuum bag, the corresponding surface of the cured laminates was relatively uneven, see Figure 8.9. Second, while void content
overall was low, the few voids which could be observed near the laminate edges tended to concentrate in the undulation regions or between adjacent tows, see Figure 8.9b. Away from the laminate edges the void content was sufficiently low that it did not affect the deformation of the specimens. Note that the laminates studied in Chapters 5, 6, and 7, were manufactured using compression molded prepgs, and did not exhibit the void concentrations observed in AP-PLY laminates manufactured using VARTM.

Figure 8.9: Laminate quality in AP-PLY specimens: (a) uneven top surface of triaxial AP-PLY laminate, and (b) void concentrations at tow boundaries in triaxial AP-PLY laminate.

Figure 8.10 plots the force-displacement curves of the AP-PLY laminates at impact energies of 30 J and 50 J. Despite the minor differences in the internal architectures of the specimens relative to the impact location (±4 mm), the resulting force-displacement curves were found to be highly repeatable. These results are in agreement with previous studies where even larger variations in the impact location still produced repeatable results [10]. The authors acknowledge that specimens with larger heterogeneities, such as wider tows and tow gaps (distance left between tows placed in the same pass), might present repeatability issues. Additional experimental studies with more extensive experimental campaigns, beyond the scope of this project, are required to fully understand the influence of the impact location on the low-velocity impact response of AP-PLY laminates plate.

The XP laminate exhibits the softest response of the three configurations at both the high and low impact energy. The load at maximum deflection for the 30 J impacts of the XP, TRI, and QI specimens are approximately 11 kN, 12 kN, and 12.5 kN, respectively. Notably, when the impact energy is increased to 50 J, the peak loads in the TRI and QI laminates rise to approximately 16 kN, while the XP laminate achieves its maximum load capacity at 13 kN.

To investigate the response of the different AP-PLY laminates in terms of energy, Fig. 8.11 plots the coefficient of restitution (COR) for each configuration [244, 18]. The COR represents the ratio of the rebound velocity of the impactor
Figure 8.10: Force vs displacement plots for 30 J (a,b,c) and 50 J (d,e,f) impacts on (a,d) cross-ply, (b,e) triaxial, and (c,f) quasi-isotropic AP-PLY composites.

The previous results suggest there is more damage accumulation in the XP laminates subjected to 50 J impacts, a supposition validated by the delamination footprints of the specimens, illustrated in Figure 8.12. Fig. 8.13 quantifies the surface area of the delamination footprint as function of the impact energy. All AP-PLY specimens exhibited elliptical delamination footprints of approximately the same size at an impact energy of 30 J in agreement with the coefficients of restitution reported in Figure 8.11, the discrepancies between the configurations were within the experimental scatter. At the higher impact energy of 50 J, the size of the XP delaminations increases significantly compared with the other configurations. The shape of the delamination also changes from the original elliptical footprint into a
cross-shaped delamination, following the orientations of the perpendicular tows. This result can be explained by more closely analyzing the internal architecture of the AP-PLY laminates. In the XP specimens, a maximum of two layers are interlaced, compared with 3 or 4 interlaced layers in the TRI and QI laminates. Furthermore, the tows that are interlaced in a XP laminate are always orthogonal to one another, while in the TRI and QI laminates, the interface angles between tows vary. As illustrated in Figure 8.3, this results in increasing homogeneity as the number of tow orientations rises. As a consequence of their heterogeneity, and the large bending stiffness mismatch between interlacing tows, the XP laminates are more prone to delamination than the TRI or QI laminates.

At both 30 J and 50 J, matrix cracks can be observed in the rear faces of the impacted specimens, see Figure 8.14. Fiber fracture was not observed in the rear face of the specimens even at the higher 50 J impact energy. The matrix cracks formed along the orientations of the tows in the laminates, both within tows and along the boundaries between them. The greater number of orientations in which matrix cracks can form in the TRI and QI laminates may facilitate energy dispersion through intralaminar matrix cracking rather than delamination, perhaps explaining the discrepancy between the damage footprints of the AP-PLY laminates at high energy levels.

The residual compressive strengths of the AP-PLY laminates, as a function of impact energy, are illustrated in Figure 8.15. The undamaged compressive strengths determined using the short block compression tests were found to be low relative to the theoretical compressive strength of conventional
Figure 8.12: Delamination footprints of AP-PLY laminates impacted at impact energies of 30 J and 50 J, measured using ultrasound non-destructive inspection.
Figure 8.13: Impact footprint as function of the impact energy

Figure 8.14: Cracks in rear faces of AP-PLY specimens impacted at 50 J.
Figure 8.15: Residual strength vs impact energy for the 3 different AP-PLY composites.

unidirectional composite plies. This can be attributed to the high misalignment of the fibers in the undulation regions, which promote fiber kinking, see Figure 8.16. Similar results were previously obtained by Nagelsmit [10].

Figure 8.16: Images of XP short block compression specimens at (a) 0 load, and (b) after failure. Note the fiber curvature in the specimens and the resulting kink band in the broken specimen.

The CAI strength of a composite is generally governed by the buckling
resistance of its sublaminates, which are formed when impact loading causes interfacial damage between plies [99]. In principle, the through-thickness reinforcements present in AP-PLY laminates delay the onset of delamination and arrest its propagation through a specimen. The QI and TRI AP-PLY laminates exhibit relatively high CAI strengths. The XP CAI strength is affected to a greater extent by the presence of delamination. At an impact energy of 30 J, the residual strength of the XP laminate is 38.0% lower than the undamaged compressive strength, while for the TRI and QI laminates, which both exhibit larger damage footprints, the reductions in compressive strength are 18.1% and 16.7% respectively. A possible explanation for this counter intuitive result is that while the footprint may be smaller, the through thickness distribution of the debonded interfaces must also be considered. As mentioned previously, in the XP laminates, a maximum of two plies are interlaced. As a result, the sublaminates in a XP laminate are thinner than in the other configurations. As such, the delaminations in a XP specimen may be spread out over a larger number of interfaces, resulting in less stable sublaminates. Second, since the XP laminates contain proportionally more tows aligned with the loading direction, they are more sensitive to damage because the fiber misalignment resulting from impact damage has a proportionally larger effect on the strength of the material. At 50 J, the reductions in compressive strength relative to the undamaged strength were 49.1%, 26.0%, and 20.6% for the XP, TRI, and QI AP-PLY laminates respectively, following similar trends.

The hypothesis that sublamine stability is the primary determinant of CAI strength is further supported by post-mortem photographs of the broken specimens. From Figure 8.17 it is clear that the XP specimens fail through the formation of a global shear band. The individual sublaminates fail simultaneously, lacking the strength to support loads individually. In addition, the crack path, highlighted in red, is a fairly straight line across the surface of the specimen passing through the impact point. In contrast, the TRI and QI specimens exhibit more tortuous crack paths that follow, to an extent, the tow boundaries in the laminate. The longer crack length results in increased energy dissipation. The cross sections of the fracture surfaces indicate that the failure mechanisms were more distributed in the QI and TRI laminates. Instead of a global crack, the sublaminates which formed as a result of impact damage failed individually and became entangled, further increasing the energy dissipated during the compressive loading.

Figure 8.18 compares the numerical predictions of the force-displacement response of the three different AP-PLY laminates with the experimental results. For the TRI and QI laminates, the correlation between the simulations and the experimental results is fairly good. The softening of the stress-strain response with the accumulation of damage, and the resulting peak load are both well captured. The discrepancies between the numerical and experimental peak load for the TRI laminate are 16.5% and 14.4% (30 J and 50 J), while for the QI laminate they amount to 12.0% and 10.2% for the
Figure 8.17: Crack path and cross-section of fracture plane in (a) cross-ply, (b) triaxial, and (c) quasi-isotropic CAI specimens impacted at 30J.

Figure 8.18: Comparison of experimental and numerical force vs displacement plots for 30 J (a,b,c) and 50 J (d,e,f) impacts on (a,d) cross-ply, (b,e) triaxial, and (c,f) quasi-isotropic AP-PLY composites.
30 and 50 J impacts respectively. The behavior of the XP AP-PLY laminate is not as well captured by the numerical models. Although the initial response of the specimens is accurately represented, the discrepancies between the numerical and experimental results grow as deflection rises, suggesting that the model may not capture all of the damage occurring within the specimens. This is most notable at the higher impact energy of 50 J, where the predicted peak load is significantly (26.3%) higher than the experimental average. Examining the initial elastic stiffness of the AP-PLY laminates, it is clear that the lower elastic stiffness of the XP specimens compared with the TRI and QI specimens is accurately represented by the numerical models. This indicates that there are damage mechanisms in the XP laminates that are not accurately represented by the numerical modeling framework. As previously discussed in Chapters 6 & 7, the assumption of isostrain conditions in the unit cells may have affected the accuracy of the numerical models. The results suggest that there is insufficient matrix damage accumulation in the AP-PLY laminates. Under isostrain conditions, in elements containing tow and resin constituents, the pure resin volume will be subjected to lower strains than in a more realistic model which more accurately captured the strains on each constituent. As a result, matrix damage may trigger at higher element strains than expected, resulting in less matrix damage accumulation. A micromechanical model which accounts for the different strain states in individual constituents of unit cells might generate more accurate results but is beyond the scope of the present work.

Another damage mode which is evidently not represented perfectly is delamination. At high energy, the experimental delamination footprint of the XP laminates far exceeds the predictions of the numerical model. Since the present model does not permit delamination within undulation region unit cells, the cohesive surfaces defined in the model are discontinuous. While creating tie constraints between undulation regions replicates, to an extent, the effect of through thickness fiber reinforcement, it is an approximation of a complex interface with out-of-plane curvature. A fully microscale model, such as the one proposed by Li et al., may generate more accurate predictions of the delamination footprint [29, 184]. However, as mentioned previously, the computational cost and meshing issues involved in such an approach may limit its effective application to the simulation of drop weight tower tests. In addition, as previously mentioned, matrix plasticity is not implemented in the numerical model, which may have further reduced the accuracy of the predicted force-displacement responses.

The experimental unloading curves differ from the numerical predictions. There are several explanations for this discrepancy. First, as discussed in the work of Lopes et al., in the experiments, the impactor may lose contact with the specimens before their unloading is complete. As a result, the force-displacement curves may indicate higher permanent indentation values than realistically attained. Due to damping in the simulations, there is less loss of contact between the impactor and the specimen [140]. Second, the
force of gravity was omitted in the simulations, which may have affected the rebound velocity of the impactor [133]. Finally, the discrepancy between the experimental and numerical permanent indentation may be the result of the omission non-linear matrix behavior in the numerical model, which is beyond the scope of the present work and will be investigated in future studies [195, 99, 245].

In terms of failure modes, the numerically predicted damage mechanisms correlate well with those observed in the experiments. Matrix cracking and delamination are the dominant failure modes governing specimen behavior. Fiber fracture was not observed in the experiments at the impact energies tested and is not present in the numerical model results. Cross-sections and images of the rear face of the simulated specimens, are presented in Figure 8.19 illustrating damage in each laminate. Note that since the experimental specimens were subsequently tested to failure in CAI tests, it was not possible to compare the numerical and experimental post-impact cross-sections. However, the cracks in the rear faces of the specimens observed experimentally and their numerical predictions are in good agreement.

![Figure 8.19: Matrix damage contours on the cross-sections (above) and rear faces (below) of the (a) cross-ply, (b) triaxial, and (c) quasi-isotropic AP-PLY laminates.](image)

With respect to delamination, the quality of the correlation between the numerical and experimental results follow the same pattern as the force-displacement curves, see Figure 8.20. The numerical delamination footprints of the TRI specimens differ from the experimental results by 4.6% and 10.5% at 30 J and 50 J respectively. For the QI specimens the discrepancies are -2.8% and 5.5%. While the delamination footprint of the XP laminates is captured fairly well at 30 J (-15.0% discrepancy), at 50 J the damage footprints measured from the experimental specimens far exceed the footprint predicted by the numerical model (by -51.8%). This partly explains
Figure 8.20: Delamination footprints of AP-PLY laminates impacted at impact energies of 30 J and 50 J, measured using ultrasound non-destructive inspection.
the divergence of experimental and numerical force-displacement curves for these specimens. As discussed previously, a microscale model may more accurately capture the delamination of these specimens, but is far more difficult and computationally expensive to implement, and is more prone to mesh interpenetration issues. An additional explanation for the discrepancy between the numerical and experimental delamination footprint for the 50 J XP impacts, is that the fracture energies of interfaces between tows are a function of the mismatch angle \[246\], and 0°/90° interfaces typically exhibit the lowest fracture toughness. The variation in fracture toughness was not implemented in the presented numerical modeling methodology.

![Figure 8.21](image)

**Figure 8.21:** Comparison of AP-PLY and baseline numerical force vs displacement plots for 30 J (a,b,c) and 50 J (d,e,f) impacts on (a,d) cross-ply, (b,e) triaxial, and (c,f) quasi-isotropic AP-PLY composites.

Comparing numerical simulations of the AP-PLY laminates with simulations of baseline, non AP-PLY laminates provides insights into the differing deformation mechanisms and the affect of the internal architecture on damage formation and propagation. Figure 8.21 compares the force-displacement behavior of AP-PLY and baseline laminates at both 30 J and 50 J. Generally, the impact response of the AP-PLY laminates is very similar to that of the baseline composites. However, for the QI composites, the baseline laminates’ peak loads are noticeably higher than their AP-PLY equivalent. This suggests that the QI AP-PLY laminates dissipate more energy through intralaminar damage than the baseline composites. It is possible that due to the large number of through thickness connections in the AP-PLY QI laminates (compared with the XP and TRI laminates), delamination is inhibited such that intralaminar matrix damage is exacerbated, resulting in the softer response.
Comparing interlaminar damage, it is clear that the baseline laminates’ delamination footprints are universally larger than those of the AP-PLY laminates, see Figure 8.20. Since the force-displacement response of the AP-PLY laminates is either similar to or softer than the response of the baseline laminates, the large delamination footprint of the latter suggests that the AP-PLY laminates promote intralaminar damage growth, while limiting the formation and propagation of delamination. Consequently, since interlaminar damage has a greater effect on CAI strength than intralaminar matrix cracking, it is likely that the ability of the AP-PLY laminates to inhibit delamination — in favor of more significant matrix cracking — affords them higher residual compressive strengths than their conventional counterparts. This hypothesis is supported by the results of Nagelsmit, who reported improvements in CAI strength of up to 15% for AP-PLY laminates (compared with their baseline alternatives) laminates[10].

8.6 Conclusions

The impact response and residual compressive strength of three different AP-PLY laminates was investigated experimentally and numerically. Cross-ply, triaxial, and quasi-isotropic AP-PLY laminates were manufactured using dry fibers (Hexcel IMA) impregnated with resin (Hexcel RTM-6) using VARTM. Specimens from each laminate were subjected to 30 J and 50 J impacts, and were subsequently tested to failure in compression.

Of the three configurations, the cross-ply laminates exhibited the softest response and the lowest peak loads. The lower coefficient of restitution of the cross-ply laminates at high impact energy, compared with the other AP-PLY configurations, indicated greater damage accumulation in the former. Aside from their lower elastic stiffness, the greater heterogeneity of the cross-ply laminates, and the large mismatch angle between tows, were identified as possible causes of the reduced maximum load capacity.

In compression after impact tests, the triaxial and quasi-isotropic laminates performed similarly, while the cross-ply specimens exhibited lower strengths. This was attributed partly to the smaller proportion of through thickness reinforcement in the cross-ply laminates, increasing their susceptibility to sublaminate buckling.

The low velocity impact experiments were simulated using a multiscale numerical modeling framework. Model predictions of the force-displacement response were in good agreement with the experimental results for the triaxial and quasi-isotropic composites, but were less effective in capturing the behavior of the cross-ply specimens, possibly due to the underestimation delamination at high impact energies. Delamination damage was effectively captured by the numerical modeling framework for all but the high energy (50 J) cross-ply specimens, for which the numerical model underestimated the footprint. A microscale model which represents the through thickness tow
undulations as solid continua would likely produce more accurate estimates of delamination footprints, but this approach has limitations of its own and is beyond the scope of the present work.

The numerical model was subsequently used to simulate the impact response of conventional angle ply alternatives to the AP-PLY laminates. The baseline specimens exhibited similar force-displacement responses to the AP-PLY specimens. The fact that the predicted delamination footprints of the baseline specimens were significantly larger than those of the AP-PLY specimens, suggested that the latter inhibited delamination in favor of increased intralaminar damage accumulation. As a consequence, the AP-PLY composites are theorized to exhibit greater CAI strengths.
Chapter 9

Conclusions and Future Recommendations

9.1 Conclusions

9.1.1 Contribution to knowledge

In line with the objectives of this thesis outlined in Section 1.3, this section summarizes the main contributions to knowledge made by the present work.

- Developed a numerical framework able to accurately and efficiently represent AP-PLY laminate behaviour. The proposed methodology is the first to be able to automatically generate arbitrary three-dimensional AP-PLY laminate geometries and simulate their response to different kinds of loading, while accounting for the effect of the tow undulations on the overall laminate response.
  - Allows for a better understanding of the deformation mechanisms of AP-PLY laminates
  - Creates opportunities for design and optimization of AP-PLY laminate structures

- Investigated in detail the effect of tow undulations on the mechanical response of AP-PLY laminates. Strain concentrations resulting from AP-PLY tow undulations were investigated using the aforementioned numerical modelling framework, and using digital image correlation. In addition, x-ray computed tomography was used to investigate matrix cracking and delamination propagation in laminates with tow undulations.

- Investigated the in-plane tensile properties of AP-PLY laminates at high strain rates. While AP-PLY laminates have been previously studied under quasi-static tensile loading, the effect of AP-PLY tow undulations on the high strain rate tensile response of composite laminates was previously...
undocumented. The novel experimental approach, involving the use of an extremely large split-Hopkinson bar to study carbon fiber composites with large RVEs may be of interest to researchers outside the realm of AFP and AP-PLY.

9.1.2 Conclusions

- Tow undulations in AP-PLY laminates were studied using in-situ X-ray computed tomography. Under tensile loading, cracks developed in the tows orthogonal to the loading direction, resulting in non-linear stress-strain response. Ultimately the presence of the tow undulations in cross-ply AP-PLY composites created stress concentrations which reduced their strength relative to conventional laminates in which the fibers are all aligned in-plane.

- The quasi-static tensile response of baseline and AP-PLY cross-ply and quasi-isotropic laminates was characterized. Laminate stiffness was found to be unaffected by the AP-PLY architecture, regardless of the AP-PLY configuration. The effect of the AP-PLY process on laminate strength was found to depend on the stacking sequence. Cross-ply laminates exhibited sensitivity to the stress concentrations present at through thickness undulations, reducing their strength compared with conventional angle-ply composites. In contrast, the strengths of quasi-isotropic specimens were found to exceed those of their conventional counterparts, suggesting the through thickness reinforcements effectively act as crack arrestors.

- The high strain rate tensile behaviour of AP-PLY composites was investigated using split-Hopkinson bar experiments conducted at the European Laboratory for Structural Assessment. Cross-ply and quasi-isotropic AP-PLY and baseline composites were subjected to tensile loads at a strain rate of approximately $30s^{-1}$. As in the quasi-static regime, the effect of the AP-PLY process on laminate stiffness was found to be minimal. Furthermore, the moduli of the AP-PLY composites were independent of the strain rate. The effect of the AP-PLY architecture on laminate strength was found to depend on the dominant failure mechanism. The quasi-isotropic laminates exhibited marginally higher strengths than the baseline equivalent laminates, as a result of the crack arresting effect of the through thickness undulations, which promoted tow debonding rather than tow splitting. In the cross-ply laminates, the stress-concentrations which governed failure at low strain rates were found to be less significant due to reduced matrix cracking at high strain rates. As a result, AP-PLY cross ply laminate strengths were only marginally different from baseline cross-ply laminate strengths.

- The low velocity impact behaviour and compression after impact strength
of three different AP-PLY composites was evaluated. Cross-ply, triaxial, and quasi-isotropic AP-PLY composites were first subjected to 30 J and 50 J impacts, and then tested to failure in compression to determine their residual strength. The cross-ply laminates exhibited the softest response, the lowest compression after impact strengths, and at high energies, the largest delamination footprint and lowest coefficient of restitution. These characteristics were attributed to their greater heterogeneity compared with the quasi-isotropic and triaxial composites, the large mismatch angle between tows, and the smaller number of interconnected plies.

- A multiscale numerical modeling framework was developed to capture the behaviour of AP-PLY composites at relatively low computational cost. Rather than modeling the tow undulations as solid continua, the numerical framework used an approach based on the division of AP-PLY laminates into unit cells. The volume fractions of constituents in these unit cells could be varied to replicate the properties of different regions in the AP-PLY composites, e.g. regions containing through thickness tow undulations. A continuum damage mechanics material model was developed to capture the initiation and evolution of damage in each constituent. Cohesive surfaces were used to account for interlaminar damage. The material model was implemented as a Fortran material subroutine (Abaqus VUMAT), while the geometric pre-processing scripts were written in Python. Both the material model and the geometry creation programs are publicly available on GitHub.

- Numerical predictions of laminate stiffness and strength under quasi-static and dynamic tensile loads were found to be in good agreement with the experimental values. The model effectively predicted the location and orientation of crack planes, and the formation of matrix cracks within a specimen.

- Parametric studies were conducted using the numerical modeling framework to investigate the effect of specimen size, tow gaps, and the out-of-plane fiber orientation in a laminate on the tensile mechanical properties of cross-ply and quasi-isotropic composites. Laminate mechanical properties were found to be consistent for all specimen sizes, suggesting that the properties of AP-PLY laminates can be determined using approximate RVEs. Increasing the size of the gap between tows placed in a single pass was found to increase quasi-isotropic laminate strength, while cross-ply laminate strength and stiffness were unaffected. Steeper tow undulations were found to increase laminate strength, but did not affect laminate moduli.

- Simulations of the low velocity impact experiments were conducted using the same numerical modeling framework. While the numerical model effectively captured the force-displacement response of the triaxial and quasi-isotropic composites, predictions of the cross-ply
laminate response were less accurate. The discrepancy was attributed to the underestimation of fiber failure. Similarly, while numerical predictions of interlaminar damage were consistent with experimental results for the quasi-isotropic and triaxial laminates, and for the cross-ply laminates at 30 J, the delamination footprint of the cross-ply laminates impacted at 50 J was underestimated by the numerical models. The discrepancy as attributed to the variation in fracture toughness with tow interface angles.

To compare the impact response of AP-PLY and conventional laminates, the numerical modeling framework was also used to simulate the response of baseline laminates with the same stacking sequence as the AP-PLY laminates. From analysis of the force-displacement response, it was evident that the AP-PLY laminates inhibit interlaminar damage to a greater extent than the baseline composites. The AP-PLY laminates inhibition of delamination results in more substantial intralaminar damage, which may reduce the peak of the force-displacement curve. Since delamination damage is the primary determinant of compression after impact strength, the ability of the AP-PLY laminates to inhibit this damage mode was theorized to improve their impact tolerance.

9.2 Future Work & Recommendations

Further experimental studies As a consequence of their novelty, there are only a few experimental studies characterizing the properties of AP-PLY composites. This problem is exacerbated by the cost and complexity of manufacturing AP-PLY composites, and the many different possible configurations. Establishing a reliable dataset of AP-PLY composite properties would aid the development and calibration of numerical models, and facilitate a deeper understanding of this family of materials. Specific examples of potential experimental studies include: high strain rate compression, characterization of the mixed-mode fracture toughness, and studies on the fatigue properties of AP-PLY composites, among others.

In addition to coupon level studies, manufacturing and testing a complete structure made up of AP-PLY laminates is critical to understand whether the behaviour observed at smaller length scales can be replicated at the component level.

Hybridization While Rad et al. previously studied hybrid AP-PLY laminates [160], their investigations focused solely on laminates consisting of full AP-PLY layers combined with conventional plies, see Chapter 3. Hybrid laminates in which the AP-PLY process is applied only to small regions of a laminate — in which high toughness is highly desirable — are worth investigating. By targeting specific regions to reinforce the benefits of the AP-PLY architecture
can be exploited while minimizing the associated increases in manufacturing time and cost.

**Numerical model improvements**  The assumption of isostrain conditions in unit cells, and the use of volumetric stress averaging to determine the homogenized stress in an element, limit the ability of the model to capture micromechanical damage effects with complete accuracy. In addition, the use of unit cells prevents the realistic replication of delamination within undulation regions. Future developments of the modeling framework could investigate the use of alternative methods to represent strain within unit cells, and implement methods such as XFEM or the phantom node approach to allow for delamination within elements containing two distinct constituents. Furthermore, the inability of the presented numerical model to correctly represent the non-linear response of the matrix was acknowledged as a limitation in Chapter 4. The numerical model does not presently account for differences in the normal and shear fracture energies of composite laminae. The implementation of these material behaviours may improve model predictions of laminate behaviour, e.g. predictions of permanent indentation in laminates subjected to low velocity impact tests.

**Optimization** As a result of the modularity and flexibility of the AP-PLY preforming process, the design space for this family of composites is extremely large. This complicates the selection of an optimal AP-PLY configuration for specific structural applications. Future work could address this issue by developing an optimization framework capable of identifying the ideal stacking sequence to minimize cost, manufacturing time, weight, and so forth, while meeting design specifications. Artificial intelligence based approaches will likely be required to efficiently determine optimized stacking sequences, given the large number of variable parameters.
Appendix A

Appendices

A.1 Determination of in-situ parameters

Determining the in-situ strengths discussed in Section 4.4 necessitates the definition of the shear response factor $\beta$. According to Hahn and Tsai, the non-linear shear response of a unidirectional transversely isotropic composite laminate can be approximated using the following polynomial [247]:

$$\gamma_{12} = \frac{1}{G_{12}}\sigma_{12} + \beta\sigma_{12}^3$$  \hspace{1cm} (A.1)

where $\sigma_{12}$ and $\gamma_{12}$ represent the shear stress and strain, $G_{12}$ represents the shear modulus, and $\beta$ represents the non-linearity of the shear stress-strain response. When $\beta = 0$ the response is completely linear. The $\beta$ parameter was determined by fitting the aforementioned polynomial to experimental in-plane shear curves. The experimental stress-strain curve and its numerical approximation are illustrated in Figure A.1. The value of the beta parameter which minimized the discrepancy between the experimental and numerical results was $4.72 \times 10^{-8}$.

A.2 Analytical Deflection Model

Since all of the laminates in this study are both balanced and symmetric, the $B$ components of their ABD matrices are all equal to zero, and they do not exhibit bending-twisting coupling (i.e. $D_{16} = D_{26} = 0$). As such, it is straightforward to approximate the structural response of the specimens to out-of-plane impact loading using composite laminated plate theory.

Neglecting out-of-plane stresses, equilibrium conditions for a simply supported rectangular plate with side lengths $L$ and $W$ (Figure A.2), subjected to an out-of-plane point load $p_z$, dictate that:

$$D_{11} \frac{\partial^4 w}{\partial x^4} + 2 (D_{12} + 2D_{66}) \frac{\partial^4 w}{\partial x^2 \partial y^2} + D_{22} \frac{\partial^4 w}{\partial y^4} = p_z$$  \hspace{1cm} (A.2)
Figure A.1: Experimental shear stress-strain response of VTC401 and numerical approximation using polynomial expression.

Figure A.2: Schematic of a composite plate subjected to a point load. Adapted from [242].
Where $D_{ij}$ are elements of the plate’s bending stiffness matrix $D$, and $w$ represents the plate deflection. This expression is valid only for small values of $w$. The general solution to Equation A.2 takes the form:

$$w = \sum_{m=1}^{M} \sum_{n=1}^{N} \sin \frac{m\pi x}{L} \sin \frac{n\pi y}{W}$$  \hspace{1cm} \text{(A.3)}

Where $M$ and $N$ are the number of terms in the solution. Solving, as in Kassapoglou et al. [242], we obtain:

$$w = \sum_{m=1}^{M} \sum_{n=1}^{N} \frac{4F}{LW} \sin \frac{m\pi x}{L} \sin \frac{n\pi y}{W} \sin \frac{m\pi x}{L^2} \sin \frac{n\pi y}{W^2} + D_{11} \left(\frac{m\pi}{L}\right)^4 + 2 \left(D_{12} + 2D_{66}\right) \frac{m^2n^2\pi^4}{L^2W^2} + D_{22} \left(\frac{n\pi}{W}\right)^4$$  \hspace{1cm} \text{(A.4)}

If, as in the drop weight tower experiments, the point load is applied at the center of the plate, then Equation A.4 can be simplified by substituting $x = L/2$ and $y = W/2$.

$$w_{\text{max}} = \sum_{m=1}^{M} \sum_{n=1}^{N} \frac{4F}{LW} \sin^2 \frac{m\pi}{2L} \sin^2 \frac{n\pi}{2W} \sin \frac{m\pi x}{L} \sin \frac{n\pi y}{W} \sin \frac{m\pi x}{L^2} \sin \frac{n\pi y}{W^2} + D_{11} \left(\frac{m\pi}{L}\right)^4 + 2 \left(D_{12} + 2D_{66}\right) \frac{m^2n^2\pi^4}{L^2W^2} + D_{22} \left(\frac{n\pi}{W}\right)^4$$  \hspace{1cm} \text{(A.5)}

Consequently, the curvatures, stresses, and strains can be calculated using regular composite laminated plate theory.

### A.3 Quasi-static Tensile Characterization Strain Maps

Digital image correlation was used to record complete strain maps for AP-PLY specimens tested to failure under quasi-static tensile loading. A single strain map, of a cross-ply AP-PLY laminate under tension, is presented in Chapter 6 (Figure 7.11). For completeness, additional strain maps for the cross-ply and quasi-isotropic AP-PLY specimens are reproduced here, in Figures A.3 and A.4 at the same nominal strain of 1.1%. Due to the large file size of complete DIC image sets (thousands of images per test), data from only a selection of specimens was retained for further post processing. Furthermore, note that no strain maps were saved for the baseline specimens, as there were no significant strain gradients observed in the specimens during testing. Two different lenses were used to record the cross-ply specimens, which is why there are differences between the zoom level of the strain maps in Figure A.3.
Figure A.3: Strain maps of cross-ply AP-PLY specimens under quasi static tension, imaged at 1.1% nominal strain. Note the strain concentrations occurring at the tow undulations. Subfigure (a) was captured using a different lens from subfigures (b) and (c), which is why the image is more zoomed in and has a smaller field of view.

Figure A.4: Strain maps of quasi-isotropic AP-PLY specimens under quasi static tension, imaged at 1.1% nominal strain. Note the uniformity of the strain on the surface of the specimens compared with the cross-ply specimens in Figure A.3.
A.4 Dynamic Tensile Characterization Strain Maps

As in the quasi-static regime, two dimensional digital image correlation was used to record complete strain maps for AP-PLY specimens tested to failure under dynamic tensile loading. Strain maps at two nominal principal strain levels were presented for both cross-ply and quasi-isotropic AP-PLY laminates in Chapter 7. A complete set of strain maps is reproduced here in Figures A.5 and A.6 at the same nominal strain levels. Since the dynamic experiments were very short in duration, there were fewer frames to analyze using DIC, this reduces the size of each dataset significantly. Note that different fields of view were used for the cross-ply specimens, resulting from different test setups on different days of testing. The smaller field of view used for specimens 1 through 4 makes it more difficult to capture the strain concentrations occurring at the specimen shoulders.
Figure A.5: Strain maps of cross-ply AP-PLY specimens under dynamic tensile loading, at 0.6% and 0.8% nominal principal strain.
Figure A.6: Strain maps of quasi-isotropic AP-PLY specimens under dynamic tensile loading, at 0.8% and 1.0% nominal principal strain.


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[20] C. Edmond, “If airlines were a country they’d be one of the world’s top 10 greenhouse gas emitters,” 2019.


