

CRANFIELD UNIVERSITY

Amit Ramji

Damage Tolerance Enhancement of Thermoset Composites
Modified with Thermoplastic Veil Interleaves

School of Aerospace, Transport and Manufacturing
PhD in Aerospace Engineering

PhD Thesis
Academic Year: 2016 - 2019

Supervisor: Dr. Yigeng Xu
Associate Supervisors: Dr. Mehdi Yasaee and Prof. Phil Webb

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This thesis is submitted in partial fulfilment of the requirements for
the degree of PhD

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ABSTRACT

A feasible method of overcoming the stringent certification and ecological sustainability objectives within the aerospace sector is to substitute conventional engineering alloys with high-performance lightweight composite materials for primary structures. Although, widespread utilisation of laminated composites are constrained by the significant reduction of the residual compressive strength stemming from a lack of through-thickness reinforcement leading to interface delamination. This work presents the detailed investigations on the characteristics of the fibre-matrix interactions in conjunction with ply interface enhancements; to develop new knowledge for damage tolerance enhancement of thermoset composite laminates modified with thermoplastic veil interleaves.

The influence of fracture initiation and propagation is the focus of this work, through the experimental exploration of various failure mechanisms. The failures of toughened carbon-fibre/epoxy laminates with poly-phenylene sulfide (PPS) veil interlayers are considered under; inter-laminar, intra-laminar, fracture migration and low-velocity impact damage.

Conservative test standards for the interlaminar fracture resistance present veil interleaves as having a 48% enhancement on the interlaminar fracture parameters for longitudinal propagation of unidirectional (UD) fibre reinforced polymer (FRP) laminates. However, interleaving enhancements are revealed to be over-projected when fracture propagation is transverse to the interfacial fibre orientation with enhancements being <2%. Under mode I loading, intra-laminar fracture plane migration is the dominant damage mechanism for transverse yarn interfaces, which trigger migration away from the toughened interface.

In contrast, woven lamina cannot permit through-thickness migration due to yarn interlacing, where 136% for longitudinal bias and 20% for transverse biased enhancements are apparent using 20 gm⁻² PPS. Furthermore, reductions of interfacial fibre orientation bias sensitivity were demonstrated for veil interleaved interfaces. Similarly, under mode II loading, transverse biased interfaces exhibited up to 75% improvement in delamination resistance with gradual growth characteristics resulting from the veil and transverse fibre interactions.

Interleaved laminates were subsequently explored for low-velocity impact damage, emphasising on veil distributions within the laminate. Interleaving the outside interfaces transformed the damage from inter-laminar to intra-laminar and increased the delamination threshold load (DTL) by 9% or entirely removed DTL occurrences for heavily interleaved layups. The veil damage mechanisms resulting in improved fracture resistance and enhanced damage tolerance are due to the inclusion of lower modulus PPS fibres. These discrete thermoplastic regions within the laminate provide a cushioning effect while increasing the localised ductility resulting in increased fibre-matrix interactions at the fracture process zone.

KEYWORDS

Carbon fibre composite

Damage tolerance enhancement

Delamination damage mechanics

Fibre bridging

Fracture toughness

Impact behaviour

Interphase

Matrix cracking

Non-woven Nano veils

Veil interleave toughening

ACKNOWLEDGEMENTS

I would like to thank all those who made this project possible. I am particularly appreciative to my supervisors; Dr. Yigeng Xu, Dr. Mehdi Yasaee and Prof. Phil Webb, who provided the opportunity, motivation, support and guidance throughout my research and development of this project.

I would also like to thank Dr. Marzio Grasso for his collaboration, in-depth reviews, support and guidance while preparing and publishing valuable papers resulting from this work. In addition, I thank Prof. Phil Irving from whom I have learnt an indispensable amount regarding the topics of testing and evaluation.

Furthermore, I am incredibly thankful to Technical Fibre Products for supplying the PPS veils used in this project. Alongside, Jim Hurley and Jarryd Braithwaite for their assistance in specimen manufacture and testing respectively. James Watson and Jamie Bristow from the Cranfield Impact Centre were also a significant part of the project during the impact testing and provided their valuable time and dedication towards the completion of this work. Additionally, I would like to thank Barry Walker for all the NDT sessions performed in his lab, his decades of knowledge and experience in equipment, testing and Land Rovers!

A special thanks to my parents, siblings, including my niece and nephew for all the dinner preparations, carrying out my share of chores and being pleasant distractions while writing the thesis and associated papers. I am especially grateful to my parents for their sacrifices and struggles they have made for the benefit and support of my learning and career; it is one of the fundamental reasons for my accomplishments so far. A further dedicated acknowledgement is to all my friends and their families, who have demonstrated an abundance of patience and support during my endeavours.

Last but not least, I would like to thank all the staff of SATM at Cranfield University for accepting and awarding the PhD studentship. It is my hope that Cranfield University continues to support, provide resources and fund vital research projects like this in the future.

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LIST OF ABBREVIATIONS

3ENF	Three-point bend end notched flexure
4ENF	Four-point bend end notched flexure
5HS	5-Harness Satin Weave
ABD	ABD Matrix for CLT
ASTM	American society for testing and materials
BVID	Barely Visible Impact Damage
CAI	Compression after Impact
CBT	Classical beam theory
CF	Carbon Fibre
CFRP	Carbon Fibre Reinforced Polymer
CLC	Combined Loading Compression
CLT	Classical Lamination Theory
CNLB	Centre notched laminate beam
CNT	Carbon Nano-Tube
DCB	Double Cantilever Beam
DMTA	Dynamic Mechanical Thermal Analysis
ELS	End loaded split
ENF	End Notched Flexure
FRP	Fibre Reinforced Polymer
FVF	Fibre Volume Fraction
GF	Glass Fibre
GFRP	Glass Fibre Reinforced Polymer
gsm	Grams per square metre
HP-RTM	High-Pressure Resin Transfer Moulding
IFT	Interlaminar Fracture Toughness
ILSS	Interlaminar Shear Strength
ISO ISS	International organisation for standardisation
MBT	Modified beam theory
MMB	Mixed-mode bending
MMELS	Mixed-mode end-loaded split
MWCNT	Multi-Walled Carbon Nano-Tube
NCF	Non-Crimp Fabric
NDE	Non-Destructive Evaluation
NDI	Non-Destructive Inspection
NDT	Non-Destructive Testing
NPC	Non-pre cracked
NWCT	Non-woven carbon tissue
NWGT	Non-woven glass tissue
NWPT	Non-woven polymer tissue
NWT	Non-woven tissue

OHC	Open-Hole Compression
OHT	Open-Hole Tension
PA6	PolyAmide 6 (Nylon)
PA66	PolyAmide 66 (Nylon)
PBI	Poly-benzimidazole
PBT	PolyButylene Terephthalate
PC	Precracked
PE	PolyEthylene
PEEK	PolyEtherEtheKetone
PEI	Polyetherimide
PEN	PolyEthylene Naphthalate
PES	Polyethersulfone
PET	PolyEthylene Terephthalate
PP	PolyPropylene
PPS	Poly-Phenylene Sulfide
PTFE	Polytetrafluoroethylene
PU	Polyurethane
PVAL	Polyvinyl alcohol
PVB	Polyvinyl butyral
R-Curve	Resistance Curve
RI	Resin Infusion
RTI	Resin Transfer Infusion
RTM	Resin Transfer Moulding
SCB	Single Cantilever Beam
SEM	Scanning Electron Microscopy
SERR	Strain Energy Release Rate
SLB	Single leg bending
TDCB	Tapered double cantilever beam
Tg	Glass Transition Temperature
TP	Thermoplastic
TS	Thermoset
UD	Unidirectional
UTS	Ultimate compressive strength
VARTM	Vacuum Assisted Resin Transfer Moulding

NOMENCLATURE

G	Strain energy release rate
G_c	Critical strain energy release rate
G_P	Apparent crack propagation resistance
G_I	Strain energy release rate in mode I
G_{II}	Strain energy release rate in mode II
G_{IC}	Critical strain energy release rate in mode I
G_{I_Eff}	Apparent / Effective crack propagation resistance in mode I
G_{IIC}	Critical strain energy release rate in mode II
G_{IIP}	Apparent crack propagation resistance in mode II
G_T	Total critical mixed-mode strain energy release rate
V_f	Fibre volume fraction
RW	Resin weight
h	thickness of beam specimen
b	Width of beam specimen
L	Specimen length
a	Crack length
a_0	Initial crack length
P	Load
δ	Load point displacement
C	Specimen beam compliance (δ/P)
C_{MMB}	MMB specimen beam compliance
A_I	Mode I crack length correction factor (DCB specimen)
A_{II}	Mode II crack length correction factor (ELS specimen)
E_{11}	Longitudinal modulus of elasticity
E_{22}	Transverse modulus of elasticity
E_{33}	Through thickness modulus of elasticity
E_f	Flexural modulus of elasticity
G_{12}	Shear modulus of elasticity
ν_{12}	Poisson's ratio (displacement in direction 1 and contraction in direction 2)
K_{int}	Interface stiffness
K_I	Maximum cohesive interface mode I opening stiffness
K_{II}	Maximum cohesive interface mode II shearing stiffness
σ_{int}	Interface strength
σ_E	Maximum interface strength
σ_I	Maximum cohesive interface mode I opening strength
σ_{II}	Maximum cohesive interface mode II shearing strength

1 Introduction

1.1 Background

Fibre Reinforced Plastics (FRP) have been continually increasing in usage since the past half-century, owing to their higher specific strengths and stiffness. Today, the growth of the composites industry has inflated to the extent that the global composites market is predicted to be worth \$40 billion by 2023 [1]. This growth is often attributed to the applicability of composite materials within high-strength aerospace applications (including others, e.g. automotive, consumer products and energy sector) owing to its ability to introduce significant weight savings and emission reductions. However, the implementation of composites to a broader range of markets, for instance; wind energy, marine, oil and gas, construction and sports equipment accounts by volume as a majority of the market share [1,2]. Carbon Fibre Reinforced Polymers (CFRPs) are the desired material for use in advanced sporting equipment and motorsport applications for accessories and lightweight structures. The parallel increases in usage lead to dramatic growth in the global composites industry, making them a viable investment option. CFRPs are undoubtedly the most popular for high performance and lightweight applications. Increasing demand as a result of increased performance, coupled with a reduction in costs, has made CFRP materials a desirable material option for a broader range of industries involving everyday consumer products.

Although CFRP composites are a viable option for extreme lightweight aerospace applications, impeding factors involving their post-damage performance has limited its widespread application. Composite materials are applied conservatively for structures involving human safety, resulting from a lack of detailed awareness of the fracture mechanisms and predictability of failures compared to conventional isotropic materials. Aerospace composites, in particular, are prone to frequent and often undetectable impact damage. Their steady and ever-increasing use in aircraft structures requires improved design methodologies and life cycle predictions. Low interlaminar toughness is an inherent property of non-woven and two-dimensional woven composite laminates where it is accountable for the mechanical degradation of an impact site.

This work endeavours to investigate, evaluate and understand the methods to increase the damage tolerance abilities of laminated composites, particularly during low-velocity impact (LVI). An LVI event is highly significant in aerospace applications, as damage is often undetectable on the surface, resulting in the need for specialist Non-Destructive Testing (NDT) equipment to scan the damage extent.

Composites have a reduced tolerance to impact, which is a crucial problem for the future development and deployment within lightweight structures. LVI events can transpire at any point during the manufacture, servicing and maintenance of a structure. As a result, toughening techniques have been developed to improve CFRP's performance and post-impact mechanical behaviour such as their in-plane compressive strength associated with interlaminar performance.

1.2 Research Motivation

Advanced composite materials, such as those used in primary structure applications within the Aerospace, Automotive and Maritime sectors, are broadly increasing in popularity. Although composites offer many benefits and are an attractive material option, the post-damage disadvantages mentioned above impede their widespread deployment. Composites possess low compression strength, are susceptible to delamination and generally have a weak impact tolerance. To fully exploit their beneficial capabilities, the brittle characteristics, defect intolerance, and early fracture rates are some of the many areas being addressed by composite designers and research groups.

The use of composites brings about intrinsic anisotropy within the material microstructure, leading to complex loading and multiple interacting failure modes. The significant failure mechanisms for composites are organised into the following clusters: inter-laminar, intra-laminar and trans-laminar modes. Unlike in isotropic materials, which fracture normal to the direction of principal stress; composite fractures are often made up of a combination of failure modes [3].

In comparison to metallic alloys, composites offer excellent mechanical advantages; however, such traits are sacrificed while under longitudinal compressive loads. Whilst subjecting a composite structure to in-plane compressive loads, delamination formation is a significant concern as the

structural strength is locally diminished. Locally lessened strength in a composite structure is caused by the load-carrying fibres undergoing micro-buckling. Additionally, it can be caused by drastic differences in fibre orientation between lamina, leading to failure of the overall composite structure.

Crack sensitivity is a frequently used parameter when examining the defects within the microstructure and often under LVI. Composite designers use crack sensitivity and fracture toughness under three modes to relate the crack growth potential and energy to fracture of a composite structure to design often over-conservative structures. The key focus of this research is to improve impact characteristics with the introduction of thermoplastic fibres in-situ with conventional thermoset composites. Although research on thermoplastic additives is not a new concept, the understanding of their role in the micro-mechanics of fractures is still not fully understood concerning their actual fracture behaviour, relative fibre-matrix interactions and strain concentration regions.

LVI damage is significant for aerospace applications due to its undetectable nature, frequency of occurrence and potentially catastrophic consequences. This project targets to advance the knowledge of damage mechanisms and enhance the damage tolerance by interleaving laminated composites for LVI. Additionally, it will explore the feasibility of understanding ideal interface positions for the non-woven veil placement. The interface predictions and properties of unidirectional (UD) composites and woven fibre composites can be later adapted with angle and orientation factors in three-dimensional reinforcements by ultimately commingling within fibre tows.

The LVI damage response and fracture mechanisms of interleaved composite laminates will be investigated experimentally, alongside the standard; flexural, mode I, and mode II mechanical testing practices. Variable parameters such as laminate thickness, load-carrying fibre material and resins will be limited to pre-impregnated materials which are comparable to those within the aerospace and automotive structures. Additional parameters such as lay-up configurations, loading and curing will be customisable in each application using established methods of composite sizing and processing.

Toughening methods include; fibre surface treatments [4–12], commingling with tougher fibres [13–19], interleaving [20–26] and through-thickness Z-pinning [27–32]. Intermingling and interleaving thermoplastic fibres have demonstrated to improve the impact damage resistance of polymer matrix composites by creating non-woven areas of fibre interaction and enhanced bridging. Such bridging mechanisms are an inherent feature of composites and essential to the dissipation of energy during loading. By using interleaving techniques, it is possible to introduce the characteristics of thermoplastics locally to modify the interlaminar properties at bespoke locations within a laminated structure.

The influence of Z-pinning has also shown improvement in the delamination resistance under mode I & II [27–32]. However, combinations of Z-pinning and interleaving has never been previously investigated. The combination may provide exciting results as the thermoplastic interleaves would reduce the propagation distance for a given crack opening, allowing the Z-pins an opportunity to improve the through-thickness behaviour. The concept of Z-pinning in combination with thermoplastic toughened laminates is a topic beyond the scope of this work; however, the combined effects are thought-provoking due to the veil potentially reducing the strain conditions.

The interface or more appropriately termed interphase [13,33] is assessed in detail at the initial stage of this project; being an inherent source of problems when designing composites. The interphase region of thermoplastic veil interleaving is not a fully established topic where publications generally suggest a careful and complete consideration is required for toughened composites. For interleaved and commingled composites, interface regions require further research based on the recent peak of research interest [17,19,34–36]. The current literature illustrates the importance of the interphase; however, specific commingling and interleaving literature are scarce. The interleaving method is introduced in section 2.10.4 and fibre level commingling in section 2.10.3, along with the historical development and current status of knowledge.

Toughening using interleaving is used as an intermediate step to understand the interlaminar behaviour of the interface while using thermoplastic additives. The test materials, detailed configurations, and procedures are located across various research papers within Chapters 3-4.

1.3 Aim and Objectives

1.3.1 Aim

Develop knowledge for enhanced damage tolerance design of composite structures in the aerospace industry using interleaved veils.

1.3.2 Objectives

1. Review current materials and methods for toughness enhancement of laminates.
2. Understand toughening mechanisms by investigating the effect of interfacial fibre orientation and interleaving on delamination resistance.
3. Understand the impact performance of composite laminates by varying the through-thickness veil distribution.
4. Provide the industry with further knowledge for the enhanced damage tolerance design of laminated composite structures.

1.4 Thesis overview

This thesis delivers the findings from the research activities concerning the *“Damage Tolerance Enhancement of Thermoset Composites Modified with Thermoplastic Veil Interleaves”*. It consists of seven chapters where research activities are arranged into a series of papers, where the key sections are illustrated in Fig.1. Chapter 1 provides an overview for the motivation connected with the research endeavours along with describing the overall project aim and associated objectives. Subsequently, Chapter 2 delivers a complete assessment of the concepts, theory, practices and significant outcomes in the area associated with fracture arrest and damage tolerance enhancement methods to meet the first objective.

Chapter 3 consists of four research papers considering the interfacial influence of interlaminar fibre orientation of UD and 5HS CFRP laminates when combined with veil interleaving to meet the second objective. The fracture toughness investigations can primarily identify the compatibility of the materials and curing conditions. To minimise the influence of the increased inter-ply thickness of the interface, a low areal density for veil interleaves are selected.

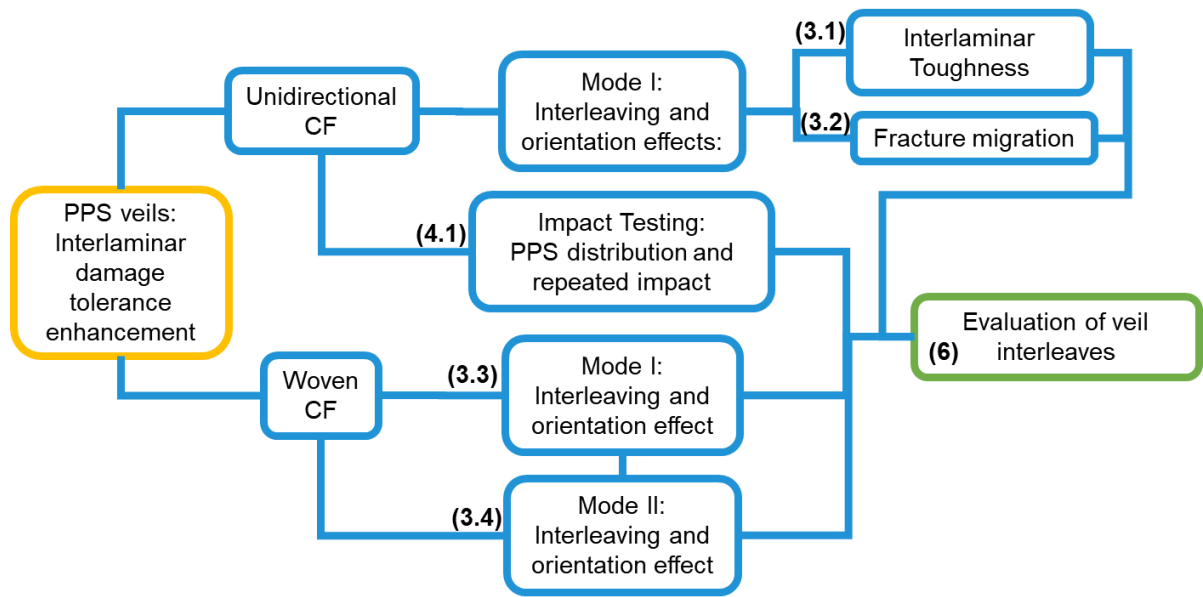
Section 3.1 considers the *“Effect of Interfacial Fibre Orientation and PPS Veil on Delamination Fracture Resistance of UD CFRP Laminates under Mode I*

Loading". Likewise, Section 3.2 considers the *"Delamination Migration in Multidirectional CFRP Laminates under Mode I Loading"*. Section 3.3 explores the *"Effect of Interfacial Fibre Orientation and PPS Veils on Fracture Resistance of Woven CFRP Laminates under Mode I Loading"*. While section 3.4 examines the *"Influence of Interfacial Fibre Orientation and PPS Veils on the Fracture Resistance of Woven CFRP Laminates under Mode II loading"*.

Chapter 4 consists of a research paper exploring the influence of veil distribution on UD CFRP laminates under single and repeated LVI to meet the third objective. The variation of interleave quantity and position within the laminate allows studying the damage cross-section to identify differences and observe the failure mechanisms. Repeated impact tests can also be used to identify any differences in further damage progression when impacted at the same and shifted locations. Section 4.1 considers the damage in UD laminates, specifically; *"Influence of Veil Distribution on the Delamination of Multidirectional CFRP Laminates under Single and Repeated Low-Velocity Impact."*

Chapter 5 summarizes and discusses the applicability of thermoplastic additives, in particular, veil interleaving. Chapter 5.1 describes the research methodology used within this project to determine the deepest method of exploration for the topic of toughening traditional thermoset composite laminates with thermoplastic inter-layers.

The research review and key results are concluded in Chapter 6 against the aims and objectives initially set out to meet the fourth objective. The chapter provides a summary of the added value and understanding concerning the damage tolerance enhancement capabilities using thermoplastic veils. Chapter 7 contains the topics where the further scope of work is envisioned for future toughening of thermoset CFRP's.



(N) = Section / Chapter

Fig.1 Thesis overview.

2 Literature Review

2.1 Introduction to composite materials

A typical composite material is manufactured with a minimum of two components of vastly differing mechanical and physical properties. When correctly harmonized together, composite materials produce a material with far superior performance characteristics than the individual components. Advanced composites offer a lower density, improved specific strength and stiffness, excellent fatigue life and good electrochemical resistance that are not achievable by the separate constituents [37–40].

The constituents are combined to form composite materials which has distinct interfaces between each other, often referred to as the interphase. The performance advantages as above has allowed composite materials to become a significant category of engineering material with global annual production of composite materials increasing dramatically in recent years. The various disciplines and sectors show increasing composites usage by 5-10% per annum with a predicted market value of \$40 billion by 2023 [1,2].

The aircraft industry has contributed to the growth of advanced composites, demonstrated within Airbus's XWB and Boeing's Dreamliner platforms. Aerospace composite materials provide a proven method to simultaneously improve the structural behaviour, reduce parts, weight and enhance the resultant fuel efficiency.

2.2 The categorisation of composite materials

Composite materials are made up of distinct discontinuous regions (fibres, flakes, particles, pellets, etc.) surrounded by a continuous portion (matrix). The discontinuous constituents are typically the reinforcement which are stronger and stiffer in comparison to the continuous portion.

Distinct material property differences between reinforcement and matrix warrant two major grouping methods being commonly applied in composites. The first being the matrix material type: polymer, metal or ceramics. The second being the geometry of the reinforcement being a fibre, particulate, pellet and flake reinforced composite. Further groupings include the reinforcement distribution;

for example, fibres being unidirectional (UD) or woven into multi-directional fabrics. A further adaptation for the groupings is for hybrid composite materials made up of multiple reinforcement materials with further differing properties.

Although the higher specific strength and stiffness of reinforcements are an advantage, the matrix remains necessary owing to the brittle characteristics of typical reinforcement materials. The matrix participates in load transfer between reinforcement (typically fibres) through the interphase region and encapsulates the reinforcement in their intended alignments while also preventing degradation from operational and environmental factors. While fibres restrict cracking along the fibres, matrix cracking, inter-laminar and intra-laminar failure remain a concern and therefore is the focus of this research.

The anisotropic characteristics of fibre reinforced composites arise from discontinuous regions between the matrix and reinforcement components. The low transverse or through-thickness strength must be considered for composite structures, as poor interlaminar resistance can be detrimental to their application. The detailed study of failure mechanisms and difficult detection of internal damage are aspects worth considering, which ultimately lead to costly and problematic repairs.

2.3 Thermoset Vs Thermoplastic

The benefits of having a high specific-stiffness and strength within conventional thermoset matrix laminates result in strong and lightweight structures. However, thermoset matrix composites often suffer from abrupt brittle failure rather than delayed and gradual failure. Thermoplastic matrix materials offer tougher characteristics based on their increased ductility, however with decreased stiffness and in-plane strength.

Thermoplastic matrix composites also permit unlimited storage life at room temperature, have a shorter fabrication time for high volume applications and offer post-formability (thermoforming). Thermoplastic matrix composites can also be assembled by solvent bonding or induction welding, have simple handling requirements (no tackiness hence low contamination risk), can be recycled and reprocessed, with good impact toughness [41].

The limitations for thermoplastic composites are; restrictions on part thickness due to matrix viscosity, low creep resistance, low-temperature resistance, and difficulty in welding and joining of parts. Recent advances in wind turbine blade manufacturing call for the use of wing-like structures using spars and ribs for manufacturing large turbine blades [41]. One of the reasons for the limitations and using airframe type structures is that solid root sections of the airfoil cannot be manufactured in thermoplastic due to component thickness and heat transferability limitations. The welding and joining of parts can however be accomplished using sacrificial heating coils or pads placed between the weld surfaces to join the parts with the addition of pressure [41].

Nevertheless, the growth of thermoplastic matrix composite materials is increasing where the growth of thermoplastics is often attributed to the wide-ranging variety of structures within the aerospace, automotive and consumer product applications [1]. Strict and conscientious demands in these sectors must be met, especially for materials critical to public and environmental safety. The safety considerations must include; high oxidative and thermal stability, good toughness, solvent resistant, low dielectric conductivity, high strength and stiffness.

Advances in thermoplastic composite components in the aerospace sector can be observed by example in the Airbus A380, A340-600 and A400M for wing leading-edge applications. Fernández *et al.* [42] have demonstrated the successful manufacture of leading-edge wing structures and stiffened panel skins using autoclave co-consolidation. The literature describes advantages by using thermoset matrix composites, including; improved toughness and damage resilience, swift fabrication cycles, and plausibility of assembly by induction welding. Entirely thermoplastic matrix materials are however not the focus of the project but their desired ductility could be implemented in hybrid composite materials.

2.4 Composite laminates

Composite materials in their laminate form are the typical arrangement used in the aerospace sector owing to their superior mechanical properties, consistency and established manufacturing methods. However, the understanding and treatment of interphase failure mechanisms is a significant topic of concern. The

inhibiting trait of internally damaged composites and difficult detection in most situations requires a myriad of Non-Destructive Testing (NDT), becoming costly during the service life of a structure. Another approach to resolve the issues and ultimately design for these eventualities is by establishing a deeper understanding of factors affecting the failures. The detailed understanding of failure mechanisms ultimately lead to hybrid and toughened composite laminates which have recently demonstrated toughness enhancements. The toughness enhancement methods are reviewed and explored within this research to improve the understanding and verify the level of enhancements achievable to continue the broader applicability of composites in the future.

2.4.1 Laminate characterisation

Laminated composites are manufactured by stacking layers in various orientations and being infused within a continuous matrix. Their modern use has been appreciated within many consumer products, such as building materials and other stacked ply materials. However, the focus of this work will be limited to fibre reinforced composite laminates used mostly in aerospace, automotive and energy applications. The typical types of fibre reinforced architecture are presented in Fig.1.

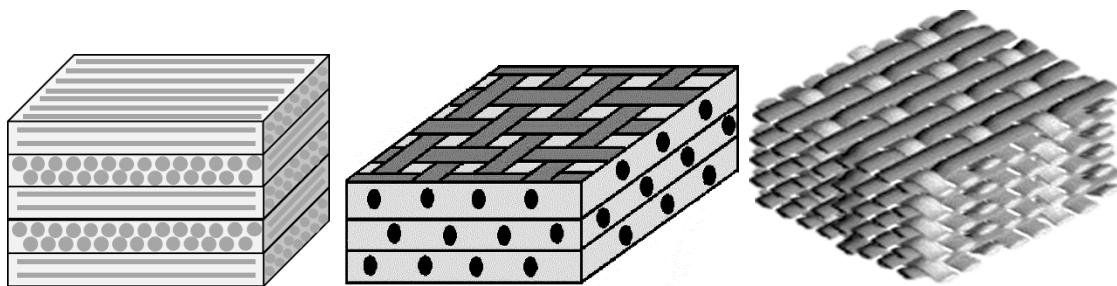


Fig.1 Typical fibre reinforced laminate architectures for: UD (left), Plain Weave (middle) and 5-Harness-Satin (right).

Alternative layup configurations other than those illustrated in Fig.1 are also extant, including bi-axial, various twill patterns, satin weaves, and stitched combination fabrics. The use of each architecture is dependent on the application, with each having different mechanical properties and analysis methods.

Manufacturing processes for laminated composites also vary depending on their application, geometry and mechanical requirements. The use of an internal core is possible in addition to combining the various fabrics with the caveat that resins and interphases maintain compatibility.

2.4.2 Composite laminate layups

Supplementary to the various reinforcement fabrics available, reinforcement contents, volume fractions, electrochemical and fire-resistant properties can also be configured accordingly to suit the desired application. The major advantage of composite laminates is the ability to adjust the mechanical and structural performance such as stiffness, strength and impact response. The ability to form structural parts that are extremely stiff in bespoke directions and versatile in other directions is achievable by adjusting the fibre orientations and stacking arrangement. The relative inter-ply orientation is also a vital consideration, as internal failures attributed to the presence of miss-matched plies and localised strain concentrations typically lead to failure propagation.

2.5 Impact of Composite Laminates

Impact damage tolerance is a crucial consideration for composites, particularly for safety-critical applications stemming from the detrimental decline of structural strength under post-impact loading. Impact damage can occur at any time where its detection possibilities are minimal until near-catastrophic failure and often overcome by embedding light-sensitive dyes [43]. Under Low-Velocity Impact (LVI), composites can undergo internal damage which are typically barely visible [44–47] on the surface from tool drops, stone chips, knocks by operators and extreme weather conditions. LVI is sometimes considered more significant based on the detection difficulties, whereas high velocity or high energy impacts are known events and mostly exhibit surface damage. The underlying damage within a composite laminate if undetected can catastrophically fail, resulting from the internal stress raiser coupled with the inability to transfer shear stresses from the initial damage. Improving the understanding of these typical failures and methods to improve the resilience of composite laminates forms part of the objectives for this research.

In comparison to failures of traditional isotropic materials, the failure mechanisms within composite laminates are more elaborate and complex. Under LVI, the damage modes may be at inter-laminar, intra-laminar, trans-laminar or combinations of such levels with interacting failure regions. Consequentially from LVI and associated damage detection difficulties, LVI will be one of the major focus of this study in combination with exploring methods of toughening and improving the damage resistance.

2.5.1 Low-velocity Impact

LVI remains relatively open to interpretation, as the transition region between low-velocity and high-velocity is often debated by many research groups. The commonly accepted velocity for the maximum velocity transition to be classified as a low-velocity is 10 m/s [48]. Localised damage can occur from the impulse type loading during an impact event, where the laminate's feedback response is dominated by 'stress-waves' propagating within the material. Impact severity is hence dependent on stress wave propagations, where high velocities having short impact durations prevent a global structural response. The inertia effect of thicker and larger laminated components can result in localized damage or penetration depending on the energy and material combinations. LVI events can impose a complete structural response due to increased impact durations, thus allowing elastic energy absorption with some localised damage. The localised damage which is often barely visible is the focus of this work due to its challenging nature of detection.

Additional considerations are also placed on the impact-energy rather than the impact velocity alone. The striker mass and carriage during an impact test or real-life impact event is undoubtedly dependant on the size and mass of the object [49–51]. For this investigation, LVI will be followed as defined in ASTM 7136/M [52] with considerations of energy, velocity and impactor dimensions.

2.5.2 Damage tolerant design

The overarching goal from this research is to evaluate the approach used for the Damage Tolerant Design (DTD) of composites. Damage, as mentioned above, can arise from; tool impacts during assembly and service, surface imperfections, cut-outs, holes, fastening features, and foreign object impact. Each of these

events or features can have severe consequences on the future strength of the structure as a result of damage propagation and reduced in-plane and mechanical performance. DTD has been applied within the aerospace sector and it is supplementary to the safe-life design methodology. Safe-life design methods ensure operating stresses are maintained below the material fatigue strengths using Load-Cycle (S-N) curves. DTD was founded through fracture mechanics and focused on crack growth processes by determining the Stress Intensity Factor, k (SIF).

In summary, DTD for a material consideration should have the ability to resist fracture from pre-existing damage for a given period; essential for components prone to safety-critical failure involving property or loss of life. DTD is, however, a conservative method of analysis; where a sizeable initial crack is assumed, and component life before crack initiation is not considered. Fracture mechanics methodologies could be extended to include smaller crack sizes, providing an improvement to the estimated life with increased accuracy and avoiding over conservatism.

Consideration of DTD is beneficial for aerospace sectors; to plan service intervals, reduce weight, predict detectable cracks, plan for cracking rather than oversizing, and design against non-inspectable cracking within primary structures. There are obvious differences in the DTD of composites compared to their metallic counterparts, primarily the non-linearity of crack growth rates and propagation lengths. The inherent non-linearity in composite crack growth is due to bridging effects, namely; filament sliding, delamination, breakages, asymmetry (in laminates), internal or surface imperfections, and many other processing factors. Composites are known to have unique crack growth-rates within the material microstructure due to the presence of multiple load paths. However, composite failures are non-linear and the damage is difficult to predict as a result of their inherent anisotropy.

Another consideration of over-conservatism is where fracture toughness is concerned for coupon level material characterization. Standards used to define the fracture toughness of laminated coupons often require the use of plies with fibre orientations aligned to the growth path. ASTM D5528/13 [53] and ASTM D7905 [54] suggests that “test laminates being unidirectional with delamination growth occurring in the 0° direction”. It also states that “toughness values obtained by this test method for delamination growth at $0^\circ/0^\circ$ interfaces may not

be representative of the toughness required for delamination growth at interfaces with different relative ply orientations". The procedure, although conservative, cannot capture the exact behavioural characteristics of multi-directional laminates and propagation across transverse (off-axis) plies.

The in-plane compressive performance with the continuation of loading is a limitation of laminated composite materials. The level of in-plane compressive strength is quantified by Compression-After-Impact (CAI) [55] or Open-Hole-Compression (OHC) test methods, to evaluate the post-impact residual strength remaining within the laminate. Impact damaged laminates often contain fibres that are non-supported by the matrix and highly sensitive to micro-buckling. As a result, the compressive damage performance is often an over-conservative and critical consideration used for structural aerospace components. The through-thickness importance being significant such that three-dimensional woven and through-thickness stitched materials are developed [56–58]. Woven laminates, however, often suffer from premature compression failure due to their inherent tow interlacing architecture, hence are mainly applied in non-safety critical components [59].

The mechanisms with their exploration of enhancements associated with impact damage affecting the residual strength of laminates have been considered in great detail by many authors [17,20,23,35,59–61]. For example, a key consideration is the fibre orientation, where it was demonstrated that 0° fibres buckle effortlessly than laminates with 90° dominated fibres [62]. Larger damaged areas can also initiate a significant loss in the laminate's compressive strength. By increasing the resistance to damage initiation and propagation (increasing G_{IC} and G_{IIC}), the reduction in damage area and decreased likelihood for advanced failure is feasible.

The method to enhance the damage tolerance by simply tailoring the layup configuration was demonstrated by Hitchen *et al.* [63]. It was demonstrated that delamination initiation energy increased when 45° laminae were present towards the impact surfaces and when increasing the number of dissimilar interface plies. Dissimilar ply interfaces can hinder the propagation at the interface and either arrest the crack or trigger fracture migration and thus allow propagation in a new direction [64]. Similarly, CAI tests by Rhead *et al.* [65] demonstrated that laminates with the most robust resilience at failure are those which exploited

stacking sequences. Their work also described the stacking exploitation which prevented delamination damage towards the surface and therefore inhibited the buckling-driven delamination propagation.

During a delamination event, the laminate is locally distributed into sub-laminates with no local constraints. Delaminated regions are then susceptible to buckling under in-plane compression due to their reduced effective thickness as with slender beams undergoing buckling and owing to their reduced interlaminar shear strength [66]. CAI failure of the laminate involves a share of global buckling, localised buckling of the thinner 'sub-laminates' [67] and 'micro-buckling of the fibres' [68]. The 'micro-buckling' within laminates generally results in inter-laminar to intra-laminar matrix fracture, and local buckling facilitates the delamination growth; hence the buckling driven failure of axially loaded fibres [66,69,70]. The residual compressive strength is mainly influenced by the laminate's local buckling resistance, constraints within the CAI test fixture, presence of resin-rich or defect regions at the inter-ply interface and membrane elastic coupling effects [69].

The damage initiation process during impact is a crucial stage to isolate the fundamental material properties which influence the damage resistance and post-impact performance. The impact damage resulting from different energy levels is usually categorised into three separate levels of damage (Fig.2) [71]. Progressive damage modelling and drop weight tests have demonstrated that various regions exist in the classification of damage. Region 1 is when energy levels are up to but not exceeding the impact damage threshold – commonly referred to as Delamination Threshold Load (DTL) [45,72,73]. Region 2 is where the impact energy can cause matrix cracks and delamination. Finally, region 3, which is beyond the scope of this research due to the focus on BVID, is when energy levels are great enough to cause extended fibre breakage and penetration. For this research, the low-velocity impact is referring to levels where the damage does not span towards through-thickness penetration and remains within region 2.

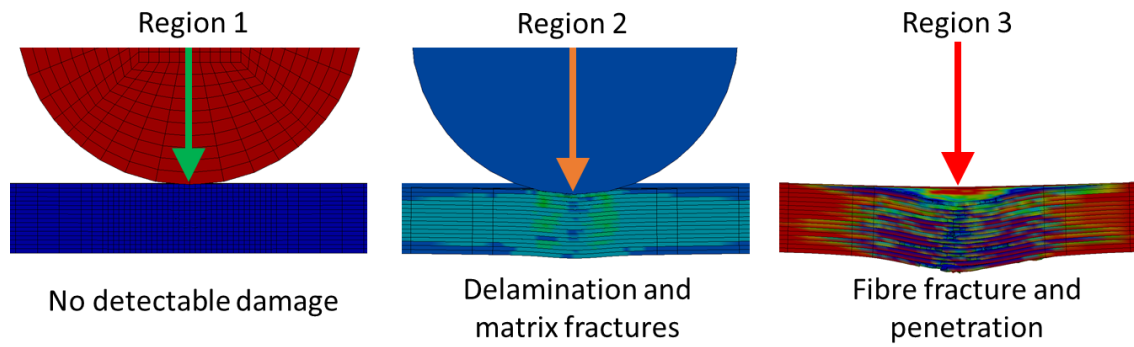


Fig.2 Classification of impact energy level after damage – adapted from [64].

Composite laminates suffer from two key failure modes during a LVI event and should be the first themes to investigate at a coupon level. These modes are; fibre-matrix interface failure and intra-laminar failure being the matrix cracking. The second being inter-laminar failure, which is the delamination between layers. Under LVI, a mixture of these failure modes can occur and are a result of the stress intensity exceeding the matrix capability, where matrix failure during LVI is the most common. Under high velocity and ballistic impacts, the dominant failure mechanisms are large-scale fibre breakages resulting from brittle characteristics of fibres combined with the high impact energy dissipation during a shorter impact time.

2.6 Key failure mechanisms in composites

Introduced earlier was the significance of delamination and fractures in laminated composites resulting from out-of-plane impact, and found to be detrimental while under post-impact loading. These impact-forces and energy dissipation propagate laterally, resulting in in-plane shear and transverse stress within the laminate, alongside bending stress components depending on the laminate thickness. As the composites are made up from mainly two components; a continuous region – the matrix and discontinuous regions - reinforcement fibres, some portions of the regions are unable to absorb and dissipate the energy, leading to the strain energy being released. The release of strain energy in a composite takes the form of crack initiations and propagations.

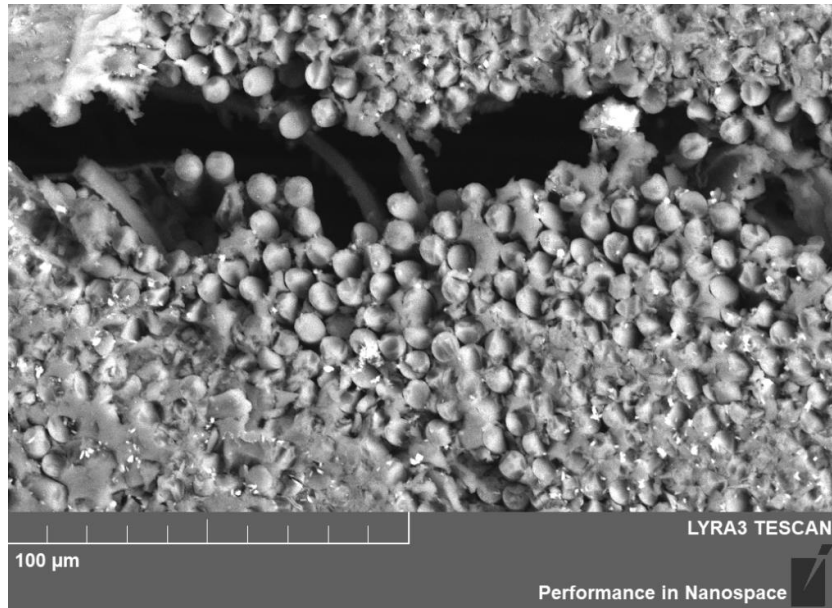


Fig.3 Typical fibre de-bonding during Intra-laminar failure.

During an impact event, transverse shearing stresses located at the upper laminae result in matrix fractures where fibre de-bonding takes hold (Fig.3). The formation of tensile cracks is subsequently induced when the normal (out-of-plane) stresses exceed the tensile strength of the inter-ply region. Additionally, shear failure fractures can be formed resulting from the transverse shear stress within the laminate. Tensile cracks are more commonly induced in thin laminates being triggered in the back-face layers due to excessive membrane-type behaviour and transverse deflection.

The prediction of damage behaviour and fracture distribution of matrix related fractures is non-trivial, where examples of such failures are provided in Fig.4 - Fig.5. Typically, thick laminates exhibit beam-type behaviours; as they are stiff, less compliant, hence introduce shearing fractures in the upper laminae resulting from increased contact forces and a lower transverse deflection. Cantwell *et al.* [48] and Abrate [74] arrived at similar conclusions in the early 1990s regarding laminate thickness and its geometry effect on fracture type and distribution. They discovered that the localised contact stress triggers matrix fractures under the impact site. The damage growth subsequently propagates from top-down away from the impact site and forms a 'pine-tree' fracture pattern (Fig.5). In contrast, as introduced previously, thin laminates have bending related stress components at the reverse side of the laminate that triggers matrix cracking at the back face

with local micro-buckling and breakages at the impact site. The characteristics of thin laminate damage mostly occurring on the back face are therefore difficult to detect within structures. For thick laminates, surface indentation is usually observed and can be inspected without NDT scanning techniques.

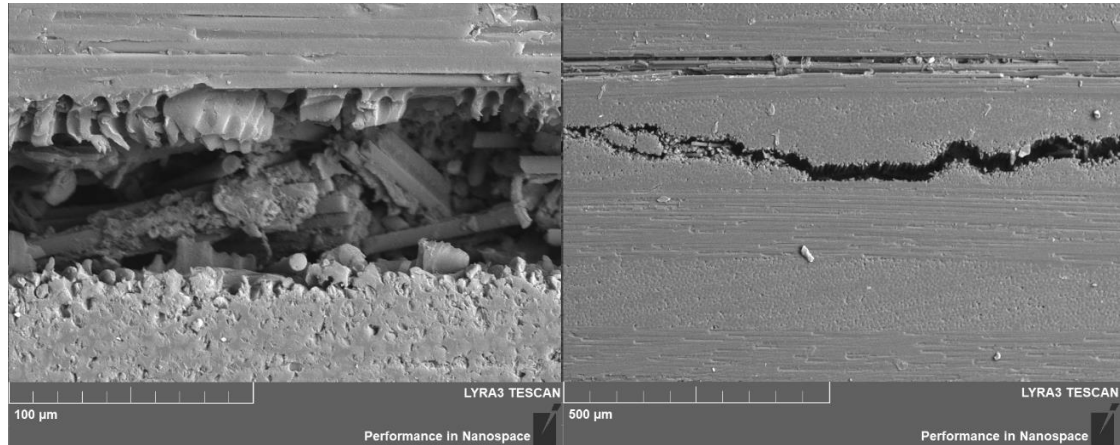


Fig.4 Post-Impact examples of Inter-laminar (left) and Intra-laminar failure (right).

Fig.5 presents an illustration of the typical damage distribution of fractures and delamination, which are commonly observed after an impact event. The impact resistance capability for absorbing impact energy is consequently related to the planar dimensions of the damaged area [75]. A larger damage area results in large and slender sub-laminates, which can buckle under in-plane compression loading.

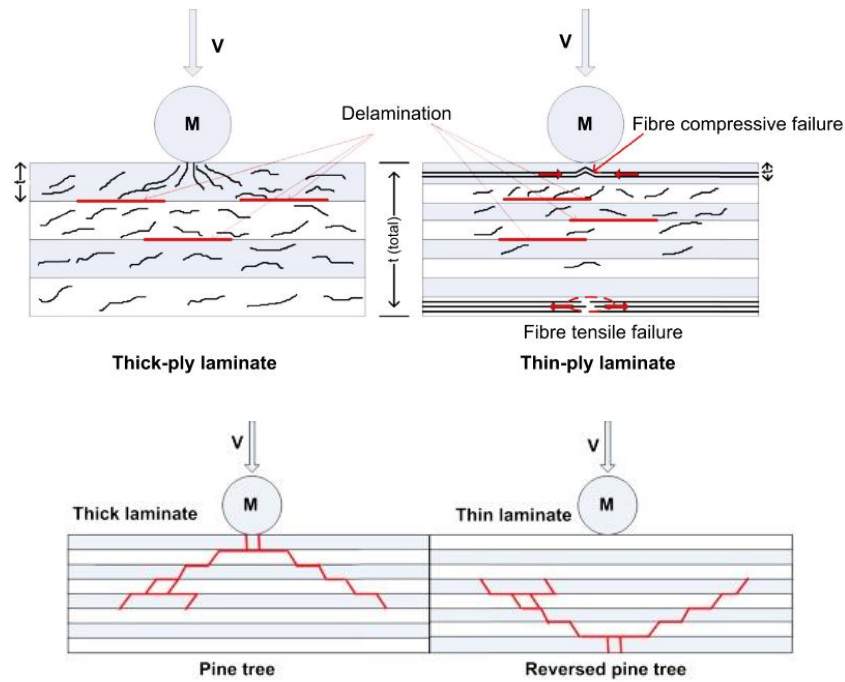


Fig.5 Typical damage patterns for thick and thin laminates [76].

2.6.1 Failure Criteria

Composite laminates have three distinct principal directions for a particular orthotropic layup configuration. The resulting principal stress plane of the laminate may have an alternative principal strain direction as a result of the layup configuration or fibre architecture. Composite laminates can also have a variety of strengths in different directions, owing to their constituents and reinforcement directionality as introduced in section 2.3. Based on the anticipated anisotropy of composites, defining the failure criteria can become more complexed than for isotropic materials. As a result, the maximum strength must also be defined in each direction which may not align with the critical loading condition of a component. Laminates must, therefore, have failure criteria determined from the evaluation between maximum loading stress fields and the allowable stress in that particular direction.

The basic limit-criteria simply accounts for failure that occurs when the laminae stress exceeds corresponding directional strengths [77]. For consideration of the basic failure limit-criteria, five strength parameters are considered; 'tensile' and 'compressive' strengths in the longitudinal (fibre) direction (σ_{xt} & σ_{xc}), 'tensile' and 'compressive' strengths in the transverse direction (σ_{yt} & σ_{yc}), and 'in-plane shear' fracture stress (σ_s).

If one considers σ_1 , σ_2 and τ_{12} as the longitudinal, transverse and shear stresses, the failure is considered when this limit is reached or exceeded by the stress limits as follows:

$$\text{Under tension: } \begin{cases} \sigma_1 \geq \sigma_{xt} \\ \sigma_2 \geq \sigma_{yt} \end{cases} \quad \text{Eqn.1}$$

$$\text{Under compression: } \begin{cases} |\sigma_1| \geq \sigma_{xc} \\ |\sigma_2| \geq \sigma_{yc} \end{cases} \quad \text{Eqn.2}$$

$$\text{Under shear: } |\tau_{12}| \geq \sigma_s \quad \text{Eqn.3}$$

Similarly, the maximum strain criteria [58] is considered to occur when strain components exceed the ultimate strain corresponding to the relative direction, expressed by:

$$\text{Under tension: } \begin{cases} \varepsilon_1 \geq \varepsilon_{xt} \\ \varepsilon_2 \geq \varepsilon_{yt} \end{cases} \quad \text{Eqn.4}$$

$$\text{Under compression: } \begin{cases} |\varepsilon_1| \geq \varepsilon_{xc} \\ |\varepsilon_2| \geq \varepsilon_{yc} \end{cases} \quad \text{Eqn.5}$$

$$\text{Under shear: } |\gamma_{12}| \geq \varepsilon_s \quad \text{Eqn.6}$$

where, ε_1 , ε_2 , and γ_{12} is the longitudinal, transverse strains, and shear strains in the material principal plane. Where the notations following ultimate longitudinal and transverse tensile strains (ε_{xt} & ε_{yt}), ultimate longitudinal compressive and transverse compressive strains (ε_{xc} , ε_{yc}) and ultimate shear strain (ε_s).

In reality, the stress and strain components along the material principal axis are usually undetermined. However, piecewise stresses in the non-principal directions are obtainable by translating with angle θ relative to the principal load and material direction, the maximum criteria are resolved. The maximum strain criterion is similarly rotated at an angle φ to establish the major and minor Poisson's ratios (ν_{12} and ν_{21}).

The maximum stress and strain criteria are preliminary methods which have seen modifications and enhancements to better predict the transition to failure. The theoretical predictions and experimental results may differ significantly in performance, resulting from the earlier simplifications and lacking the consideration of interacting failures. Accordingly, the adaptations of these basic-limit criteria are necessary for detailed composite laminate analysis and design. Further groups of limitation classification methods are the tensorial, polynomial

and quadratic interactive criteria [78,79]; these assume the failure being caused by interacting combinations of the stress components.

The polynomial criteria are adapted to consider the anisotropy of the composite laminate, which considers the interactions amongst the various fibre and matrix failure types. The ‘Tsai-Hill’ criterion [80] is a direct adaptation of the ‘von-Mises’ criterion [81]; therefore, cannot predict the initiation of ply failure. The ‘Tsai-Hill’ criterion can neither distinguish the difference between through-thickness and ply stresses of thick laminates [82]. The ‘Tsai-Hill’ criterion describes the failure occurring when the below relation is true:

$$\frac{\sigma_1^2}{X^2} - \frac{\sigma_1\sigma_2}{X^2} + \frac{\sigma_2^2}{Y^2} + \frac{\tau_{12}^2}{S^2} \geq 1 \quad \text{Eqn.7}$$

The ‘Tsai-Hill’ criteria have been adapted to interactions in tension and compression, considering the stress and strength to evolve into the ‘Tsai-Wu’ quadratic failure criterion [83]. However, ‘Tsai-Wu’ criteria require biaxial test data for the terms $F_{1/2}$. The ‘Tsai-Wu’ criterion states an adaptation of the ‘Tsai-Hill’ from the above and presented as:

$$F_1\sigma_1 + F_2\sigma_2 + F_3\tau_{12} + F_4\sigma_1^2 + F_5\sigma_2^2 + F_6\tau_{12}^2 + 2F_7\sigma_1\sigma_2 \geq 1 \quad \text{Eqn.8}$$

$$\text{where } F_1 = \frac{1}{\sigma_{xt}} + \frac{1}{\sigma_{xc}} ; F_2 = \frac{1}{\sigma_{yt}} + \frac{1}{\sigma_{yc}} ; F_3 = 0 ; F_4 = -\frac{1}{\sigma_{xt}\sigma_{xc}} ; F_5 = -\frac{1}{\sigma_{yt}\sigma_{yc}} ;$$

$$F_6 = \frac{1}{\sigma_s^2} \quad \text{and} \quad F_7 = \frac{1}{2\sigma_{yt}} [1 - (F_1 + F_2)\sigma_{yt} - (F_4 + F_5)\sigma_{yt}^2]$$

‘Physical separate mode’ criteria have also evolved, which considers two conditions: for fibre and matrix failure, where interactions can be considered. The Hashin failure criteria determine failure occurring when the following relations are apparent for [77,81,83,84]:

$$\text{Longitudinal Tension: } \left(\frac{\sigma_1}{\sigma_{xt}}\right)^2 + \alpha \left(\frac{\tau_{12}}{\sigma_s}\right)^2 \geq 1 \quad \text{Eqn.9}$$

$$\text{Longitudinal Compression: } \left(\frac{\sigma_1}{\sigma_{xc}}\right)^2 \geq 1 \quad \text{Eqn.10}$$

$$\text{Transverse Tension: } \left(\frac{\sigma_2}{\sigma_{yt}}\right)^2 + \left(\frac{\tau_{12}}{\sigma_s}\right)^2 \geq 1 \quad \text{Eqn.11}$$

$$\text{Transverse compression: } \left(\frac{\sigma_2}{2\sigma_{s_{23}}} \right)^2 + \left[\left(\frac{\sigma_{yc}}{2\sigma_s} \right)^2 - 1 \right] \frac{\sigma_2}{\sigma_{yc}} + \left(\frac{\tau_{12}}{\sigma_{s_{23}}} \right)^2 \geq 1 \quad \text{Eqn.12}$$

‘Physically-based criteria’, where each failure mode is described by specific equations as per Hashin and Rotem [79] has assigned limits for each failure mode to be considered: longitudinal (fibre direction) tension and compression, transverse (matrix) tension and compression. As presented above, limit, quadratic and polynomial criteria account for failure directionality, but physically based criteria consider the nature of failure and are deemed more valuable. A combination of limit, quadratic, polynomial, empirical and physically-based criteria need to be used to fully appreciate the real laminate failure criteria.

2.6.2 Initiation of cracks

An inherent local anisotropy problem arises in fibre-reinforced composite laminates resulting from the discontinuity and strain capability between matrix and fibres. Although the matrix is continuous, the study of interface regions are the driving-force for present and future research endeavours to eventually, enhance the fibre-matrix interactions. The stacking configuration is another consideration where grouped or dissimilar interface angles can decrease the general through-thickness strength and toughness throughout the laminate during an impact event. The tendency for laminated composites to fail at the interfaces where non-aligned fibrous layers are present is a major concern when considering the residual strength and future loading of the post-delaminated sub-laminates.

It has been established that composites can be susceptible to impact damage events within their service life from various sources. The resulting detrimental damage made up of; delaminations, fibre breakage and matrix cracking will reduce the performance capability and lead to premature failure if undetected. The detailed mechanisms and enhancement methodologies will be reviewed in this chapter. The through-thickness damage consisting of micro-buckling, matrix fractures, localized bending and the various laminar failure modes are also described in detail within this section. The understanding of the detailed source of fractures and proposing enhancements allow composites laminates to improve their performance characteristics and allow further widespread applicability.

2.6.2.1 Matrix failure

The failure of a laminate is typically caused by the matrix when the damage energy is low and slow enough as under LVI. The failure mechanisms within the matrix originate mainly when the laminae stress has transverse or shear components. The matrix failure can also take place when stress is longitudinal to the fibres under a non-uniform loading which triggers shear, buckling or pull-out forces at the fibres. The matrix failures are influenced vastly by the matrix properties; for example, brittle, compliant, have defects or rich concentration areas. The types of matrix materials used in composites are discussed in section 2.8 in more detail. However for laminates, the common materials are thermosetting (epoxy, polyester, polyurethane, phenolic cyanate ester) or thermoplastic (polypropylene, polyamide-nylon) [85].

Considering the matrix from a fracture initiation perspective; thermoset resins partake a brittle type failure compared to thermoplastics. However, thermoset materials are profoundly stiff and attractive over the elastic thermoplastics. The brittle matrix undoubtedly affects the behaviour of the overall composite, where the failure features and type of failure in the composite can be affected.

The formation of micro-fractures of the matrix occurs when the laminate is under transverse tensile loading, or when combined with in-plane shear. These micro-fractures form mostly at interfaces in the bulk matrix owing from the localised fibre-matrix strain distribution. This strain distribution is the key element when considering the initiation and propagation of fractures, where significant research efforts are underway (Section 2.10).

The typical failure of composites with a brittle matrix are presented in Fig.6(a) and ductile failure in Fig.6(b). For brittle matrix fracture observed normally in non-toughened fibre-reinforced composites, the fracture area is localised and has a clearly defined fracture plane. Ductile fracture is generally observed in toughened resin systems or where thermoplastic additives are incorporated. Ductile failures have an irregular plane with regions where fibre-matrix separation exists or remains attached with signs of plastic deformation, cusps and irregular resin remnants.

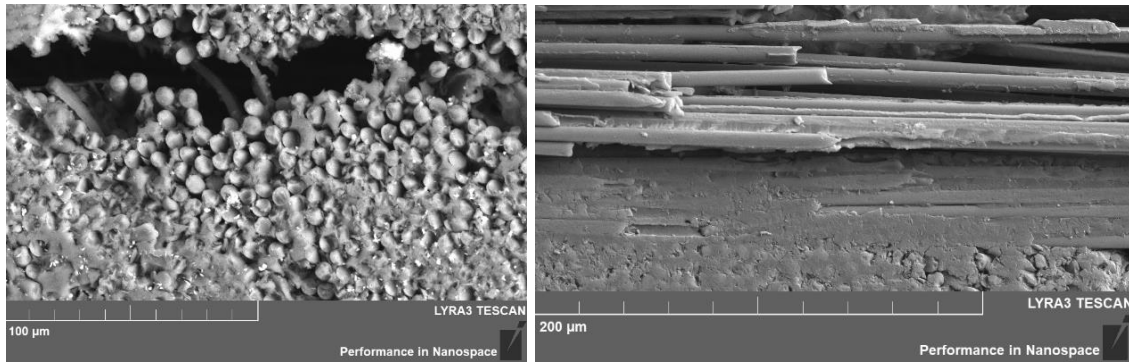


Fig.6 (a) Typical brittle fracture (left) and (b) Ductile fracture (right).

2.6.2.2 Fibre micro-buckling and kinking

During compressive loading, micro-buckling and kinking is the deformation of single fibres within the composite which can influence the formation of matrix fracture and delamination. A further factor is the shear stiffness of the matrix influencing the probability for micro-buckling [86]. Defects introduced during manufacturing including; fibre waviness, residual stresses, fibre misalignment and porosity can affect the laminate's vulnerability to micro-buckling.

Kinking is the term describing a large rotation of fibre tows causing a shear deformation of the matrix. As a result, fibre breakages can initiate at the kink regions [86] depending on the severity of the rotation. The unstable nature of kinked fibre formation and failure makes it difficult to simulate and replicate due to the microscopic scales of the damage. However, under in-plane compression, kinking of the fibres can develop from the fibre misalignment in a sequence of rotation and subsequently shearing stress within the matrix. If the shear stress in the matrix is high enough to cause micro-cracks, failure or yielding, fibres are no longer supported where further fibre rotation and failure occurs.

2.6.2.3 Local bending

The consideration of local bending failure is by assuming the plate to be a simply supported plate where the longitudinal tensile stress is on the bottom and compressive on the upper portion of the plate. The neutral axis is the site for maximum shear stress. Though delamination caused by local bending occurs when this shear stress exceeds the critical shear stress value of the matrix at any through-thickness position. Matrix crack formations can be induced, and the subsequent stress concentrations promote detrimental delamination at the interface positions. During a LVI event of thin laminates, the previous scenario

takes hold where fractures are generated as a result of local bending, where the characteristic damage is shown in Fig.5. This type of damage occurs when delamination and shear-triggered matrix fractures arise at the neutral plane, with a likelihood of tensile triggered matrix fractures occurring towards the laminate's reverse side.

The plate thickness and layup configuration affect the damage size, and both factors will affect the global stiffness and interfacial delamination magnitude [71,75,87,88]. The bending consideration of delamination between adjacent plies is demonstrated as the flexural stiffness mismatch by Liu [87]. It was also demonstrated as a method of reducing the damaged area and stiffness mismatch, the angle difference between plies must be kept to a minimum. The compiled delamination forms an elongated or elliptical shape, where the major axis of the ellipse follows the longitudinal direction of the fibres at the lower half of the interface [89].

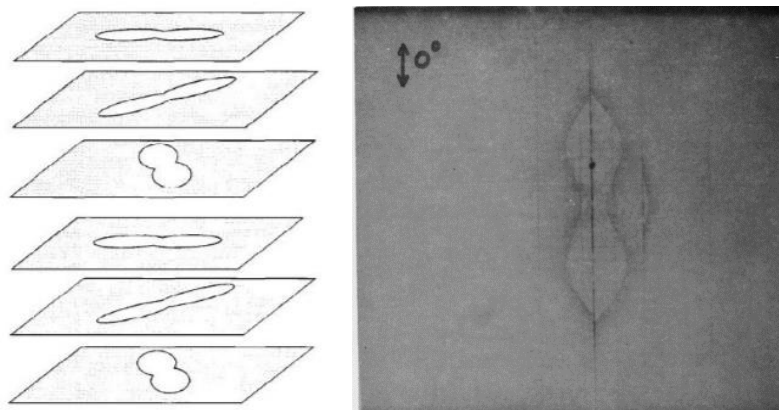


Fig.7 Peanut shape delamination pattern (left) [90] and X-Ray of delamination in graphite/epoxy plate [89] (right).

It is commonly accepted that delamination at most UD fibre interfaces will grow along the longitudinal direction of the lower ply [63,89,91]. Propagations in the transverse direction can restrict growth, where the formation of matrix cracks and fracture migration takes hold, subsequently triggering delamination further at the interface below (Section 3.2).

2.6.3 Crack propagation

There are various scenarios where the propagation of cracks can occur within a composite laminate. For example, during compressive loading, a thin laminated panel can only maintain its stiffness until it eventually buckles. The buckling strain limit of the laminate is influenced by the composite matrix, fibre components, geometry, boundary and loading properties. In section 2.6.2, the initiating mechanisms which trigger interlaminar cracks were examined. It was also established that once a kink, internal defect or damage is introduced, the composite strength is reduced considerably. Locally, the loss of stiffness resulting from the damaged region will eventually allow reduced strength and buckling under compressive strains. Buckling failure is known to be an unstable failure under in-plane compression; however, the factor affecting global buckling is the formation of intra-laminar and inter-laminar failure modes.

The focus on improving regions which permit large scale damage propagation is key to implementing the future enhancement capabilities of composite laminates. For developing the understanding of inter-laminar crack propagation, it is first essential to differentiate the failure-inducing modes as illustrated in Fig.8. The first is an opening mode or inter-laminar failure mode, the second being the in-plane shearing mode and the last being the out of plane shearing mode or mixed bending and shearing mode.

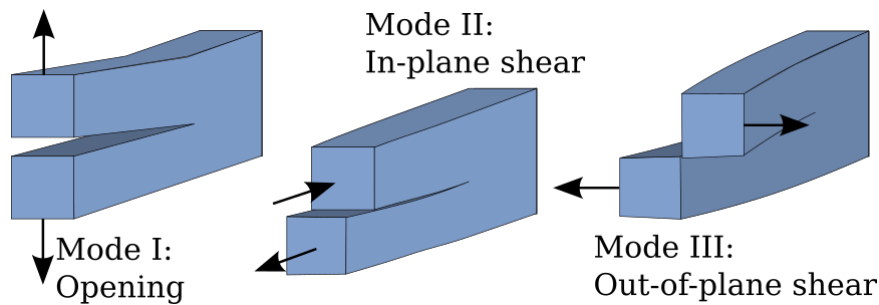


Fig.8 Crack loading failure modes: “Mode I, II and III” – Inkscape vector graphics adaptation [92,93].

For mode I opening failure consideration, the fracture toughness or critical strain energy release rate ($SERR_{crit}$) is determined by standardized experimental methods. The pure mode I fracture toughness of the laminate - G_{IC} is established using the ‘DCB’ test configuration, as presented within Fig.9 [53].

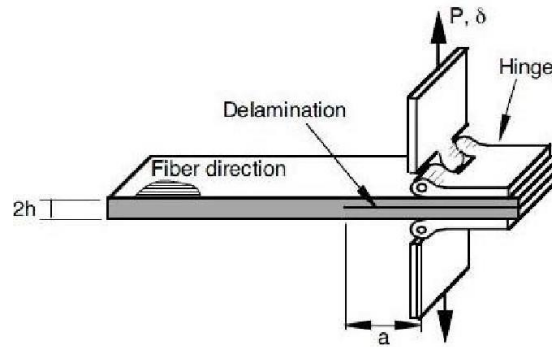


Fig.9 DCB Mode I Fracture toughness test (ASTM D5528) [53].

For mode II shearing failure, the fracture toughness - G_{IIC} tests are conducted under 3/4-point ENF tests (Fig.10 [54]). The load is introduced as an out-of-plane flexural load to develop the crack from an initial delamination or artificial PTFE film pre-crack. The resulting crack extends due to shear strains near the neutral axis and high shear stresses at the fracture tip. The ENF-tests are generally unstable owing from the unstable nature of the shear stress distribution and crack potential energy under bending deformation. The ENF tests can therefore only yield the G_{IIC} initiation values, unlike the mode I test, which provides resistance profiles (R-curves).

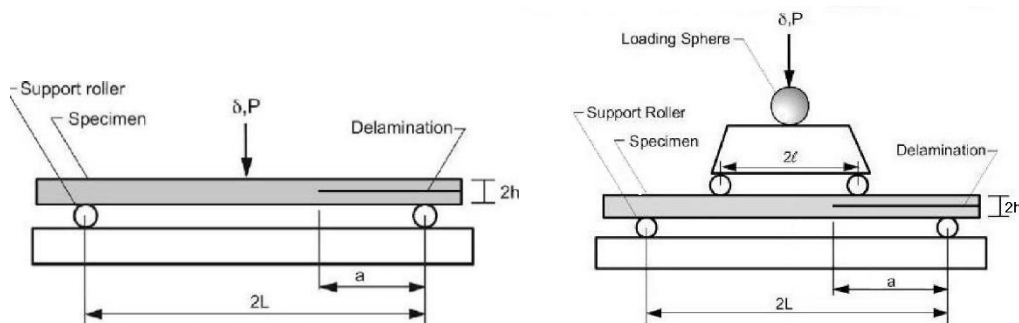


Fig.10 Mode II Fracture toughness 3ENF and 4ENF test (ASTM D7905) [54].

Similarly, under mixed-mode I/II fracture toughness - G_{IC}/G_{IIC} , the test configuration is carried out using a Mixed-mode bending (MMB) test as presented in Fig.11 [94]. Furthermore, the edge-cracked torsion (ECT) test is frequently applied when pure mode-III fracture toughness - G_{MC} , is required for laminate characterisation [95].

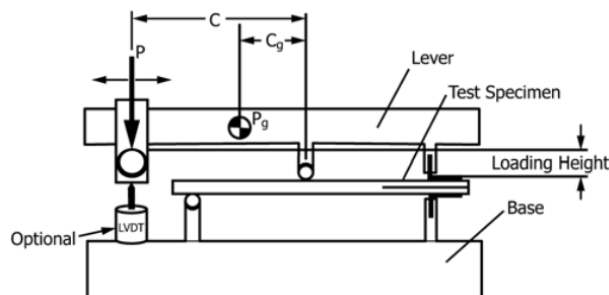


Fig.11 Mode III Fracture toughness using MMB test (ASTM D6671) [94].

2.6.3.1 Inter-laminar fracture propagation

The delamination in the form of inter-laminar fractures is one of the fundamental problems for composite laminates. Their structural and inter-laminar resistance is considered by the failures under transverse tensile and inter-ply shear-stress. The increase of inter-laminar stress occurs when the existence of abrupt changes in geometry and material exist, including; thick inter-laminar interfaces, mismatched fibre orientations; excessive transverse tensile or in-plane shear loading. Delamination develops as an inter-laminar failure when a stress or strain concentration of a laminate is put under continued loading with a transverse or in-plane shear loading from compression. Further delamination induced failure can subsequently emerge, under combined compression and bending stresses resulting from the fractured ply regions. Improving the interlaminar failure is one of the core objectives of this project, in particular, the understanding of potential toughening methods and their mechanisms. The inter-ply interface is discussed further in section 2.10.3, where the brittle interfaces can be enhanced using thermoplastic interlayer toughening techniques.

2.6.3.2 Intra-laminar fracture propagation

The process of intra-laminar fracture propagation is similar to inter-laminar failure; however, transpire at the fibre-matrix interface and not between lamina. Intra-laminar fractures often require more energy than inter-laminar depending on the fibre orientations of the interface. The increase in energy can be explained from the increased fibre-matrix interactions and friction where the detached fibres contribute and absorb further energy if under shear loading [96]. The exact mechanisms of intra-laminar failure have been studied by many groups and are developed further in section 2.10.3. The toughness of the matrix is a deciding factor if intra-laminar failure will initiate, where partially attached fibres on each side of the fracture can suggest fibre-matrix interface failure. In-plane shearing, tension and matrix fractures are the primary reasons for propagation of fractures within the laminae. Internal failure of the laminae brings rise to residual stress features which can promote additional strength reductions and lead to the propagation.

2.6.3.3 Mixed interlaminar fracture toughness

The study of damage within laminated composites has had substantial advancements for the modes described above. The focus on pure mode I & II studies have also been considered where the material components are modified for the fracture analysis of UD laminates. The need, however for realistic mixed-mode I & II loading is becoming more apparent, owing from the real-life fracture of composites being predominantly mixed-mode. There are, nevertheless, some previous studies on multidirectional and woven laminate fractures under mixed-mode [30,97–102]. MMB tests are regarded as the standard approach in the evaluation of the toughness for wide-ranging directional configurations, as suggested in ASTM D6671 [94].

2.7 Delamination Threshold

The delamination of a composite laminate can arise under a variety or combination of three fracture modes presented previously in Fig.8. It has been established by many authors [39,45,50,51,72,88,103–109] that delamination initiation can be associated to the delamination threshold load (DTL) for many material combinations. It is also generally accepted that delamination should not initiate if the peak impact-force is below the DTL. Prediction methods for delamination initiation are reasonably accurate and essential for damage characterisation parameters of composite laminates under LVI. There remains, however the question on the validity of DTL presence for woven laminates which cannot propagate significant delaminations due to variations in laminae fibre orientations owing to yarn interlacing. Additionally, for toughened resin systems, a more gradual damage growth is expected to be observed at certain impact energy levels. The interfacial fibre orientation combined with the presence of thermoplastics is the most interesting aspect of the research.

2.7.1 Delamination initiation prediction by experimental methods

Predicting the occurrence of delamination and the general response of composite laminates during impact is compulsory to designing structures which are resistive to such damage. Initially, the impact energy threshold was trialled to predict the delamination trigger energy. However, as suggested in [88,110,111], a universal energy threshold cannot be detected; hence the emphasis on the impact-force profile can provide an improved understanding. Numerous research efforts have been completed to establish the damage threshold, leading researchers to notice an abrupt load-reduction within the impact-force profile (Fig.12(a)).

The first significant load-reduction within the impact-force history represents a significant loss of laminate stiffness owing to the unstable and abrupt damage propagations [88,112]. Schoeppner *et al.* [88] exposed three distinct segments of impact damage from the impact-force profile. The first segment is described as the linear structural response and is mostly quasi-static, where no noticeable damage is formed. The second segment can have an oscillatory impact-force profile where the curve does not represent a DTL event, though a slight deviation in the slope can exist. This deviation in slope before the DTL event can suggest some energy absorption has taken place even when no damage can be observed by NDT scanning methods. The introduction of matrix fractures can occur in segment two, which does not affect the laminate stiffness greatly [110,113]. However, matrix fractures as described in sections 2.6.2 - 2.6.3 are initiation sites for the development of delamination, fibre buckling and fibre fractures.

The third segment occurs when the DLT event is discernible, and a distinct stiffness loss is apparent. The detection of delamination at this level can be established by conventional NDT scanning methods. Beyond segment three is significant further damage comprised of; delamination, matrix fractures and larger scale fracture propagations. Fig.12(b) shows a representative load-displacement profile of the impact event, the DTL is visible and a deviation in slope is also apparent. The enclosed area within the load-displacement is the dissipated and absorbed energy during the impact event. The differences in these charts and between materials and layup configurations can vary significantly. For example, as demonstrated in Chapter 4, the differences are apparent in the impact-force history for UD laminates with multiple impacts and thermoplastic veil interleaving.

When the DTL position can be delayed or eliminated, a successful toughening effect on the catastrophic behaviours of damage propagation is apparent.

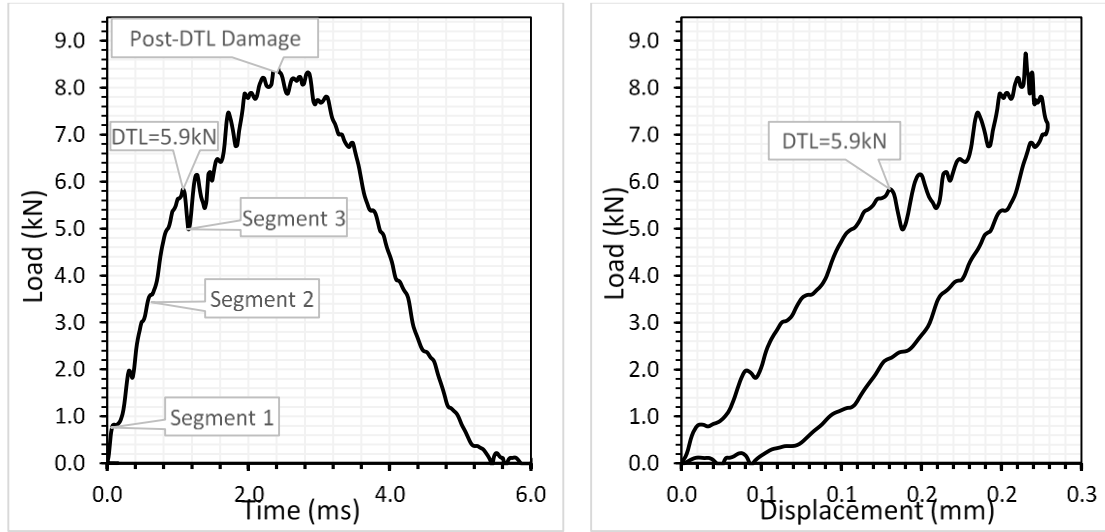


Fig.12 Identification of DTL for [0/90]_{7s} laminate impacted at 25J with 25mm impactor: (a) impact-force profile, (b) load-displacement history.

2.7.2 Delamination Initiation - Analytical method

To envisage the damage threshold, early models by Davies *et al.* [114] calculate the critical impact-force using the following relation:

$$P_{cr}^2 = \frac{8\pi^2 E t^3 G_{IIc}}{9(1-\nu^2)} \quad \text{Eqn.13}$$

Where the flexural modulus is denoted as 'E', laminate thickness as 't', the critical strain energy release rate under mode-II fracture is G_{IIc} (as per section 2.6.3 and 3.4), and the Poisson's ratio is ν .

The Davies model was determined from mode II fractures by mid-plane shearing with a simply supported plate boundary condition. Using this model, the laminate is considered as isotropic where delamination is assumed amalgamated through the thickness and located at the mid-plane, with no considerations on multi-plane fractures. It has been demonstrated through experimental work that the above method, despite being simplistic, can predict the DTL. However, the method may not be precise nor produce any valuable suggestion for the DTL for toughened or thick laminates or those with complex interfacial ply configurations.

The variation of DTL predictions against laminate thickness have been explored previously [88,115], where its dependency is affected by $t^{3/2}$. The investigation was conducted mainly on carbon fibre reinforced epoxy laminates with various thicknesses at various impact energies.

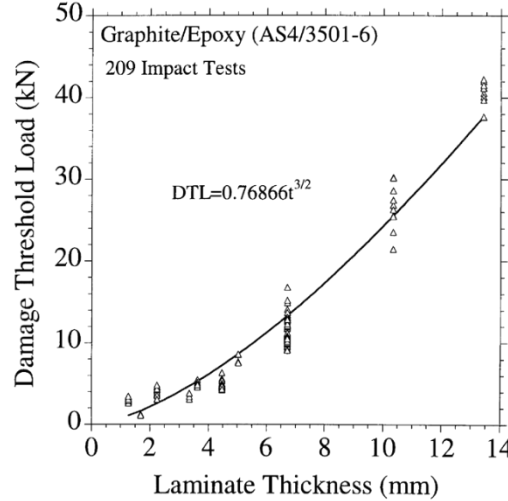


Fig.13 DTL dependence on laminate thickness [88].

2.7.3 Repeated impact after Delamination Initiation

The impact-force profile under secondary impact at the same location demonstrates interesting findings to understand progressive damage mechanisms, as demonstrated by Xu *et al.* [45,116,117]. It has been reported that if a primary impact of the laminate has reached the DTL, no further damage is observed if the laminate is impacted again under the same or lower falling-mass energy. The fading of the repeat load-drop would suggest that no further delamination was introduced, with a smooth secondary impact-force profile. Additionally, a greater maximum impact-load was also observed during the second impact, despite being impacted at the same energy. The perceived improvement to the toughness can be due to the elastic behaviour at failed locations and the matrix crack voids.

The variation of impact energy levels and laminates of differing thickness have produced results as those shown in Fig.14 and explored further in Chapter 4 with similar results. The suspected residual stress, larger contact area from the initial indent, the presence of matrix cracks and delamination from the first crack may be potential explanations for this behaviour. The delaminated region and their sub-laminates could be a reason for an increased contact force, in addition to the smoothing of the repeat impact curve being linked to the deformation of thinner

sub-laminates. Occurrence or the likelihood of same location repeated impacts may not occur in real life. However, the damage evolution and toughening methods can be better understood and is a topic of further exploration in Chapter 4.

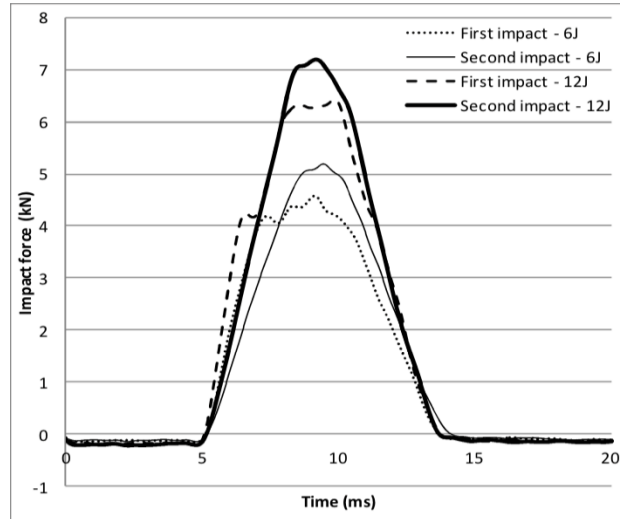


Fig.14 Effect of secondary impact on DTL [117].

2.7.4 Numerical Simulation of Delamination

Extensive efforts are evident for predicting the damage initiation and growth behaviour, especially the modelling of delamination for composite laminates. The development of mesoscale models into tiebreak and cohesive zone modelling have occurred over the past 30 years. Several advanced models and element formulations are described with their advantages and applications [118]. Tay [119] reviewed the advancement of simulating delamination and also found tie-break contacts and cohesive zone element modes being the most common to simulate the delamination in laminates.

Simulation of delamination initiation and propagation was also demonstrated by monitoring the transfer of force and displacement between coincident nodes by using tie-break nodes [120]. For tie-break methods the delamination path is confined to the pre-defined adhesive interface; where interface nodes initially tie the material on each ply surface as coincident nodes. However, tie-break contact methods cannot simulate matrix fractures and are susceptible to stress estimation errors during the simulation. Additionally, the friction behaviour of fibres partially attached cannot be accounted as a node is either connected or disconnected with no partial bond friction capabilities. The tie-break model allows a quick simulation solution to understand the extent of damage as illustrated in Fig.15.

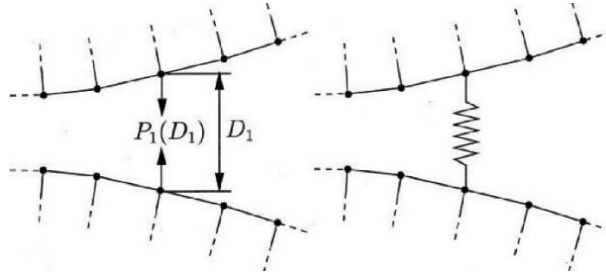


Fig.15 Illustration of the tied penalty formulation [120].

The cohesive-zone-model (CZM) [121] was initially introduced for modelling crack formation and growth between aggregates in concrete. Unlike the tie-break methods, CZM treats the damage formation and propagation as a gradual separation at the cohesive boundary with the inclusion of interface adhesion forces. CZM can favourably identify the adhesion forces during the fracture of the material components when applied between layers of varying fibre orientations to gradually simulate the interface delamination failure [70,78,102,122–126]. The CZM can similarly be applied within the laminae to consider in-plane matrix fractures using traction-separation laws and a phase-field approach [127,128].

A study of the literature shows an assortment of traction-separation laws being available for the CZM [129–131]; these can be classified into bi-linear [132,133], polynomial [129], trapezoidal [134,135], parabolic [136], and exponential methods [126,137].

The various adhesive traction laws for the CZM utilize the displacement or relative separation between the cohesive nodes [129]. The area beneath the load-displacement profile is determined as the delamination energy required to separate the cohesive zone consisting of plastic deformation and brittle fractures. Upon reaching the maximum forces and further introducing damage, a linear softening takes hold resulting from the progressive separation of the cohesive zone. For each cohesive element within the zone, the adhesive traction force converges to zero towards the end of the softening stage and is considered as entire separation with no interaction or complete delamination. Similar to the virtual crack closure approach as below, the unloading of the zone follows a reduced stiffness linear relationship after the delamination.

The virtual crack closure technique (VCCT) is a multi-step process initially proposed by Rybicki and Kanninen [138] and implemented to identify the fracture energy release rates within a simulation model. VCCT methodology adopts the reasoning that the sum of energy dissipated during crack growth is equal to the energy required to close the crack using an energy conservation approach [139,140].

2.8 Structural methods to inhibit fractures

2.8.1 Fibre orientation manipulation

The in-plane manipulation of fibre angles has shown some positive results in resisting delamination growth and thus influencing fracture propagation under mode I [141,142], mode II [143] and impact [51,71,75,87–89,91]. Individually placed variations in fibre orientation were demonstrated to alter the crack propagation behaviour; hence is considered a method worthy of closer investigation. Variable fibre angles can alleviate the necessity to terminate plies; however, they can reduce the stiffness of a component and may suffer fibre buckling related concerns. An alternative to randomly oriented fibres and maintaining the overall component stiffness and strength is the use of thin fibrous webs or chopped fibres.

2.8.1.1 Discontinuous fibres

The effect of fibre orientation under mode I loading by Shetty *et al.* [144] found that $0^\circ/0^\circ$ interfaces had the least delamination resistance whereas $90^\circ/90^\circ$ interface resistance had the greatest. ASTM D5528 [53] suggests the use of fibres oriented in the crack direction, hence the use of off-axis plies is an adaptation of the standard. The work did not specify the exact details of the through-thickness crack location within the laminate, therefore is worth further investigation for the exact damage mechanisms. Robinson *et al.* [145] also reported that fracture propagation along a non- $0^\circ/0^\circ$ interfaces would migrate from the defined interface of the DCB specimen, thereby increasing the effective fracture toughness. It was also demonstrated that if the off-axis deviation is reduced, the increase in toughness is also reduced. Additionally, it was established that the G_{IC} of a $+45^\circ/-45^\circ$ is greater than a $0^\circ/0^\circ$ interface but not significantly. Once again, other than fibre bridging descriptions, the accurate and

detailed fibre-matrix level interactions have yet to be explored. The fundamental reasoning for off-axis fibres affecting the toughness and crack deviation is due to increased energy dissipation resulting from the pull-out of individual fibres.

Interlaminar fracture investigations with the use of abrupt fibre orientation changes were conducted by Kim *et al.* [146]. It was also recorded that ply mismatch angles strongly affected the crack front and promoted deviation of the crack path. Introducing abrupt changes of the fibre orientation near the approaching crack path has been demonstrated in the past 30 years with the use of buffer or softening strips (section 2.10.2). However, the application of buffer or softening strips is considered as a last resort for interlaminar regions, owing to the likelihood of often triggering undetectable localised stress raisers.

Investigations using 45° oriented buffer strips within a cross-ply laminate showed in-plane crack hindering features at the strip locations [147]. Similar buffer strip tests by Sendeckyj [148] revealed in-plane crack inhibiting features with the orientated buffer strips. Both investigations, unfortunately, showed no further exploration on the exact microscopic damage mechanisms and the strain rate boundaries where the strips would be most effective.

It is apparent from the literature that fibre orientation ahead of delamination crack front could indeed resist the fracture growth. The introduction of these discontinuous laminae also brings about ply terminations, resulting in reduced loading capability hence deterioration of in-plane stiffness and strength properties of the composite laminate.

Another alternative to using discontinuous plies for variable fibre orientations with terminations could be the use of woven fabrics which can have a variety of interface orientations. The prospect of inhibiting fracture growth in 2D woven laminates is believed to have different failure mechanisms to that of UD plies based on the interlacing and undulating fibre architecture. For example, weave architectures using satin weaves have an inherent orientation bias on the interface which could be manipulated to control the growth of fractures (Chapter 3).

2.8.1.2 Continuous fibres

To support the variety of loading conditions, unlike isotropic materials, composites require a layup made of directionally stacked lamina. Composite laminates are often tailored to suit the loading, stiffness and buckling requirements. Preliminary and current laminates utilised composite plates by modifying fibre orientation for each layer [149,150]. The UD tapes of varying fibres were combined to enhance the stiffness contour providing the best buckling and compression resistance in comparison to a previous UD stacking sequences.

Finite element analysis methods are a well-established means to determine the fibre placement in the optimal loading paths. Aircraft structures, in particular, have a necessity for lower mass and pressure stable structures. The directional tailoring of laminae allows the reduction of the pressure pillowing phenomena [151] alongside having bespoke stiffness and strength requirements in critical locations. The use of 'variable stiffness composite panels' are made possible by modifying the fibre orientations for each ply where significant analysis methods have been developed [151–159]. Buckling failure improvements up to 53% were recorded in comparison to cross-ply panels resulting from extensive experimental and analytical work to develop the variable stiffness composite panels [160]. However, altering the layup fibre orientation varies the stiffness and hence their modal response compared to the baseline cross-ply configuration. The mode shapes were examined using Moiré shadow interferometry in the early development of laminates [160]; however, calculations can now be applied within finite element modelling.

The thoughtful placement of fibres within a composite can expectedly result in increased strength; Kelly *et al.* [161–164] observed bolted joint strength increase by up to 170% by steering fibres aligned to load paths around bolt holes. However, no further study has looked at the use of tow steering with considerations on delamination and fracture propagation. From the information that the IFT of interfaces with off-axis plies is greater than $0^\circ/0^\circ$, it may be possible to introduce fibre steering to modify the delamination towards prearranged areas. The approach could use mismatched plies in areas where damage suppression is required where inspection is not possible, and use orientations as necessary for areas where delamination is less critical. The downside of these arrangements is unfavourable alterations to other mechanical properties of the laminate performance that will need to be reconsidered.

2.8.2 Shaped fibres

Generally, fibrous reinforcement materials are considered as having a rounded cross-section. The existence of alternative fibre cross-sections is possible, which can affect the mechanical adhesion and pull-out strength governed by the fibre-matrix interphase. Usage of triangle-shaped fibres (Fig.16a) manufactured using drawn fibre extrusion techniques have demonstrated improved buckling resistance [165]. The approach may be a result of increased compaction and a higher fibre volume ratio which can ultimately reduce the potential for micro-buckling and therefore increase the overall compression resistance.

Tests on 6-lobed shaped fibre (Fig.16b) may increase the IFT compared to conventional circular section fibres [166]. This could be a result of the tortuous intra-laminar fracture path owing to the decrease of interface stress near the shaped fibre resulting from increased fibre surface area.

Alternatively shaped fibre reinforcements considered include; 2-lobed peanut-shaped glass fibres [167–169] (Fig.16c), ribbon-shaped carbon fibres [170] and C-shaped carbon fibres [170–172], (Fig.16d). Although these innovative fibre reinforcement shapes have been conceptualised, their study focusing on interlaminar fracture resistance is not completed, but micromechanical models have been developed [167]. The likelihood of scarce activity on these topics is undoubtedly due to the manufacturing difficulty and costs. It may indeed be possible to mass-produce glass and thermoplastic fibres in these novel shapes, but high strength fibres like aramid and carbon may prove difficult.

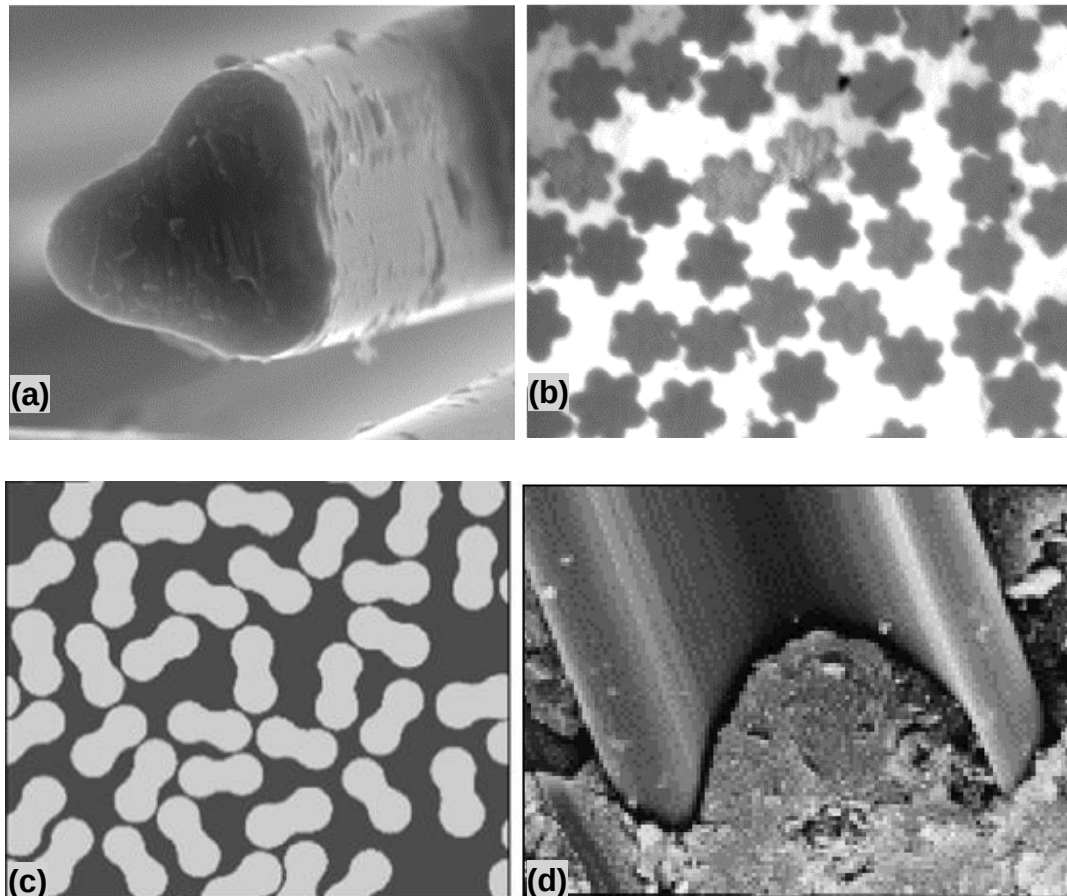


Fig.16 (a) Triangular shaped glass fibres [173], (b) Six-lobed shaped glass fibres [166], (c) 2-lobed peanut-shaped glass fibre [167], (d) C-Shaped carbon fibre in epoxy resin [171].

The local modification of the strain using shaped fibres by increasing interface surface area is a potential solution worthy of further investigation. The shaped fibre methodology can introduce residual-stress zones within a laminated component in predefined locations; however, the effects on fracture propagation is currently unknown. From the fracture mechanics standpoint, the shaped fibres promote a torturous fracture path of the matrix, where potential G_{IC} and G_{IIC} could be increased significantly due to a change in fracture mechanisms and pathways. Lobed-fibres, on the other hand, may introduce voids or trapped regions between lobes where matrix resins cannot bond with the fibre; these are, however manufacturing and surface sizing constraints to consider in the future. If the fibre-matrix wettability could be addressed, especially for 6-lobed and higher lobe count fibres, the potential for shaped fibres can offer the ability to inhibit the detrimental fracture propagation often associated with impact damage.

2.8.3 Three dimensional reinforced composites

Three-dimensional (3D) fibre reinforced composites are manufactured using textile manufacturing methods, namely; stitching/tufting, braiding, weaving alongside through-thickness knitting [34,56,174,175]. The advantages of 3D textile composites are that they can be manufactured into complex pre-forms and subsequently infused with resins within a mould to provide far superior through-thickness qualities. 3D textile reinforced composites exhibit added superiority and their widespread uses are apparent in contrast with traditional 2D laminates. 3D textile reinforced laminates have seen little implementation into the aerospace industry due to slow development and limited research efforts. The factors inhibiting the widespread deployment of 3D textile reinforced composites are likely due to scientific understanding, technical capabilities of large-scale manufacturing and financial constraints.

To understand the enormous task of reviewing 3D textile fabrics, it is important first to understand the various weave types and manufacturing techniques. Knitted textile fabrics as those shown in Fig.17, consist of individual interlocked looped unit cells. As a result, 3D fibre architectures allow for an entire range of advantages, including exceptional shape forming capability without crimping or folds [175,176].

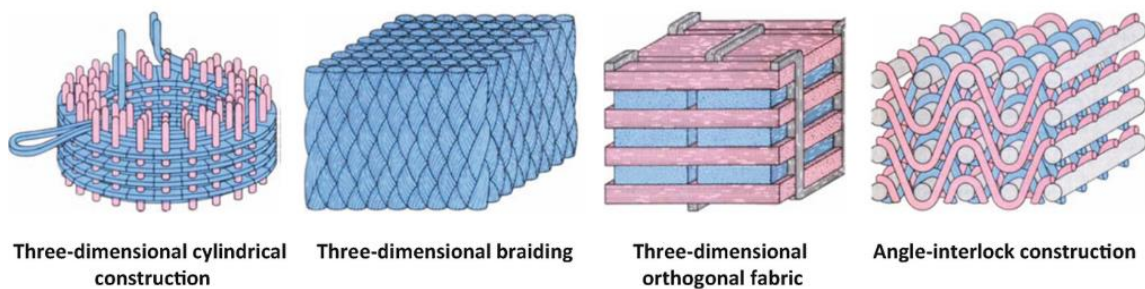


Fig.17 Typical 3D preform weave types [177].

The limited characterisation of specific textile preforms have shown a considerable decrease in tensile strength and fatigue life using 3D stitched glass yarns with vinyl ester resin [178]. These reductions are thought to be attributed to the localised fibre alignment and damage caused by the stitching and weaving process. These regions can spawn resin-rich pockets which can reduce strength and stiffness despite appearing tougher due to the 3D fibre orientations. Preliminary studies on a form of 3D textile using stitched yarns to enhance the

through-thickness behaviour have shown crack redirection and arresting capabilities [110-113]. The substantial effect of 3D stitching on the mode I IFT was studied by Dransfield *et al.* [179]. It was demonstrated that an increase of 1500% was possible in G_{IC} for CFRP fabrics using carbon stitching yarns, 2800% with Kevlar yarns and 800% using glass fibre yarns. The prospect of stitching laminated structures at first appears as an ideal solution to the inter-laminar and intra-laminar failure traits commonly observed in UD and 2D woven laminates. However, their use throughout the structure would not be practical nor beneficial, but when stitched locally within a laminate could allow for crack inhibiting regions.

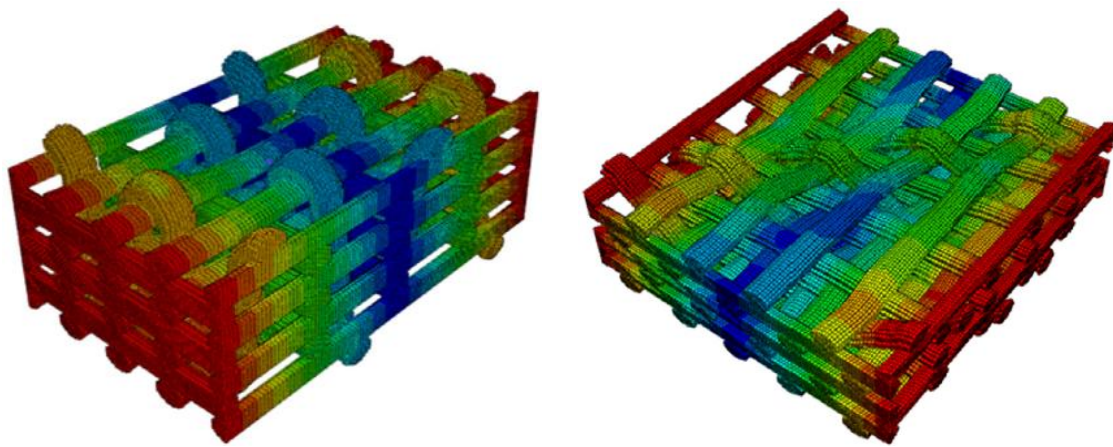


Fig.18 (a) 3D Orthogonal Interlock, (b) 3D Through thickness angled Interlock [180].

An alternative method for conceiving crack inhibiting features similar to stitching is the 3D braiding process which involves yarn interlocking. The textile preforms developed using the braiding process (Fig.18) are correspondingly infused with resin to create a 3D textile reinforced composite. It has been demonstrated that textile reinforced composites using 3D braiding also have a significant impact and high compressive strength [56,181–185]. The braided textiles also performed exceptionally with increased interlaminar toughness owing to the through-thickness fibres [176,186,187]. Considering that 3D braiding is an expensive and lengthy process; their applicability within aerospace structures is not wholeheartedly feasible plainly for improving their fracture inhibiting properties. However, their use in smaller reinforced components can demonstrate significant strength and toughness benefits compared to for example short fibre reinforced injection moulded parts [176].

From the 3D weaving methods in Fig.17, the simple through-thickness stitching is likely the most feasible method for inhibiting delaminations in specific locations where necessary on the laminated structure. However, another apparent prospect is the use of 3D woven fabric composites with interleaved toughening materials for improving through-thickness characteristics.

As a result, similar impact studies by Abounaim *et al.* [58] demonstrated the improvements in 3D textile composite laminates by flat knitting PP fibres (Fig.19). The technique demonstrated spacing opportunities without filler cores, improved stiffness and provided greater impact resistance. Laminate composites using intermingling is however different, where thermoplastic fibres can be blended within the fabric and the stitched yarns. This study introduced elastic thermoplastic PP fibres which also demonstrated benefits. The prospect of intermingling is reviewed within section 2.10.3.

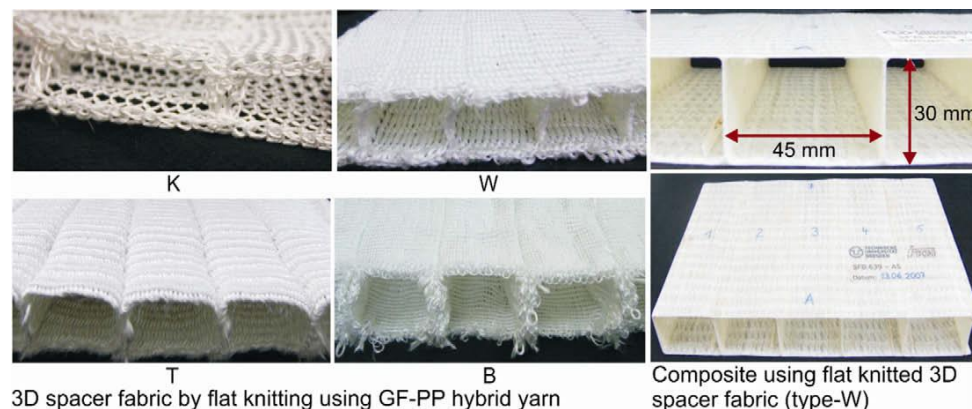


Fig.19 3D spacer fabric utilising thermoplastic yarns [58].

The inter-laminar and intra-laminar propagations commonly observed in 2D laminates have been inhibited by a significant factor owing to the through-thickness 3D interlocking. The 3D braided fabric reinforcements also have higher impact damage resistance and suspected reductions in notch / hole sensitivity than 2D laminates. Their wide-spread application is, however, restricted by manufacturing difficulties involving expensive tooling, time constraints, current technical development and economic viability.

2.8.4 Z-Pinning

An alternative method to inhibiting planar fractures against 3D weaving techniques is through Z-pinning; a similar approach to through-thickness stitching; however, without large needle pierced regions. The z-pinning process involves fine diameter reinforcement fibres to be pinned through the laminate thickness (prior to curing) often using an ultrasonic impactor. The pinned laminate can have improved through-thickness enhancements in mode I & II with z-pin pull-out being the primary failure mechanism, as shown in Fig.20.

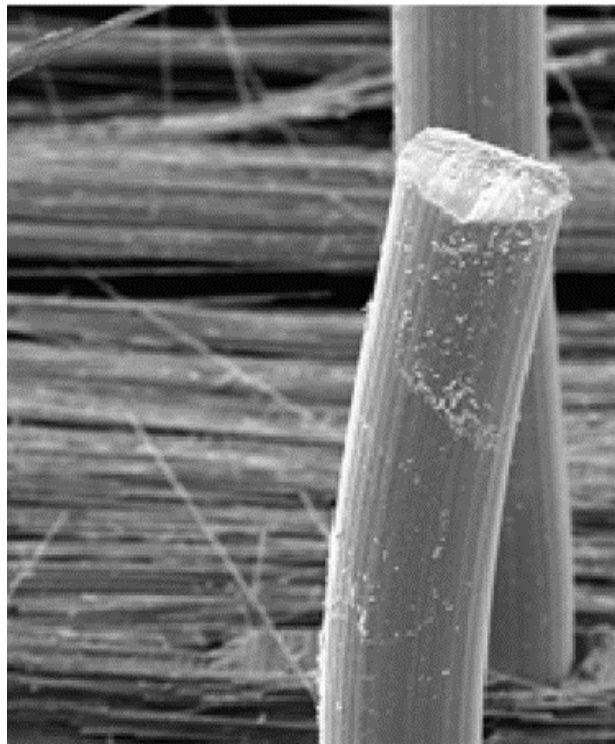


Fig.20 Pull out failure of Z-pin [29].

A detailed review of z-pins used for the purpose of composite laminates was completed by Mouritz [32,188]. It is worth noting, as with all through-thickness toughening techniques, Z-pins have shown various enhancements and drawbacks:

The benefits of using Z-pins include; enhancements in IFT under modes I, II and mixed (II/III) [27–29,189–191], can potentially reduce LVI damage, increase compression after impact strength [28,192] and have crack arresting potential in bespoke locations [188].

The drawbacks of Z-pins include; triggering of microcracks, fibre waviness, inconsistent pinning angles and fibre breakage of the parent laminate [188]. Z-pins have also been shown to have resin-rich regions at pinned locations, can reduce the in-plane compression and tensile strength [188,193]. Z-Pins can impair the in-plane fatigue life due to local stress concentrations and micro-buckled fibres [188].

The enhancements by using Z-pinning have demonstrated improvements as presented above. However, interleaving with combinations of Z-pinning have never been previously investigated. The combination could deliver interesting outcomes; the idea being that interleaving could reduce the crack propagation distance while allowing Z-pins to improve through-thickness strength. The concept of Z-pinning in combination with toughened interleaved laminates are thought-provoking but beyond the scope of this work.

2.9 Crack arrest mechanisms

The topic of interlaminar crack arrest methods can be separated into two major groups with many candidate solutions for the damage manipulation and suppression within composite laminates. These groups include material and structural approaches. Example solutions can include any of the following:

- Modification of local elastic properties with thermoplastic additives.
- Interleaving with tougher inter-ply layers.
- Intermingling / Commingling of fibres with tougher fibres.
- Manipulation of interfacial fibre orientations and shapes.
- Manipulation of through-thickness fibre interlocking / braiding / through-thickness stitching.

Each crack arrest technique introduced above will likely have variable performance characteristics where one cannot simply use the compatibility as a suitable definitive for toughening. The global suitability of the technique, the effect on overall stiffness and strength, manufacturing effort, designer flexibility, and overall comparison against the virgin laminate are of significance. The materialistic approach is considered in section 2.10, whereas a structural approach is considered in section 2.8. Section 2.12 summarises the most suitable approach to crack arrest while considering the practicalities for immediate implication into current laminates and subsequently, the practical findings in Chapters 3-4.

Hindering or arresting a propagating crack in laminated composites is achieved by introducing locations which are better capable of absorbing energy. Unlike many other topics relating to composites, there is relatively limited research available on the exact toughening mechanisms. As introduced in section 2.6, the formation and propagation of damage for composite can be divided into three types; inter-laminar, intra-laminar and trans-laminar cracks. The focus of this project will be on the inter-laminar and intra-laminar damage as they are often difficult to detect, hence the most detrimental to the safety and reliability of a structure.

Composite laminates can accomplish crack arrest by introducing ductile and elastic regions within the laminate which are able to absorb energy. For composites, these regions require a higher strain energy release rate (SERR), G , to restrict the propagation of the fractures and overall increase the critical SERR, G_c . This term may be called the effective or apparent crack propagation resistance, $G_{C_Effective}$. The methods and techniques subsequently discussed have previously demonstrated an increase in the $G_{C_Effective}$ of laminates. Thus, their understanding and potential use as crack arrest techniques will be explored experimentally (Chapters 3-4). The methods explored and considered experimentally shall consequently demonstrate that interleaved materials are the most promising techniques to consider for the direct modification of conventional fibre reinforced thermoset matrix systems.

2.10 Crack arresting materials

Damage resulting from the low-velocity impact have been especially significant in the aerospace sector where flaws can be barely visible with detrimental internal damage. The capacity of the damaged laminated structure to sustain future load decreases as the damaged area increases [17,23,46,73]. The damage and failure mechanisms of composites are often described as a 'fracture mechanics' inquest as appose to a 'mechanics of materials' inquest as described by Dassios [194]. The damage size has a critical role in the progressive failure compared to the influence of the material strength. Dassios [194] explained the fibre pull-out mechanisms being useful to understand the progression of composite failure mechanisms.

During an impact event, energy is absorbed under distinct mechanisms occurring instantaneously and simultaneously within the laminate. The matrix micro-cracks and local fibre delamination interact with each other and subsequently resulting as impact triggered failure.

The damage tolerance of laminated composites has demonstrated improvement by altering the constituent reinforcing fibres or toughening the matrix [22]. Additionally, the in-plane compression strength after impact (CAI) is influence by the interlaminar fracture toughness (IFT), suggested by Cantwell *et al.* [195], hence enhancements to the IFT will lead to enhancements of the CAI strength.

2.10.1 Local (elastic) property modification

Modifying the local material properties by inducing residual stresses is a method of manipulating crack growth and its directions. Generally, local modifications are necessary for material systems with highly dissimilar material properties such as metal matrix or infused particle composites. Residual stresses during the curing cycle can arise from differences in thermal expansion coefficients, which have shown to alter propagation directions as residual stress concentrations could promote or inhibit crack movement [196]. Based upon this approach, deliberate altering of the local properties are permissible by introducing localised features, material additives, coatings and by toughening with foreign materials.

2.10.1.1 Resin / matrix variations

Introduction of resin pockets within the laminate is a method of altering the local properties within the composite. The idea of matrix pockets has been applied in areas of civil engineering, for example within aerated concrete; where aggregates are combined with fillers to reduce mass, create porosity and ultimately allow for these regions to limit the larger crack propagations. Appreciating this concept, resin pockets would have different elastic and thermal expansion coefficients than the bulk material encasing the fibres. Pocket regions can introduce localised residual stress and stress concentrations which influence the interlaminar fracture toughness. The effect albeit non-conventional could provide an interesting approach to adding lighter regions in laminates as with honeycomb or foam cores used in sandwich panels.

The effect of laminate thickness was established as influential on the impact-force and hence fracture toughness of a composite in section 2.7. Plastic deformation zones ahead of a crack tip for thin ply interfaces can allow interactions with the fibre boundaries, promoting de-bonding, kinking and buckling. Increased fibre traction can trigger crack interaction at the fibre-matrix interface leading to adhesion failure and subsequent propagating damage. On the other hand, if interface thickness is increased, the plastic yield zone forward of the crack tip would have less interaction with the fibres. The resulting fracture would remain inside the interface resulting in cohesive inter-laminar failure of the matrix interface. However, this presumption may also suggest that overall toughness could be reduced as fibre interactions would absorb higher energies under the fibre-matrix pull out explanations by Dassios [194]. The thickened regions would inherently increase the separation between laminae, allowing the laminate to have a stiffened behaviour due to increased thickness and trigger failure at a different through-thickness location.

Thickened interfaces have a similar approach to having resin pockets within the laminate, hence allowing the laminate to deform elastically. Thickened interfaces and modification of inter-ply regions can be achieved using either interleaved materials or commingled fibres. Another method of crafting void spaces is the use of nano-tubes; the dual purpose of these vascular systems mean they could participate in thermal management and self-healing networks [43,197–200]. Research groups have successfully demonstrated that propagating fractures can be hindered or arrested when intersecting with these vascular networks [200–202].

Other methods to introduce modifications to the inter-ply resin interface is by discontinuing fibres and ply orientations. The generated gaps and concentrations between fibres can help to divert the propagation or maintain localised damage. The presence of these regions is apparent in woven laminates, where the variation of interfacial fibre orientation and resin quantity at the intersections already exist. These resin variations can alter the interaction of the resulting stress concentration, and propagation can be kept localised, be diverted or prevented entirely.

2.10.1.2 Fibre ratio manipulation

Manipulation of the fibre volume fraction, V_f of a laminate can be accomplished by adapting the curing pressure, using viscous resins or by controlling the curing volume using moulds. The influence of V_f on the G_{IC} of a composite has been investigated by many authors [203–205]. It was established that increasing the V_f does not affect the G_c but increases the propagation resistance, G_p due to extra fibre bridging interactions. Davies *et al.* [204] also explored the influence of V_f on composites and concluded that lower V_f results in higher G_{IIC} . The results were attributed to resin content variables hence influencing the crack tip plasticity. The variations observed may not be universal under all circumstances as fibre-matrix interphase and localised properties have a crucial role in the laminate. Therefore, these observations require further investigation for; multi-directional fibre orientations, various material systems and through the use of toughening additives.

2.10.1.3 Parent matrix manipulation

The modification of interface regions is possible by simply using additional resin at the interface between laminae and increasing the separation of the interface. Singh [206] and Masters *et al.* [207] demonstrated significant improvement in the G_{IC} and G_{IIC} by increasing interface thicknesses where apparent increases up to 70% and 200% for each were achieved. Although resin thickness is difficult to control during the manufacturing, the plastic yield zone in front of the fracture front was described as an absorbing mechanism. It has been established that fracture through a plastic zone would exhibit necking of the resin remnants and the toughener, with the presence of resin cusps and a stick-slip behaviour during fracture [24,50,196,208–210]. The thickness of the plastic region determines the

through-thickness crack propagation response and the stick-slip process, as illustrated in Fig.21. A thicker interface can allow the crack to wander and allow plastic deformation before fracturing, not possible if the zone has minimum thickness with reduced plastic performance.

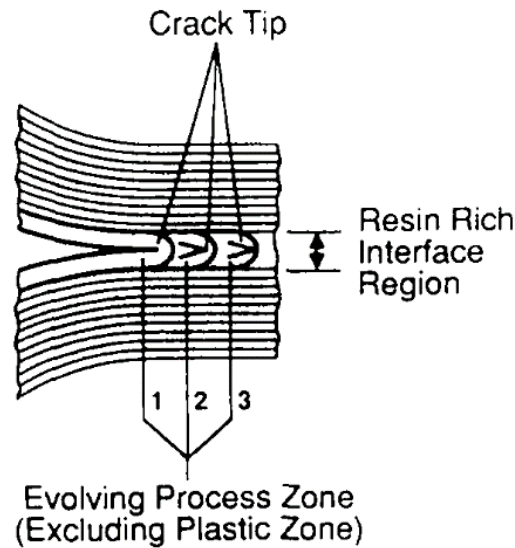


Fig.21 Illustration of yield zone and propagation interaction with thickened interface [196].

Other methods of toughening the composite laminate are by additions of rubber particles or thermoplastic polymers into the liquid epoxy. The method results in phase separation and dispersion during curing and results in a matrix consisting of epoxy and dispersed rubber regions. As expected, the dispersion, volume, particle size, and distribution govern the overall enhancement of resins. Additionally, nylon particles have also been used as toughening materials, where the particles have reportedly increased energy absorption [211–213]. The apparent fracture resistance in mode I was observed to have varied responses resulting from nylon particle inclusions [10,214].

Entirely thermoplastic resins have also been explored as tougher matrix materials owing to their excellent stability under heat and pressure (2.3). However, entirely thermoplastic matrix materials can have drawbacks, based on manufacturing, large part assembly and stiffness related difficulties. For matrix toughness, polyetheretherketone (PEEK) reinforced resins and additives have demonstrated respectable performance in interlaminar fracture enhancement [196,215,216]. Albeit their ease of implementation in the composites manufacturing process is in development since reviewed in 1989 where their use remains relatively limited [217].

2.10.1.4 Foreign particulates and fibres

In addition to the introduction of matrix modifications through additives, the use of short fibres, polymer fibres or even nano-fibres are effective methods of providing improved toughness. For example, non-woven carbon tissue, among other materials, have been demonstrated as suitable interleaves for the mode I & II fracture toughness enhancement as explored by Lee *et al.* [218–220]. The carbon tissue was found to have outstanding formability due to its low thickness and allowed unaligned short fibres to be distributed across the inter-layer, which increased the resistance. The increased mode I and mode II fracture toughness with changes to the growth locations have been demonstrated by Lee *et al.* The cohesive interleave failure led to the enhancement of G_{IC} by 28% and G_{IIIC} by 260% [218–220].

It is apparent from the detailed study of literature that particle bridging through their plasticity, crack migration, yield induced shear, bending and microcracking have been considered as toughening mechanisms [21,24,206,211,213,221–224]. The fracture surface morphology should be further investigated in all cases for the exact toughening mechanisms alongside alternative fibre orientations.

Current advances in nano-scale fibre developments have sparked several studies for their implementation as interleaved reinforcements. Nano-fibres made from Carbon, PolyEtherKetone, Nylon-66, Polysulfone and Silicon Carbide have generally demonstrated toughness enhancements in G_{IC} and G_{IIIC} , yet their detailed fracture morphology has not been consistent, suggesting their reinforcement capabilities can fluctuate.

Manufacturing methods of nano-fibre interleaves can vary which produce misaligned interleaved layers with various thickness distributions depending on the method. Developments in novel alignment techniques for orienting nano-fibres in the thickness direction and used as interleaves have been attempted [224–231]. The alignment of nano-fibres after curing of the laminate is yet to be confirmed and explored further. However, the ‘aligned’ nano-fibres as interleaves in the composite resulted in enhanced fracture toughness of 76% and 300% for mode I & II.[226].

2.10.1.5 Treatment of fibre surface, coatings and sizing

The surface preparation of fibres can significantly influence the adhesive strength to the matrix and thereby affect the performance of the composite [12,232]. The adhesion of fibres to the matrix interface can be disseminated into five topics; matting absorption and wettability, chemical reactivity, electrostatic attraction, chemical bonding and mechanical interlocking [232]. Comprehensive studies on the influence of fibre to matrix adhesion have been performed by many researchers [6,8,11,233–237]. Interestingly the reported enhancements of G_{IC} and G_{IIC} resulting from adhesion improvements [6] using 100-200nm of epoxy surface coating on the fibres can be achieved. It was also demonstrated that interlaminar shear strength (ISS) could improve up to 19% and 160% in mode I & II, respectively [234].

Polyvinyl alcohol (PVAL) treatment of carbon fibres is an oxidative chemical process performed by Hu *et al.* [238]. The study revealed a minimal benefit on the G_{IC} however presented extensive fibre bridging leading to an increased crack propagation resistance. A similar sizing study by Deng *et al.* [5] on epoxy coatings after an electrochemical oxidation process on carbon fibres had shown little benefit on the G_{IC} however the propagation resistance G_{IP} was increased for mode I by 30% [5]. Once again, identical studies using polyamide (PA) toughening using carbon fibres resulted in small changes in G_{IC} , however, a large increase was correspondingly observed in the crack propagation resistance, G_{IP} of 53% [239].

As the development of nano-technology moves forward, the creation of fuzzy fibres by nano-tube grafting techniques onto the fibre surfaces have revealed to be highly effective in improving single fibre pull out strengths by 60% [240]. The approach is similar to that used in rebar tendons for concrete reinforcements, where lobed features improve the surface traction forces of the reinforcement. However, for composite fibres involving costly processing and grafting complexity means there are merely a few inter-laminar and intra-laminar studies on these hierarchical fuzzy fibres [9,240–245]. The stretched breaking techniques applied to thermoplastic fibre processing in section 2.10.3 could be applied to reinforcement fibres. However, the practicality of modifying individual fibres cannot be currently justified for large-scale aircraft and spacecraft structures.

The improved adhesion behaviour using fibre surface treatments have demonstrated de-bonding strength improvements at the fibre level. Ultimately, surface treated fibres can improve the toughness as observed from the increased G_{IC} and G_{IIC} . Although the individual processing of fibres is an unlikely large-scale solution, strategically positioned treated fibres could manipulate the fracture propagation and direction, hence providing scope for crack inhibiting capabilities. Strategic placement of treated fibres has a similar modification approach to strategically placed foreign materials. These intentionally placed materials can be introduced in bespoke locations without significant detriment to the global performance of the component as with interleaved materials.

2.10.2 Buffer strips

The application of buffer materials for composites has been in existence since the 1970s, and was mainly applied in situations of high-stress concentrations, near cut-outs, holes, elevated temperature zones and other stress risers [246–255]. The concept makes use of strips of typically fibrous materials embedded within the laminate as interleaves or at ply boundaries, as shown in Fig.22. On the contrary, another approach uses softening strips, where material strips having a lower elastic modulus have shown to arrest cracks during tests [250].

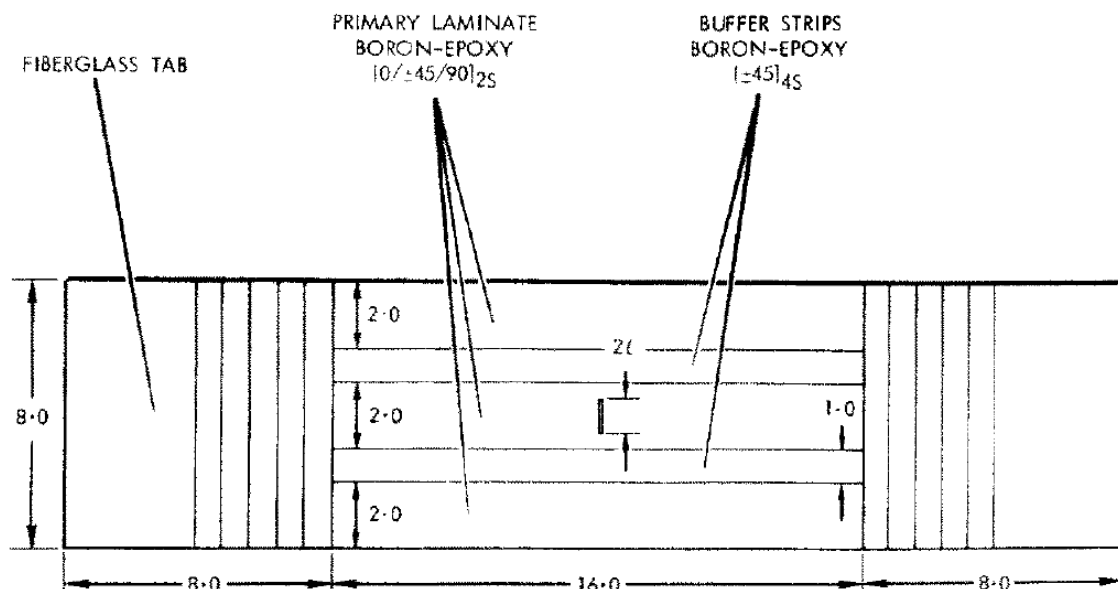


Fig.22 Panel with centre slit cut-out and two buffer strips [256].

Buffer strips and their application appear to aid crack arrest by reducing the local elastic properties or adding ductility around stress concentration regions such as holes or cut-outs. When buffer strips are implemented as interleaves, they can be formed into any shape and located at critical locations within the laminate. An interesting application would be strips or disks interleaved at locations of bonded inserts or regions where drilled holes and fasteners are placed.

2.10.3 Intermingled Hybrid Composites

2.10.3.1 Interphase of commingled composites

Surface tailoring and sizing is a method used by composite manufacturers with many approaches to execute them; where some are described previously (2.10.1.5). The processing aspects and tailored surfaces of commingled glass fibre reinforced polypropylene composites (GF/PP) have been investigated by E. Mäder *et al.* [13]. Customized X-shaped and Y-shaped thermoplastic fibres (Fig.23) have been developed to enhance the consolidation during laminate curing. The spinning process for X-shaped and Y-shaped thermoplastic constituents improved the cooling rates and offered control in the sizing of the interface.

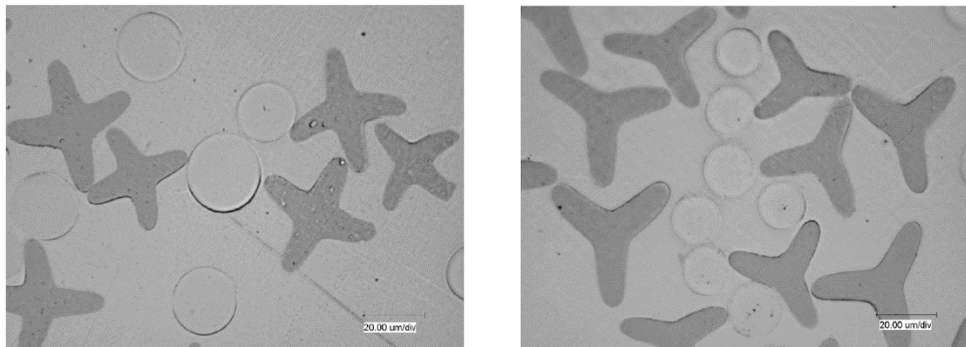


Fig.23 Shaped tailoring for improved interphase: (a) X-Shape and (b) Y-Shape [13].

2.10.3.2 Hybrid Composites

The use of thermoplastic matrix composites involves high temperature, pressure, consolidation difficulties in part thickness and consistency, as mentioned in section 2.8 with the manufacturing of large wind turbine blades [41]. An alternative to developing toughened composite materials using completely thermoplastic matrix materials is to use intermingling techniques (Fig.24-Fig.27)[16]. Hybrid composites in their fibrous form prove beneficial to future composite applications, with the manufacturing ease of thermoset and toughness capabilities of thermoplastics.

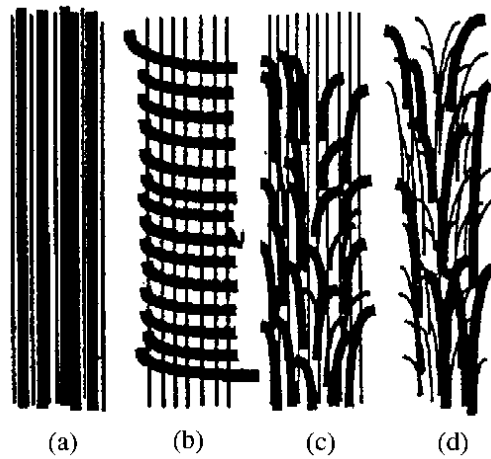


Fig.24 Types of Hybrid Yarn Intermingling [16]: (a) commingled, (b) co-wrapped, (c) core spun and (d) stretch-broken yarn.

The introduction of thermoplastic fibres into thermosetting composites has been contemplated for some time; it was first announced to be used in laminate composites by M. D. Campbell in 1981 [257]. Campbell's patent includes the idea of producing impact-resisting composites, with energy-absorbing elements helically wound around load carrying elements. However, the idea originated in 1980; R. F. Shannon [258] obtained a patent for commingling ceramic fibres for fire-resistant materials and laminate boards. It was not until 1994 when R. Sassa [15,259] described the use of polytetrafluoroethylene (PTFE) fibres for commingling with fibreglass using air-jet texturing (Fig.26). The commingled fibreglass from Sassa's patent promised improved resistance to flexing, abrasion and chemicals [15,259]. In 2003 Duncan *et al.* [260] envisioned the use of thermoplastics in continuous fibre form. Using the pulltrusion process, pellets of thermoplastic additives and glass filaments were consolidated into reinforced pulltrusions. Furthermore, Jo *et al.* [261] secured rights for the use of commingled continuous glass fibres with thermoplastic for in-line extrusion manufacturing. The significance of these notions advances the field towards hybrid composites to be ultimately implemented into yarn-level hybrid laminate panels.

There are various methods of manufacturing hybrid yarns; be it commingling, co-wrapping, core-spinning, or stretch breaking (Fig.25). Alagirusamy *et al.* [16] and Selver *et al.* [17] have studied the method of maintaining the level of mixing throughout the constituent materials by air-jet texturing during the manufacture of hybrid yarns. From the literature, it is apparent that commingling has the most potential for thoroughly blending the constituents, leading to flexible, soft and

drapable yarns [16]. The method used for commingling fibres uses rapidly moving jets of air to generate entanglement of filaments. Understanding the manufacturing method of commingling is essential, as the distribution of constituents is a variable known to be significant in the final product (Fig.27). The initial multi-filament yarn (Fig.25) can control the ratio of thermoplastic to load-carrying filaments. The distribution and homogeneity of fibre dispersion are influenced by the air-jet nozzle pressure, flow and orifice dimensions. Homogeneity of the intermingled yarns allows for complete impregnation of matrix material and consequently leads to homogeneous composite components.

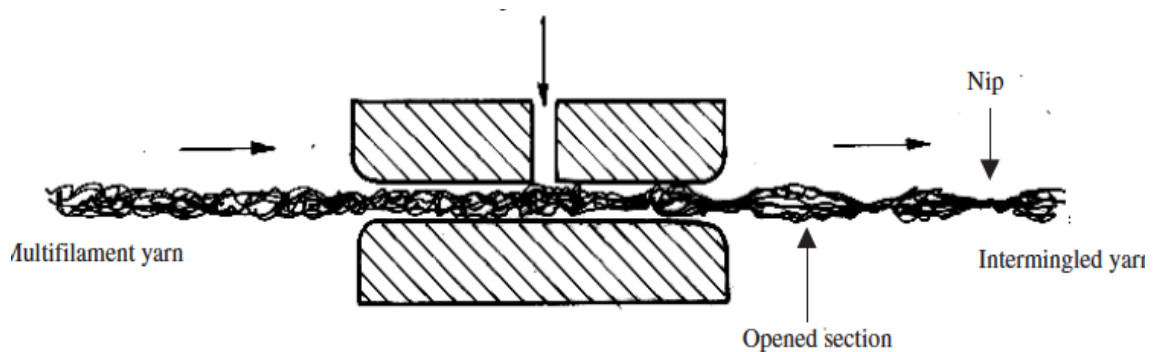


Fig.25 General intermingling process [16].

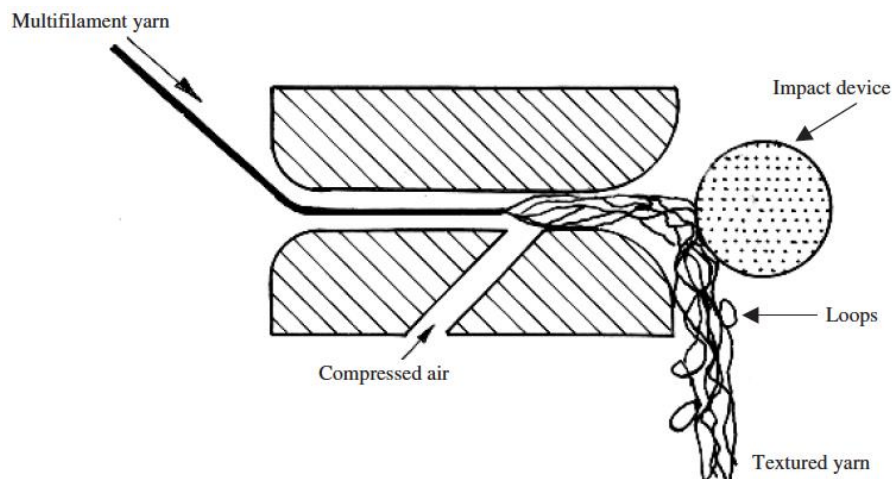


Fig.26 Air-jet texturing process [16].

Since its introduction, commingling of thermoplastic yarns in various forms moved to the commercialisation of pre-impregnated fabrics for unidirectional and woven composites. Companies such as Cytec, Sigmalex, Formax and SGL Group have shown interest in the commingling of thermoplastic filaments with glass, carbon or aramid fibres.

Paul J. Hogg [22] describes the principle of commingling thermoplastics within thermoset composites and documents improvements to the impact toughness. The investigation also demonstrates how commingling of thermoplastic fibres with structural fibres exhibit superior damage tolerance compared to non-toughened virgin laminates. Improved mode I & II interlaminar toughness can also be dramatically enhanced when commingling with Polyamide (PA/Nylon) fibres.

The study of energy absorption in composites is fundamental for impact scenarios; being the only mechanism to slow down fractures and penetration as described in section 2.5. Hogg [22] and Selver [17] demonstrate the improvement of energy absorption compared to regular glass fibre composites. Hogg also demonstrates that dissolved thermoplastic surrounding the load-carrying fibres carries no benefit to impact resilience; only in fibre format are observable benefits. Commercially available materials such as PRIFORM by Cytec (Fig.27) have been used in their investigation, consisting of glass fibres commingled with thermoplastic Polypropylene fibres (GF/PP) within a thermoset matrix. Additional commingled UD and woven fabrics include Twinex®, CURV®, PURE®, SIGRAFIL®, SigmaRF® and COMFIL® by various manufacturers.

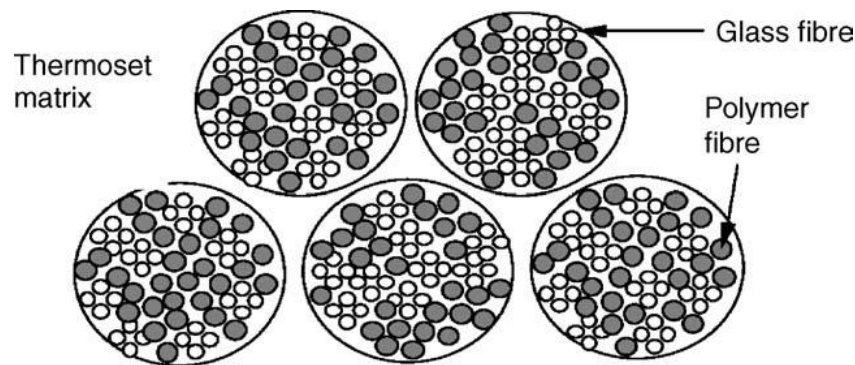


Fig.27 “PRIFORM” by Cytec – Hogg [22].

Moving further towards the importance of interphase in composites, a study by Zhandarov and Mäder [262] studies the characterization of fibre/matrix interfacial strength. Their work into micromechanical testing and fibre pull-out techniques is an excellent method to follow for the detailed analysis of commingled composites. Their work discusses various advantages and disadvantages of the stress and energy-based models for interfacial de-bonding and adhesion. An interesting parameter is the adhesion pressure; a stress-based criterion for interfacial failure. A two-stage model has been developed, which considers the adhesion pressure and de-bonding with an imperfect interface as a subsequent analysis step [262].

The work of Zhandarov *et al.* [262] is useful for this research; the principles remain the same for commingled composites; however, with additional consideration of discontinuous fibrous thermoplastics.

Recent advances again by Zhandarov and Mäder [263], make use of alternative methods using the maximum force as functions of fibre embedded lengths in their pull-out models. Their work studies nanotube carbon coatings; a coating and treatment method used to improve surface and interfacial adhesion, wettability and homogeneity. The alternative method considers non-linearity of the de-bonding progression and is deemed more accurate when estimating the interfacial toughness.

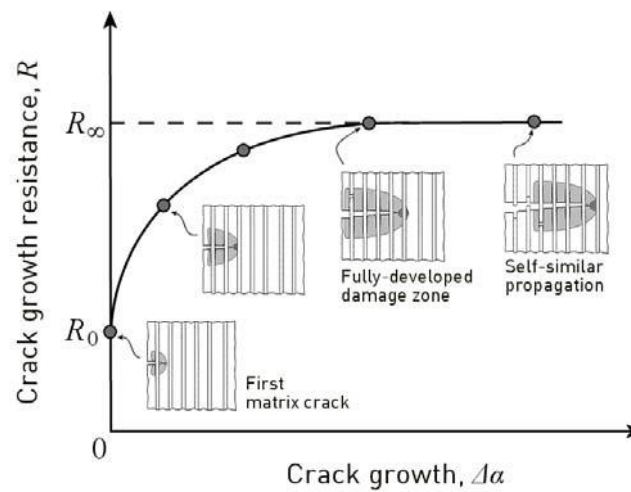


Fig.28 Typical crack growth resistance curve resultant of energy dissipation [194].

Failure mechanisms perpendicular to the fibre direction involve four main types of damage: matrix fractures, fibre-matrix de-bonding, frictional fibre pull-out and fibre fracture. Frictional fibre pull-out and fibre-matrix de-bonding are demonstrated as the most efficient energy dissipation mechanism [126]. Observing the progressive failure mechanism in Fig.28, Lin *et al.* [126] explains three stages towards failure (Fig.29): Stage I, a linear response corresponding to perfect fibre/matrix de-bonding. Stage II being an initiation of matrix fractures, leading to a non-linear force/displacement relation. Stage III is where the crack approaches critical length and instability evolves, leading to abrupt de-bonding, frictional sliding and fibre pull-out.

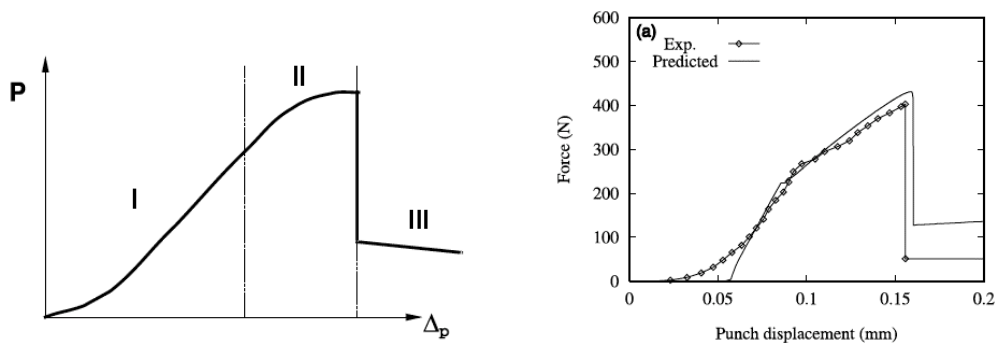


Fig.29 Load-displacement curves for single fibre push-out [126] (a) Typical and (b) Predicted/Experimental.

One can note from the above progressive failure knowledge (Fig.28 to Fig.29), that localised elasticity by commingled thermoplastics can be a solution to enhance the damage tolerance capabilities of composites. If the crack tip approaches a local region of higher elasticity, as within commingled thermoplastic fibres; the crack tip propagation would undoubtedly have a retardation effect. The relatively higher elastic fibre will induce de-bonding and fibre pull-out in the future pathway of the crack – the most effective mechanisms for energy dissipation [126].

Hybrid and commingled composites are increasing in attention; publications by Hasan *et al.* [14] shows a renewed interest in the topic and evidence of gaps in the knowledge. Investigations by Hasan *et al.* [14]; largely on the influence of processing parameters for commingled hybrid yarns have been conducted, using Carbon fibre and Polyamide filaments (Nylon - DuPont® [264]) (CF/PA). Once again, air-jet intermingling (Fig.26) is used to manufacture commingled yarns for hybrid composites with results showing improved surface wettability by heat treatment of the CF; also precautionary measures to avoid damage by air-jet pressure and promising improvements in overall mechanical strength.

2.10.3.3 Intermingled Natural fibre composites

Natural fibres are also an exciting topic for research which are increasing in popularity due to their improved ballistic and recyclability paybacks. Impact studies involving natural fibres (jute, hemp and flax) intermingled with dissolved thermoplastic have been carried out by Wambua *et al.* [265]. Fully dispersed commingled thermoplastic acting as the matrix in natural fibre composites have been shown to provide no benefit in GF/PP composites by Hogg [22]. However, a thermoset matrix and the presence of elastic regions can be preferential and beneficial. Such regions can also take the form of selective toughening by

thermoplastic sheets between plies, powder coatings, heat-treated nano-coatings and veil interleaving methods.

Coatings offer benefits to adhesion and wettability of the filaments to the matrix, however, cannot be a drastic improvement for the impact properties due to their low relative content. Laminates using films of thermoplastic often cause delamination during impact; the directions of elastic regions are spread throughout the lamina and can cause premature delamination. It is therefore apparent that intermingling, specifically commingling, is the better option; as regions are locally elastic and lead to retardation of crack propagation by added localised ductility. The scope of comingled thermoplastic fibres is one which is excessively costly for immediate implementation into current laminates, stemming from the processing of raw fibre tows. Manufacturing these comingled tows into a pre-impregnated laminae is the most beneficial; however, a wealth of further study must be conducted for future applicability.

2.10.4 Interleaved materials

The method of interleaving is a technique where generally thermoplastic materials are placed between laminae to improve the inter-laminar fracture toughness of the laminated composite. The process of implementation is simple, whereby interleaved layers can be placed within the laminate (Fig.30) at locations where required without adding significant costs and complexities as with other methods. The other benefit of interleaved materials is that they can be placed in particular locations where toughness is required. For example, interleaves can be positioned at cut-outs, insert locations, fastener regions, and impact prone surfaces.

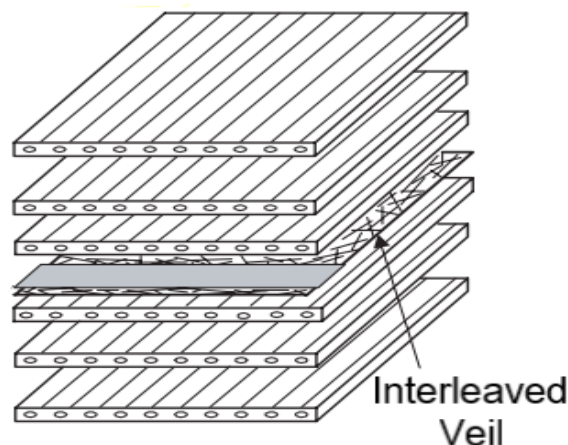


Fig.30 Interleave placed within a laminate stack adapted from [219].

The abundance of interleaving material variations in recent years calls for the categorisation into five types; additional matrix resin interleave (Section 2.10.1.1), thermoset films, short fibrous veils, polymer particulates, and thermoplastic films. Interleaving may behave as a fracture suppressant where the reported enhancements were discussed for inter-laminar related failures. Therefore, the interleaving prospect can be especially useful to improve the impact damage tolerance and in-plane compression strength in comparison to non-interleaved materials. Interleaving has demonstrated significant enhancements to the fracture toughness of laminates when combined with fibre reinforced laminates; however, the scope for further work will become apparent in the subsequent sections.

Nonwoven veils made of carbon fibre tissue were incorporated into CFRP laminated composites as reported by Cheong and Lee [218]. Their testing revealed the hybrid layup had 30% higher tensile strength than virgin CFRP at 1.8% of the strain level. Lee and Noguchi later focused on shear behaviour of nonwoven carbon veils [218] which reported a shear strength 10% higher for hybrids than CFRP. An additional observation by Lee *et al.* [218] exposed a benefit of veil interleaves is an ability to absorb fracture energy. Consequently, Lee *et al.* further explored interleaves within CFRPs under mode II interlaminar fracture toughness tests [219] and also for mode I [220]. Under mode II [219], UD CF/epoxy pre-preg laminae with carbon veils of 12 gm⁻² areal mass were manufactured as 3-ENF configurations. In comparison to non-interleaved CFRP, the results highlighted an influence of short fibre bridging increasing the IFT by 260% (pre-crack at centre) and by 205% (pre-crack in lower position).

In the latter study by Lee *et al.* for mode I [220], It was reported that enhancements were merely 8% higher than the control when using 6 veil layers and 28% higher when using 12-80 layers. However, the exact damage mechanisms were not studied or reported why the differences were minimal between 12-80 layers. Subsequently, Lee *et al.* [266] similarly explored the viability of carbon tissue interleaves and carbon nano-tube veil within laminated CFRPs. Lee *et al.* [266] established that the mode I interlaminar toughness decreased 5% for the carbon tissue veil, whereas the carbon nanotube veil had an increase of 353% against the control CFRP. Under mode II, IFT increased 194% for the carbon tissue interleave, also 246% using the carbon nanotube interleave in comparison to the control.

Subsequently, Lee *et al.* [267] proposed techniques for void evasion by manufacturing carbon veil enhanced hybrid CFRP pre-pregs. Lee *et al.* [267] also suggested another method to decrease voids is by the addition of extra epoxy below each veil, which expectedly resulted in a decrease of voids. However, to study the interlaminar interactions across various veils used in laminated composites with various resin systems, Kuwata and Hogg [268–270] have provided a more preferential and significant contribution to the veil interleaving knowledge.

Kuwata *et al.* [268–270] studied a variety of interleaved veils: 2xPolyesters, 2xHybrid (polyester/carbon fibre) and 1xHybrid (polyamide/carbon fibre). The most prominent results under mode I was when combining UD plies with polyester and polyamide veils which resulted in higher toughness using a vinyl-ester resin system. Kuwata *et al.* [268] established that thermoplastic fibrous veils were the most worthwhile for mode I fracture energy absorption resulting from the improved fibre-matrix interactions.

In comparison, mode II tests under the 4-ENF configuration demonstrated the reliance on fibre bridging and increased influence of resin properties. Consequently, the fracture types under mode II loading involved further complexity than expected and advocated further studies [269].

Beckermann *et al.* [25] also explored the influences of nanofibre veils on the interlaminar toughness. It was established that the use of 4.5 gm^{-2} PA66 was the most effective for the greatest IFT, where improvements in IFT were reported to be 156% (under mode I) and 69% (under mode II) against the control samples.

Tsotsis [271] also applied interleaving techniques to improve the interlaminar toughness properties; however, with the addition of thermoplastic fibrous veils, as advocated by Hogg *et al.* [22]. The outcomes resulted in Tsotsis securing US patent [272] rights which proposed manufacturing combinations of interleaving thermoplastic veils and stitching of laminates that can preferentially absorb low-energy impact. The method increases mode II interlaminar toughness using interleaves and also mode I by through-thickness tow stitches. The combination of toughening techniques are subsequently applied to the Boeing-Dreamliner 787 [272]; however, there is no clarity on the variability of interfacial reinforcement orientations.

2.10.4.1 Nonwoven veil interleaves

This research endeavour focuses on the more profound understanding of the mechanisms of interleaved veils under mode I & II and their respective IFT parameters to ultimately enhance the impact damage tolerance. Therefore, it is fundamental that a deeper understanding of nonwovens are developed for a more detailed understanding of damage mechanisms.

Interleaving of composite laminates using nonwoven veils appears as an alternative to the intermingling process described in section 2.10.3. Considering the ease of implementation into composite laminates and using higher elasticity materials for improvements of the IFT and fracture behaviour makes veil interleaving highly attractive. This technique combines the advantages of being non-disruptive to the overall laminate stiffness and allows the modification of bespoke locations within the laminate.

Nonwoven veils can be made up of a web of randomly oriented natural or synthetic fibres and held together by fibre-to-fibre friction or adhesion [273]. Veils are applied extensively within automotive applications for surface finishes, consumer products and military applications for items including but not limited to: fibrous filters, abrasive backing, cable sheathing and insulation, among many other applications [274–279].

2.10.4.2 Nonwoven veil manufacturing

The process for manufacturing nonwoven tissue or veils is segregated into three main categories: dry-lay with air as the carrying media for web formation, a wet-lay process using a fluid carrier that permits an anisotropic distribution of fibres and a heated / pressurised extrusion method, commonly known as electrospinning.

The most common method is the wet-lay process for manufacturing veil interleaves used in composites, owing to their ability to maintain an even dispersion and distribution of fibres. The process can also allow for binders and additives to be used, which can dissolve or combine with the parent matrix without creating interphase adhesion problems. The wet lay method is also widely used for paper manufacturing where fine and short fibres can form various thicknesses of sheets in highly accurate formats.

The wet lay process employs a viscous slurry where discontinuous fibres are mixed and processed with a defined fibre weight dilution, in addition to adding a de-flocculent and coagulant agent to prevent the fibres from forming premature clusters within the solution. The fibrous slurry solution is distributed through a moving carrier screen or substrate where the excess liquid is drained often by vacuum and heated blower fans. Consequently, the nonwoven veil is combined into a webbed veil by spraying powdered or liquid binders. The nonwoven web is then carried within a heated roller press to consolidate the spray-on binders. Subsequently, it is passed through a series of forced-convection drying ovens to dehydrate the web of any residual moisture. Lastly, the veil web is examined for quality, often using non-contact optical transmission or laser triangulation methods and subsequently stored in cool and dry conditions [274]. The importance of storage conditions become further significant for moisture soluble binders which can deteriorate and breakdown the web fibres if moisture is prematurely introduced during storage.

In addition to the brief description of the wet lay process, there are several variables which affect the fibre dispersion, preferential and machine fibre directions. There are also numerous methods available to bind the nonwoven web dependant on the desired application.

2.10.4.3 Interleaved thermoplastic veils

An intermediate step towards hybridising composites and intermingling thermoplastics is the use of wet-laid thermoplastic veils between reinforcement plies of a composite laminate. Resin rich regions, miss-matched fibre orientations and the introduction of interlaminar stress concentrations result in polymer composites being inherently feeble between plies. Non-woven thermoplastic veils can be incorporated in these inter-ply regions; they require no dispersion into the matrix resin [16] and do not significantly affect other mechanical properties [257].



Fig.31 PPS veil interleave material – Technical Fibre Products [280].

The use of non-woven veils for interleaving CF composites has been carried out by Lee and Noguchi [218–220], who found enhancements in fracture toughness by interleaving polymer veils for mode I & II. Many similar studies have been carried out [22–26,223,281]; however, the exact damage mechanism and optimisation of the hybridising has not been fully explored. Additionally, the plateau of enhancement has not been explored, with consideration of exact veil interface locations, density, applicability in multiple layers and the relative interfacial fibre orientation effects.

For the incorporation of veils into autoclave cure pre-preg systems, the outgassing of PPS is unlikely with the curing temperatures. Thermogravimetric analysis (TGA) by Naffakh *et al.* [282] demonstrates that mass-loss initiation occurs at temperatures exceeding 450°C, which is much greater than most laminate curing and service temperatures. The glass transition temperature of PPS with various crystallinities initiates at 85°C and a change in storage modulus occurring at 80°C; thereby improving the chance of compaction during the temperature ramp-up stage of the cure.

The consideration of detailed finite element modelling of nonwoven webbed veils requires idealisation of the dispersed web structure from the chaotic 3D fibrous structure. A simple idealisation can be through an assumption of consistent thickness where a random 2D fibrous network is conceivable. Software combinations using MSC DigiMat in conjunction with Fluent CFD for fibre mapping is permissible. For either 3D or 2D models, the veil fibres are idealised as a random distribution and separated from each other. The fibre centre of mass can be idealised as the centre of a finite unit cell element with their corresponding internal

fibre orientations being randomly orientated or mapped if orientation bias is present from the manufacturing process. For the consideration of a global interleaved application model requires the use of an equivalent cohesive zone model (CZM). The CZM can be applied using affluent coupon test data to characterise the interleave layer as a single interface.

From the reported studies and current trends for toughening techniques, interleaved thermoplastic layers offer an encouraging technique to enhancing the impact damage tolerance of composite laminates owing to their enhancements on the IFT. The flexibility of interleaved materials is an attractive and versatile option which can provide the initial point for interlaminar crack arrest and damage tolerant enhancement studies.

2.11 Damage mechanics for Interleave Toughened Composites

A method identified to achieving improved interface fracture toughness and higher failure strains is by hybridizing composites by combining brittle fibres with ductile fibres [283,284]. The larger contrast in fibre failure strain can lead to greater enhancements by increasing energy absorption through increased bridging accompanied by plastic failure of the interleave layer. The presence of these high-strain ductile materials within the composite laminate can enhance the fracture behaviour by two distinct characteristics. The first being the local stress fluctuations where the stress intensity at the crack front is lower than the surrounding region, and secondly due to a complete lack of stress intensity when a crack encounters the compliant interface and hence undergoes crack re-nucleation [285] and potential migration.

Hossain *et al.* [286] studied the fracture behaviour of layered materials and demonstrated a substantial increase in toughness. The enhancements and associated fracture mechanisms were a result of contrasting variations in the elastic moduli and fracture toughness of the layered constituents. The toughening of the layered materials were attributed to stress fluctuations in the layers which re-nucleated as the crack approaches from the compliant ductile region to the stiff layer. The fracture through a ductile region would typically exhibit nanofiber bridging and microcracking followed by progressive failure through plastic deformation or necking of the toughening fibres [223,224]. For toughened hybrid interleaved composites, the same damage mechanisms appear valid; where fracture re-

nucleation and migration can present apparent toughening of the material as the fractures navigate the local anisotropy and contrasting failure strain of the composite constituents. The appreciation of detailed damage mechanisms involved when interleaving ductile thermoplastic veils requires investigation into the various failure modes (Fig.8) to distinguish the primary benefits from the secondary outcomes.

2.11.1 Inter-laminar Failure

The study on interlaminar crack growth with interleaved veils has been a significant avenue to increase the understanding of damage mechanisms in toughened composite laminates. Research efforts by many [25,36,223,224] found that the onset, pre-crack and propagation G_{IC} values for laminates could vary depending on the interlayer material, areal density and strain capability. It was found that non-interleaved control configurations exhibit the typical brittle failure usually observed with epoxy resin systems, whereas interleaved specimens exhibit evidence of necking and ductile failure of the interlayer. Beckermann *et al.* [25] demonstrated the brittle fracture surface observed after progressing through ductile regions where the crack migrated out of the toughened interlayer and into the untoughened lamina interface regions.

2.11.2 Intra-laminar Failure

The incorporation of interleaved materials in laminated composites often have secondary effects by re-nucleating and migrating fractures to non-ductile regions due to the contrasting strain limits and localized stress conditions. The resulting fracture toughness and behaviour appear as apparent or effective enhancements with fractures typically moving into intra-laminar regions. Under mode II loading conditions the fractures occur resulting from combined shear and tensile microcrack growth (Fig.8). Although the exact toughening mechanisms and their explanations are not fully expansive, it is apparent from the study of fracture surface morphologies that thermoplastic nanofibers in veil interleaves can bridge microcracks forming within the interlayer during mode II fracture [25]. These bridging characteristics can absorb the energy required for further fracture, thereby enabling the interleaved laminate to have an effective toughening at the fibre level and requiring increased energy due to individual fibre friction contributions. The multi-directionality of laminae can also significantly influence the fracture behaviour, where the effects of

interfacial orientation are expanded in Chapter 3. The fibre level interactions brought about from interleaving between lamina can provide improved fracture characteristics and resist the microcrack opening resulting in improved mode II fracture toughness.

2.11.3 Low-velocity Impact Damage

The ultimate goal from this research endeavour is to improve the low velocity impact performance of laminated composites by enhancing the interlaminar toughness due to being a limiting factor for increasing composite applications. The interleaving approach allows bespoke regions of laminated structures to be enhanced for damage tolerance, however, the position and placement within the laminate is often a variable which is dependent on the impact energy, impactor size and laminate thickness and component dimensions. The behaviour of fracture damage in composite laminates for this research will pursue those which cause internal barely visible damage (section 2.5) due to their detrimental and difficult detection traits. The damage caused by such impacts results in the need to enhance the inter-laminar and intra-laminar fracture resistance and increasing the overall toughness. During an impact event, the interface regions experience significant strain concentrations which cause laminae to delaminate. Often interlaminar fracture behaviour also experiences intra-laminar fractures (Fig.5) which cannot directly be addressed using interleaves, however, these fractures often have treacherous pathways requiring greater impact energies. By improving the interlaminar fracture resistance or the strain energy release rate, improved impact damage tolerance can be achieved through the stacking of filler or toughening interleaves.

The position of interleaves within the laminate also has significance where the total laminate thickness, alongside the impact energy, impactor shape and laminate dimensions determine where the damage would occur within the laminate (Fig.5). The need for understanding the influence of veil at the interfacial region of multi-directional laminates along with through-thickness veil distribution is apparent and explored experimentally in Chapter 4.

2.12 Summary

Whilst the global behaviour of light-weight composite laminates provides extraordinary performance; the inter-laminar failure behaviour consequently results from mode I & II type fractures. These microscopic fractures ultimately develop into enlarged delamination regions under bending, compression, impact and in-plane shear loading. The appreciation of interleaved veils with consideration of their detailed interlaminar influence become significant for enhancing the damage tolerance laminates. This chapter has provided the necessary understandings of damage tolerance, including IFT in composites and principles of damage tolerance enhancement using nonwoven veil interleaves. The testing procedures for assessing the mode I & II fracture toughness and impact testing were described. Thereafter, various examples for improving the IFT were provided with a consensus that veil interleaving was as demonstrated as the most practical and encouraging toughening technique.

The damage characteristics of interleaved composite laminates were subsequently described while highlighting uncertainties and variables which could be encountered. The need for understanding the fracture behaviour in multi-directional laminates was also highlighted while leading onto the effectiveness and uncertainties relating to the through-thickness veil distribution. The work fundamentally leads to the exploration of many important laminate configurations and conditions to be explored in Chapters 3 and 4.

3 Effect of veil interleaves on Mode-I & Mode-II fracture resistance:

3.1 *Effect of Interfacial Fibre Orientation and PPS Veil on Fracture Resistance of UD CFRP Laminates under Mode I Loading.*

Pages 71-93

3.2 *Delamination Migration in Multidirectional CFRP Laminates under Mode I Loading.*

Pages 93-114

3.3 *Effect of Interfacial Fibre Orientation and PPS Veils on Fracture Resistance of Woven CFRP Laminates under Mode I Loading*

Pages 116-135

3.4 *Influence of Interfacial Fibre Orientation and PPS Veils on the Fracture Resistance of Woven CFRP Laminates under Mode II loading.*

Pages 136-155

3.1 Effect of Interfacial Fibre Orientation and PPS Veil on Delamination Fracture Resistance of UD CFRP Laminates under Mode I Loading

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Abstract: Interlaminar fracture toughness characterizes the resistance of a material to delamination initiation and growth, it is a crucial parameter in damage tolerance design and analysis of laminated composite structures. This paper presents an experimental investigation on the combined effect of interfacial fibre orientation and thermoplastic veil interleaves on the fracture resistance of laminated unidirectional (UD) carbon fibre reinforced polymer (CFRP) under Mode I loading. Double cantilever beam (DCB) specimens with interfacial fibre orientations of 0/0, 0/45, 0/90, 45/90 and 90/90 at the midplane were examined to obtain the effective resistance in two groups: one with an interleaved veil at the midplane and a control group without veil. It was found that for specimens without the veil, the measured effective propagation resistance at the 90/90 interface is 194% greater than that of the 0/0 interface. Enhancement of the fracture resistance using the interleaved PPS veil is fibre orientation-dependent, with 0/0 interfaces showing the greatest improvement of 48.6% in propagation fracture resistance. Fracture surfaces have been examined under Scanning Electron Microscope (SEM) for samples with various interfacial conditions to identify the failure characteristics. Fibre bridging and fracture migration are found to be closely affected by the interfacial fibre orientation and interleaving, influencing the propagation fracture resistance. Results also demonstrate that the fracture toughness determined from 0/0 interfaces following ASTM D5528-13 is over-conservative in design optimisation and simulation. Furthermore, the toughness enhancements of the veil assessed on the 0/0 interface are over-projected for interfaces with alternative fibre orientations.

Keywords: Polymer-matrix composites (PMCs) – A; Delamination – B; Fracture toughness – B; Damage tolerance – C; Fractography – D

3.1.1 Introduction

Laminated composite structures have many advantages compared with metallic counterparts, being lightweight, having improved in-plane mechanical properties and design flexibility for components with bespoke structural performance. However, a weak laminae interface due to the lack of through-thickness reinforcement is a major concern for laminated fibre reinforced composites. A compromised interface relating to poor interlaminar toughness G_{IC} and G_{IIC} for Mode-I and Mode-II loading conditions can trigger early fracture at the interface, leading to the formation of delamination. It is widely accepted that laminated composite structures are highly susceptible to out-of-plane impact damage, introduced during the manufacture, operation or maintenance of a component. *“Even when no damage is observed on the surface at the point of impact, barely visible impact damage (BVID) such as matrix cracking and delamination can occur and contribute up to 60% loss in a laminated composite component’s compressive strength”* [45]. Additional factors contributing to delamination tendencies include stress concentrations, voids, resin-rich regions, and other material or manufacturing defects. Subsequent propagation of the delamination during service often leads to an undetectable compromised structure with a substantial reduction in mechanical strength, posing a significant threat to the safe continued operation.

Many efforts have been made by researchers to improve the fracture resistance and overall delamination threshold; mostly by modifying the material constituents. The surface treatment of reinforcement fibres, commingling, interleaving and through-thickness reinforcement have all revealed various levels of success in improving the interlaminar fracture toughness (G_{IC}). Among them, tough thermoplastic inter-layers positioned between the laminae of the composite has shown to significantly increase the overall fracture toughness [2-13]. Planar toughening methods like these, allow for the exploitation of the tough and ductile inter-layer with marginal increases of the interface thickness. The fracture and growth behaviour of thermoplastic interleaved laminates are related to the deformation radius at the delamination front, alongside peeling and bridging of fibres behind the delamination front. It is accepted that a larger radius delamination front associated with a toughened matrix can hinder the growth and the path of the propagating delamination. The interleave material has a significant influence on the stress concentration and opening profile at the front of the delamination and hence resistance to the initiation and growth of the delamination.

It is worth noting that most research activities considering mode I fracture toughness of a given UD fibre/matrix material system were predominantly conducted using double cantilever beam (DCB) samples with a 0/0 midplane interface according to ASTM 5528-13 [53], however with some exceptions [141,142,145,287–289]. The effect of interleaving veils on the fracture resistance is subsequently assessed by adding a polyphenylene sulfide (PPS) veil layer to the 0/0 midplane interface. The results obtained are hence only valid for the interface of 0/0 interfacial fibre orientation when the delamination is initiated and propagated along the pre-cracked 0/0 interface.

However, actual laminated composite components have interfaces of different fibre orientations and often involve multiple delaminations propagating through the thickness of the laminate. The influence of interleaved veils on the propagation resistance may also vary with the fibre orientation at the interface. There remains uncertainty related to fractures in interleaved multi-directional laminates that may migrate and avoid the toughened interlayer entirely. These are essential technical questions for the design and manufacturing optimisation of laminated composite components and yet received minimal attention, particularly for the combined effect of fibre orientation and interleaved veil on the effective fracture resistance. This paper presents a detailed experimental exploration to understand the combined effect of interfacial fibre orientation, and PPS interleaved veils on the fracture resistance and propagation behaviour of UD CFRP laminates under mode I loading.

3.1.2 Materials and Experimental Procedures

Double cantilever beam specimens, as illustrated in Fig.32, were manufactured for the experimental investigation with specimen geometry following ASTM 5528-13 [53].

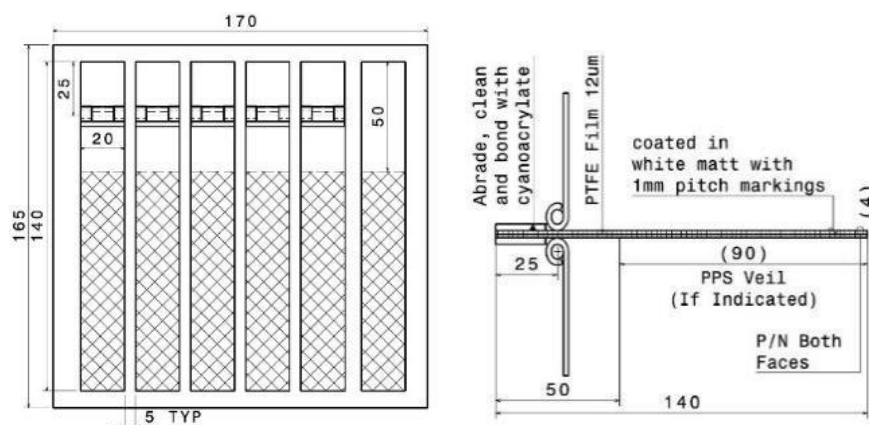


Fig.32 (a) Geometry of laminate plate for specimens, and (b) DCB specimen assembly.

For all specimens, the pre-impregnated CFRP material was supplied by SHD Composites, consisting of UD carbon 3K Toray T300 fibres, with MTC510 toughened epoxy system of 37% resin weight [290]. It has an areal density of 150gm^{-2} , a nominal ply cure thickness of 0.143 mm and relevant mechanical properties listed in Table.1. Each non-interleaved (control) sample for a specific midplane interfacial fibre orientation had a veil interleaved counterpart for comparison. The veils supplied by Technical Fibre Products are commercially available; *Optiveil OP-49-48 PPS* [280], with a nominal areal density of 10 gm^{-2} and thickness of 0.09 mm. *Optiveil OP-49-48* was studied due to its lightweight thermoplastic qualities, compatibility with the MTC510 epoxy resin system, along with the flexibility to be introduced at bespoke locations within a laminated structure. The veils are manufactured using a wet-lay process with chopped strand PPS fibres of $10\text{ }\mu\text{m}$ diameter and 6 mm length in a styrene acrylic binder and has a dissimilar manufacturing process to the electrospinning process studied in other literature.

Table.1 Mechanical properties of MTC510- UD150-HS-37% RW [290].

Property	Result	Standard
Fibre Density [g/cm^3]	1.8	N/A
Matrix Density [g/cm^3]	1.21	N/A
Fibre volume fraction [%]	53.4	N/A
Cured ply thickness [mm]	0.143	N/A
0° Tensile strength [MPa]	2112	BS EN ISO 527-5
0° Tensile modulus [GPa]	110	
0° Poisson's ratio	0.34	
90° Tensile strength [MPa]	54	
90° Tensile modulus [GPa]	8.2	
90° Poisson's ratio	0.01	
0° Compression strength [MPa]	988	EN 2850 Type B
0° Compression modulus [GPa]	105.1	
90° Compression strength [MPa]	200	
90° Compression modulus [GPa]	9.3	
In-Plane shear strength $\pm 45^\circ$ [MPa]	99	ASTM D3518
In-Plane shear modulus $\pm 45^\circ$ [GPa]	3.6	
Interlaminar shear strength 0°[MPa]	85	BS EN ISO 14130

As presented in Fig.32, five specimens were fabricated under the same manufacturing conditions, providing four consistent repetitions for each interface configuration listed in Table.2. The DCB specimens were rectangular with the dimensions of 140(L)×20(b)×4mm(h). For the interleaved specimens, a layer of non-woven veil was placed at the midplane, adjacent to the PTFE release film. The twenty-eight UD CFRP layers were cold consolidated under vacuum during the layup and subsequently placed on a flat steel plate with peel film, caul plate, breather fabric and cured within a vacuum bag in the autoclave. The panels were cured in the autoclave under pressure at 620 KPa (90 psi), with a target bag vacuum of less than 5 KPa and temperature set to 100°C for 4 hours as prescribed for MTC510 [290].

Table.2 Specimen configurations for DCB tests.

Midplane Specimen Code	Layup Configuration
90/90 90/90 +veil	$[(0, 90)_7]_s$ $[(0, 90)_7, \frac{veil}{2}]_s$
0/0 0/0 +veil	$[(90, 0)_7]_s$ $[(90, 0)_7, \frac{veil}{2}]_s$
90/0 90/0 +veil	$[0, 90]_{14}$ $[(0, 90)_7, \frac{veil}{2}, (0/90)_7]$
90/45 90/45 +veil	$[(0, 90)_6, (-45, 90, 45, 0), (90, 0)_6]$ $[(0, 90)_6, (-45, 90, \frac{veil}{2}, 45, 0), (90, 0)_6]$
0/45 0/45 +veil	$[(90, 0)_6, (-45, 0, 45, 90), (0, 90)_6]$ $[(90, 0)_6, (-45, 0, \frac{veil}{2}, 45, 90), (0, 90)_6]$

The fracture toughness of the specimen was assessed under the same principle as specified in ASTM D5528-13 [53] and BS ISO 15024:2001 [291]. The deviation predominantly being the mid-plane interface configuration, required for focusing on the effect of fibre orientation and veils on the fracture resistance. Unlike the 0/0 fibre orientation specified in ASTM D5528-13 [53], five interfacial fibre orientations as shown in Table.2 were introduced at the midplane of the DCB specimen; which are 0/0, 90/90, 0/90, 90/45 and 0/45 respectively. A fully symmetric and quarter symmetric laminate provides the maximum bending stiffness and does not suffer membrane–bending coupling effects relative to the middle plane. The layups above are symmetric on the outer majority of plies, hence membrane coupling effects are thought to have minimal influence. The coupling and twisting effects are subsequently considered using numerical simulation techniques specifically focussing on the mixed-mode contribution ratio.

The specimens were clamped via cyanoacrylate bonded hinges into vice grip fixtures and axially aligned with the load path. A printed scale bar affixed to the specimen allowed visual observation of the delamination front on a white painted edge. The tests were conducted using a dual column universal test machine (Instron 5965) equipped with a calibrated 5 KN load cell. The crosshead displacement rate was set to a constant 2 mm/min. The test process was captured on video using a 150 mm macro lens and tripod-mounted DSLR camera system shown in Fig.33.

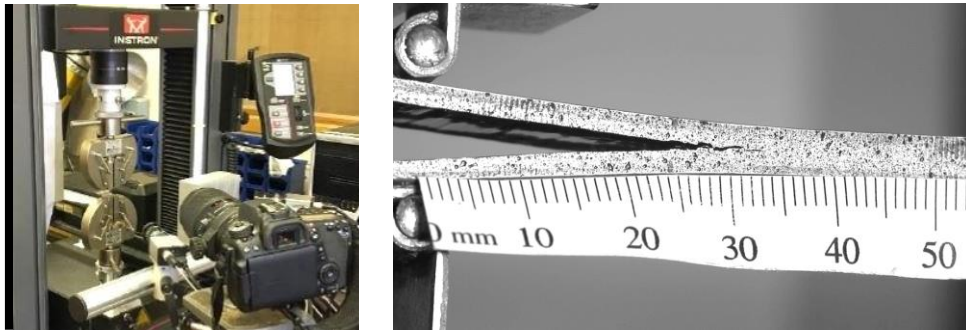


Fig.33 (a) Test setup, (b) Loaded specimen.

The onset of delamination and subsequent propagation was visually recorded on the edge of the specimen whilst obtaining the corresponding load-displacement from the test machine. The specimens were continuously loaded until sufficient data points were obtained or until specimen failure. Delamination length measurements (a) were subsequently correlated to the reaction load (P) using the video recording, constant displacement rate and the loading initiation time. The mode I fracture toughness, G_{IC} , was calculated using the modified compliance calibration method [53]:

$$G_I = \frac{3P^2 C^{2/3}}{2A_1 b h} \quad (\text{Eqn.14})$$
 where P is the applied load, b is specimen width, h is the specimen thickness, A_1 is the slope of the least-squares plot of the delamination length normalised by specimen thickness, a/h as a function of the cube root of compliance $C^{1/3}$. C is the ratio of the load point displacement to the applied load, δ/P .

3.1.3 Results

3.1.3.1 Effect of interfacial fibre orientation and veil on load-displacement

The enclosed area of the load-displacement curve obtained in the DCB tests reflects the specimen resistance to delamination and is therefore closely related to the fracture resistance.

Fig.34 presents the consistency of load-displacement curves of all repeated test specimens. There are distinct differences in the enclosed area (Table.3 - Table.4) of the load-displacement curve among different interfacial fibre orientations, demonstrating the significant influence of interfacial fibre orientation on fracture resistance. For the specimens without the veil shown in Fig.34(left), the 90/90 midplane specimen has the highest resistance to delamination while the 0/0 midplane specimens has the lowest resistance. The 0/90 interface propagation migrated from the pre-cracked mid-plane into the 90° layer; therefore, the effective traction load appears consistent with the 90/90 sample set.

The shape differences of the load-displacement curves are also worth noting. The 0/0 midplane specimens exhibit a stable reduction of the load after the peak force of the initial elastic deformation. Interestingly other specimens exhibit some further load carrying capacity beyond the peak force of the initial elastic deformation. Fig.34(right) shows the load-displacement behaviours of the specimens with the interleave veil. In comparison to the curves in Fig.34(left), the veil interleaved configurations in Fig.34(right) generally improves the resistance to delamination. However, the beneficial effect of the veil on fracture toughness and resistance is dependent on the fibre orientation at the midplane. The specimen with the 0/0 midplane benefits the most, while other midplane orientations appear to receive varying levels of enhancement resulting from the veil. Examination of the fracture surface morphology in the following sections explains the effect of interfacial fibre orientation, and veil interleaves on the resistance to delamination to shed some light on the major factors influencing the apparent toughness.

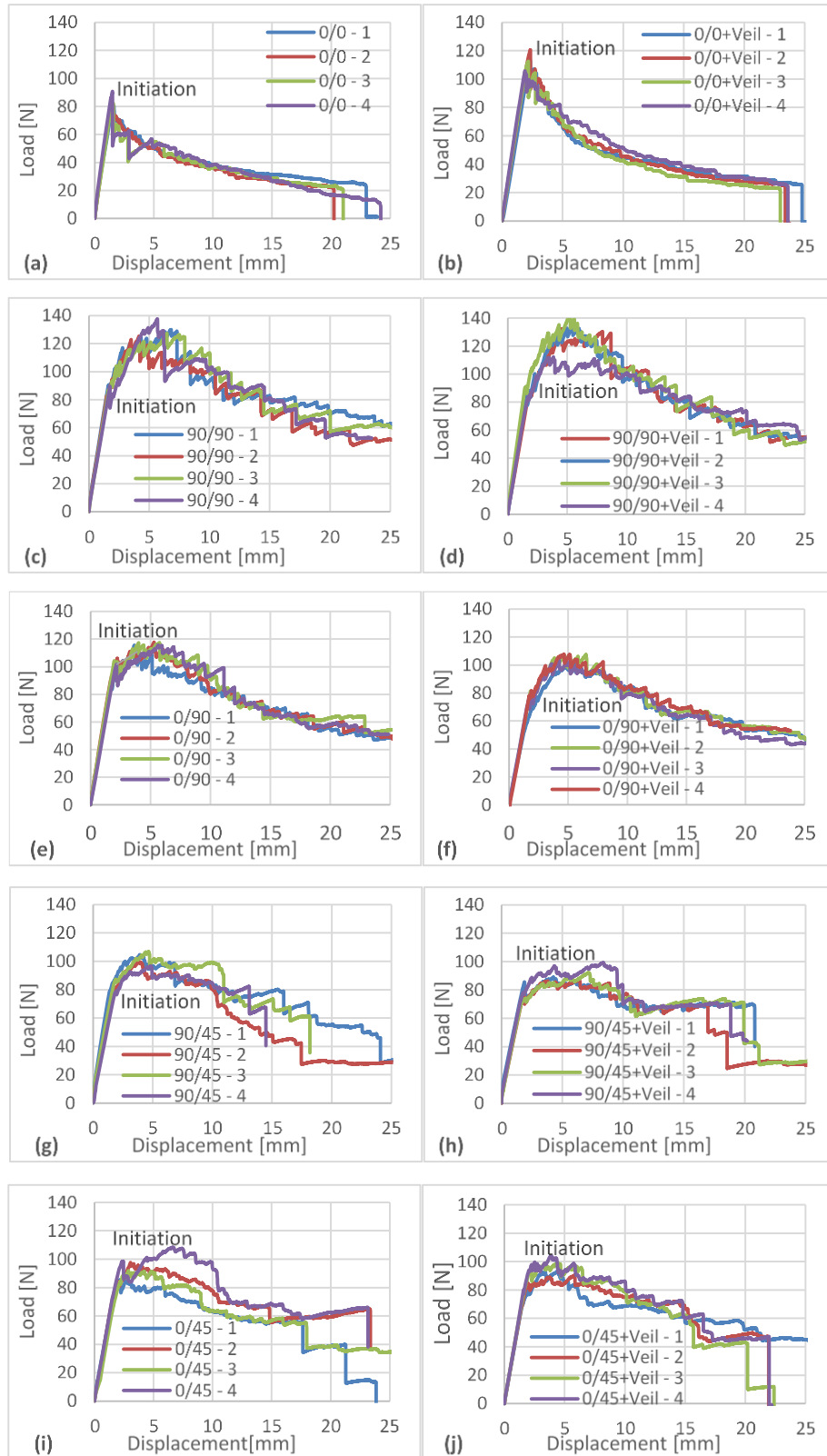


Fig.34 Load-displacement variation curves: (left) without veils, (right) with veils.

3.1.4 Effect of interfacial fibre orientation and veil on delamination resistance

The conventional method of processing load-displacement data is to use the mode-I strain energy release rate calculation introduced above and subsequently considered. Another approach to capture the global energy absorbed, by calculation of the area under the load-displacement curve from Fig.34. The area under the initial linear curve is the elastic potential energy being applied during the displacement. Some portion of the elastic potential energy is transformed into the fracture energy during the release of the stored elastic energy, however, some energy remains as elastic potential energy. The total absorbed energy results are listed in Table.5 - Table.6 and constrained to the initial region of crack opening displacement (0mm-15mm) for consistency. Additionally, the sample's standard deviation is also calculated to consider the change in energy dissipated between the specimen repetitions.

Table.3 Global Fracture Energy with veil consideration.

	Energy, E (Area Under Load-Displacement curve) (No Veil) [J]		Energy per unit Crack Growth [J/mm]	Energy, E (Area Under Load-Displacement curve) (+10gm ⁻² veil at midplane) [J]		Energy per unit Crack Growth (+10gm ⁻² veil at midplane) [J/mm]	
Midplane	Mean	StdDev	Mean	Mean	StdDev	Mean	δE/mm
0/0	222.4	29.6	13.9	271.8	14.6	17.0	22%
90/90	466.6	10.6	29.2	479.3	21.7	30.0	3%
0/90	424.9	10.0	26.6	386.0	8.4	24.1	-9%
90/45	379.0	36.7	23.7	367.1	13.6	22.9	-3%
0/45	360.7	36.3	22.5	367.1	19.5	22.9	2%

Table.4 Global Fracture Energy with orientation consideration.

Midplane	Energy, E (Area Under Load-Displacement) – (No Veil)		Energy, E (Area Under Load-Displacement) (10gm ⁻² veil at midplane)	
	Mean [J]	δE Vs. 0/0	Mean [J]	δE Vs. 0/0 + Veil
0/0	222.4	0%	271.8	0%
90/90	466.6	110%	479.3	76%
0/90	424.9	91%	386.0	42%
90/45	379.0	70%	367.1	35%
0/45	360.7	62%	367.1	35%

Note: stiffness variation due to mid-plane variation not considered, FEA for mixed-mode stiffness contribution computed in Table.7

The load-displacement curves in Fig.34 are processed further to obtain the resistance curves (R-curves) using the compliance calibration method, derived from the modified beam theory (MBT). R-curves of the five interfacial fibre orientations are presented in Fig.35, showing the influence of fibre orientation and veil interleaves on the apparent resistance to delamination. The inset images in Fig.35 are scaled fracture surface images highlighting migration positions relative to the crack length.

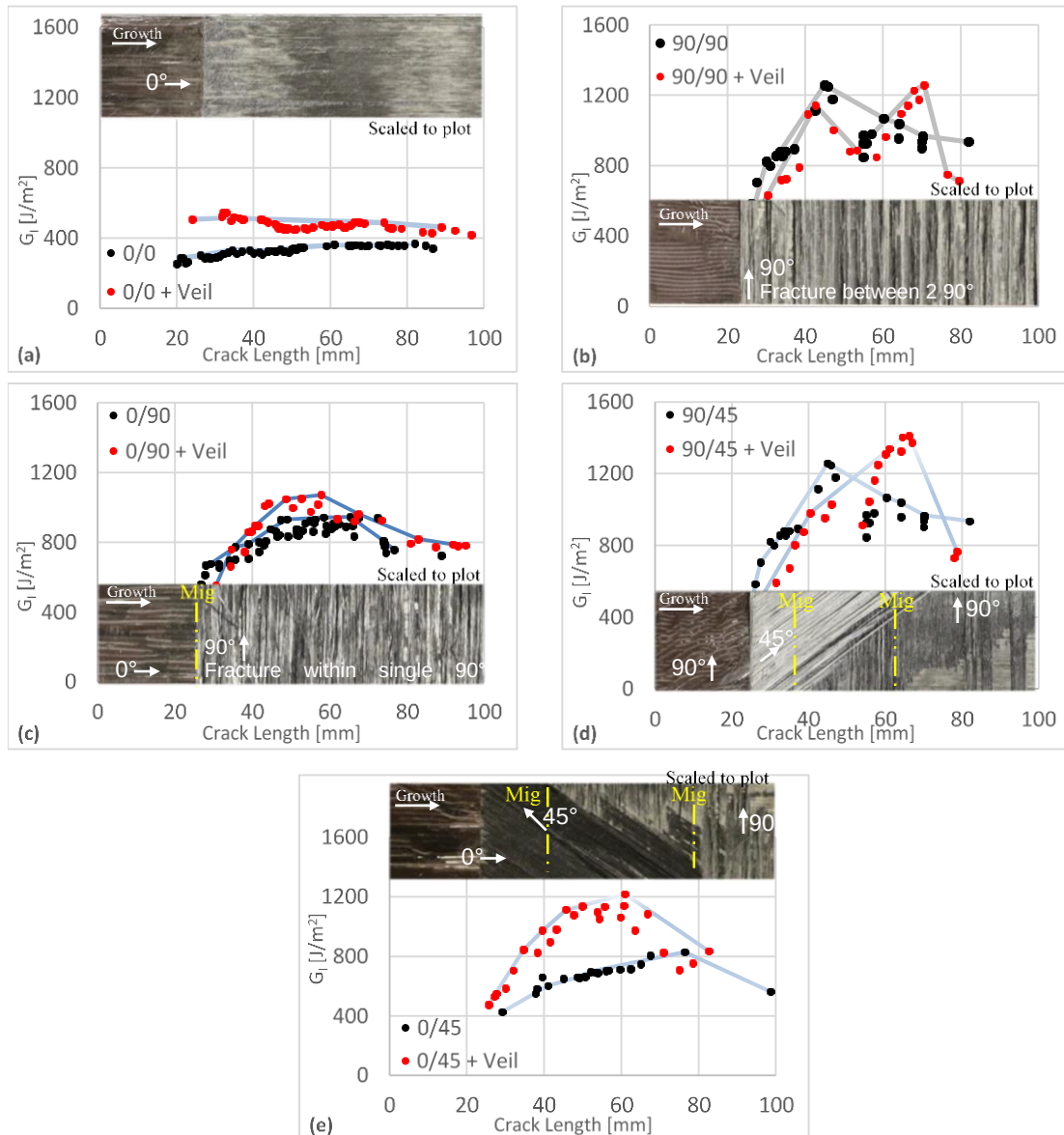


Fig.35 R-Curves for interfacial fibre orientations: (a) 0/0, (b) 90/90, (c) 0/90, (d) 90/45, and (e) 0/45.

Each graph in Fig.35 contains experimental data from representative specimens where markers in black represent the control specimens without veil, while red markers represent specimens with the veil interleaved midplane. Results in Fig.35 show that fibre orientation has a significant effect on resistance to delamination with the 0/0 interface being the least tough and the 90/90 interface appearing the toughest. Apart from the 0/0 interfacial fibre orientation as shown in Fig.35(a), the delamination resistance for the remaining four interfacial fibre orientations (Fig.35(b–e)) varies significantly in magnitude with the growth of the delamination.

These large variations in the apparent fracture toughness reflect the deviations in fracture plane position due to migration as observed in the inset photos within Fig.35. The variations reflect the real characteristics of the process of delamination growth at the interface of different fibre orientation, dissimilar to that observed for delamination growth at the 0/0 interface. The inset photos show the correlation of the fracture resistance to the surface morphology at various delamination lengths. They demonstrate that variation of the effective fracture resistance is closely related to the complex global interlaminar and local intra-laminar growth interactions at different interfacial fibre orientations due to fracture migration, where the delamination in the initial fracture plane wanders into adjacent interfaces [78,143,292–295].

For the DCB sample with the 0/0 midplane fibre orientation (Fig.35(a)), the growth of the delamination is dominated by the expected interlaminar delamination growth without fracture migration. The delamination growth of the 90/90 midplane fibre orientation shown in Fig.35(b) is however dominated by global intra-laminar growth within the 90° layers as recurring and undulating fracture migrations. Growth of the delamination for the remaining three fibre orientations (Figs.4(c–e)) shows various levels of fracture migrations leading to interlaminar and intra-laminar growth. The considerable variation of the resistance to delamination at the interface with differing fibre orientation is related to the through-thickness fracture migrations triggered at various positions. The fracture migration details are presented in the microscopic examination of the fracture morphology in the subsequent section.

Results in Fig.35 demonstrate that veil interleaved specimens have improved fracture resistance in general for most orientation configurations, with the apparent benefit being observed in the specimens of 0/0 and 0/90 orientations. For the other three fibre orientations, PPS veil interleaves appear to have limited benefits on the fracture resistance in comparison with samples without the veil. It is also worth noting that results for fibre orientations other than for 0/0 do not represent the direct influence of PPS veils on fracture toughness and must be considered as an effective-toughness. This is due to the evidence shown in the inset photos in Fig.35, where the delamination paths are revealed to avoid the interface with the veil layer as a result of fracture migrations being triggered.

A summary of the test results for the influence of interfacial fibre orientation and PPS veil on the propagation fracture resistance is presented in Table.5 - Table.6

Table.5 Mode I G_I (*Effective G_I due to intra-laminar migration)

Midplane	No Veil [J/m^2]			10gm ⁻² veil at midplane [J/m^2]			δG_{I_Prop}
	Sample $G_{I_Min/Max}$	G_{I_Prop} Mean	G_{I_Prop} StdDev	Sample $G_{I_Min/Max}$	G_{I_Prop} Mean	G_{I_Prop} StdDev	
0/0	285 / 369	325.6	29.7	418 / 544	483.7	41.5	+48.6%
90/90*	584 / 1272	957.5	146.0	632 / 1258	973.3	196.5	+1.7%
0/90*	664 / 942	799.4	126.6	760 / 1071	884.3	128.7	+10.6%
90/45*	552 / 1281	756.6	126.7	593 / 1409	813.0	154.6	+7.5%
0/45*	425 / 828	609.2	46.8	458 / 1210	766.1	178.3	+25.8%

Table.6 Mode I G_I (*Effective Propagation G_I due to intra-laminar migration)

Midplane	G_{I_Prop} (No Veil)		$G_{I_C_Prop}$ (10gm ⁻² veil at midplane)	
	Mean [J/m^2]	δG_{Ic} Vs. 0/0	Mean [J/m^2]	δG_{Ic} Vs. 0/0 + Veil
0/0	325.6	0%	483.7	0%
90/90*	957.5	+194%	973.3	+101%
0/90*	799.4	+146%	884.3	+83%
90/45*	756.6	+132%	813.0	+68%
0/45*	609.2	+87%	766.1	+58%

Results show fibre orientation has a significant effect on fracture resistance with the 0/0 interface having the lowest delamination resistance. Fracture resistance at the 90/90 interface is 194% greater than at the 0/0 interface for specimens without the veil. Thermoplastic veils improve the fracture toughness in general, but its improvement is fibre orientation-dependent with the 0/0 interface showing the greatest improvement of 48.6%. Propagation fracture resistance at the 90/90 interface is twice that of the 0/0 interface with the veil. It should, however, be recalled that G_{I_Prop} values for 0/90, 90/45 and 0/45 interfaces are not the pure mode I fracture toughness at the interface but are effective* G_{I_Prop} parameters due to significant fracture migration tendencies. Delamination migrations into a neighbouring weaker interface indicate that the actual fracture resistance at these interfaces may have higher toughness than those presented above. The direct numerical value of fracture toughness for the 0/90, 90/45 and 0/45 cannot be concluded as a result. The values reported in Table.5 - Table.6 are the mean effective propagation resistance extracted in the regions where migration was minimal (between 25 mm – 40 mm), which represent the actual resistance to fracture and should be used for design optimization and modelling when migration effects are not considered.

3.1.5 Fracture morphology at interfaces of various fibre orientations

The DCB fracture toughness tests were recorded to examine the fracture behaviour with a macro lens attached to a DSLR camera as shown in Fig.33. Representative images of the side view of the delamination growth are shown in Fig.36. The sample with 0/0 fibre orientation (Fig.36(a)) has a smooth delamination path while the sample with 90/90 fibre orientation (Fig.36(b)) has a rough and undulating delamination path due to fracture migration and requiring higher fracture propagation energy (Fig.34 and Table.5 - Table.6). The density of fibre bridging varies with the interfacial fibre orientation, and the delamination tends to migrate from the midplane to the neighbouring weak interface during its growth for samples of 0/90, 90/45 and 0/45 interfacial fibre orientations (Fig.35(c-e)). Examination of the side view of the specimens highlight key differences resulting from interfacial fibre orientation and subsequently studied in further detail.

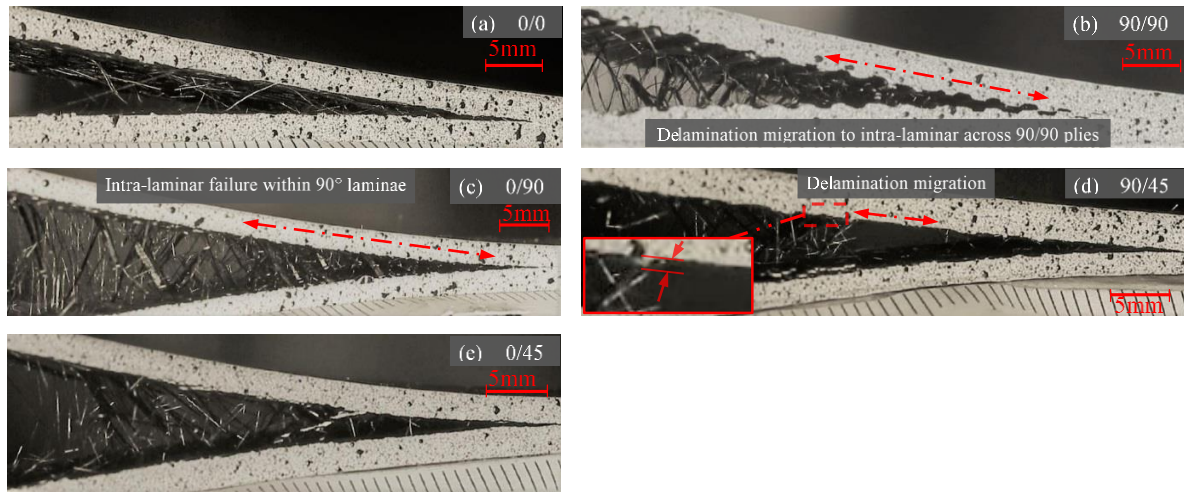


Fig.36 Delamination growth of specimens: (a) 0/0, (b) 90/90, (c) 0/90, (d) 90/45, and (e) 0/45.

The limited improvement of fracture resistance for samples with PPS veils for interfacial fibre orientations other than 0/0 can be explained by the fact that the veil may improve the magnitude of fibre bridging, but the delamination migrated into an alternative interface without the veil. This is supported by the fracture features of the 90/90 specimen (Fig.36(b)) where fracture paths avoid the interleaved interface and propagation continues above or below the 90/90 midplane, predominantly as intra-laminar failure. A stick-slip behaviour is found in the majority of non-0/0 specimens due to crack wandering and being affected by combinations of inter-laminar, intra-laminar and migration mechanisms. Results can therefore be considered as effective fracture resistance parameters with significantly higher fracture energy required for propagation.

Upon completion of the mode I fracture tests, specimens were dissected and placed within the SEM to examine the morphologies of the fracture surfaces. Fig.37 presents the fracture surface morphology of the specimens with 0/0 fibre orientation at the midplane. Both non-interleaved (Fig.37(a)) and interleaved (Fig.37(b)) specimens exhibit dominant interlaminar delamination growth at the 0/0 midplane. Closer examination of the fracture morphology reveals that interleaved specimens have a coarse and irregular texture (Fig.37 (b)) compared to that of the non-interleaved specimen (Fig.37(a)), thus explaining the 48.6% improvement in fracture resistance observed by the addition of veils. This is consistent with results in literature for enhanced interface bridging and resistance [22,23,25,26,223]

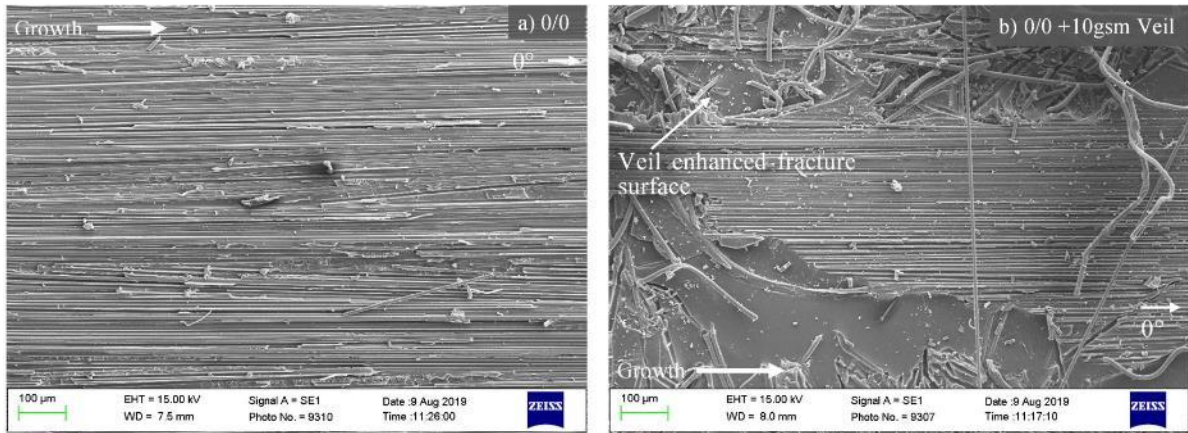


Fig.37 Fracture surface of 0/0 midplane specimens: (a) without veil; (b) with 10gm⁻² veil.

The fracture surface morphology for specimens with the 90/90 midplane fibre orientation is shown in Fig.38. Unlike the 0/0 specimens which exhibit dominant interlaminar delamination growth, both the non-interleaved (Fig.38(a-b)) and interleaved (Fig.38(c-d)) 90/90 specimens reveal a mixture of intra-laminar growth within the 90° layers. The intra-laminar growth within the 90° layer is dominant in 90/90 specimens as evident in Fig.38(a & c), with local interlaminar delamination growth also apparent in Fig.38(b & d). Remnants of resin regions and fibre peeling can be observed on the surface, indicating the further potential for improvement of the fibre-matrix surface interphase. Interestingly the interlaminar delamination grows intermittently along the neighbouring 90/0 interface with 0° fibres visible in both interleaved and non-interleaved, the fracture path does not strictly follow the pre-cracked 90/90 midplane.

The resulting fracture surface morphology of the interleaved and non-interleaved specimens has similar features, were the enhancement by interleaved veils on fracture resistance was not as influential as the 0/0 specimens. The veil provided little enhancement and slightly increased fibre bridging for the 90/90 interface, increasing the effective fracture resistance by 1.7% compared with the non-interleaved specimen. The interfacial fibre orientation is more influential on the mode I fracture resistance than interleaved veils as ply orientation normal to the fracture path triggers fracture path migration. However, the effect of veil on improving mode II fracture resistance may be greater than mode I as the enhanced fibre bridging will trigger increased fibre peeling and fibre fracture. Enhanced fibre-matrix interactions such as these, will result in greater resistance to the in-plane shear of fracture surfaces at the interface.

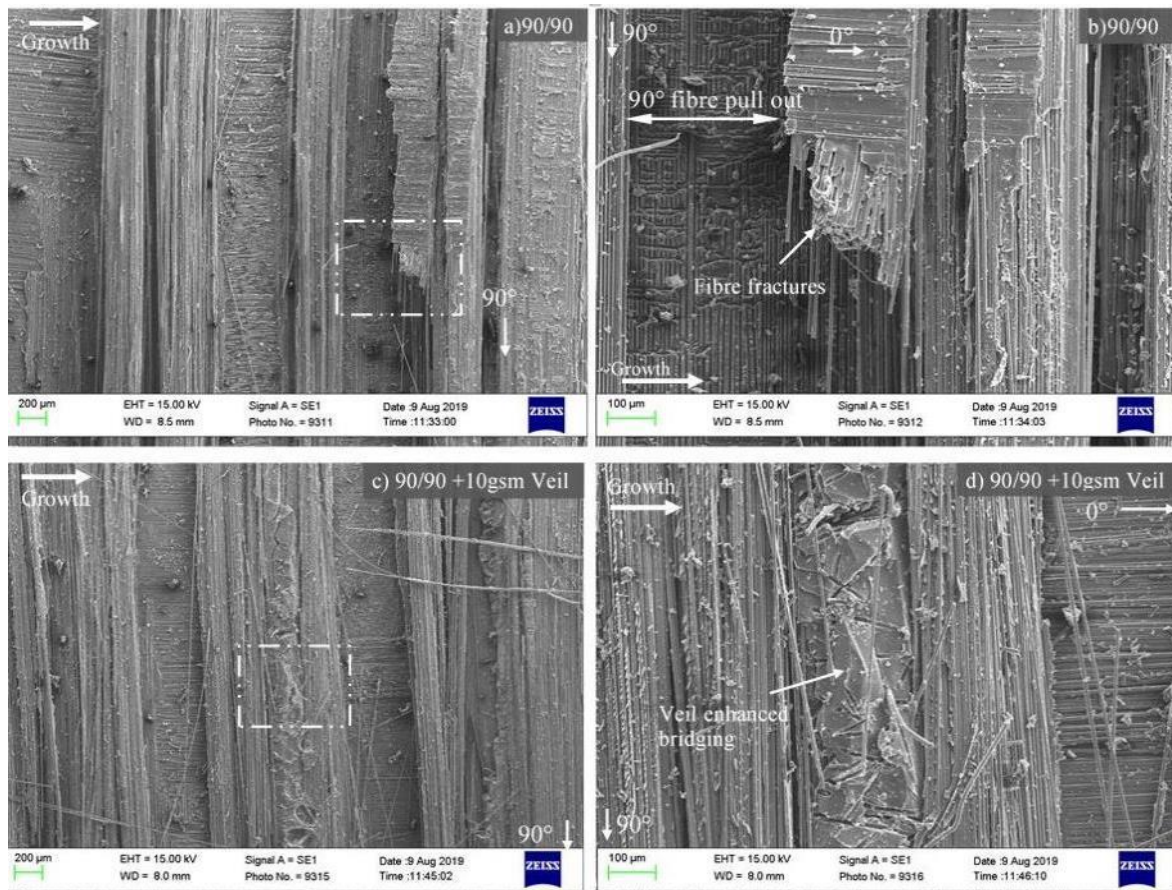


Fig.38 Fracture surface of 90/90 midplane specimens: (a-b) without veil; (c-d) with 10gsm^{-2} veil.

Fracture surface morphology for specimens with the 0/90 midplane fibre orientation is shown in Fig.39. Similar observations to the 90/90 specimen are evident, both the non-interleaved (Fig.39(a-b)) and interleaved (Fig.39(c-d)) 0/90 specimens exhibit a mixture of dominant intra-laminar growth and locally limited interlaminar delamination growth. The localised interlaminar delamination however occurs at the midplane, which is different to the 90/90 specimen. The presence of cusps [292,295] and remnants of matrix remaining on each surface indicates that surface treatment of the fibres has room for improvement of the fibre-matrix interphase. Compared with the non-interleaved specimen, the effective fracture resistance of the interleaved specimen is improved by 10.6% by the addition of interleaved veils at the 0/90 interface. It can also be observed in Fig.39(d) that the interleaved fracture surface has coarse and prominently irregular surface compared to the non-interleaved specimens (Fig.39(b)), suggesting that interleaving indeed affected the resistance to delamination growth. The 10.6% improvement in effective fracture resistance should be investigated further, considering the true global failure mode across the entire sample set was intra-laminar.

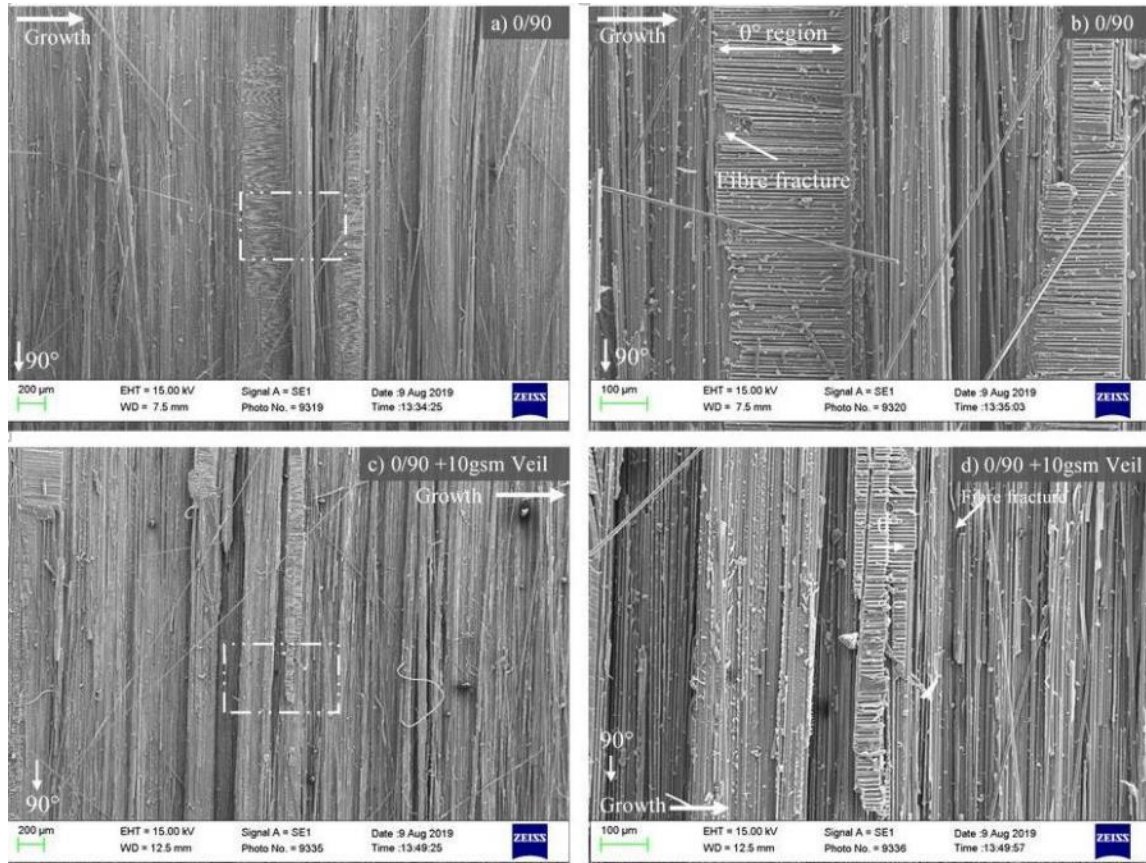


Fig.39 Fracture surface of 0/90 midplane specimens: (a-b) without veil; (c-d) with 10gsm⁻² veil.

Fracture surface morphology for specimens with the 90/45 midplane fibre orientation is shown in Fig.40. Both non-interleaved (Fig.40(a-b)) and interleaved specimens (Fig.40(c-d)) did not delaminate at the pre-cracked 90/45 interface. Delamination migration took place immediately after the PTFE release film and grew within the 45° layer until approximately 40mm from the hinge pin (Fig.32& Fig.35(d) - Fig.36(d)). Thereafter, delamination grew within the neighbouring 90° and subsequently the 0° interface (inset image on Fig.35(d)). Remnants of resin traces are also observable in Fig.40(b & d). The morphology of the interleaved and non-interleaved specimens has similar features and enhancement by veil interleaves on the effective fracture resistance, however, is very limited (7.5%). Nevertheless, it must be noted the true toughness value between 90/45 is undetermined due to fracture migrations but remains an important effective parameter, suitable for use in low fidelity numerical models.

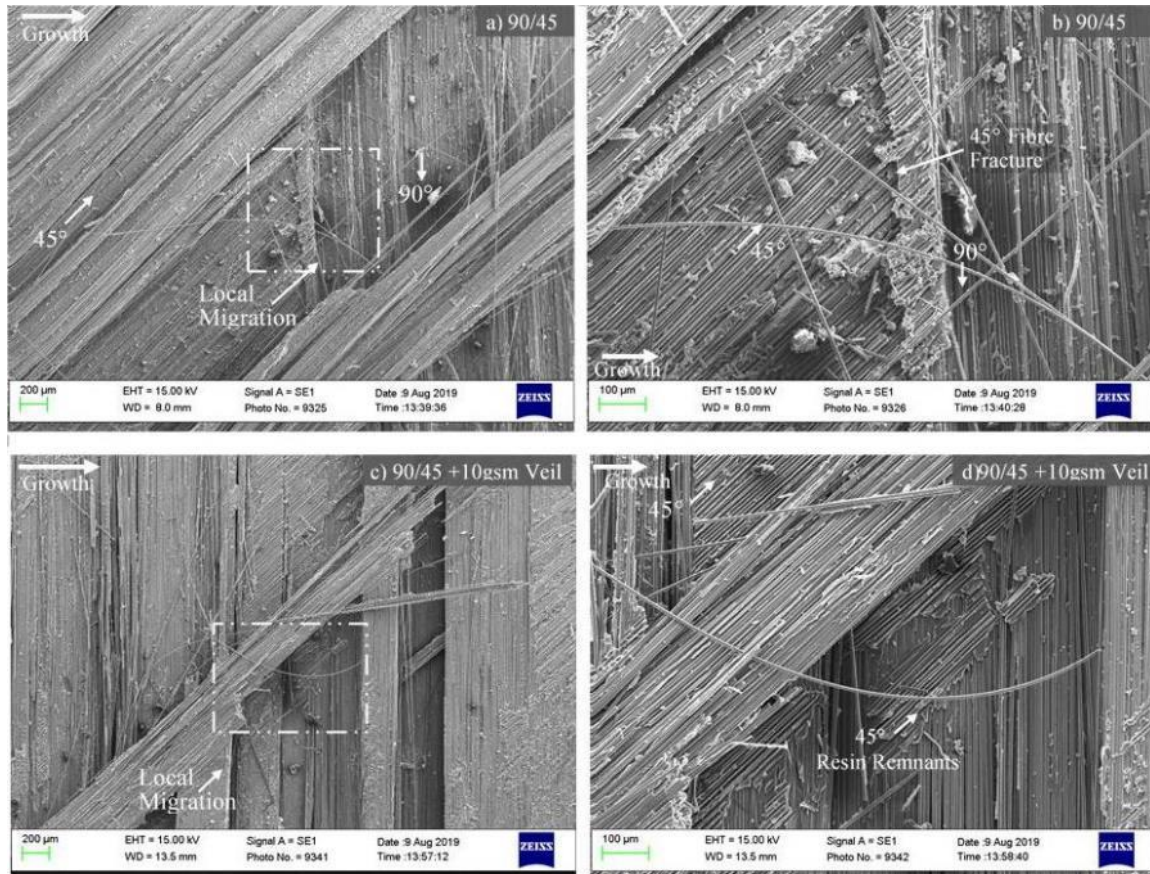


Fig.40 Fracture surface of 90/45 midplane specimens: (a-b) without veil; (c-d) with 10gm⁻² veil.

Fracture surface morphology for specimens with the 0/45 midplane fibre orientation is shown in Fig.41. Similarly to 90/45 specimens, for non-interleaved (Fig.41(a-b)) and interleaved specimens (Fig.41(c-d)), delamination did not take place at the pre-cracked 0/45 interface. Crack plane migration took hold immediately after the PTFE release film and grew within the 45° laminae, progressed through combinations of 45° and 90° laminae as an intra-laminar failure from 25-40 mm and entirely 90° until 64 mm where delamination occurs at the interface with the neighbouring 0° laminae (Fig.35(e) & Fig.36(e)). The morphology of interleaved and non-interleaved specimens has similar features, with fracture migration, inter-laminar and intra-laminar failures. The interleaved however having a coarse and significantly irregular surface (Fig.41(d)) compared to the non-interleaved midplane (Fig.41(b)), resulting in the apparent 25.8% enhancement. The true toughness value is undetermined, and the associated migration tendencies provide scope for future work.

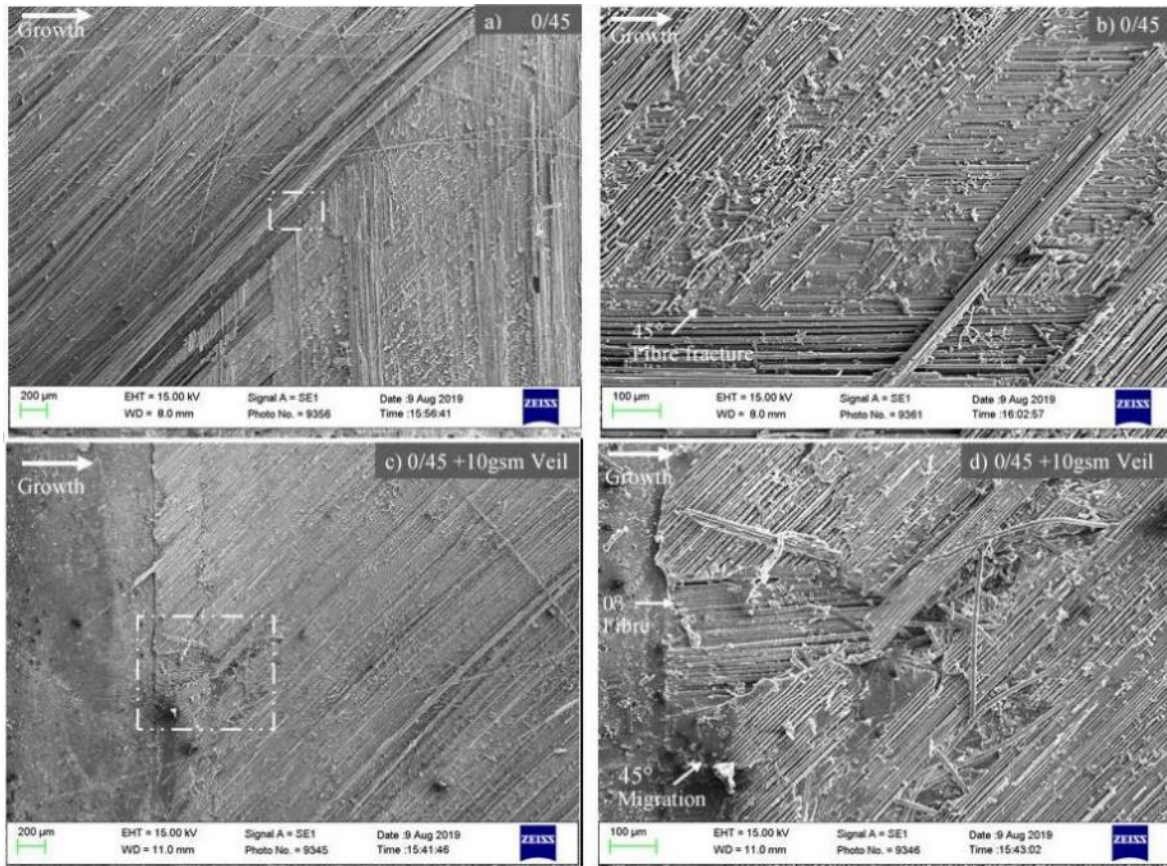


Fig.41 Fracture surface of 0/45 midplane specimens: (a-b) without veil; (c-d) with 10gm⁻² veil.

3.1.6 Discussion

The process of delamination growth of UD CFRP laminates in a multi-directional layup has been affected by many factors; among which the interfacial fibre orientation has a major influence on fracture migration and subsequent interactions between interlaminar and intra-laminar crack growth. Apart from the 0/0 specimens, where the interlaminar delamination grows along the interface at the midplane, the remaining four fibre orientations exhibit complex fracture morphologies and stick-slip behaviours due to a combination of localized intra-laminar, inter-laminar, localised mixed-mode and fracture migration mechanisms (Fig.35 & Fig.39-Fig.41). For example, both non-interleaved and interleaved 90/90 specimens exhibit a coarse delamination path, where bundles of fibres act in hindrance to the delamination growth. The delamination front deviated around these bundles, in the overall propagation direction. The front of the delamination during the propagation phase is subjected to opening mode only.

A 3D Finite Element Model (FEM) has been implemented to quantify the mixed-mode contributions arising from the off-axis fibre orientation. The implemented FEM is shown in Fig.42. The layup was defined using ANSYS Composite PrepPost (ACP) by assigning the orientation and material properties to each layer and using the Virtual Crack Closure Technique (VCCT) to model the delamination with interface elements. VCCT calculates energy-release rate with the assumption that the energy needed to separate a surface is the same as the energy needed to close the same surface.

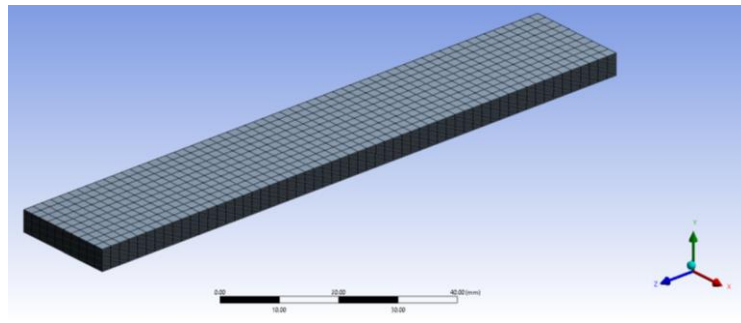


Fig.42 Finite Element Model of the DCB specimen.

The model was used to simulate each of the tested interfaces considering the initial delamination length of 25 mm. For the interfaces with a 45° ply orientation, two additional delamination lengths of 50 mm and 75mm were considered to assess if the increased length would increase the twisting effect. The deformed configurations corresponding to the layup 90/45 for the aforementioned delamination length values are shown in Fig.43.

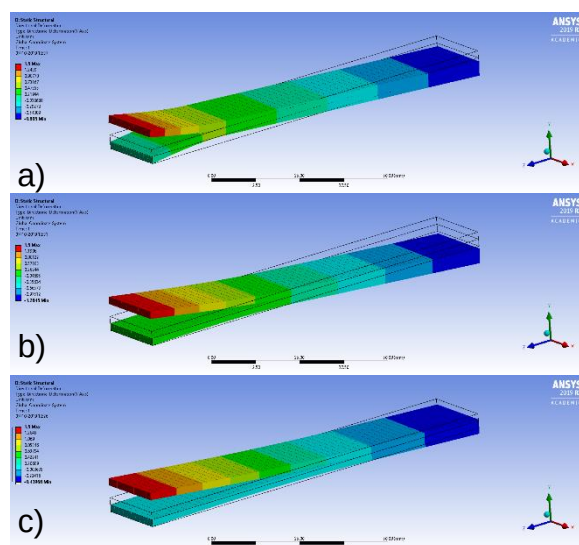


Fig.43 Contour plot of the displacement for three delamination lengths: a) 25 mm, b) 50 mm and c) 75 mm.

The average values of the energy release rate G for mode II and mode III were normalised by the average value of the mode I for the same crack length. The values computed with the FEM are reported in Table.7. The maximum value of the ratio in Table.7 is 2.96%, indicating minimal effects of the 45° orientation layers on the energy release rate and mode I energy absorption being dominant.

Table.7 Ratio of the Energy Release Rates for three delamination length values.

Midplane	25 mm		50 mm		75 mm	
	G_{II}/G_I	G_{III}/G_I	G_{II}/G_I	G_{III}/G_I	G_{II}/G_I	G_{III}/G_I
0/0	5.55E-07%	2.30E-09%	//	//	//	//
90/90	8.42E-07%	8.80E-11%	//	//	//	//
0/90	0.57%	5.52E-04%	//	//	//	//
90/45	2.00%	-0.01%	2.10%	-0.01%	2.18%	-0.01%
0/45	2.96%	0.03%	1.99%	0.24%	1.96%	0.31%

For the 0/90 sample, fracture migration took place after delamination initiation at the 0/90 interface and migrated into the 90° layer, then proceeded as dominantly intra-laminar failure. The specimen with the 45° layer at the midplane often exhibited strong interactions between intra-laminar delamination and interlaminar delamination, as shown in Fig.35 & Fig.40-Fig.41.

As presented in Table.5 - Table.6, the addition of a veil interlayer increased the fracture propagation resistance of the laminate in general. The improvement on the effective fracture resistance is, however, fibre orientation-dependent, varying from 48.6% for the 0/0 specimen to 1.7% for the 90/90 specimen. The true G_{IC} fracture toughness value for the 0/90, 90/45 and 0/45 specimens cannot be concluded as the delamination was made up of intra-laminar and migration regions (Fig.35 and Fig.39-Fig.40). The deviation in cracks occur as a result of localised mixed-mode, and are not truly mode I. For true mode I, single cracks must be dominant, however by varying the interface fibre orientation, localised multiple split or bifurcating cracks can be found to propagate in the general growth direction. As a result, the global absorbed energy results in Table.3 - Table.4 demonstrate an alternative approach to quantifying the behaviour when migration is present in multidirectional laminates.

The G_{IC_prop} results in Table.5 - Table.6 are obtained using test data in regions where delamination was between the pre-cracked planes (between 25 mm and 40 mm), subsequently averaged across repetitions of the same specimen configurations. These average results may not reflect the true fracture resistance for the interface of the given fibre orientation due to fracture migration but provide a good indication regarding the influence of interfacial fibre orientation on the effective propagation fracture resistance. It must be noted that for true two-dimensional consideration, the 0/45 and 90/45 interface will have similar growth behaviour, for example, under impact scenarios.

Based on the results of the current work, continued testing of non-interleaved and interleaved laminates in the prescribed 0/0 orientation as per ASTM 5528-13 [53] shows little benefit in understanding the complex behaviour of interlaminar and intra-laminar delamination processes. Fracture toughness measured from the 0/0 specimen is highly conservative according to the results in Table.5 - Table.6. ASTM 5528-13 [53] is also overly conservative by suggesting the use of 0° dominated specimens for G_{IC} values. From the results shown herein, the ASTM standard is indeed providing a conservative approach in suggesting the layup configuration of 0° plies for coupon testing for engineering applications.

However, it is worth noting that the effective fracture toughness for interfacial fibre orientations other than 0/0 are significantly higher. Therefore, it is crucial to understand the fracture behaviour for different fibre orientations to explore the full potential of real components made of multi-directional CFRP laminates. For accurate numerical simulation of delamination, a highly detailed ply-by-ply model capable of capturing the delamination migration will be needed. However, this is not computationally efficient when modelling large composite structures. In this instance, it may be more practical to use effective-toughness parameters produced here to 'simulate' the delamination growth within a stack of plies. Therefore, the current results have close industrial relevance in design optimisation. Knowledge developed for the influence of interfacial fibre orientation combined with interleaved veil on the effective fracture toughness will lead to reliable design and simulation tools for the prediction of mechanical performance of hybrid laminated composite structures.

3.1.7 Conclusions

Experimental investigations have been conducted on the effect of interfacial fibre orientation and thermoplastic veil interleaves on the fracture toughness of UD CFRP laminates under mode I loading. A summary of conclusions can be collated from the results of the study as follows:

- Interfacial fibre orientation has a substantial influence on resistance to delamination initiation and propagation of UD CFRP laminates. Specimens with midplane fibre orientations other than 0/0 exhibit further load carrying capacity beyond the initial elastic deformation and have significantly higher effective fracture toughness than specimens with the 0/0 midplane fibre orientation.
- G_{IC} derived from the specimen with 0/0 midplane fibre orientation as per ASTM 5528-13 leads to conservative damage tolerance design and analysis of multi-directional CFRP laminates. Effective resistance parameters for corresponding interfacial fibre orientations should be used for more accurate and efficient design optimizations and numerical modelling of laminated components.
- PPS veils at the interface improve the fracture resistance; however, the improvement is fibre orientation-dependent. Toughening using veils on the G_{IC} can be over-projected if based on 0/0 midplane, therefore incorrectly prescribed for multi-directional laminates.
- Fracture surface morphology shows associations of the effective resistance with interfacial fibre orientation and interleaved veils. Dominant migration tendencies were found for 90/90, 0/90, 90/45, and 0/45 mid-plane orientations.

The research on the combined effect of alternative layups and interleaved veils on fracture resistance of UD CFRP laminates provide detailed insights into the behaviours and failure mechanisms of inter-laminar, intra-laminar and fracture plane migration of laminates. Details of failure mechanisms and their interactions must be further understood and employed in design and simulation processes to optimise the performance of laminated composite structures.

REFERENCES

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3.2 Delamination Migration in Multidirectional CFRP Laminates under Mode I Loading.

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Abstract: This paper focuses on the effect of interfacial fibre orientation and interleaved veil on the delamination migration of carbon fibre reinforced polymer laminates under mode I loading. Double cantilever beam specimens with midplane interfacial fibre orientations of 0/0, 90/90, 0/90, 0/45 and 90/45 were tested under two conditions: one with interleaved thermoplastic polyphenylene sulfide veil at the midplane and one without. Results show that, except for the 0/0 configuration, all other orientations exhibit varying levels of migration associated with the interfacial fibre orientation and veil interleaving. The apparent fracture toughness determined with the modified compliance calibration method is closely related to the delamination migration and hence a structural energy dissipation measure dependent on interfacial fibre orientation and the interleaved veil. Distributions of the fibre and matrix materials around the delamination front are found to be closely related to the delamination migration behaviour along its path. The experimental observation and rationalisation presented in this work provide further knowledge regarding delamination migration and its correlation to the apparent fracture toughness, which is of direct relevance to the damage tolerance design of laminated composite components.

Keywords: Delamination – B; Fracture toughness – B; Damage tolerance – C; Fibre bridging – C; Fractography – D

3.2.1 Introduction

Delamination is often associated with the lack of through-thickness reinforcement and stiffness mismatch between the adjoining layers of a composite laminate. It poses direct threats to the load-carrying capacity and service life of laminated composite components. Various methods have been developed to characterize delamination under cyclic or quasi-static loading using linear elastic fracture mechanics theory. The majority of these methods use pre-cracked laminate specimens that grow delamination at a single interface with measurements of the critical strain energy release rate as the fracture toughness of the laminate. This

approach has led to several test standards and significant advances towards damage tolerance design and analysis of laminated composite structures [98,145].

However, practical laminated composite components have multidirectional layup configurations and often involve multisite and interacting damages. Multidirectional carbon fibre reinforced polymer (CFRP) laminates typically exhibit multiple delamination cracks at several interfaces under fatigue load or low-velocity impact. Instead of propagating at its initial interface, the delamination may grow through the thickness, joining up with delamination at a neighbouring interface or kinking out of the original interface into a neighbouring interface on its own.

The transition of delamination growth between different interfaces of a composite laminate is known as delamination migration and is commonly observed in notched or bonded composite components under low-velocity impact, fatigue or thermal loading. It is closely associated with the damage tolerance capability of laminated composite structures and hence triggers a series of research activities to understand the governing mechanisms of delamination migrations under various material and loading conditions [294–297]. A test configuration using Single Cantilever Beams (SCB) has been developed by Ratcliff *et al.* [6] for controlling width-wise stable migrations. Under the SCB configuration, as shown in Fig.44, the loading and stress conditions change for materials around the delamination front depending on the relative position of the loading point to the delamination front. Controlled migrations occur along the principal plane where the maximum tensile stress is acting on.

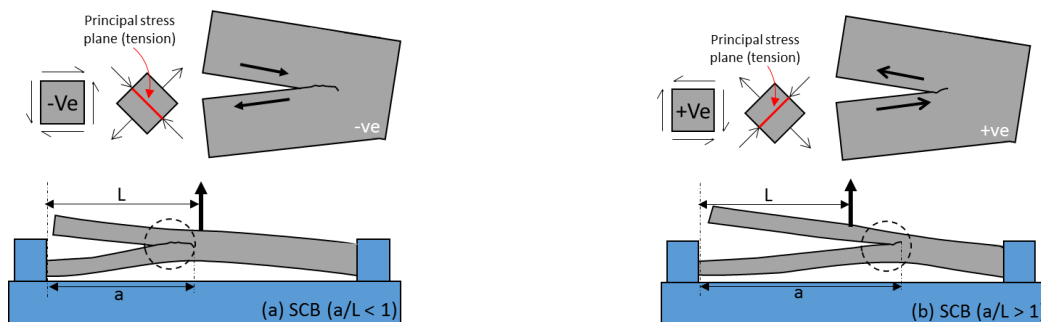


Fig.44 Stress conditions around the delamination front of the SCB: (a) loading point ahead of the delamination front, (b) loading point behind the delamination front.

Ratcliff *et al.* [6] demonstrated through the SCB testing of cross-ply CFRP laminates that delamination at the 0/90 interface migrated to a neighbouring “90/0 interface by kinking through the 90° ply stack. The transition from the delamination at the 0/90 interface into the 90° ply stack was gradual whilst the transition of the kinked crack into the 90/0 interface was sudden”. These migration behaviours provide valuable experimental evidence in validating analytical and numerical models aimed at predicting delamination migration. Subsequent research [78,102,141,143,292,293,298–300] have further advanced the knowledge of delamination migration.

It is, however, worth noting that delamination migration in multidirectional laminates involves complex interactions of the delamination front with the surrounding fibre and matrix materials. Most research in the literature focus on the studies of delamination migration of cross-ply laminates under the SCB test configuration; where loading conditions may not be accurate representations of practical laminated components. Detailed investigation on the influence of a wide range of interfacial fibre orientations on delamination migration is minimal. Moreover, while substantial efforts have been made to characterise the toughening effect of interleaved veils for composite laminates [20–23,25,26,96,223], little has been done to understand the effect of interleaved veils on delamination migration, especially for the combined effect of interfacial fibre orientation and interleaving. The present work aims to fill the gap through detailed experimental investigations of double cantilever beam (DCB) specimens with interfacial fibre orientations of 0/0, 90/90, 0/90, 0/45 and 90/45 under two conditions: one with polyphenylene sulfide (PPS) veil interleave at the midplane and one without. Combined effects of interfacial fibre orientation and PPS veil on delamination migration and apparent fracture toughness have been investigated to develop further knowledge for effective damage tolerance design of laminated composite components under mode I loading.

3.2.2 Materials and Experimental Procedures

The pre-impregnated CFRP material was supplied by SHD Composites, consisting of UD carbon T300 fibres, with MTC510 toughened epoxy system of 37% resin weight [290]. It has an areal density of 150gm^{-2} and nominal ply cure thickness of 0.143mm with relevant mechanical properties listed in Table.8. Each non-interleaved sample for a specific midplane fibre orientation had a veil interleaved counterpart for comparison. The veils supplied by Technical Fibre Products are Optiveil OP-49-48 [280], with a nominal areal density of 10gm^{-2} and thickness of 0.09mm. Veils were manufactured using a wet-lay process with chopped strand PPS fibres of $10\mu\text{m}$ diameter and length of 6mm in a styrene-acrylic binder.

Table.8 Mechanical properties of MTC510--UD150-HS-37%RW [290].

Property	Result	Standard
0° Tensile strength [MPa]	2112	BS EN ISO 527-5
0° Tensile modulus [GPa]	110	
0° Poisson's ratio	0.34	
90° Tensile strength [MPa]	54	
90° Tensile modulus [GPa]	8.2	
90° Poisson's ratio	0.01	
0° Compression strength [MPa]	988	EN 2850 Type B
0° Compression modulus [GPa]	105.1	
90° Compression strength [MPa]	200	
90° Compression modulus [GPa]	9.3	
In-Plane shear strength $\pm 45^\circ$ [MPa]	99	ASTM D3518
In-Plane shear modulus $\pm 45^\circ$ [GPa]	3.6	
Interlaminar shear strength 0°[MPa]	85	BS EN ISO 14130

Five specimens were fabricated under the same manufacturing conditions, providing four repetitions for each layup configuration listed in Table.9. The DCB specimens were rectangular with the dimensions of 140(L)×20(B)×4.2mm(H). For the interleaved configurations, a layer of veil was placed at the midplane, adjacent to the PTFE film insert. The twenty-eight UD CFRP layers were cold consolidated under vacuum during the layup and subsequently placed on a flat steel plate with release film, breather fabric and cured within a vacuum bag in the autoclave. The panels were cured in the autoclave under pressure at 620 KPa (90 psi), with a target bag vacuum of 29" Hg (gauge) and temperature set to 100°C for 4 hours as prescribed for MTC510 [290].

The apparent fracture toughness of the specimen was assessed with the modified compliance calibration method as specified in ASTM D5528-13 [53]. The deviation predominantly being the midplane interface configuration, which is required for the study of the effect of fibre orientation and veils on the strain energy release rate and migration mechanisms. Unlike the 0/0 fibre orientation, five interfacial fibre orientations, as shown in Table.9 were introduced at the midplane of the DCB specimen.

Table.9 Layup configurations for DCB specimens.

Specimen Code	Layup Configuration
90/90	$[(0, 90)_7]_s$
90/90 +veil	$[(0, 90)_7, \frac{\overline{veil}}{2}]_s$
0/0	$[(90, 0)_7]_s$
0/0 +veil	$[(90, 0)_7, \frac{\overline{veil}}{2}]_s$
90/0	$[0, 90]_{14}$
90/0 +veil	$[(0, 90)_7, \overline{veil}, (0/90)_7]$
90/45	$[(0, 90)_6, (-45, 90, 45, 0), (90, 0)_6]$
90/45 +veil	$[(0, 90)_6, (-45, 90, \overline{veil}, 45, 0), (90, 0)_6]$
0/45	$[(90, 0)_6, (-45, 0, 45, 90), (0, 90)_6]$
0/45 +veil	$[(90, 0)_6, (-45, 0, \overline{veil}, 45, 90), (0, 90)_6]$

The specimens were clamped via cyanoacrylate bonded hinges into vice grip fixtures and axially aligned with the load path. A printed scale bar affixed to the specimen allowed visual observation of the delamination front on a white painted edge. The tests were conducted using a 5kN dual-column servo-electric universal test machine (Instron 5965). The crosshead displacement rate was set to a constant 2mm/min. The test process was captured on video using a 150mm macro lens and tripod-mounted DSLR camera system shown in Fig.45.

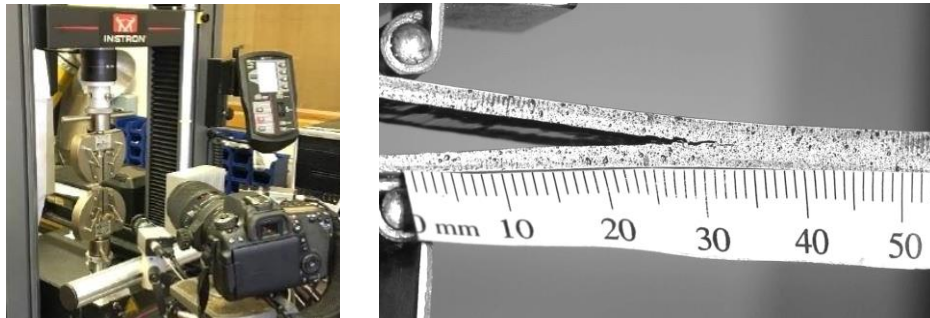


Fig.45 (a) Test setup, (b) Loaded specimen.

Fig.45(a) displays the test set-up with Dantec 400 DIC and Fig.45(b) displays the speckle pattern on the edge of the specimen used for the digital image correlation (DIC) analysis. The onset of delamination and subsequent growth were identified by cross-referencing the video recording with the loading rate and the recording time. The resolution of the images is within $\pm 0.5\text{mm}$, which is accurate enough to deduce the crack tip according to the ASTM standard [53]. The onset of delamination and subsequent growth have been verified by correlating these events to the abrupt load-drops of the load-displacement histories of the specimens in Fig.46.

The specimens were continuously loaded until sufficient data points were obtained or specimen failure. Delamination length measurements were obtained by correlating the video recording with the constant displacement rate. The mode I strain energy release rates, G_I , were calculated using the modified compliance calibration method [53]:

$$G_I = \frac{3P^2C^{2/3}}{2A_1bh} \quad \text{Eqn.15}$$

where P is the applied load. C is the compliance determined by the ratio of the load point displacement to the applied load, δ/P . A_1 is the slope of the least-squares plot of delamination length normalized by specimen thickness, a/h , as a function of the cube root of compliance, $C^{1/3}$. b is the specimen width, and h is the specimen thickness.

3.2.3 Results

3.2.3.1 Delamination migration and load-displacement curves

Fig.46 shows the representative load-displacement curves of DCB specimens with five interfacial fibre orientations. Effect of delamination migration is reflected by the change in the local slope of the load-displacement curve associated with the intermittent delamination deflection and growth in the migration process. This is supported by the noticeable fluctuations of the load-displacement curves in Fig.46 for all configurations except the 0/0 specimen. Specimens with the 0/0 interfacial fibre orientation do not exhibit significant load fluctuation as the delamination growth is continuous in the absence of delamination migration. The

small load fluctuation for the 0/0 specimen is related to localised fibre pull-out, fibre breakage and delamination propagation along the 0/0 interface. Further evidence can be found in the fractographic examination in section 3.2.5.

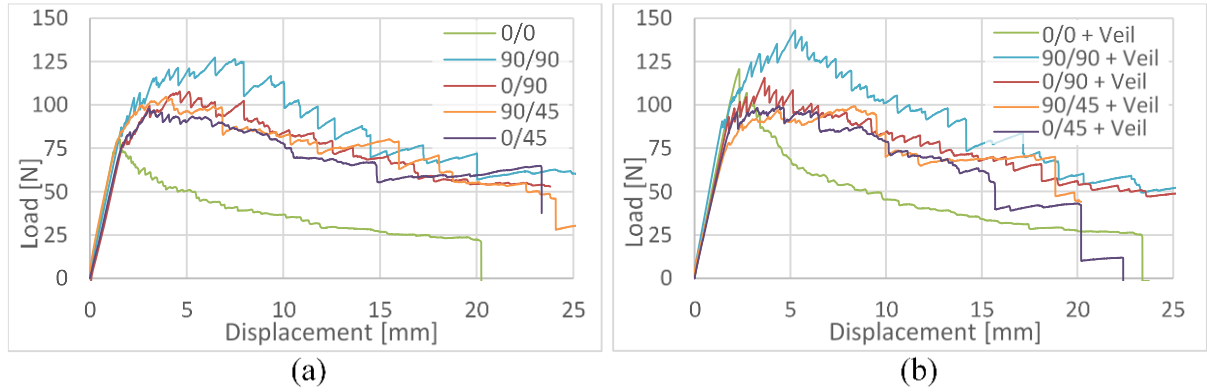


Fig.46 DCB load-displacement curves: (a) without veils, (b) with interleaved veils.

The load-displacement behaviour of specimens with the veil interleave is shown in Fig.46(b). In comparison to the curves in Fig.46(a), the veil interleave generally improves the resistance to delamination growth. The beneficial effect of the veil is, however, dependent on the fibre orientation at the midplane. Specimens with the 0/0 configuration showed the most benefit from the veil, while other midplane orientations appear to receive varying levels of enhancements. Levels of load fluctuations in interleaved specimens are similar to those of specimens without the veil.

3.2.4 Delamination migration and apparent fracture toughness

Fig.47 shows the resistance curves (R-curves) derived from the load-displacement curves of repeated DCB tests using the modified compliance calibration method detailed in [53]. The intermittent nature of the delamination migration leads to large variations in the calculated strain energy release rate (G_I) for specimens of 90/90, 0/90, 90/45 and 0/45 interfacial fibre orientations.

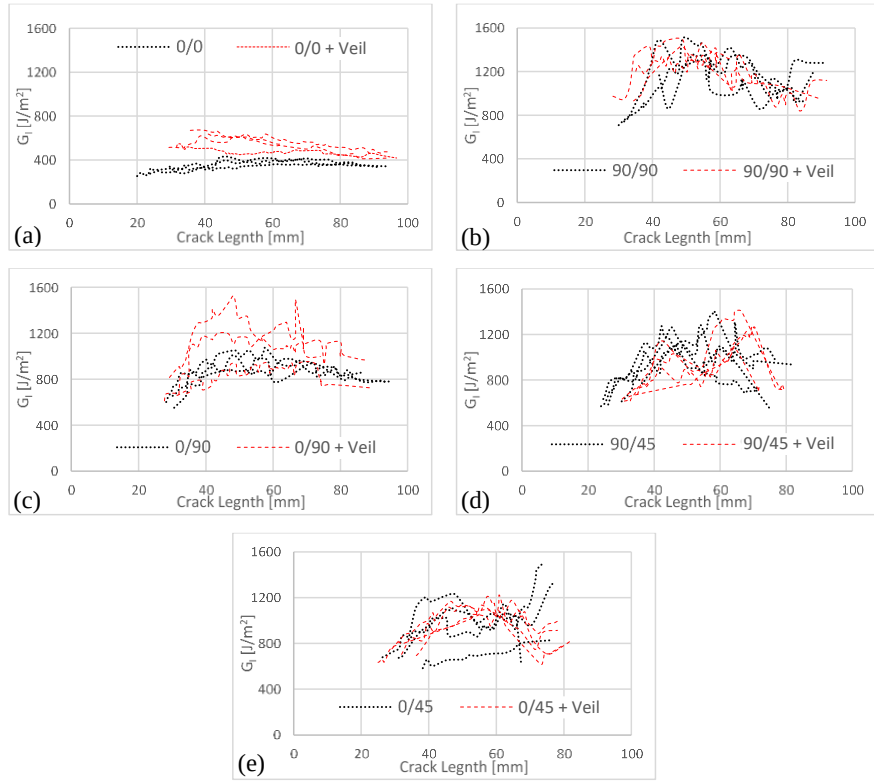


Fig.47 R-curves with interfacial fibre orientations of: (a) 0/0, (b) 90/90, (c) 0/90, (d) 90/45, and (e) 0/45.

The 0/0 specimen results (Fig.47(a)) are grouped closely with minimum variation, indicating continuous delamination growth along the 0/0 interface with no delamination migration. The PPS veil has an apparent beneficial effect on the fracture resistance at small delamination length (<80mm). The R-curves for specimens with other four midplane configurations (Fig.47(b)-(e)) show varying levels of enhancement of fracture resistance, which is related to varying levels of migrations to be detailed in section 3.2.5. For the 90/90 configuration (Fig.47(b)), although the R-curves show a small increase in fracture resistance with PPS interleaves, similar large variations of G_I for both veiled and unveiled specimens indicate that delamination migration occurred and the delamination was governed by the same mechanism. R-curves for 0/90 configuration (Fig.47(c)) exhibited noticeable improvement of the veiled specimen over unveiled specimen on fracture resistance at small delamination length. R-curves for 90/45 (Fig.47(d)) configuration shows no clear distinction between unveiled and veiled specimens, indicating similar resistance to delamination growth. This is associated with immediate delamination migration at the front of the PTFE film during the test. Instead of propagating along the 90/45 midplane where the veil was added, the

initial delamination created by the PTFE film insert tends to migrate immediately into either the 90° layer above or the 45° layer below the midplane interface. This triggers large variations of the fracture resistance shown in Fig.47(d) for this configuration, which is directly related to the delamination migration behaviour to be detailed in section 3.2.5.4. The veil, therefore, has a negligible effect on the R-curves of veiled and unveiled 90/45 specimens. R-curves for 0/45 (Fig.47(e)) configurations show a small increase in fracture resistance with PPS interleaves. Section 3.2.5.4 shows the delamination migration behaviour of 0/45 configuration.

Table.10 summarises test results of delamination resistance of the DCB specimens in terms of strain energy release rate G_I . Considering large variations in calculated strain energy release rate due to irregular delamination migration; mean values of repeated tests over the delamination length between 30mm and 60mm were presented to facilitate the assessment of the overall effect of interfacial fibre orientation and PPS veil on delamination resistance. Results show interfacial fibre orientation has significant effects on delamination resistance with the 0/0 interface having the lowest toughness. The delamination resistance at the 90/90 interface is 3.2 times of that at the 0/0 interface for specimens without the veil.

Thermoplastic veils improve the delamination resistance in general, but its improvement is fibre orientation-dependent with the 0/0 interface showing the greatest improvement of 41%. Delamination resistance at the 90/90 interface is 2.4 times greater than of the 0/0 interface with the veil. It is, however, worth noting that only the 0/0 delamination resistance reported in Table.10 represents the fracture toughness G_{IC} for the 0/0 interface. The delamination resistance values for 90/90, 0/90, 90/45 and 0/45 interfaces are not the pure mode I fracture toughness G_{IC} for these interfaces due to delamination migration. They represent apparent fracture toughness for delamination at these interfaces over the length of 30 – 60mm. The fact that delamination avoids these interfaces indicates that the true G_{IC} values for these interfaces are higher than the apparent fracture toughness shown in Table.10.

Table.10 Delamination resistance of DCB specimens.

Midplane	SERR (No veil) [J/m ²]		SERR (With veil) [J/m ²]		Effect of veil
	Mean	StdDev	Mean	StdDev	
0/0	369	62	522	63	+41%
90/90*	1164	163	1236	141	+6%
0/90*	843	236	974	145	+15%
90/45*	801	184	810	199	+1%
0/45*	746	184	792	182	+6%

*Apparent fracture toughness due to delamination migration.

3.2.5 Mechanisms of delamination migration in CFRP laminates

It is expected that delamination migration is influenced by the local stress field around the delamination front. The interfacial fibre orientation and the veil interleave will affect the local stress field and hence the migration behaviour. Upon completion of the DCB tests, optical microscopy was used to characterise the delamination path and fracture surfaces. The fractographic information is combined with the analysis of the local distribution of fibre reinforcement and epoxy matrix materials around the delamination front to understand the mechanism for delamination migration in multidirectional CFRP laminates.

3.2.5.1 Delamination path for DCB specimen with 0/0 midplane

Fig.48(a) is a representative side view of the delamination growth path of the DCB specimen with the 0/0 interfacial fibre orientation at the midplane. Stable delamination growth along the 0/0 interface was observed for both the interleaved and non-interleaved specimens, which is expected as the 0° fibre at the interface prevents delamination migration. There is, however, a distinct difference in the morphology of the fracture surfaces between the interleaved and non-interleaved specimens. Compared with the non-interleaved specimen (Fig.48(b)), the veil-interleaved specimen (Fig.48(c)) exhibited a coarser fracture surface with increased fibre pull-out and bridging. These fractographic findings are consistent with the test results in sections 3.1 & 3.2 showing noticeable benefits of PPS veil on delamination resistance. The fracture toughness of the interleaved specimen is 41% higher than that of the non-interleaved specimen (Table.10). The force-displacement curve in Fig.46(a) exhibited an elastic response up to 80 N for non-interleaved specimens, whereas veil interleaved 0/0 specimens had a peak force of 120 N as shown in Fig.46(b).

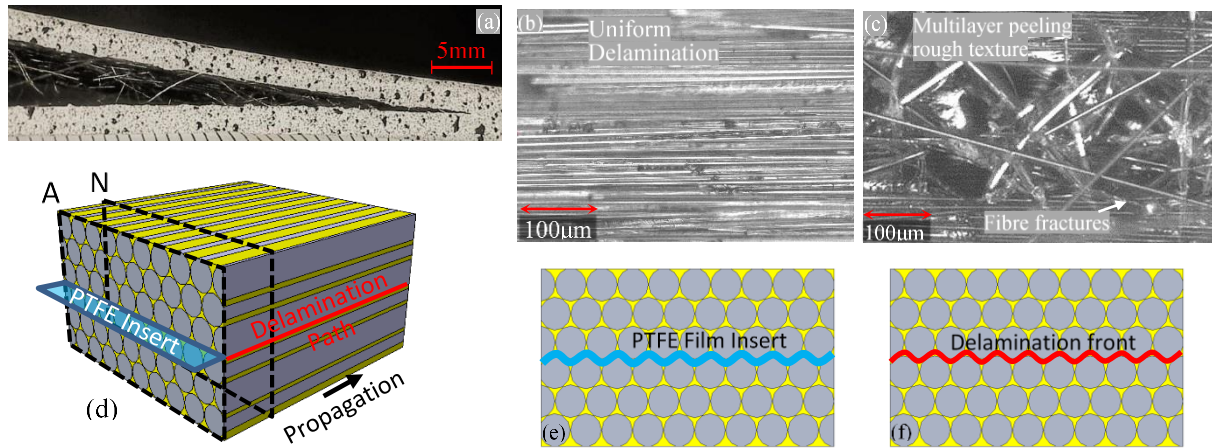


Fig.48 Delamination path for 0/0: (a) side view, (b) fracture surface without veil, (c) fracture surface with veil, (d) 3D illustration of delamination path, (e) A-A section view, (f) N-N section view.

Fig.48(d) is the 3D illustration for the distribution of fibre reinforcement (grey), and epoxy matrix material (yellow) along the delamination path (red) of the 0/0 DCB specimen. Delamination migration from the 0/0 midplane interface is prevented due to the existence of reinforcement fibres in the propagation direction. Fig.48(e) is the section view showing the width-wise delamination profile and the distribution of the fibre and matrix materials at the plane where the PTFE insert finishes. The actual waviness of the delamination profile is very small considering the diameter (5 – 10 µm) of the carbon fibre and the delamination growth is confined within a narrow region on the face in Fig.48(d)). Fig.48(f) presents the section view after the delamination has propagated along the 0/0 interface for some distance. The delamination profile and distribution of the fibre and matrix materials are the same in Fig.48(e & f), demonstrating the same resistance during delamination growth and rationalising the stable and continuous delamination growth behaviour observed in the 0/0 DCB tests (Fig.46). While multiple delamination damages are commonly observed at separate interfaces above and below the midplane, delamination migration and multi-layer delamination coalescence through the thickness of the laminate cannot occur for 0/0 interleaved and non-interleaved specimens. The dominant failure mode is an interlaminar failure along the 0/0 interface for 0/0 specimens.

3.2.5.2 Delamination path for DCB specimen with 90/90 midplane

Fig.49(a) is a representative side view of the delamination growth path of the DCB specimen with the 90/90 interfacial fibre orientation at the midplane. Unstable and undulating delamination growth was observed for both the interleaved and non-interleaved specimens, exhibiting delamination migration tendencies through the thickness of the 90/90 DCB specimen. Localised fibre fracture was observed as the 90° bundles were ‘pulled-out’ from the laminae matrix and remained partially bonded to each surface. The force-displacement curve in Fig.46 exhibited an elastic response up to 127 N for non-interleaved 90/90 specimens whereas veil interleaved 90/90 specimens had a peak force of 143 N. The 6% increase in apparent fracture toughness shown in Table.10 for veil-interleaved 90/90 specimens can be attributed to the enhanced fibre pull-out and bridging of the veiled interleaved specimen (Fig.49(c)) compared with the non-interleaved specimen (Fig.49(b)).

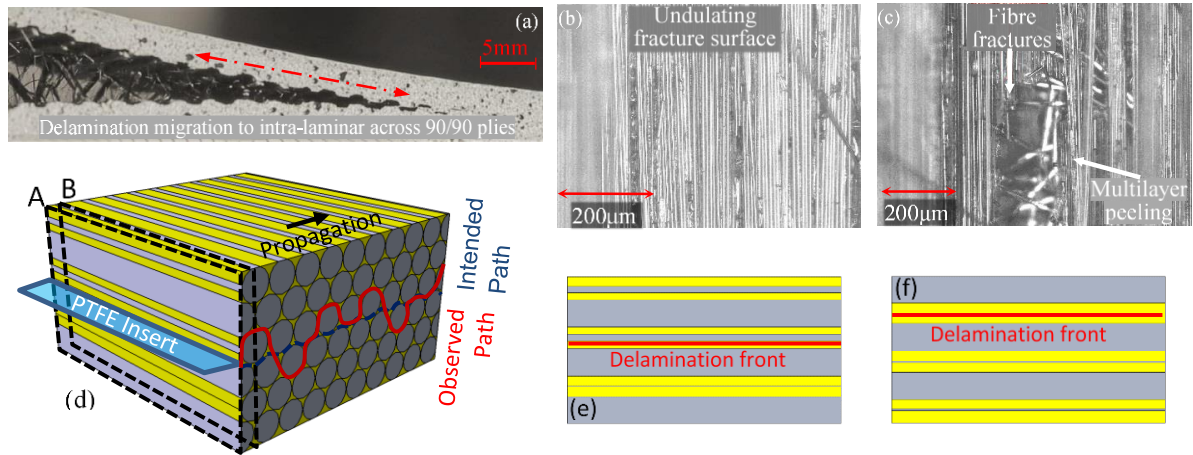


Fig.49 Delamination path for 90/90: (a) side view, (b) fracture surface without veil, (c) fracture surface with veil, (d) 3D illustration of delamination path, (e) A-A section view, (f) B-B section view.

Fig.49(d) presents the 3D illustration for the distribution of fibre reinforcement and matrix materials along the delamination path of the 90/90 DCB specimen. Fig.49(e) illustrates the sectional view showing the through-thickness distribution of the fibre and matrix materials at the A-A plane. Fig.49(f) depicts the section view after the delamination has propagated to the B-B plane. There is a distinct difference in local distribution of fibre and matrix materials between planes A-A and B-B, affecting the local stress field around the delamination front and hence

the delamination resistance (Fig.46). It is expected that the resin-rich regions are prone to the initiation and coalescence of micro matrix cracks and hence trigger delamination deflection and migration. The migration path in Fig.49(d) is determined based on the analysis of the variations of the resin-rich regions around the delamination front and test observations (Fig.49(a)). Migration from the 90/90 midplane is frequently accompanied with direction changes, reverting towards the midplane where the maximum applied stress exists under mode I loading. As a result, the delamination will propagate along the observed undulating red path, not along the ideal blue path at the 90/90 midplane interface shown in Fig.49(d). The delamination migration introduces mixed-mode delamination growth and reduces the effective driving-force due to delamination deflection. The delamination resistance measured from the 90/90 DCB specimen is, therefore, the apparent fracture toughness and is considerably higher than that of the 0/0 DCB specimen (Table.10). The dominant failure mechanism for 90/90 interleaved and non-interleaved DCB specimens is the intra-laminar failure of the 90° laminae.

3.2.5.3 Delamination path for DCB specimen with 0/90 midplane

Fig.50(a) is a representative side view of the delamination growth path of the DCB specimen with the 0/90 interfacial fibre orientation at the midplane. Similar to the observation of 90/90 specimens, delamination growth is unstable, but the level of undulation is less significant in 0/90 as the delamination strays within a single 90° lamina. It was observed across the repeated tests that the delamination migrated into the 90° layer after the PTFE insert, demonstrating that the neighbouring 90° intra-laminar interface was weaker than the 0/90 interface and the measured delamination resistance in Table.10 is an apparent fracture toughness of the 0/90 DCB specimen. The force-displacement curve exhibited an elastic response up to 105 N for non-interleaved 0/90 specimens (Fig.46(a)), whereas veil interleaved 0/90 specimens had a peak elastic load limit of 115 N (Fig.46(b)). The fracture surface for the interleaved midplane (Fig.50(c)) was coarser and irregular compared to the non-interleaved midplane (Fig.50(b)). The difference in the elastic load limit and fracture surface morphology between non-interleaved and interleaved specimens contribute to 15% improvement in apparent fracture toughness (Table.10).

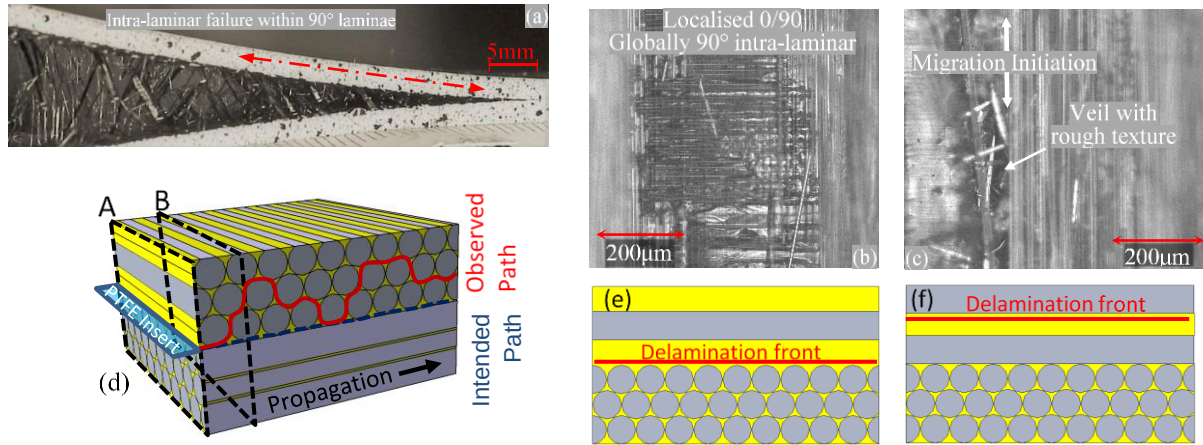


Fig.50 Delamination path for 0/90: (a) side view, (b) fracture surface without veil, (c) fracture surface with veil, (d) 3D illustration of delamination path, (e) A-A section view, (f) B-B section view.

Fig.50(d) depicts a 3D illustration for the distribution of fibre reinforcement and matrix materials along the delamination path of the 0/90 DCB specimen. Fig.50(e) illustrates a section view with the through-thickness distribution of the fibre and matrix material at the A-A plane. Likewise, Fig.50(f) illustrates the section view once the delamination has propagated to the B-B plane. Similar to the 90/90 interfaces, distinct differences in the local fibre-matrix distribution exists between planes A-A and B-B, indicating differences in the local stress field around the delamination front and hence the resistance to delamination growth (Fig.46). The delamination tends to deflect into the resin-rich area and migrate into the 90° layer, not into the 0° laminae due to the strong 0° fibres at the interface. The delamination resistance measured from the 0/90 DCB specimen is, therefore, the apparent fracture toughness and is much higher than that of the 0/0 DCB specimen (Table.10) due to delamination migration. The dominant failure mechanism for 0/90 interleaved and non-interleaved DCB specimens is the intra-laminar failure of the 90° layer.

3.2.5.4 Delamination path for DCB specimen with 90/45 midplane

Fig.51(a) is a representative side view of the delamination growth path of the DCB specimen with the 90/45 interfacial fibre orientation at the midplane. Delamination propagated mainly within the 45° laminae for both non-interleaved and interleaved specimens, not along the pre-cracked 90/45 interface. Fig.51(a) shows a region (9~10mm) where intra-laminar migration into the 90° laminae is observed. Migration into the 90° laminae occurred after the initial 45° fibres ended at the rear

edge of the specimens. Chen *et al.* [78] and Gong *et al.* [292,300] describes the width-wise delamination being non-uniform due to the ply-splits when considering 45° laminae. Similar large-scale migrations were observed across all repetitions for non-interleaved and veil interleaved 90/45 specimens. Fracture features similar to the 0/90 mid-planes were observed (Fig.51(b-c)), indicating the influence of adjacent laminae on the apparent fracture toughness. The small difference (1%) in the load limit (Fig.46) and apparent fracture toughness (Table.10) for non-interleaved and interleaved 90/45 midplanes shows a negligible effect of the veil due to immediate delamination migration which makes the delamination not to propagate along the pre-cracked 90/45 interface where the veil was inserted.

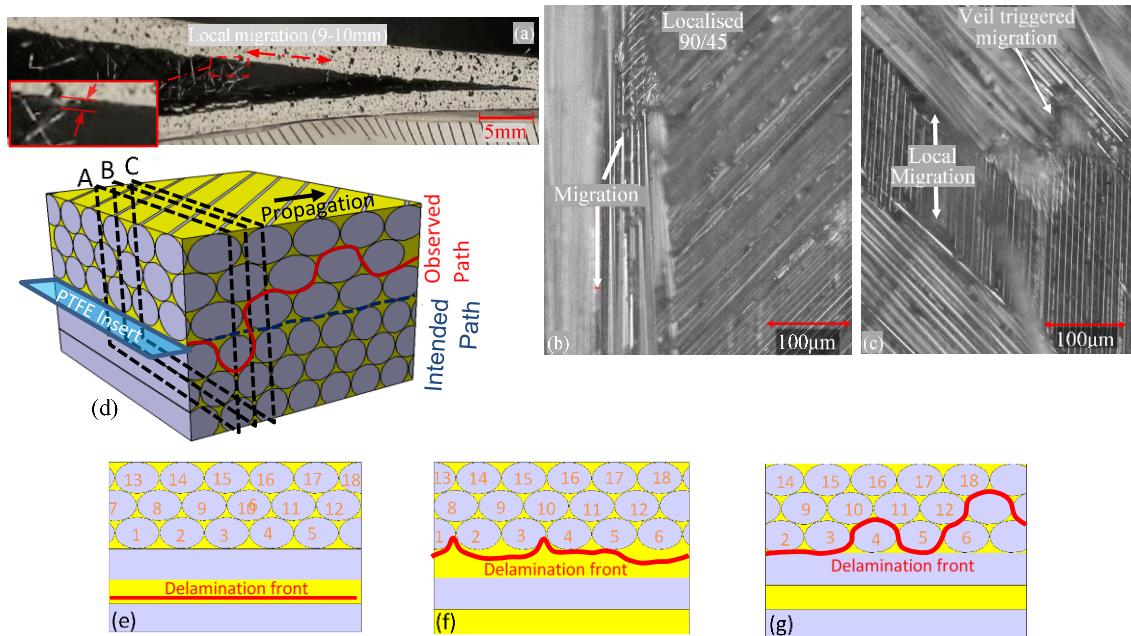


Fig.51 Delamination path for 90/45 specimen: (a) side view, (b) fracture surface without veil, (c) fracture surface with veil, (d) 3D illustration of delamination path, (e) A-A section view, (f) B-B section view, (g) C-C section view.

Fig.51(d) presents a 3D illustration for the distribution of fibre reinforcement and matrix materials along the delamination path of the 90/45 DCB specimen. Fig.51(e-g) are the section views showing the through-thickness distribution of the fibre and matrix material and the profiles of the delamination front at planes of A-A, B-B, and C-C, respectively. The distribution of fibre and matrix materials varies from plane to plane along the delamination path, which triggers delamination migration and affect the resistance to delamination shown in Fig.46 and Table.10. The dominant failure mechanism for 90/45 midplane specimens for interleaved and non-

interleaved was the intra-laminar failure of the 45° and 90° layers. Unlike the 0/0, 90/90, 0/90 specimens, the delamination front across the width of the 90/45 DCB specimen is no longer uniform. Instead, the delamination front was skewed across the width of the specimen as shown in Fig.51(g), which could be linked to the presence of the 45° ply splits in the 90/45 specimen [78,292,300]. It is, however, worth noting that the profile in Fig.51(g) only represents one of the many possible outcomes, factors such as local misalignment of fibres or defects may trigger migrations at different locations and hence lead to different delamination front profiles. Further investigation is required to understand the effect of 45° layer on the profile of the delamination front.

3.2.5.5 Delamination path for DCB specimen with 0/45 midplane

Fig.52(a) is a representative side view of the delamination growth path of the DCB specimen with the 0/45 interfacial fibre orientation at the midplane. It was observed that intra-laminar failure first occurred within the 45° laminae, then progressed through a combination of 45° and 90° plies as intra-laminar failure from ~25mm to ~40mm, and finally propagated through the neighbouring 90° until ~64mm from the hinge pin. The 6% increase in apparent fracture toughness shown in Table.10 for veil-interleaved 0/45 specimens can be attributed to the enhanced fibre bridging, and the veil triggered migration of the interleaved specimen (Fig.52(c)) compared with the non-interleaved specimen (Fig.52(b)).

Fig.52(d) presents a 3D illustration for the distribution of fibre reinforcement and matrix materials along the delamination path of the 0/45 DCB specimen. Fig.52(e-g) are the section views showing the through-thickness distribution of the fibre and matrix materials and the profiles of the delamination front at planes A-A, B-B, and C-C, respectively. Similar to what was observed for 90/45 specimens, the dominant failure mechanism for 0/45 specimens for interleaved and non-interleaved was the intra-laminar failure of the 45° and neighbouring 90° layers. The delamination front of the 0/45 DCB specimen was skewed across the width of the specimen as shown in Fig.52(f-g) due to the presence of 45° fibre splits.

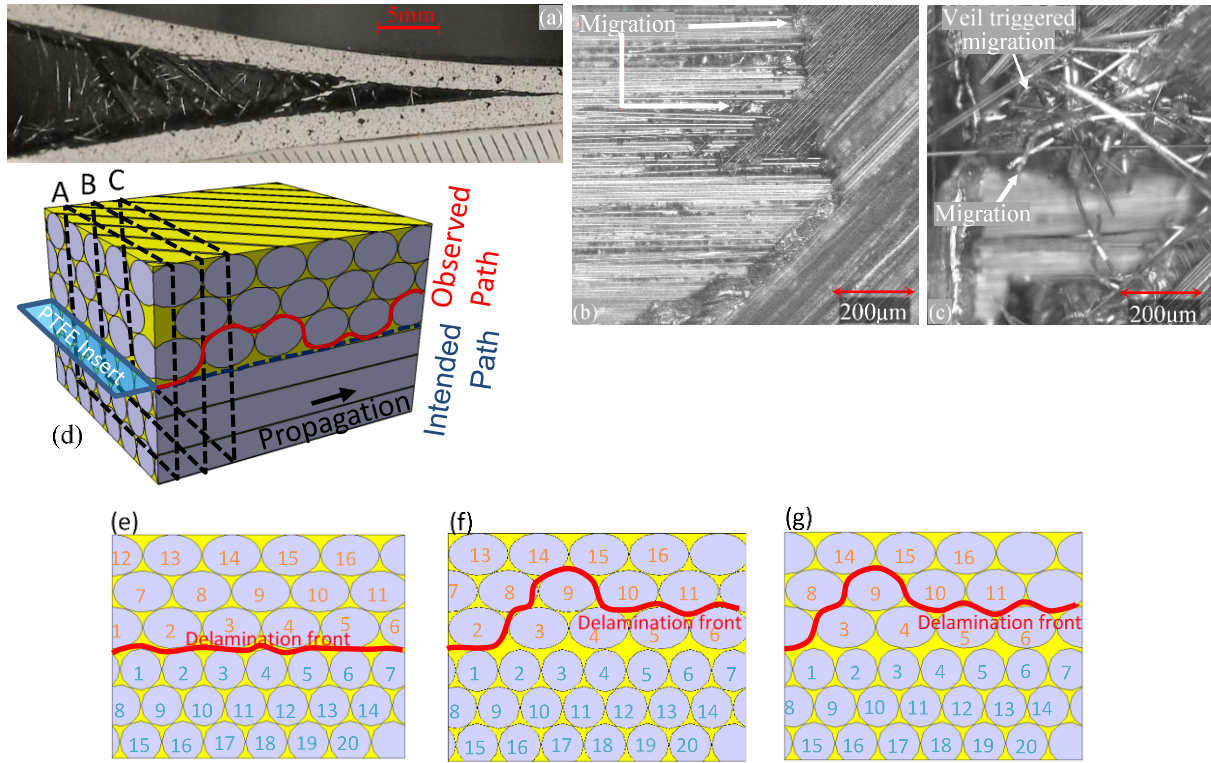


Fig.52 Delamination path for 0/45 specimen: (a) side view, (b) fracture surface without veil, (c) fracture surface with veil, (d) 3D illustration of delamination path, (e) A-A section view, (f) B-B section view, (g) C-C section view.

3.2.6 Discussion

Results of DCB tests show distinct tendencies of delamination migration in multidirectional UD CFRP laminates. Apart from the 0/0 specimens where the delamination propagates along the midplane interface, the remaining four (90/90, 0/90, 90/45, 0/45) specimen configurations exhibit complex fracture morphologies and delamination paths due to delamination migration (Fig.49-Fig.52). The delamination profile is directly related to the interfacial fibre orientation at the midplane. The 90/90 and 0/90 specimens exhibit undulating delamination paths in the lengthwise direction, but the delamination front in the width direction is uniform (Fig.49-Fig.50). The profiles of the delamination front of the 90/45 and 0/45 specimens were however skewed across the width of the specimen as shown in Fig.51-Fig.52, which are related to the presence of ply-splits of the 45° layer [78,292,300]. The bending-twisting coupling due to the inclusion of the 45° layer for the 90/45 and 0/45 configurations is a crucial aspect to check and may add further complexity to the delamination process.

However, it is worth noting that the unsymmetrical layup is necessary to investigate the effect of interfacial fibre orientation on delamination resistance at the 90/45 and 0/45 interfaces in the current study. Moreover, it is expected that the effect of the bending-twisting coupling on the global behaviour of the specimen is insignificant since the four unsymmetrical layers were confined to the midplane and constrained by dominant outside twenty-four symmetric cross-ply layers as shown in Table.9. This is supported by the experimental observation that the rotation in 90/45 and 0/45 specimens was negligible during the DCB testing. Finite element simulation has been developed to quantify the effect of the 45° layer with results showing minimal (<2.2%) mode II and mode III elements during delamination growth for the DCB specimens of 90/45 and 0/45 configurations. Details of the FE model and the numerical results are presented previously (section 3.1.3.1), focusing on the quantification of the bending-twisting coupling effect and the variation of the delamination resistance.

Delamination migration leads to the stick-slip delamination growth behaviour and contributes to the fluctuations of loads in Fig.46 and significant variations of the delamination resistance in Fig.47. The strain energy release rate measured for the specimens of interfacial fibre orientations other than 0/0 are the apparent fracture toughness representing averaged resistance to delamination growth, not the mode I fracture toughness G_{IC} value for the resistance to delamination growth along the interface. The 3D profile of the delamination surface and enhanced fibre bridging and breakage due to delamination migration lead to improved resistance to delamination growth of 90/90, 0/90, 90/45 and 0/45 specimens in comparison to the 0/0 specimen. The apparent fracture toughness of the 90/90 specimen is 3.2 times of that at the 0/0 interface for specimens without the veil. Therefore, ASTM 5528-13 [53] provides a conservative approach in suggesting the use of 0/0 DCB specimen to derive the G_{IC} of the composite laminate for engineering application but can be overly conservative in design optimisation for weight saving. Veil interleaving is considered as an established method for improving the fracture toughness of multidirectional CFRP laminates [16-24]. DCB test results (Table.10) in the current study show that PPS veil led to enhanced fibre bridging and fibre breakage (Fig.48-Fig.52) and hence improved the resistance to delamination growth.

The improvement on the apparent fracture toughness is, however, fibre orientation-dependent, varying from 41% for the 0/0 specimen to 1% for the 90/45 specimen. Considering that most studies on the effect of interleaving have been conducted on 0/0 DCB specimens [16-24], the toughening effect of interleaving veils could be over-projected and incorrectly prescribed for multidirectional CFRP laminates. DCB test results in Table.10 also show that interfacial fibre orientation has a dominant influence on delamination resistance, but the influence is reduced with the addition of the PPS veil. For specimens with the veil interleave, apparent fracture toughness of the 90/90 specimens is 2.4 times that of the 0/0 specimen, which is less than 3.2 times for specimens without the veil. It is also worth noting that the effect of interleaving veil on apparent fracture toughness exists after the delamination has migrated away from the veiled interface, indicating that the PPS veil may affect not only the material at the interface but also the neighbouring materials.

The effect of interfacial fibre orientation and interleaving veil on the apparent fracture toughness is linked to delamination migrations. A good understanding of migration mechanisms is essential yet challenging to exploit the full potential of multidirectional CFRP laminates. This has been illustrated in Fig.49-Fig.52 showing complex delamination migration paths in multidirectional CFRP laminates under the influence of many factors. Ratcliff *et al.* [6-8] argued that delamination migrated along the plane under the principal tensile stress (Fig.44) and demonstrated that the migration path could be predicted by correlating the applied load to the stress field near the delamination front. It is, however, worth noting that the test configuration in the study of Ratcliff *et al.* [6-8] is developed explicitly for controlling width-wise and stable migrations. The loading condition and stress state around the delamination front changes with the relative position of the loading point to the delamination front, as illustrated in Fig.44. Delamination migration behaviours observed in the SCB testing may not be therefore representative of multidirectional CFRP laminates under mode I loading.

Moreover, the correlation between the applied load and the stress field in Fig.44 does not account for highly anisotropic natures of multidirectional CFRP laminates and cannot rationalize the dominant influence of interfacial fibre orientations on delamination migration and fracture toughness. It is expected that

the resin-rich regions are the weak areas for the initiation and coalescence of micro matrix cracks to trigger delamination deflection and migration. Detailed analyses of the variations in local distribution of fibre and matrix materials around the delamination front in the current study (Fig.49-Fig.52) offer a meaningful and effective approach to predict delamination migration path in multidirectional CFRP laminates, which will help to develop reliable design and simulation tools for the prediction of mechanical performance of laminated composite structures. Further research is being carried out to develop an effective predictive model for the delamination migration based on the understanding of migration mechanisms. The outcome of the research will be presented in another paper, together with the quantification of the bending-twisting coupling effect and the variation of the delamination resistance.

It is worth noting that the apparent fracture toughness for a specific DCB configuration in Table.10 is a structural energy dissipation measure of the specific coupon, not a material property for the resistance to delamination propagation at the midplane. This is due to delamination migration and enhanced fibre bridging in the wake of the delamination front, making the measured apparent fracture toughness delamination length dependent. The apparent fracture toughness in Table.10 represents averaged strain energy release rate over the delamination length of 30 – 60mm for a given DCB configuration and is determined with the modified compliance calibration method which is suitable for unidirectional laminates but may not be accurate for measuring the strain energy release rate for multidirectional laminate.

For the layups with the 45° ply around the midplane, the delamination front was skewed, which means that the local features of the apparent delamination path are comparable in size to the dimensions of the specimen and are not free of boundary effects. The apparent fracture toughness values in Table.10 are hence not true material properties and shouldn't be directly used for the design and modelling of a generic multidirectional composite structure. Nevertheless, the findings reported in the current study represent reliable test data based on repeated tests of multiple specimens of the same material and geometry under the same testing conditions. They are valid for a comparative study for the effect of interfacial fibre orientation and veil interleaving on delamination migration and

delamination resistance, being the focus of the study and essential for developing advanced design and simulation tools for reliable damage tolerance analysis. Further research is however required to quantify the effect of above issues so that converged design parameters free of boundary effects can be determined for given interfaces in the design and analysis of a generic composite structure.

3.2.7 Conclusions

Effects of interfacial fibre orientation and interleaved PPS veils on delamination migration and delamination resistance have been investigated for CFRP laminates under mode I loading. A summary of conclusions can be collated from the results of the study as:

- Delamination migration takes place when interfacial fibre orientations have an element of 90° or 45°. The 3D delamination profile and enhanced fibre bridging and breakage due to delamination migration lead to improved resistance to delamination growth of 90/90, 0/90, 90/45 and 0/45 specimens in comparison to the 0/0 specimen. Strain energy release rate value derived based on the modified compliance calibration method in ASTM 5528-13 can be overly conservative in design optimisation for weight saving.
- PPS veils improve the apparent fracture toughness in general, but the improvement is fibre orientation-dependent, varying from 41% for the 0/0 specimen to 1% for the 90/45 specimen. Toughening effect of interleaving veils assessed with the 0/0 DCB specimen could be over-projected and incorrectly prescribed for multidirectional CFRP laminates. Effect of interleaving veil on apparent fracture toughness exists after the delamination has migrated away from the veiled interface, indicating that the PPS veil may affect not only the material at the interface but also the neighbouring materials.
- Delamination migration is found to be closely linked to the distributions of the fibre and matrix materials around the delamination front. The observation that delamination path can be predicted based on the analysis of resin-rich regions offers a meaningful and effective approach to rationalise the effect of interfacial fibre orientation and interleaving veil on delamination migration and fracture toughness of multidirectional CFRP laminates, which will help to

develop advanced design and simulation tools for reliable damage tolerance analysis of laminated composite structures. Further research is being carried out to develop an effective predictive model for the delamination migration based on the understanding of migration mechanisms.

- The apparent fracture toughness values presented in the paper provide a meaningful comparison for the effect of interfacial fibre orientation and interleaving veil on delamination migration and delamination resistance, which is fundamental for the design optimization of composite laminates. They are, however, not free of boundary effects and are delamination length dependent. As such, they are structural energy dissipation measures for specific specimen configurations and hence not true material properties. Further research is required to determine converged design parameters free of boundary effects for reliable damage tolerance design and analysis of a generic composite structure.

REFERENCES

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3.3 Effect of Interfacial Fibre Orientation and PPS Veils on Fracture Resistance of Woven CFRP Laminates under Mode I Loading

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Abstract: This paper presents an experimental study on the effect of interfacial fibre orientation and veil interleaves on the mode I critical Strain Energy Release Rate ($SERR_{crit}$) (G_{IC}) of woven CFRP using various interfacial fibre orientations and polyphenylene sulfide (PPS) veil densities. The double cantilever beam (DCB) specimens manufactured with 5-Harness Satin (5HS) carbon fibre reinforced polymer (CFRP) prepreg have an inherent fibre orientation bias at the interface due to the satin weave. Results demonstrate that DCB specimens without veils exhibit an increase of 2.9 times in G_{IC} at the interface with 90° dominant fibre orientation in comparison to 0° dominant fibre orientation. It is also demonstrated for the first time, that veil interleaving leads to reduced fibre orientation sensitivity under mode I, facilitates reductions in resin-rich regions, improves fibre bridging and overall enhances the G_{IC} . Fractographic features were characterised to rationalise the effect of interfacial fibre orientation and PPS veil on the G_{IC} of woven CFRP laminates.

Keywords: Hybrid – A; Damage Tolerance – B; Interface/Interphase – B; Fracture toughness – B; Fractography – D

3.3.1 Introduction

Structures made of composite laminates have many advantages compared with metallic materials, being lightweight, having improved in-plane mechanical properties and design flexibility for bespoke structural performance. A lack of reinforcement in the thickness direction is a significant concern for laminated fibre-reinforced composites, often leading to delamination and fractures [73]. The fracture mechanisms of woven ply laminates are extremely difficult to identify and investigate due to complex microstructures resulting from the interlacing and undulating fibre tows and a lack of through-thickness reinforcement [301]. Early fracture and delamination can be triggered under mode I and mode II loading due to low interlaminar fracture resistance resulting from the undulating fibre tows.

Composite structures are also susceptible to impact damage, resulting from the “manufacture, operation or maintenance of the structure. Barely visible impact damage (BVID) such as matrix cracking and delamination can occur and contribute up to 60% loss in a laminated composite component’s compressive strength” [45]. Other sources to initiate delamination include local defects such as voids, resin-rich regions, and other manufacturing defects. Delamination propagation during normal operation compromises the integrity of a structure due to substantial reductions of its load-transferring capacity.

The mode I SERR and delamination threshold can be improved by modifying the material constituents. Modifications such as fibre surface enhancements, commingling with tougher fibres, interleaving and through-thickness reinforcement have all shown various levels of success in improving the G_{IC} and damage tolerance. In particular, thermoplastic inter-layers positioned between the laminae of the composite can significantly increase the overall fracture resistance [20–23,25,26,36,208,214,223]. Planar toughening methods using interleaves allow exploitation of the tough and ductile interleave layer, which subsequently enhances the fracture resistance. Resistance to delamination initiation and growth of thermoplastic interleaved laminates is improved by the reduced stress concentration due to larger radii of the delamination front and enhanced fibre bridging behind the delamination with the presence of thermoplastic layers [223,302,303].

The use of 5HS fabrics in carbon fibre reinforced plastic (CFRP) laminates is increasing due to the two-dimensional weave being inherently damage-tolerant and having manufacturing advantages. Delamination, however, remains a high potential failure attribute where the structural performance could be undermined, especially under in-plane compression. Few studies have looked at the stacking sequence of 5HS woven composites in detail with no investigation available on the influence of interleaves on the fracture resistance of 5HS woven composites with different bias interface arrangements. This paper aims to fill the gap by presenting a detailed experimental investigation to understand the combined effect of dominant interfacial fibre orientation and PPS veils on the mode I fracture behaviour of 5HS CFRP laminates. The industrial relevance of the findings to enhance the damage tolerance with optimisation and modelling strategies of laminated woven composites are explored.

3.3.2 Materials and Experimental Procedures

The pre-impregnated 5HS CFRP material for the Double Cantilever Beam (DCB) specimens was supplied by Cytec, consisting of woven carbon 3K Toray T1000 fibres, with MTM49 [304] toughened epoxy system of 42% resin weight; an areal density of 283 gm^{-2} , a nominal ply cure thickness of 0.33 mm and related mechanical properties listed in Table.11. A total of six outer plies of 5HS carbon fibre pre-preg were prepared around a central layer of nonwoven polymer veil alongside a non-interleaved control group using the same layup (Table.12). The PPS veils supplied by Technical Fibre Products are Optiveil OP-49-48 [280], with a nominal areal density of 10 gm^{-2} and thickness of 0.09 mm. The veils were manufactured using a wet-lay process with chopped strand PPS fibres of $10\mu\text{m}$ diameter and length of 6 mm in a styrene-acrylic binder.

Table.11 Mechanical properties of MTM49-3-42%-3KFT300B40B-5H-1000 [304].

Property	Result - Cytec / authors	Standard
0° Tensile strength [MPa]	1065	ASTM D3039
0° Tensile modulus [GPa]	44.6	
90° Tensile strength [MPa]	1035	
90° Tensile modulus [GPa]	42.8	
0° Compressive strength [MPa]	640	SACMA SRM01R94
0° Compressive modulus [GPa]	59	
90° Compressive strength [MPa]	610	
90° Compressive modulus [GPa]	57	
In-plane shear strength [MPa]	108	ASTM D3518
In-plane shear modulus [GPa]	2.5	
0° Interlaminar shear strength [MPa]	64.2	ASTM D2344

For 5HS woven fabrics, the weft yarn passes over four warp yarn tows and under one, allowing for improved pliability compared with plain or twill weave fabrics, therefore making it appropriate on curved surfaces. In regards to the mechanical performance, there is little directional bias for the 5HS fabric as the reinforcement is balanced in the warp and weft directions. This is supported by the mechanical properties in Table.11, showing little difference between 0° and 90° coupon test results. It is worth noting that 5HS fabrics have an inherent dominant fibre direction at each face of the woven ply. Fig.53a illustrates the fibre directional

bias in the warp and weft direction on both faces. If considering a 5x5 segment of the 5HS fabric: Face A is dominated by (4:5 or 80%) 0° fibre orientation while Face B is dominated by (4:5 or 80%) 90° fibre orientation. During stacking of the 5HS fabric, Face B has a dominant fibre orientation at the interface perpendicular to that of Face A. Table.12 shows the layup configurations of the DCB specimens with dominant fibre orientations of 0/0, 90/90, 0/90, 90/45 and 0/45 at the midplane together with two levels of veil densities in the current study.

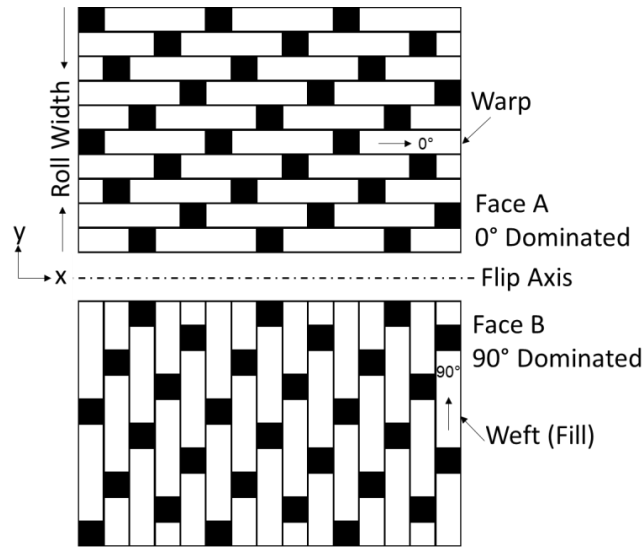


Fig.53 5HS CFRP laminae dominant fibre architecture.

Table.12 Configuration and code for DCB specimens.

Layup Configuration (Viewed from Top-Down)	Dominant Midplane Specimen Code
$[90,0]_6$	90/90 (Fig.54)
$[(90,0)_3, 1x \text{ veil}, (90,0)_3]$	10V-90/90
$[(90,0)_3, 2x \text{ veil}, (90,0)_3]$	20V-90/90
$[0,90]_6$	0/0
$[(0,90)_3, 1x \text{ veil}, (0,90)_3]$	10V-0/0
$[(0,90)_3, 2x \text{ veil}, (0,90)_3]$	20V-0/0
$[0,90]_{3s}$	0/90
$[(0,90)_3, 1x \text{ veil}, (90,0)_3]$	10V-0/90
$[(0,90)_3, 2x \text{ veil}, (90,0)_3]$	20V-0/90
$[90_6, -45, 90_5]$	0/45
$[90_6, 1x \text{ veil}, -45, 90_5]$	10V-0/45
$[90_6, 2x \text{ veil}, -45, 90_5]$	20V-0/45
$[0_6, -45, 0_5]$	90/45
$[0_6, 1x \text{ veil}, -45, 0_5]$	10V-90/45
$[0_6, 2x \text{ veil}, -45, 0_5]$	20V-90/45

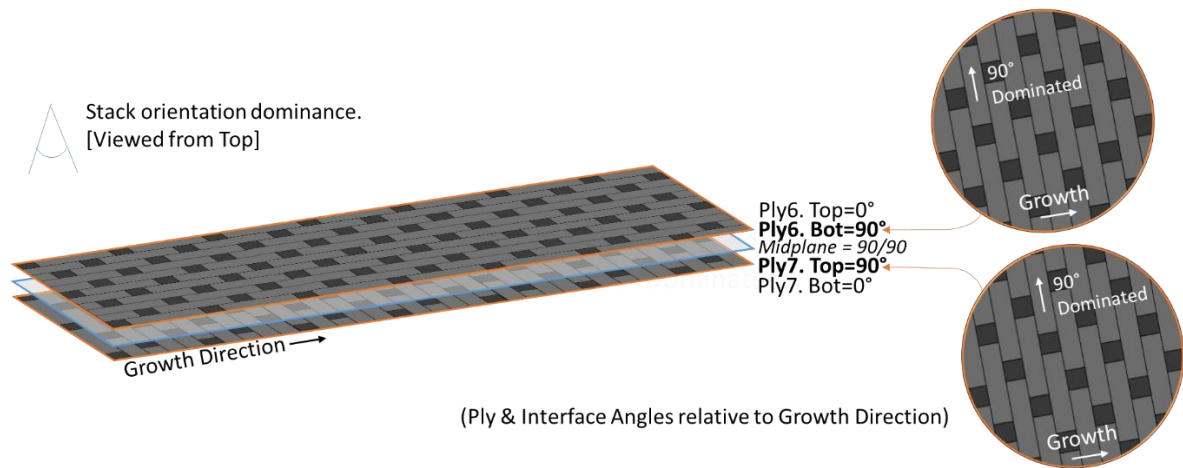


Fig.54 Example layup for 90/90 dominated mid-plane.

DCB specimens illustrated in Fig.55 were manufactured for the experimental investigation with geometry following ASTM 5528-13 [53]. As presented in Fig.55a, fifteen configurations listed in Table.12 were fabricated under the same manufacturing conditions, with five consistent repetitions for each test. The DCB specimens had the final dimensions with a length of 140 mm; width of 20 mm and thickness of 4 mm. A single layer of nonwoven veil of 10 gm^{-2} as illustrated was placed at the midplane for five layup configurations initiating at the edge of the PTFE release film. Another five configurations were produced similarly with two layers of veil building up to 20 gm^{-2} . The remainder five configurations contained no veil interleaves and only the PTFE release film for delamination initiation as a control group. Twelve 5HS weaved CF/Epoxy layers were de-bulked under vacuum and subsequently placed on a flat steel plate with peel film, caul plate, breather fabric and cured inside a vacuum bag within the autoclave. The panels were cured under pressure at 620 KPa, with a target bag vacuum of <5kPa, ramp rates of $3^\circ\text{C}/\text{min}$ and a dwell temperature of 135°C for 90 minutes as prescribed for MTM49 [304].

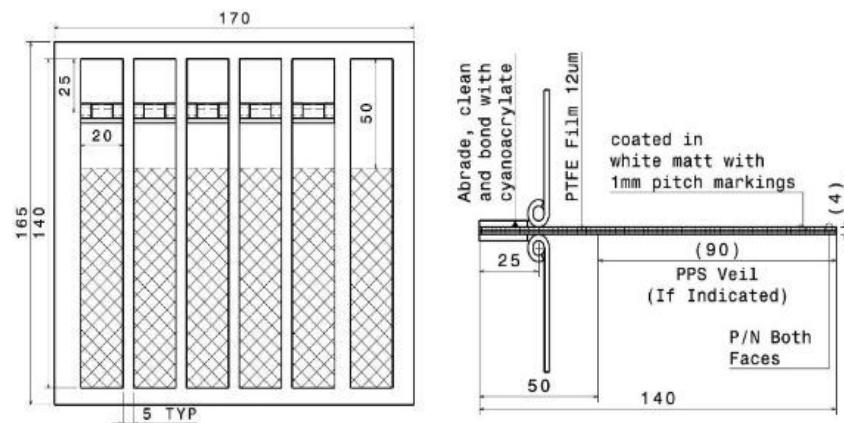


Fig.55 (a) Geometry of laminate plate and (b) DCB specimen assembly.

Under mode I loading, the G_{IC} was evaluated as specified in ASTM D5528-13 [53], however, with the modifications on layup orientation and fibre architecture. The layup configurations deviated from the standard predominantly for the midplane, being the subject of the investigation; namely for interfacial fibre orientation and veil interleaving effects on the G_{IC} . The outer laminae differences listed in Table.12 remained comparable; with fibre orientations shifting by a half-ply thickness (Face A and B of Fig.53), consequently having minimum stiffness implications. It is expected that such a small variation in stiffness would have a negligible influence on the current investigation considering the effect of interfacial fibre orientation and PPS veil on the midplane fracture resistance.

During each test, specimens were clamped onto cyanoacrylate-bonded hinges into tapered vice-grip clamps and axially aligned with the load path of the test apparatus. The specimen edge was sprayed white with a printed scale attached to allow tracking of the delamination growth during the test. Testing was conducted using a dual screw column servo-electric universal test machine (Instron 5965) equipped with a calibrated 5kN load cell. The crosshead displacement rate was set to 2 mm/min to maintain quasi-static loading conditions. The tests were also recorded on video using a 150 mm macro lens and tripod-mounted DSLR (Fig.56).

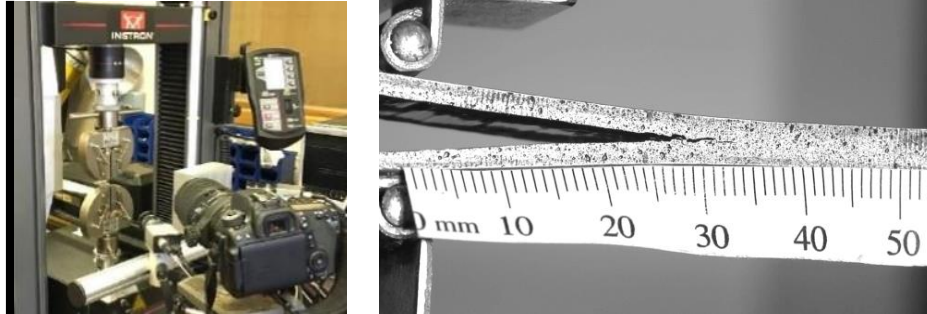


Fig.56 Experimental setup: (a) DSLR camera with 150 mm macro lens, (b) Loaded specimen.

The initiation and propagation of the delamination positions were identified with the camera and linked to the corresponding load-displacement data. The mode-I critical SERR or fracture toughness (G_{IC}) was subsequently calculated using the modified compliance calibration method [53]:

$$G_I = \frac{3P^2C^{2/3}}{2A_1bh} \quad \text{Eqn.16}$$

where P is the applied load, b is the specimen width, h is the specimen thickness, A_1 is the slope of the least-squares plot of the delamination length normalised by specimen thickness, a/h as a function of the cube root of compliance $C^{1/3}$. C is the ratio of the load point displacement to the applied load, δ/P .

3.3.3 Results

3.3.4 Effect of interfacial fibre orientation and veil on load-displacement

The enclosed area under each load-displacement curve in Fig.57 represents the resistance to delamination, hence closely related to the fracture resistance energy. The region under the initial linear portion of the curves is the elastic potential energy being transferred during the displacement. The elastic potential energy is later transformed into the fracture energy during the release of the stored elastic energy. Differences have been observed for the area under the load-displacement curve amongst the various dominant fibre orientations and interleaved configurations. Moreover, smoothing or increasing fluctuations of the stick-slip load-displacement behaviour during propagation (Fig.57), demonstrates the influence of interleaving on the delamination propagation behaviour. For the different interfacial fibre orientations, a distinct variation of forces were required to sustain the delamination growth, implying an increased level of energy dissipation during delamination propagation.

It can be observed from the load-displacement data (Fig.57) the 90/90 dominant midplane of the 5HS laminae required the greatest delamination propagation force, significantly higher than the 0/0 dominated midplane. Further differences in the load-displacement trends from Fig.57 can be observed, where the energy released during the fracture process is dissimilar among the five interfacial orientations with different veil densities. DCB specimens of 90/90 midplane have a coarse stick-slip propagation while those of 0/0 midplanes have a smooth delamination growth. For configurations other than 0/0 dominated midplanes, the local load carrying capacity and energy required for delamination propagation is increased prior to the next growth event, observable as a series of load-drops from Fig.57. The variation observed in the force and energy profiles are associated to fractographic features observed in subsequent microscopic examination in section 3.3.6.

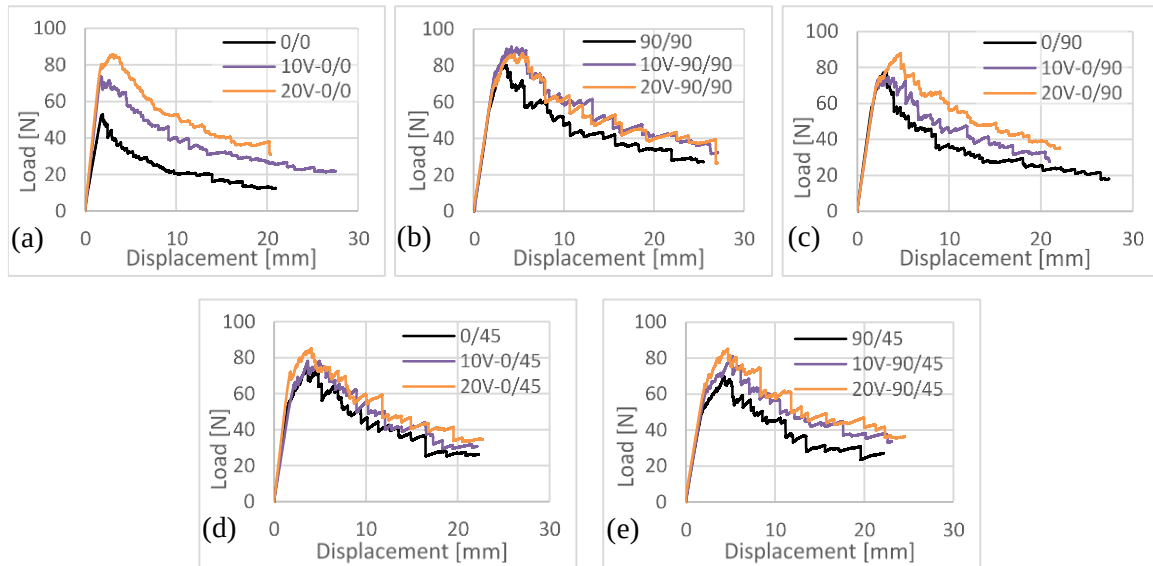


Fig.57 Load-Displacement traction plots for 5HS bias orientation related DCB tests.

A comparison of the load-displacement curves are shown in Fig.58, presenting the effect of interleaving at the midplane against the five dominant interface orientations (Table.12). Representative load-displacement plots are presented in Fig.58(a) for the midplane orientations without PPS veil interleaves. Load-displacement plots for the same interface orientations but with the addition of a 10 gm^{-2} and 20 gm^{-2} veil interleave at midplane are presented in Fig.58(b-c). For the non-interleaved specimens (Fig.58(a)), the 90/90 dominant midplane appears to have the highest resistance to delamination while the 0/0 midplane specimens have the lowest resistance. The difference in the shape of the load-displacement curves are also worth observing. 0/0 midplane specimens exhibit a stable reduction of the load after the peak force of the initial elastic deformation. Interestingly, other configurations such as the 90/90, 0/90, 90/45 and 0/45 exhibit some further load carrying capacity beyond the peak force of the initial elastic deformation.

The addition of veils influenced the load-displacement and subsequently calculated G_{IC} . Compared with the curves in Fig.58(a), it is apparent that interleaving (Fig.57(b-c)) improves the resistance to delamination in general; with most orientations exhibiting higher peak forces, increased propagation load and frequent low-magnitude growth increments. The 10 gm^{-2} interleaved specimens for all interface orientations have increased the load-carrying capacity before propagation, with significant increases in the 0/0 dominated midplane. The effect of the veil on G_{IC} is however dependent on the midplane fibre orientation. The 20

gm⁻² PPS interleaved midplanes have further enhanced load carrying capacities and reduced the inherent interfacial fibre orientation sensitivity associated with using 5HS laminae. The observations in the fracture surface morphology (Section 3.3.6) explains the cause of this behaviour with considerations on surface texture, fibre bridging, localised fibre breakage and resin-rich areas affecting the fracture resistance.

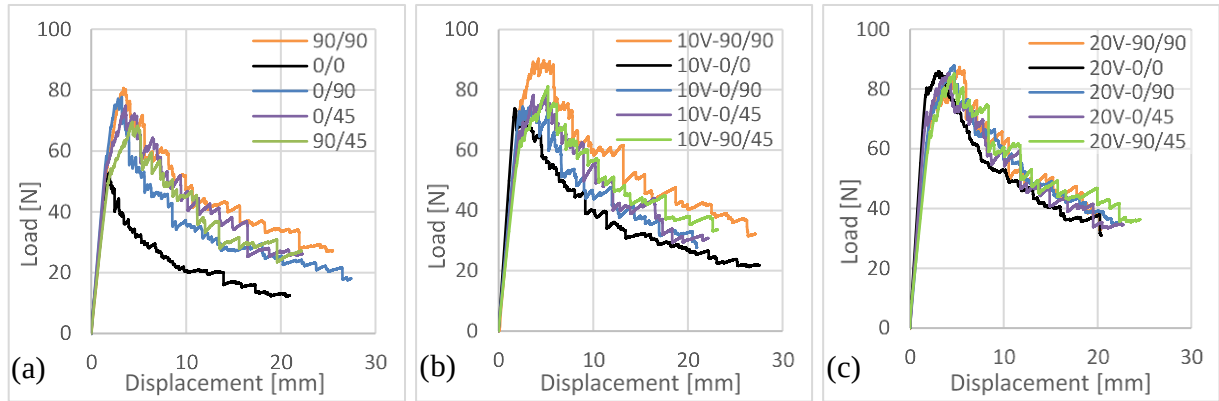


Fig.58 Load-displacement curves: (a) without veil, (b) 10 gm⁻² veil, (c) 20 gm⁻² veil at midplane.

3.3.5 Resistance curves of non-interleaved and interleaved configurations

The load-displacement curves (Fig.57-Fig.58) are used to obtain the resistance curves (R-curves) using the MBT derived compliance calibration method (Eq.1). The mode I resistance curves presented in Fig.59 showcase the influence of the dominant interfacial fibre orientations and the addition of veil interleaves on G_{IC} . For each interfacial fibre orientation, delamination resistance of specimens without veils is compared to that of specimens with veils to isolate the interleaving effects. Consequently, the veil interleaves have significantly improved the G_{IC} for the 5HS woven laminates.

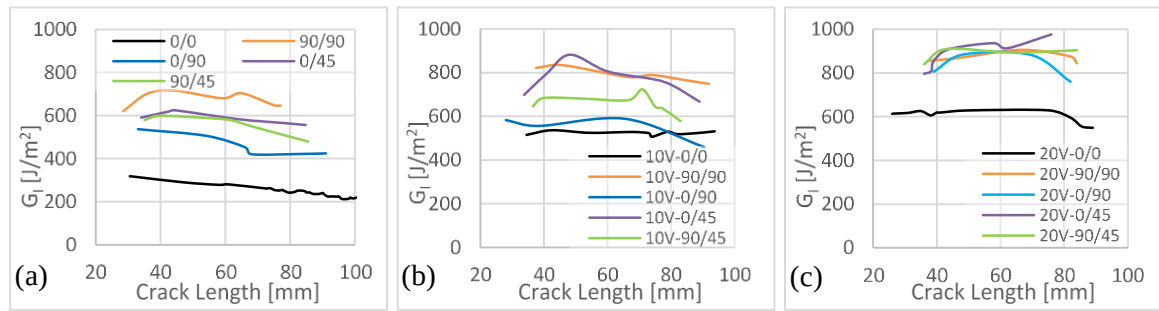


Fig.59 R-curves of 5HS CFRP (a) without veil, (b) with 10 gm⁻² and (c) with 20 gm⁻² veil for various biased midplane fibre orientations.

A summary of the test results for the influence of interfacial fibre orientation and PPS veil on the G_{IC} is presented in Table.13. The greatest benefit to the G_{IC} is for the 0/0 dominated fibre interfaces showing 111% enhancement by using 10gm⁻² veils and 136% enhancement using the 20gm⁻² veil compared to the control group. The remaining non-0/0 fibre orientations also exhibit enhancements in G_{IC} ranging between 10% to 30% for 10 gm⁻² veil interleaved and 20% to 64% for 20gm⁻² veil interleaved.

The merging of R-curves shown in Fig.59c for orientations other than 0/0 using the 20 gm⁻² PPS veil is a significant finding as it demonstrates that veil interleaving can reduce the interfacial fibre orientation sensitivity when considering fracture resistance. The reduction of fibre orientation sensitivity by using veil interleaves on the remaining G_{IC} values permits design flexibility; a substantial industrial benefit for components using biased woven CFRP laminates.

Table.13 Summary of Mode I Critical Strain Energy Release Rates

Dominant Midplane	G _{IC} (No Veil) [J/m²]		G _{IC} (10 gm ⁻² veil at midplane) [J/m²]		ΔG _{IC} 10 gm ⁻² Vs No Veil	G _{IC} (20 gm ⁻² veil at midplane) [J/m²]		ΔG _{IC} 20 gm ⁻² Vs No Veil
	Mean	StdDev	Mean	StdDev		Mean	StdDev	
0/0	259	9	547	100	+111%	611	116	+136%
90/90	745	37	820	13	+10%	896	13	+20%
0/90	524	25	582	37	+11%	858	33	+64%
90/45	606	16	685	14	+13%	869	5	+43%
0/45	608	17	790	12	+30%	886	30	+46%

3.3.6 Fracture morphology at interfaces of varying fibre orientations

The fracture surface morphology of the specimens was examined using Scanning Electron Microscopy (SEM), with specimens containing the 0/0 dominant midplane fibre direction shown in Fig.60. Non-interleaved (Fig.60a) and interleaved (Fig.60(b-c)) specimens exhibit dominant interlaminar delamination growth at the 0/0 midplane as expected. Closer examination of the fracture surface morphology reveals that interleaved interfaces exhibit a coarse and irregular texture (Fig.60(b-c)) compared to non-interleaved interfaces (Fig.a). The increased surface roughness explains the 111% and 136% improvements (Table.13) in G_{IC} by the addition of 10gm^{-2} and 20gm^{-2} veils for the 0/0 configuration and implying increased energy dissipation during propagation. The 0° fibre-matrix de-bonding is a contributor for the lower fracture resistance in the 0/0 dominated interface. However, the influence of the non-dominant 90° yarns (20% - Fig.53a) is also present, where Fig.57a exhibits a low magnitude stick-slip growth behaviour.

The enhanced fracture resistance and improved interface bridging are consistent with those observed in UD laminates utilising veil interleaves [25,26,36,223]. The interleaved fracture surfaces (Fig.60(b-c)) provides evidence of the veil enhanced roughness and fibre bridging being related to the increased energy dissipated during crack propagation.

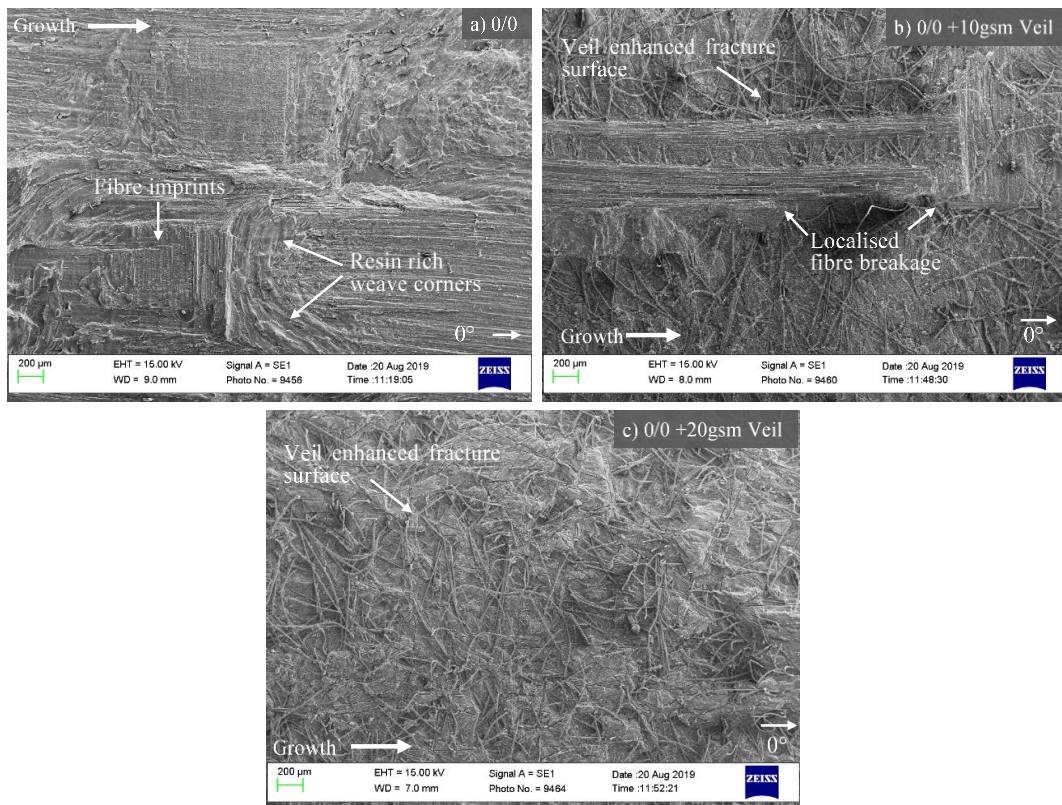


Fig.60 Fracture surface of 0/0 dominated midplane: (a) without veil; (b) 10 gm^{-2} veil; (c) 20 gm^{-2} veil.

The fracture surface morphology for specimens with the 90/90 dominant midplane fibre orientations is presented in Fig.61. Comparable to the dominant 0/0 fibre midplane, the 90/90 dominated midplane also has resin-rich zones for non-interleaved configurations and greater fibre-matrix interactions for interleaved interfaces. As a result, the G_{IC} of 90/90 compared to 0/0 dominated midplane is enhanced by 2.9 times for non-interleaved, 1.5 times for 10 gm^{-2} interleaved and 1.4 times for 20 gm^{-2} interleaved (Table.13). The arrangement of 90° fibres ahead of the delamination path provide a deterrent to the approaching delamination front as the delamination path must navigate around the 90° fibre tows. Unlike 0° dominated interfaces, the 90° dominated interfaces do not have a defined propagation path, particularly for woven 5HS laminates due to the undulating fibre tows. For UD laminates, migration is likely to be triggered with the resulting fracture mechanism being intra-laminar; however, for woven materials, the weave architecture does not allow such mechanisms to be activated under quasi-static loading conditions. The enhancements resulting from the veils are the result of improved interlaminar fracture resistance that are visible as fibre imprints, breakages, peeling, matrix striations and general surface roughness observed on the interleaved specimens (Fig.61(b-c)). These enhancements allow for design flexibility and controllability of the G_{IC} by manipulation of interfacial fibre orientation and by the addition of PPS veil interleaves.

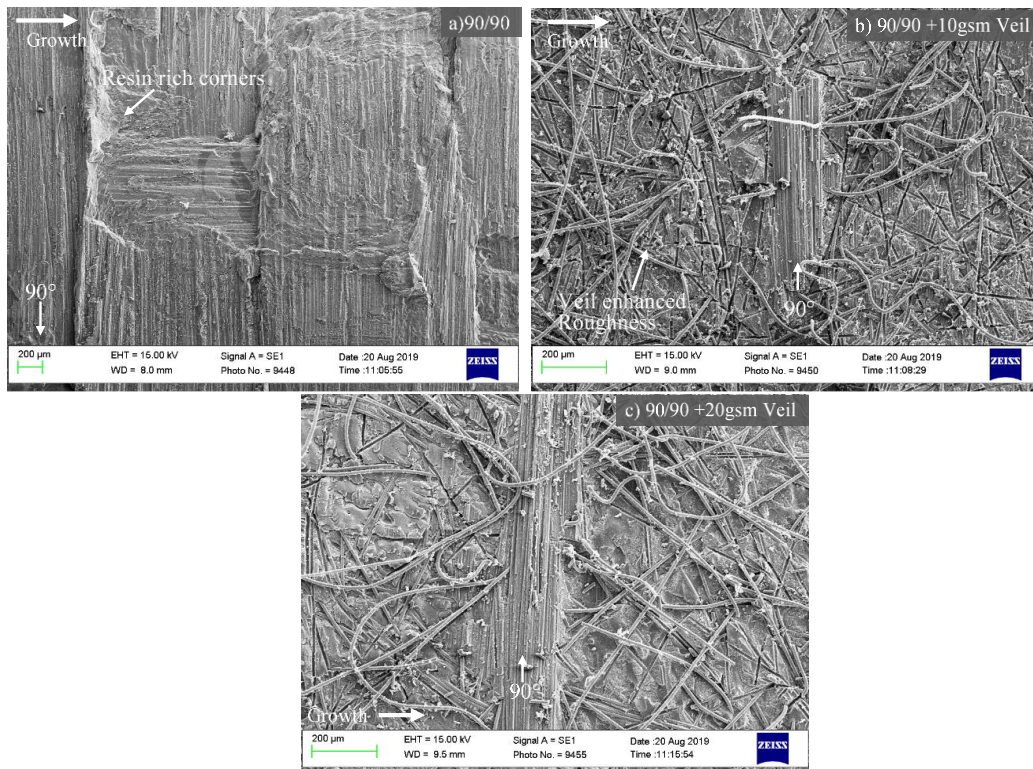


Fig.61 Fracture surface of 90/90 dominated midplane: (a) without veil; (b) 10 gm^{-2} veil; (c) 20 gm^{-2} veil.

Fracture surface morphology for 0/90 dominated midplane are presented in Fig.62, with observations similar to 90/90 dominated midplanes being visible. The non-interleaved (Fig.62(a)) and interleaved (Fig.62(b-c)) 0/90 specimens exhibit interlaminar failure with localised fibre peeling and breakage. Improvements have doubled for the G_{IC} without interleaving for the 0/90 dominated midplane compared with 0/0 midplane. Resin rich zones are present in Fig.62a, alongside resin voids in the 20 gm⁻² interleaved specimens (Fig.62c). Extended fibre fracture for the non-interleaved 0/90 dominated midplane is observed (Fig.62a). The enhancements in the G_{IC} for the 0/90 dominated midplane with interleaved veils can be explained by increased surface roughness and improved fibre-matrix interactions.

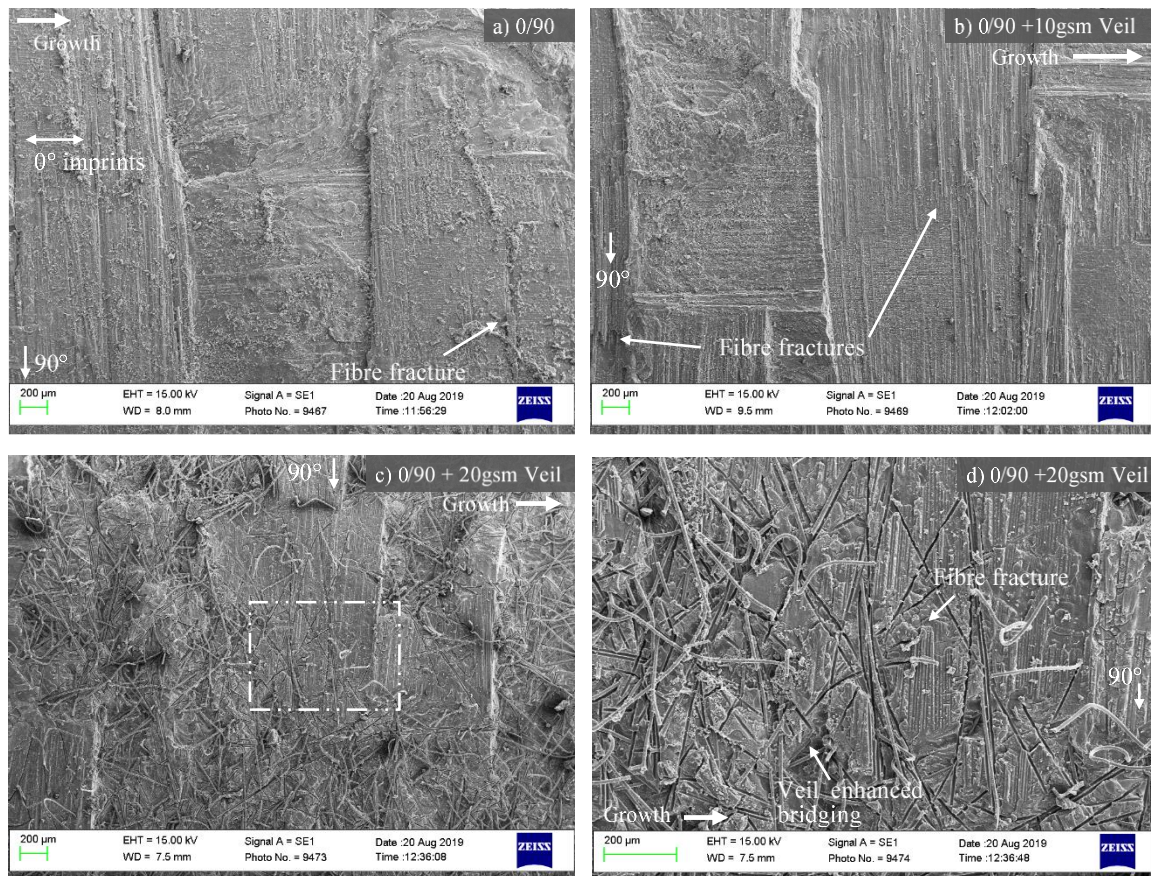


Fig.62 Fracture surface of 0/90 dominated midplane: (a) without veil; (b) 10 gm⁻² veil; (c-d) 20 gm⁻² veil.

Fracture surface morphology for the 90/45 dominated midplane is presented for non-interleaved (Fig.63(a)) and interleaved specimens (Fig.63(b-c)). The interleaved 90/45 dominated midplane specimens have greater surface roughness with fibre pull out and fibre fractures, especially in the case of 20 gm⁻² veil interleaved (Fig.63(c)). The fibre-matrix bonding and stress transfer between the angled ply layers of 90/45 allow for increased propagation resistance when using 90° fibres. The increase in fracture resistance between the 0/90 and 90/45 dominated interfaces (Table.13) is justifiable from the increased fibre level damage in Fig.63(b-c). Additionally, Fig.57(e) presents the increased traction loads for specimens with veil interleaves, further justifying the applicability of interleaving in woven ply laminates. The 90/45 dominated midplanes have increased the fracture resistance by up to 2.3 times against 0/0 dominated midplanes for non-interleaved, 1.25 times for 10 gm⁻² interleaved and 1.4 times for 20 gm⁻² interleaved. The results demonstrate that fracture resistance of laminates can be improved not only by veil interleaving but also by changing the interfacial fibre orientation. Fibre orientation can be altered to enhance the fracture resistance, provided that this does not compromise other mechanical performance being the in-plane strength and stiffness of the laminate and results in further design flexibility by improving the damage tolerance of the laminate.

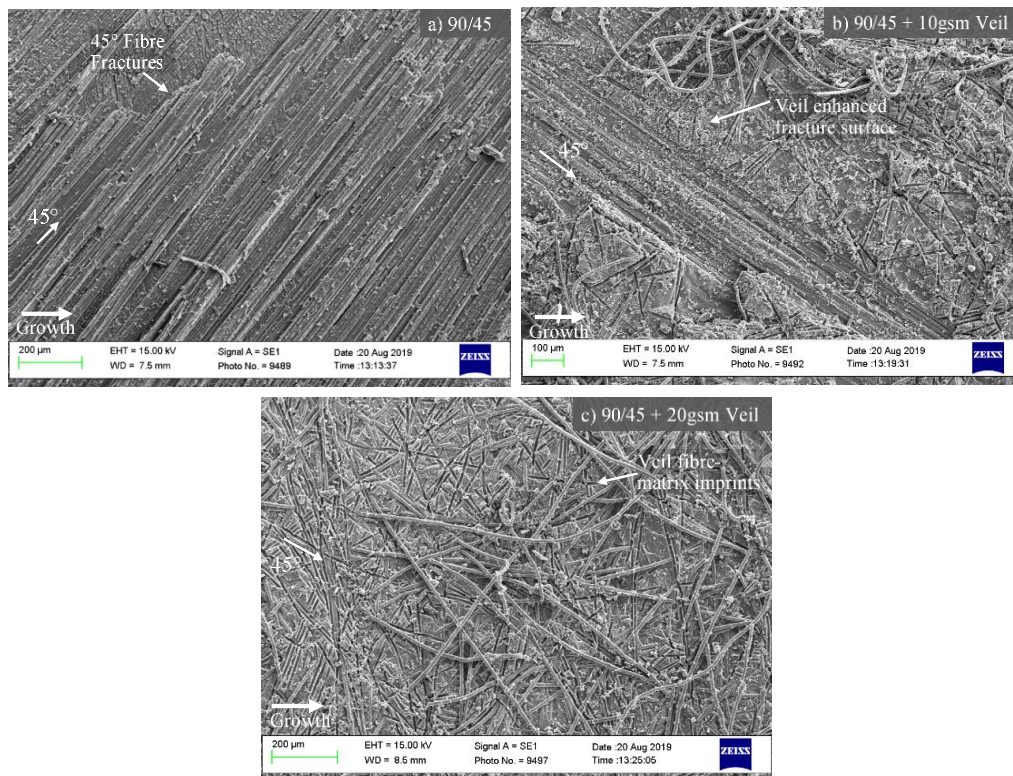


Fig.63 Fracture surface of 90/45 dominated midplane: (a) without veil; (b) 10 gm⁻² veil; (c) 20 gm⁻² veil.

Fracture surface features at the 0/45 dominated midplane (Fig.64) are similar to those at the 90/45 midplanes. The fracture surface morphology of the 10 gm⁻² interleaved specimens have increased levels of surface roughness, further fibre bridging, observable fibre breakages and fibre imprints. The 20 gm⁻² interleaved specimens exhibit further indication of increased fibre-matrix interactions. Interleaving the 0/45 dominated midplanes produced a 30% enhancement using 10 gm⁻² veil and 46% enhancement using 20 gm⁻² veil in the final G_{IC} (Table.13). In comparison to the lowest G_{IC} values from 0/0 dominated midplanes, the 0/45 dominated interfaces are 2.3 times greater for non-interleaved, 1.44 times greater for 10 gm⁻² interleaved and 1.45 times greater for 20 gm⁻² interleaved specimens.

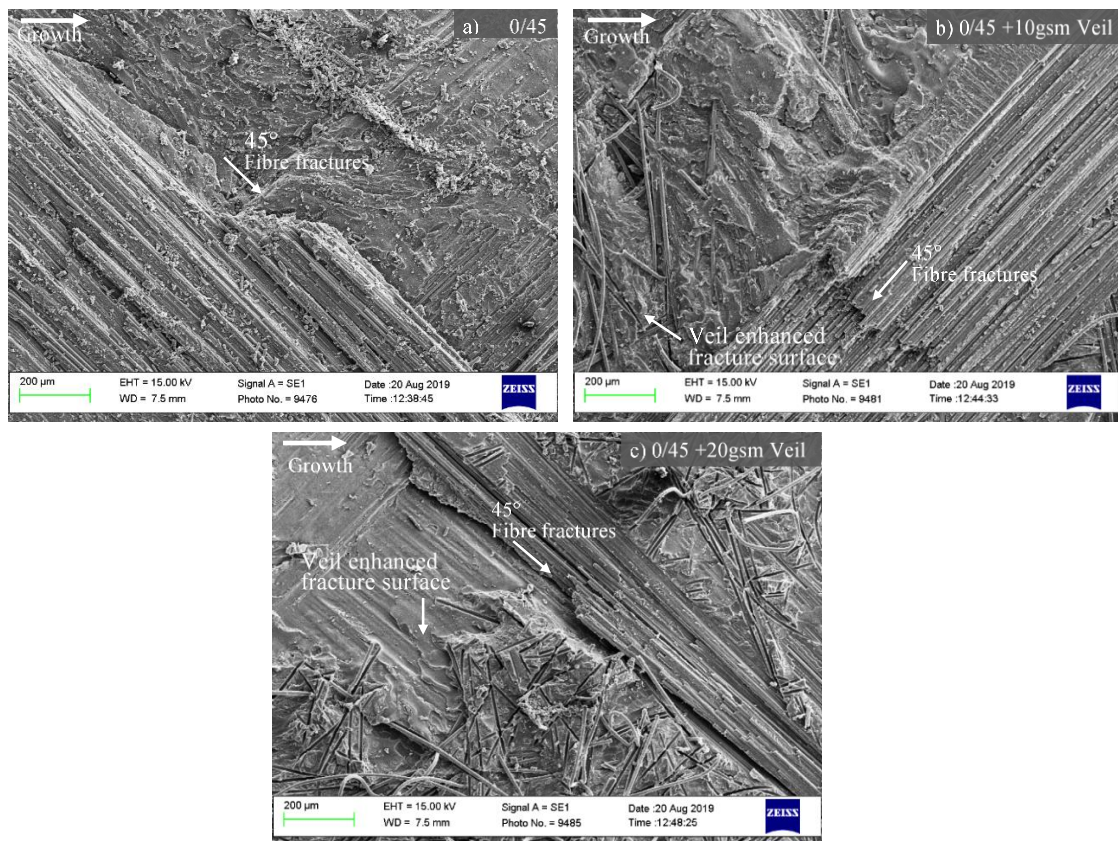


Fig.64 Fracture surface of 0/45 dominated midplane: (a) without veil; (b) 10 gm⁻² veil; (c) 20 gm⁻² veil.

3.3.7 Discussion

The combined effect of biased interfacial fibre orientation and PPS veil interleaving on the delamination growth behaviour and mode I fracture resistance of woven 5HS CFRP laminates have been investigated. Results show that the G_{IC} and delamination growth features are dependent on the interface arrangement, which are consistent with findings in the literature [305–307]. It has been demonstrated in Fig.57 that delamination propagation along the 0° biased interface has less resistance than the propagation along the 90° biased interface. Delamination propagation along the 90° dominated interfaces exhibited a stick-slip traction behaviour leading to recurrently spaced peaks in the resistance curve (Fig.57(b-c)). Similarly, interlaminar fracture with interfacial fibres aligned with a width direction component such as the 45° bias (Fig.57(d-e)) also exhibited a stick-slip behaviour.

The 5HS laminae have similar mechanical performance in both longitudinal and transverse directions (Table.11) due to the interlacing weave structure. Nevertheless, the effect of a single $\pm 45^\circ$ ply was investigated using a 3D Finite Element Model (FEM) to quantify the mixed-mode contributions arising from the off-axis fibre orientations (Fig.65). The delamination front during propagation is subjected to a dominant opening mode only as expected. ANSYS Composite PrepPost (ACP) was employed to define the layup, assigning the orientation and the material properties to each layer and using the Virtual Crack Closure Technique (VCCT) to model the delamination with interface elements. VCCT calculates the energy-release rate with an assumption that the energy needed to separate a surface is the same as the energy needed to close the same surfaces.

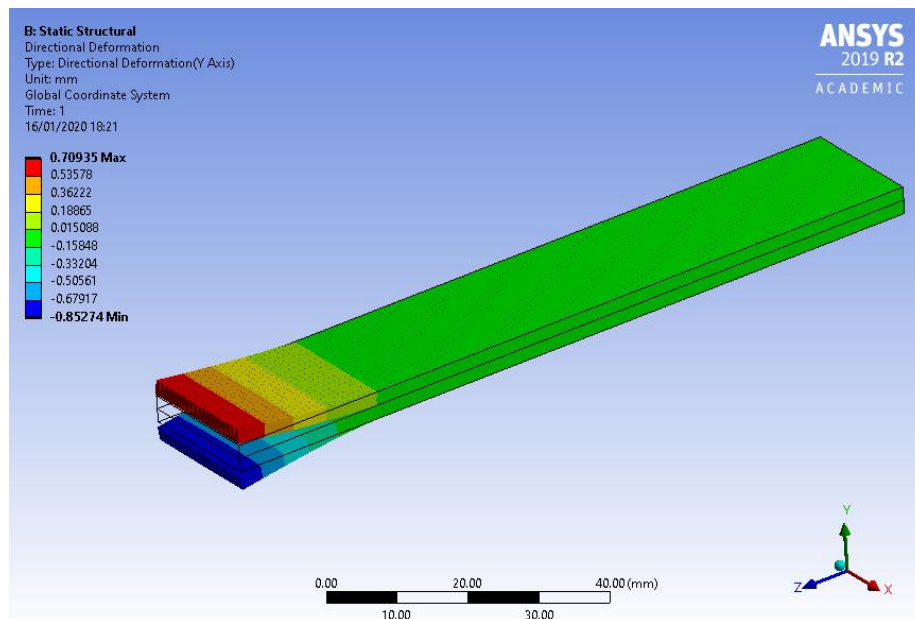


Fig.65 Finite Element Model of $[90_6, -45, 90_5]$ DCB specimen – 0/45 Biased midplane.

The average value of the energy release rate G for mode II and mode III were normalised by the average value of the mode I for the same crack length. The maximum value of the ratio in Table 14 is 2.2%, indicating very limited effects of the $\pm 45^\circ$ orientation biased layers on the energy release rate and mode I energy absorption being the dominant failure.

Table 14 Ratio of Energy Release Rate for Initial delamination length.

Dominant midplane	$a = 25 \text{ mm}$	
	G_{II}/G_I	G_{III}/G_I
0/45	1.1%	2.2%

The fibre-matrix de-bonding is the mechanism contributing to the fracture resistance in the 0° dominated interface (being the directing interface). Nevertheless, some influence of the non-dominant 90° yarns (20% - Fig.53) are also apparent from Fig.57(a) showing a lower magnitude stick-slip growth behaviour. The growth characteristics are clearly different between the various fibre arrangements based on the diversity of the load reduction events in Fig.56- Fig.57.

The addition of veils increases the local elastic load limit prior to further growth propagation in most cases. In other cases, smoothing of the stick-slip behaviour is observed and caused by the competing damage interactions resulting from either the veil interleave or dominant fibre orientation contribution. Consequently, the addition of thermoplastic interleaves demonstrated in Fig.57 by the local elastic regions between the incremental propagations are increased and beneficial in the improvement of fracture resistance and damage tolerance by increasing the interfacial ductility.

The effect of the dominant fibre orientation and veil interleaving on the G_{IC} has been summarised for the different interface configurations listed in Table.12. Comprehensive interlaminar SERR results from this investigation are presented in Table.13. The non-interleaved and veil enhanced G_{IC} parameters (Table.13) can consequently be used in low-fidelity numerical models using cohesive zone modelling techniques in future work. The use of orientation-specific G_{IC} parameters can more accurately predict the damage behaviour in laminates, compared with the single G_{IC} parameter obtained through the standard 0° DCB tests. The SERR of micro-cracks derived from the relative interface angles can be combined for an entire laminate, where a stress transfer-based stiffness reduction method can be employed to swiftly identify damage potential and likelihood. The advantages of the proposed methods allow the capability to predict the matrix fractures based on the local stresses, stiffness and orientation-specific SERR values for each lamina. Coupon studies in the search for multidirectional SERR parameters can provide a deeper understanding of the micromechanical behaviour of fibre directionality of biased woven laminae, along with the capability to study veil influenced micromechanics.

A significant outcome of this work is showcasing how the use of thick 20 gm^{-2} PPS veil reduces the interface resistance variations observed in the non-interleaved 5HS laminates (Fig.59). The novel reduction of fibre orientation sensitivity and converging G_{IC} values using 20 gm^{-2} PPS veils permits design flexibility associated with the reduced mismatch of mechanical properties of all possible interface configurations when using 5HS CFRP materials. This represents a substantial industrial benefit considering the simplification of the design and analysis of multidirectional CFRP laminate using the same set of engineering parameters and negligible loss of the in-plane mechanical performance.

Based on the above results, testing of non-interleaved and interleaved laminates following the prescribed 0° orientation as per ASTM 5528-13 [53] is inadequate to understand the behaviour of interlaminar damage and resistance of multidirectional woven 5HS CFRP laminates. The behaviour of 5HS laminates while considering various ply bias angles is different from that observed in unidirectional materials where crack jumping and migration could be triggered [102,294,295,300]. Comparatively, the global failure mode for 5HS woven laminates is primarily mode I fracture, which may have highly localised tendencies for mixed-mode II/III cusp and shear hackle type morphologies depending on the presence of interleave and fibre level interactions. For woven laminates, global intra-laminar failure behaviour leading to migration is not possible without substantial energy levels to fracture through the woven laminae, instead the global failure mode for 5HS woven laminates is a pure mode I fracture.

The G_{IC} values (Table.13) obtained from the DCB tests can be used for the orientations considered to simulate laminated structures without the need for detailed complex ply-by-ply models in future work. The current results have significant industrial relevance in the design optimisation process of composite structures utilising 5HS CFRP laminates. Knowledge developed on the influence of interfacial fibre orientation and interleaved veil on the G_{IC} leads to more accurate and reliable design and simulation tools for the prediction of mechanical performance of laminated composite structures within the aerospace and automotive sectors. Delamination damage for example during an impact event is a major concern, the applicability of angle specific SERR values means more accurate predictions can be made with predictions on the least resistant ply combinations for a given propagation direction. The reduction of fibre orientation sensitivity can provide design and manufacturing flexibility and allow the use of double-stacked plies should the stiffness be required. Furthermore, the study of angled ply biases for 5HS laminates provides a deeper understanding of the fracture mechanisms of complex and interacting failure modes during impact or loading under multiple modes.

3.3.8 Conclusions

An experimental investigation on the effect of interfacial fibre orientation and interleaved PPS veils on the mode I SERR has been conducted for 5HS weave CFRP laminates. The following conclusions can be drawn from the results of this investigation:

- 5HS laminae have distinct fibre orientation biases at its two faces. Dominant interfacial fibre orientation of 5HS laminates greatly influences the interlaminar fracture resistance. Enhancements in G_{IC} of up to 2.9 times that of 0/0 can be achieved by placing the 5HS laminae in a 90/90 dominated orientation under mode I loading.
- The areal density of the interleaved veil improves the G_{IC} ; where enhancements of 111% can be achieved using 10 gm^{-2} and 136% using 20 gm^{-2} .
- The interfacial fibre orientation bias of 5HS woven laminae can be eliminated with the use of 20 gm^{-2} interleaved veils for most interfacial fibre orientations.
- Fracture surface morphology is related to interfacial fibre orientation, veil interleaving and exhibits features including resin-rich zone reduction, fibre peeling, fibre fracture, increased surface roughness, fibre imprints and highly localised intra-laminar fracture tendencies.

Research on the combined effect of alternative layups and interleaved veils on the fracture resistance of 5HS CFRP laminates provides detailed insights into the fracture and failure behaviours of woven laminates. The failure mechanisms and interactions need to be further investigated to develop better knowledge and understanding of the enhanced damage tolerant design of woven laminated composite structures.

REFERENCES
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3.4 Influence of Interfacial Fibre Orientation and PPS Veils on the Fracture Resistance of Woven CFRP Laminates under Mode II loading.

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Abstract: In this study, the influence of interleaved thermoplastic veils on the mode II interlaminar fracture toughness (ILFT) of 5-Harness Satin (5HS) carbon fibre reinforced plastic (CFRP) composite laminates were investigated. The characterisation of delamination resistance, the effect of interfacial weave orientation and veil interleaving on the interlaminar fracture behaviour of 5HS woven CFRP laminates have been explored using End-Notched Flexure (3ENF) tests. Toughness enhancement investigations for various mid-plane interfacial fibre orientations combined with polyphenylene sulfide (PPS) veil interleaves were studied to understand the fracture mechanisms and ILFT variations. Experimental results revealed enhancements in the ILFT initiation up to 75% using veil interleaves, however, greatly dependant on the dominant interfacial fibre orientation. For directionally biased interfaces where interfacial fibres are transversely dominant (90°) to the propagation direction, greater energy is required than if propagation was along the longitudinal (0°) dominant interfaces. The fracture surface morphology of specimens with transversely oriented fibres exhibited increased fibre matrix interactions in comparison to longitudinal dominated interfaces.

Keywords: Hybrid – A; Damage Tolerance – B; Interface/Interphase – B; Fracture toughness – B; Fractography – D

3.4.1 Introduction

Laminated composite structures have many structural advantages compared with metallic materials; however, their limited through-thickness strength and toughness can lead to delamination and coalescing fractures [73]. Component fracture behaviour and damage tolerance are affected by the reduced interlaminar toughness which can trigger delamination under mode I and mode II loading. Composite structures are highly susceptible to out-of-plane impact

damage during the manufacturing, operation or maintenance of the structure and initiating a significant reduction in the performance. Subsequent delamination induced failure is a result of a combination of compressive and bending stresses caused by the delaminated plies buckling out of plane during excessive loading or impact. The damage complexity is exacerbated in woven laminae due to the interlacing and undulating fibre tow microstructures [301]. Interlaminar shear, tension and matrix cracking in laminates can cause internal delamination, leading to susceptibility for membrane-type failures of the sub-laminate and further strength reductions. Consequently, woven laminates have been limited in safety-critical applications due to their lower in-plane strength where interlacing weaves trigger local buckling. Efforts to overcome these constraints typically lead to excessively over-conservative designs with multiple load paths, however toughening the laminae interfacial regions could provide an intermediate alternative.

Delamination tendencies can arise from local features such as voids, resin-rich regions, and other manufacturing defects; where resin variations are a common occurrence in woven CFRP laminates. Research efforts to improve the interlaminar fracture toughness (ILFT) (G_{IIC}) and delamination threshold by modifying the material constituents have shown significant developments over the years. Modifications to matrix chemistry, fibre surfaces, commingling with tougher fibres, interleaving and through-thickness reinforcements have all shown improvements in G_{IIC} , delamination behaviour and damage tolerance capacity. Toughening methods using thermoplastic interleaves can allow exploitation of the tough and ductile interleave layer, subsequently enhancing the fracture toughness under mode I and mode II [21,24,27,35,36,60,224].

The use of 5HS CFRP laminates are increasing due to the two-dimensional weave possessing manufacturing advantages such as pliability on curved contours. Although, delamination remains a potential failure attribute where the structural performance could be undermined. Few studies [143,205,223,306,307] have considered the stacking sequence of 5HS woven composites in detail with no investigation available on the influence of PPS interleaves on the fracture toughness of 5HS woven composites combined with various interfacial bias fibre arrangements. The work reported here aims to fill the gap by presenting a

systematic experimental investigation to understand the combined effect of dominant interfacial fibre orientation and PPS veils on the mode II ILFT of 5HS CFRP laminates.

3.4.2 Materials and Experimental Procedures

Pre-impregnated 5HS CFRP material was selected for the 3ENF tests and supplied by Cytec. The woven pre-preg consisted of 5HS woven carbon 3K Toray T1000 fibres, with MTM49 [304] toughened epoxy of 42% resin weight; an areal density of 283gm^{-2} , a nominal ply cure thickness of 0.33mm and related mechanical properties listed in Table.15. Balanced and symmetric outer plies of 5HS woven carbon fibre were prepared around a central layer of nonwoven polymer veil with two density variations for comparison against a non-interleaved control group using the same layup. The PPS veils supplied by Technical Fibre Products are Optiveil OP-49-48 [280], with a nominal areal density of 10gm^{-2} and thickness of 0.09mm. The veils were manufactured using a wet-lay process with chopped strand PPS fibres of $10\mu\text{m}$ diameter and length of 6mm in a styrene-acrylic binder.

Table.15 Mechanical properties of MTM49-3-42%-3KFT300B40B-5H-1000 [304].

Property	Result - Cytec / author	Standard
0° Tensile strength [MPa]	1065	ASTM D3039
0° Tensile modulus [GPa]	44.6	
90° Tensile strength [MPa]	1035	
90° Tensile modulus [GPa]	42.8	
0° Compressive strength [MPa]	640	SACMA SRM01R94
0° Compressive modulus [GPa]	59	
90° Compressive strength [MPa]	610	
90° Compressive modulus [GPa]	57	
In-plane shear strength [MPa]	108	ASTM D3518
In-plane shear modulus [GPa]	2.5	
0° Interlaminar shear strength [MPa]	64.2	ASTM D2344

Unlike plain or twill weave, 5HS fabrics lead to an inherent surface fibre dominance comparable to unidirectional (UD) or biaxial fabrics, however, also contain perpendicular non-dominant fibres. In 5HS fabrics, weft yarns pass over four warp yarn tows and under one, allowing for improved pliability compared with plain or twill weave fabrics. The mechanical properties of 5HS fabrics have minimal in-plane differences between the warp and weft directions as the reinforcement fibre count is balanced. This is supported by the mechanical properties in Table.15, presenting minimal differences between 0° and 90°.

Furthermore, it is worth reiterating that 5HS fabrics exhibit different dominant fibre orientations at each surface of the woven ply, which could affect the interlaminar properties. Fig.66 demonstrates the directional bias in the warp and weft direction for each surface when considering a 5x5 unit segment of the 5HS fabric: Face A is dominated by (4:5 or 80%) 0° fibre orientations while Face B is dominated by (4:5 or 80%) 90° fibre orientations. Table.16 lists the layup configurations of the 3ENF specimens with the dominant mid-plane fibre interfaces being 0/0, 90/90, 0/90, 90/45 and 0/45 at the midplane. Additionally, the influence of PPS veil of two areal densities (10gm^{-2} and 20gm^{-2}) at the mid-plane are explored. Fig.67 provides a visual representation of the fibre directional bias at the mid-plane interface to further signify the necessity of this investigation.

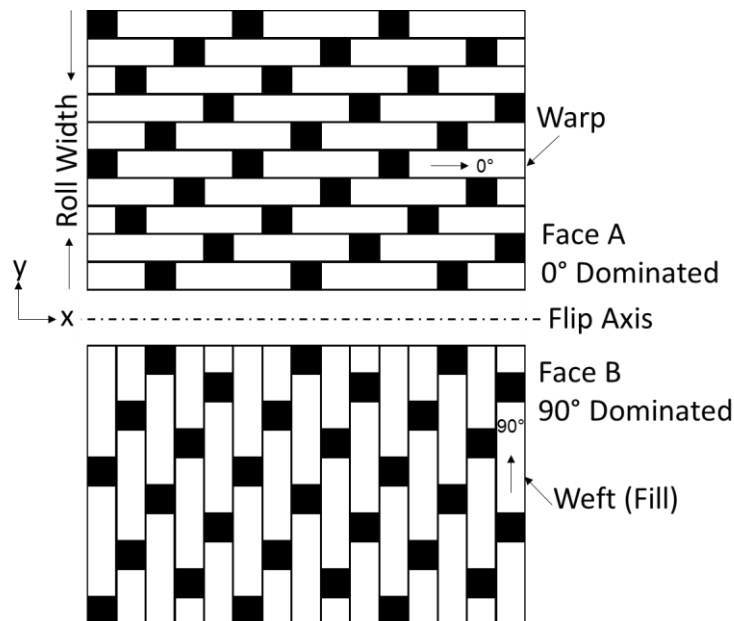
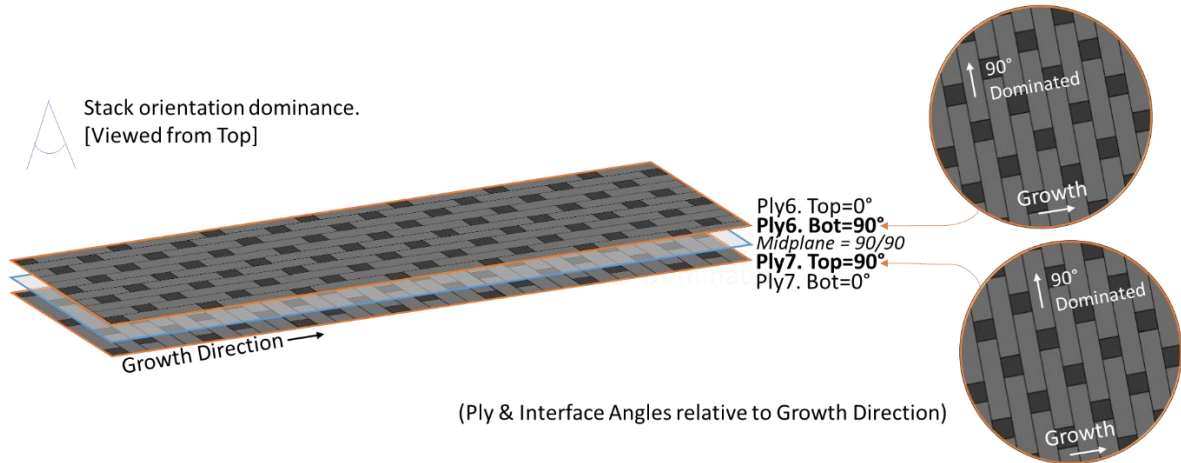


Fig.66 5HS CFRP laminae dominant fibre architecture.

Table.16 Mid-plane configurations for 3ENF tests.

Layup Configuration (Viewed from Top-Down)	Dominant Midplane Specimen Code
[90,0] ₆	90/90 (Fig.67)
[(90,0) ₃ , 1x veil, (90,0) ₃]	10V-90/90
[(90,0) ₃ , 2x veil, (90,0) ₃]	20V-90/90
[0,90] ₆	0/0
[(0,90) ₃ , 1x veil, (0,90) ₃]	10V-0/0
[(0,90) ₃ , 2x veil, (0,90) ₃]	20V-0/0
[0,90] _{3s}	0/90
[(0,90) ₃ , 1x veil, (90,0) ₃]	10V-0/90
[(0,90) ₃ , 2x veil, (90,0) ₃]	20V-0/90
[90 ₆ , -45, 90 ₅]	0/45
[90 ₆ , 1x veil, -45, 90 ₅]	10V-0/45
[90 ₆ , 2x veil, -45, 90 ₅]	20V-0/45
[0 ₆ , -45, 0 ₅]	90/45
[0 ₆ , 1x veil, -45, 0 ₅]	10V-90/45
[0 ₆ , 2x veil, -45, 0 ₅]	20V-90/45

**Fig.67 Example layup for 90/90 dominated mid-plane.**

3ENF specimens, as illustrated in Fig.68 were manufactured for the experimental investigation with the geometry following ASTM D7905-14 [54]. Fifteen configurations (Table.16) were manufactured under conditions subsequently described, with five consistent repetitions for each test and an example midplane shown in Fig.67 The 3ENF specimens had the final dimensions with the length equal to 180mm, a width of 20mm and thickness of 4mm. A single layer of nonwoven veil of 10gm^{-2} (Fig.67) was interleaved at the midplane for five layup configurations initiating at the edge of the $5\mu\text{m}$ PTFE release film. Another five configurations were produced similarly with two layers of veil building up to 20gm^{-2} . The remaining five configurations contained no veil interleaves and only the

PTFE release film for delamination initiation as the control group. Twelve 5HS weaved CF/Epoxy layers (with veil(s) - Table.16) were de-bulked under vacuum and subsequently placed on a flat steel plate with peel film, caul plate, breather fabric and cured inside a vacuum bag within the autoclave. The panels were cured under pressure at 620 KPa (90psi), with a target bag vacuum of <5kPa, ramp rates of 3°C/min and a dwell temperature of 135°C for 90 minutes as prescribed for MTM49[304].

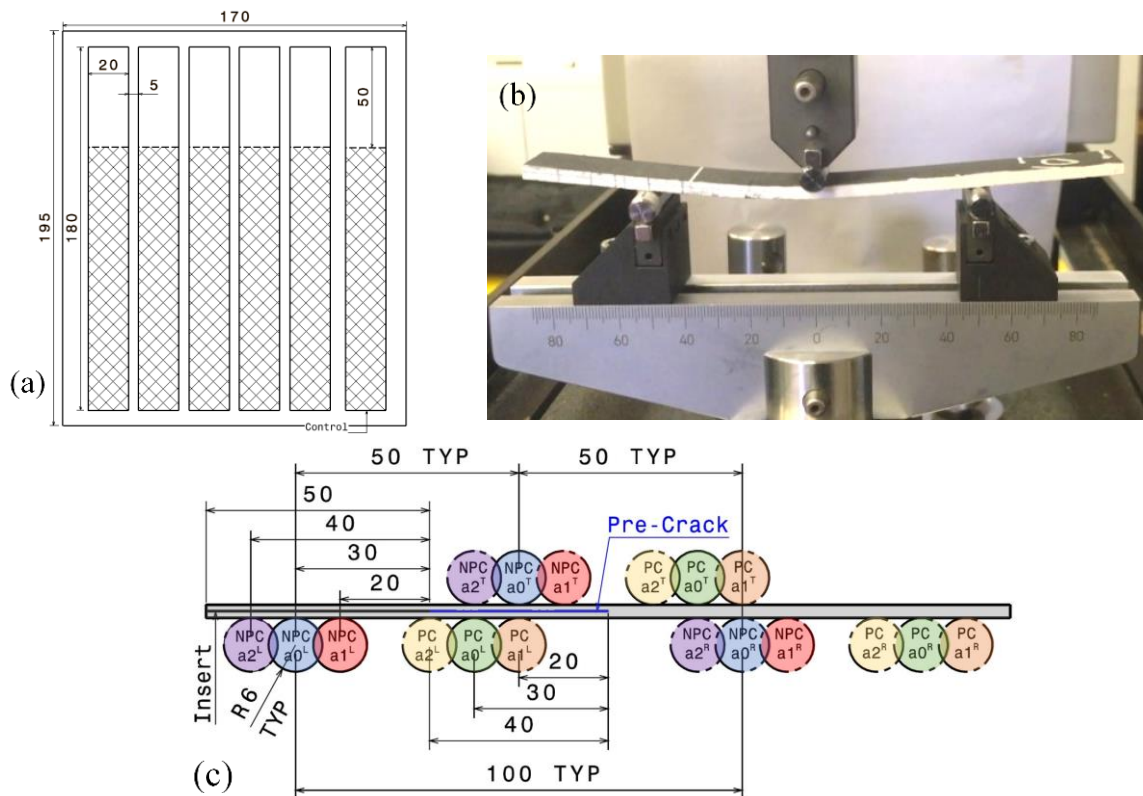


Fig.68 (a) Laminate geometry, (b) Loaded specimen, (c) 3ENF loading positions.

Under mode II loading, the G_{IIC} was evaluated as specified in ASTM-D7905/M-14 [54]. The layup configurations deviated from the standard for the fibre reinforcement type and mid-ply, namely the interfacial fibre orientation and veil interleaving effects, being the focus of the investigation.

During each test, specimens were positioned in a 3-point bending fixture with $\varnothing=12\text{mm}$ steel rollers as shown in Fig.68c. Loading in six positions provided the required information for Non-Pre-Cracked (NPC) and Pre-Cracked (PC) compliance calibration and fracture test results for each specimen. The specimen

edge was sprayed with a thin film of white paint to aid tracking of the delamination growth during specimen placement and fracture tests. Verification of crack front using ultrasonic scanning techniques were employed along with prying open specimens to verify and correlate crack length measurements. A Sonatest Veo 16:64 with a 64 linear element probe in a water tank was used for the detailed identification of delamination positions, providing accurate crack length measurements relative to the PTFE insert edge. Testing was conducted on a dual screw column servo-electric universal test machine (Instron 5965), equipped with a calibrated 5kN load cell. The crosshead displacement rate was set to 0.5mm/min to maintain quasi-static loading conditions and allow tracking of the delamination growth. The tests were also recorded using a 150mm macro lens and tripod-mounted DSLR camera for subsequent examination of the delamination events.

Compliance calibration (CC) tests were performed by loading the specimen to the peak CC force equal to 50% of the expected value of the critical force and then unloading. The loading roller was positioned so that the left roller was 20mm (a_1) and 40mm (a_2) behind the crack front (Fig.68c) respectively. The G_{IIC} values were obtained when the left roller was positioned at 30mm from the insert (NPC) and the pre-crack (PC) at the loading positions (a_0). The use of the same specimen for fracture tests has been previously demonstrated to produce accurate NPC and PC toughness parameters [54] and is followed within this study. The approach ensures that location differences between the location of the true and calculated crack tip do not affect the accuracy of G_{IIC} while utilising the compliance from the same specimen. For the fracture tests (NPC a_0 and PC a_0), the specimen is loaded until the delamination advances by visual observation on the specimen edge and by a load reduction on the force-displacement plot. The specimens were subsequently unloaded 0.5 mm/min, with force and displacement data recorded continuously at a sampling rate of 5 Hz.

For ILFT calculations, the CC coefficients from each NPC and PC tests were determined prior to the fracture tests (a_0), initiation values of G_{IIc} were later obtained from the maximum force (P). The CC coefficients, A and m , were determined using a least-squares linear regression analysis of the compliance, C , versus crack length cubed (a^3):

$$C = A + ma^3 \quad \text{Eqn.17}$$

where A being the intercept and m being the slope obtained from the regression analysis.

The mode II energy release rate, G_{II} , was determined using the compliance calibration (CC) relation specified in ASTM-D7905/M-14 [54]. Eqn.18 is the equation for G_{II} for specimens with constant width, B . Substituting the compliance relation from Eqn.17 into Eqn.18 and taking a partial derivative, yields Eqn.19, where G_{II} is a function of the CC constant, m .

$$G_{II} = \frac{P^2}{2B} \frac{\partial C}{\partial a} \quad \text{Eqn.18}$$

$$G_Q = \frac{3mP^2a^2}{2B} \quad \text{Eqn.19}$$

The mode II NPC a_0 and PC a_0 candidate fracture toughness (G_Q) was determined using Eqn.19, where m is the CC coefficient, P being the maximum load from the fracture test, a is the initial crack length, and B is the specimen width. The candidate toughness (G_Q) was determined and checked for validity as specified in ASTM D7905/M-14 [54] to be inferred as G_{IIc} .

Validity is determined if $15\% \leq \%G_Q \leq 35\%$, where $\%G_{Qj} = \left[\frac{100(P_j a_j)^2}{(P a_0)^2} \right]; j = 1, 2$ Eqn.20

The $\%G_{Qj}$ are the two values of G_Q associated with the compliance tests at $a_1=20\text{mm}$ and $a_2=40\text{mm}$ respectively. P_{a0} is the maximum load from the initiation of non-linearity of the fracture tests carried out when $a_0=30\text{mm}$ (Fig.69).

3.4.3 Results

3.4.4 Effect of interfacial fibre orientation and veil on load-displacement

The enclosed area under each load-displacement curve in Fig.69 represents the specimen's resistance to delamination, hence is closely related to the ILFT in mode II. Differences are observed in the area under the load-displacement curve among the two typical fibre orientations and veil interleaved configurations. The veil interleaved specimens reached non-linearity regions at a greater load with higher recorded displacements than for the non-interleaved specimens. Additionally, it can be observed from Fig.69 that 90° dominated mid-planes have a gradual load reduction during fracture, whereas the 0° mid-planes have a sudden growth. Between the various interfacial fibre orientations and veil configurations, a range of peak forces were observed during the fracture tests. The variations in maximum load imply a significant difference between the interfacial fibre-matrix interactions under mode II loading. As a result, also implying an increased level of energy dissipation during fracture propagation.

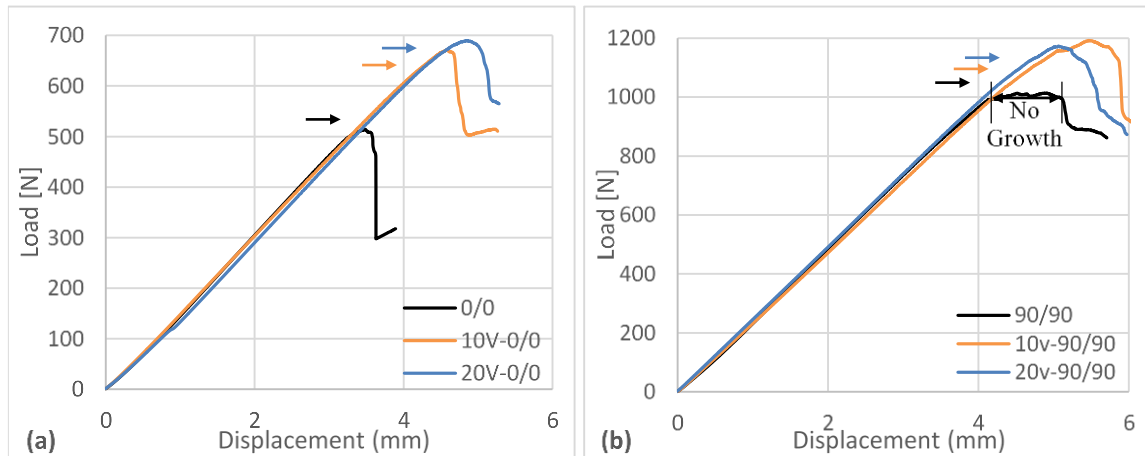


Fig.69 Typical load-displacement for PCa₀ test: (a) 0° bias, and (b) 90° bias midplane.

Note: Markers indicate the linearity end position where P_{Max} is determined.

For configurations other than 0/0 dominated mid-planes, the load carrying limit and energy required for fracture propagation is increased. Considering the shape of the load-displacement curves, it can be inferred that variations in the released energy are apparent. The shape of the load-displacement curves presents 0/0

mid-plane specimens exhibiting a sudden reduction of the load after the initial elastic deformation peak force (Fig.69a). The other configurations such as the 90/90 (Fig.69b), 0/90, 90/45 and 0/45 exhibit some further load carrying capacity beyond the peak initial elastic deformation force. The 90/90 non-interleaved specimens (Fig.69b) held a constant load with increasing displacement after the initial elastic region. This type of damage behaviour implies no propagation occurred during this stage and subsequently led to a sudden release of the stored fracture potential energy.

Similarly, other configurations exhibited gradual fracture growth trends, where veil interleaving had reduced the abrupt fracture growth behaviours. The effect of veil on G_{IIC} is however greatly dependent on the interfacial fibre orientation based on the observations in this work and previous studies [308]. The differences observed in the force and enclosed area (fracture energy) are related to damage mechanisms and subsequently presented energy within the examination of the fracture surface morphology. The specimens were also tested under 3-point flexural conditions at the tail-end before the compliance and fracture tests, giving an average flexural modulus of $E_0 \approx 42 \sim 45$ GPa.

A summary of the results for the influence of interfacial fibre orientation and PPS veil interleaves on the G_{IIC} are presented in Table.17. The greatest benefit to the G_{IIC} is within the interleaved 0/0 and 0/90 dominated fibre interfaces compared to the non-interleaved. The remaining fibre orientations also exhibit enhancements, with the minimum and maximum enhancement levels being: 21%-71% for 10gm^{-2} veil and 23%-75% for 20gm^{-2} veil interleaved.

Table.17 Influence of veil density on ILFT.

Dominant Mid-plane	G_{IIC} (0V) [J/m^2]		G_{IIC} (10gm^{-2} veil at mid-plane) [J/m^2]		δG_{IIC}	G_{IIC} (20gm^{-2} veil at mid-plane) [J/m^2]		δG_{IIC}
	Mean	StdDev	Mean	StdDev	10gm^{-2} Vs 0V	Mean	StdDev	20gm^{-2} Vs 0V
0/0	1055	84	1607	96	+52%	1630	49	+55%
90/90	1858	131	2364	119	+27%	2375	38	+28%
0/90	1007	51	1723	64	+71%	1760	12	+75%
90/45	1921	196	2327	172	+21%	2377	99	+24%
0/45	1811	181	2255	80	+25%	2228	143	+23%

A comparison of the G_{IIC} with the variation of dominant fibre orientations are presented in Table.18. It is apparent from the averaged G_{IIC} results (Table.17), there is greater variation by altering the dominant orientation of the interfacial ply than by veil interleaving alone. The fracture surface morphologies are subsequently studied for justifying the significant increases observed in the transverse dominant orientations.

Table.18 Influence of interfacial fibre orientation on ILFT.

Dominant Interface:	0/0 [J/m²]	90/90 (vs.0/0)	0/90 (vs.0/0)	90/45 (vs.0/0)	0/45 (vs.0/0)
G_{IIC} (0V)	1055	x1.76	x0.95	x1.82	x1.72
G_{IIC} (10gm⁻² veil)	1607	x1.47	x1.07	x1.45	x1.40
G_{IIC} (20gm⁻² veil)	1630	x1.46	x1.08	x1.46	x1.37

The ply bias orientation sensitivity is noticeably an influential factor in the ILFT, as presented in the G_{IIC} and load-displacement trends. Another observation is the limit and plateau of the level of enhancement when doubling the veil areal density from 10 gm⁻² to 20 gm⁻² veils (Fig.70). The convergence towards a plateau when using 20 gm⁻² veil suggests that negligible further improvement to the mode II ILFT is achievable by increasing the veil areal density from 10 gm⁻². A general reduction in G_{IIC} scatter for 20 gm⁻² veil configurations are also apparent as the fracture occurs within the defined veil-to-veil region when compared against non-interleaved and 10 gm⁻² interleaved groups. The plateau using 20 gm⁻² veils is further supported by considering the fracture surface morphologies, where fractures occur mostly within the veil interlayer. The overall observations suggest that veil interleaving indeed enhances the ILFT and delamination resistance, however, enhancements are dependent on the interfacial fibre orientation of the laminae as summarized in Fig.70.

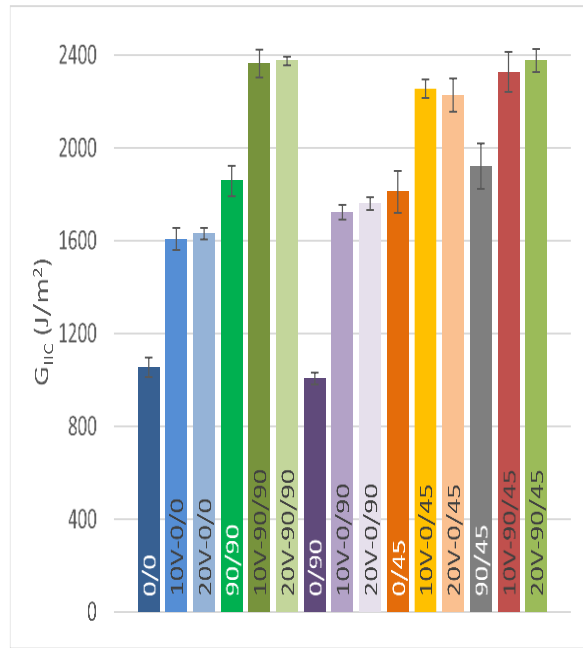


Fig.70 PCa₀ - G_{IIC} enhancements for 5HS CFRP midplanes with veil configurations.

3.4.5 Fracture morphology at interfaces of varying fibre orientations with veil interleaves

Dissemination of the results are considered from the fracture surfaces including; texture, fibre bridging observations, localised fractures, resin-rich regions and fibre orientation factors. Although the fracture surfaces (Fig.71-Fig.75) reveal highly localised fibre peeling and breakage, intra-laminar fractures and migration tendencies [308] are not possible in 5HS laminae due to the weave architecture. The mechanisms related to the differences in fracture energy are subsequently demonstrated to be associated with the frictional differences of the fibre orientation and veil enhanced bridging at the tow level. The fracture surface morphology of the specimens were examined under Scanning Electron Microscopy (SEM); a Zeiss Evo 50 Digiscan II with a 15kV beam and emitter distance between 6.5mm to 9mm.

Specimens containing the 0/0 dominant midplane are examined in Fig.71, where non-interleaved (Fig.71a) and interleaved (Fig.71(b-c)) specimens predominantly exhibit interlaminar delamination growth at the 0/0 midplane as expected. A detailed examination of the fracture surface reveals that interleaved interfaces

exhibit a coarse and irregular texture (Fig.71(b-c)) indicating the veil contribution. The increased surface roughness and fibre-matrix interactions can explain the average 52% and 55% improvements in G_{IIC} through the PCa₀ fracture test for 0° dominant interfaces (Table.17). A significant outcome from the results is the rise in G_{IIC} for 5HS laminates containing PPS veil interleaves and implies an increase in energy dissipation during propagation. The presence of broken fibres tows, resin remnants and striations indicate the veil had a significant influence on the increased G_{IIC} . The enhanced fracture resistance and improved interface bridging are consistent with those observed in UD laminates utilising veil interleaves [21,26,36,208,308]. The interleaving enhancements commonly observed in UD laminates are also applicable to 5HS laminates based on the current results; however, the relative interfacial fibre orientation should also be considered [308].

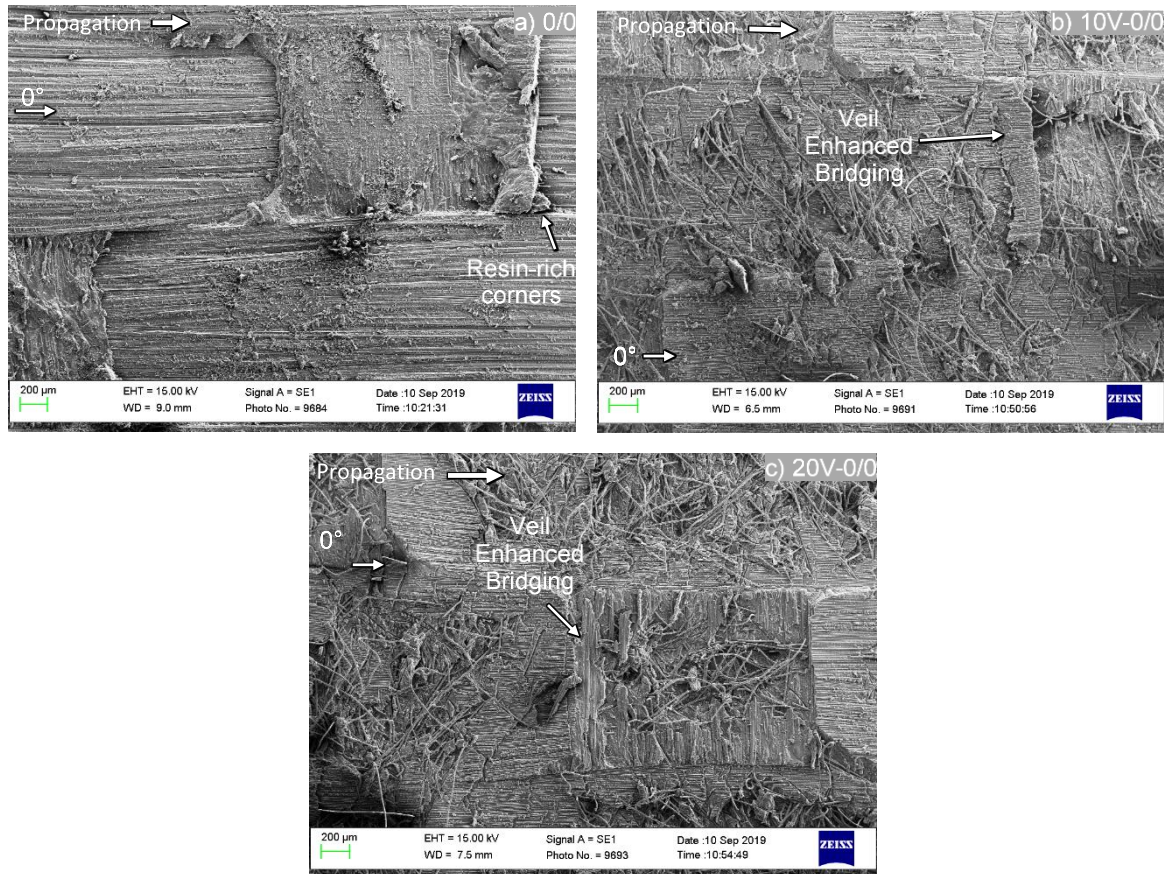


Fig.71 Fracture surface of 0/0 dominated mid-plane: (a) without veil; (b) 10gm⁻² veil; (c) 20gm⁻² veil.

Fracture surface morphology for specimens with the 90/90 dominant midplane fibre orientations are presented in Fig.73. Compared to the dominant 0/0 fibre midplane, the 90/90 dominated midplane also has sizable improvements in fracture resistance and further indication of greater fibre-matrix interactions for the interleaved interfaces. The resulting G_{IIC} of 90/90 dominant interfaces compared to 0/0 are enhanced by 1.76 times for non-interleaved, 1.47 times for 10 gm^{-2} and 1.46 times for 20 gm^{-2} interleaved (Table.17).

The arrangement of the dominant 90° fibres ahead of the delamination path can cause resistance to the approaching delamination front. Laminae interfaces with transverse fibres relative to the crack path have similar behaviours as observed in UD laminates without the global intra-laminar migration tendencies [123]. The lack of migration in woven laminae is due to the fibre weave architecture and an increased fibre friction coefficients of the transverse tows. The resulting enhancements from the veil and improved interlaminar fracture resistance are visible from the fibre imprints, breakages, peeling, matrix striations and general surface roughness of the interleaved specimens (Fig.73(b-c)). These enhancements and fracture surface observations allow further possibilities for optimisation of laminates utilising 5HS laminae by exploiting the friction differences of oriented interfaces.

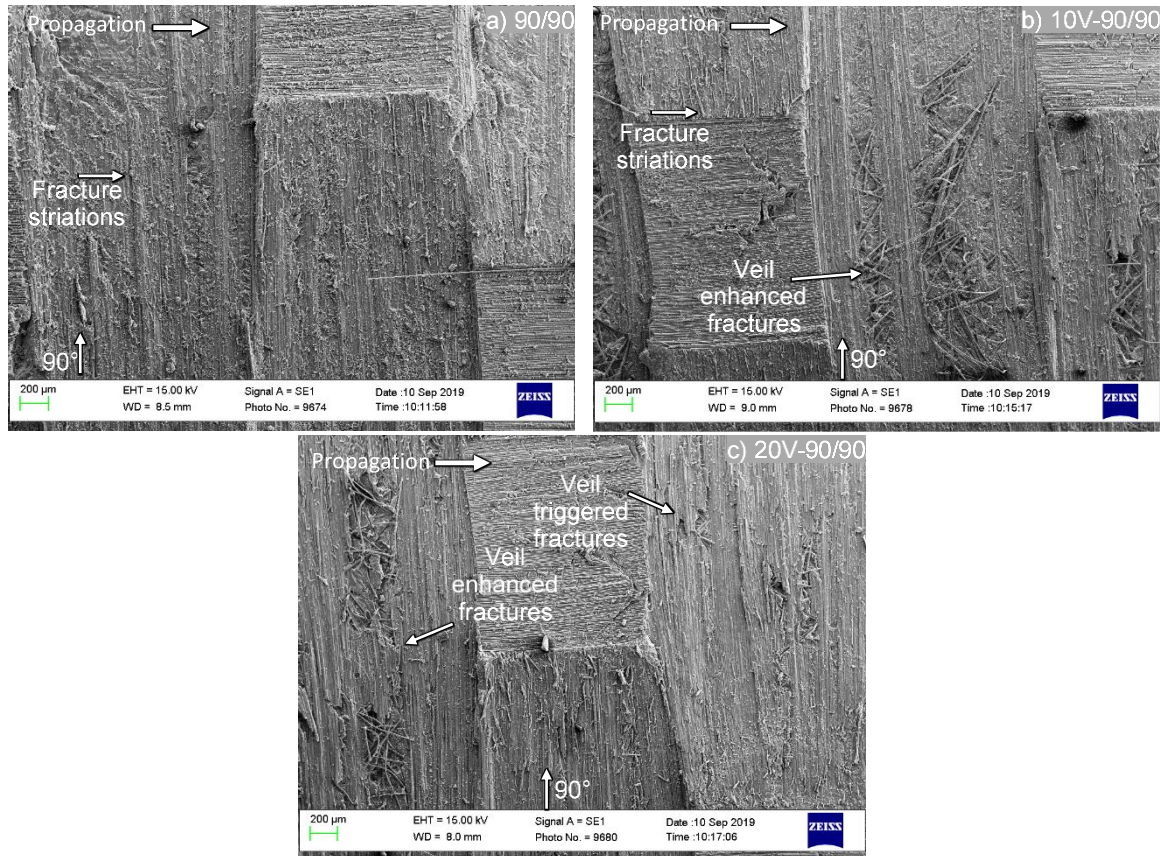


Fig.73 Fracture surface of 90/90 dominated mid-plane: (a) without veil; (b) 10gm⁻² veil; (c) 20gm⁻²veil.

Fracture surfaces for the 0/90 dominated midplane are presented in Fig 74 with observations similar to the 0/0 and 90/90 mid-planes. The non-interleaved (Fig 74(a)) and interleaved (Fig 74(b-c)) 0/90 specimens exhibit localised fibre peeling and breakages. Improvements are lower for the G_{IIC} without interleaving for the 0/90 dominated midplane compared to the 0/0 midplane, where contributions could be explained due to variation in the ply-to-ply alignments. On this consideration, resin-rich zones are also apparent in Fig 74a, effectively reducing the G_{IIC} when compared against the 0/0 and 90/90 dominated mid-planes. Extended fibre fracture for the interleaved 0/90 dominated midplane are also observed (Fig 74(b-c)), demonstrating an improvement in the fracture propagation resistance. The enhancements in G_{IIC} for the 0/90 dominated midplane with interleaved veils can be explained by the increased surface roughness, reduction in resin-rich areas and improved fibre-matrix interactions.

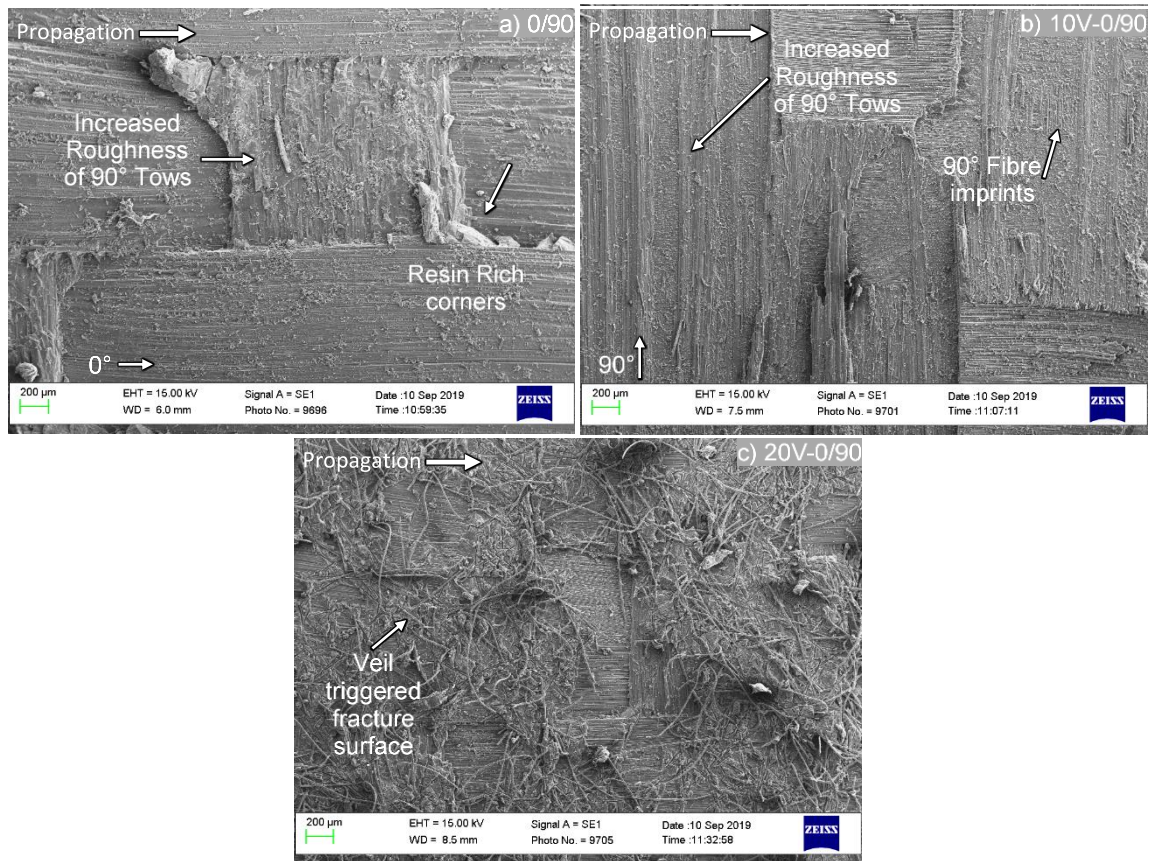


Fig 74 Fracture surface of 0/90 dominated mid-plane: (a) without veil; (b) 10gm⁻² veil; (c) 20gm⁻² veil.

The fracture surface morphology for the 90/45 dominated midplanes were also examined for the non-interleaved (Fig.75(a)) and interleaved specimens (Fig.75(b-c)). The interleaved 90/45 dominated midplane specimens have greater surface irregularity with fibre pull out and fractures compared to the non-interleaved mid-planes. Fibre-matrix bonding and shear load transfer between the angled layers of 90/45 allow for increased propagation resistance when having 90° interfacial fibres. Increases in fracture resistance for 90/45 dominated interfaces with interleaves (Table.17) is justifiable from the level of increased fibre damage (Fig.75(b-c)). 90/45 dominated mid-planes have fracture resistances up to 1.82 times greater than 0/0 dominated mid-planes for non-interleaved, 1.45 times for 10 gm⁻² and 1.46 times for 20 gm⁻² interleaved. The source of the increase (Fig.70) when using the off-axis 45° and 90° dominated plies at the interface can trigger local tendencies for mixed-mode II/III cusp type features as shown in Fig.75b. The results demonstrate that fracture resistance of laminates can be improved not only by veil interleaving but also by changing interfacial fibre orientations if no other compromises emerge relating to the mechanical properties and stiffness of the laminate.

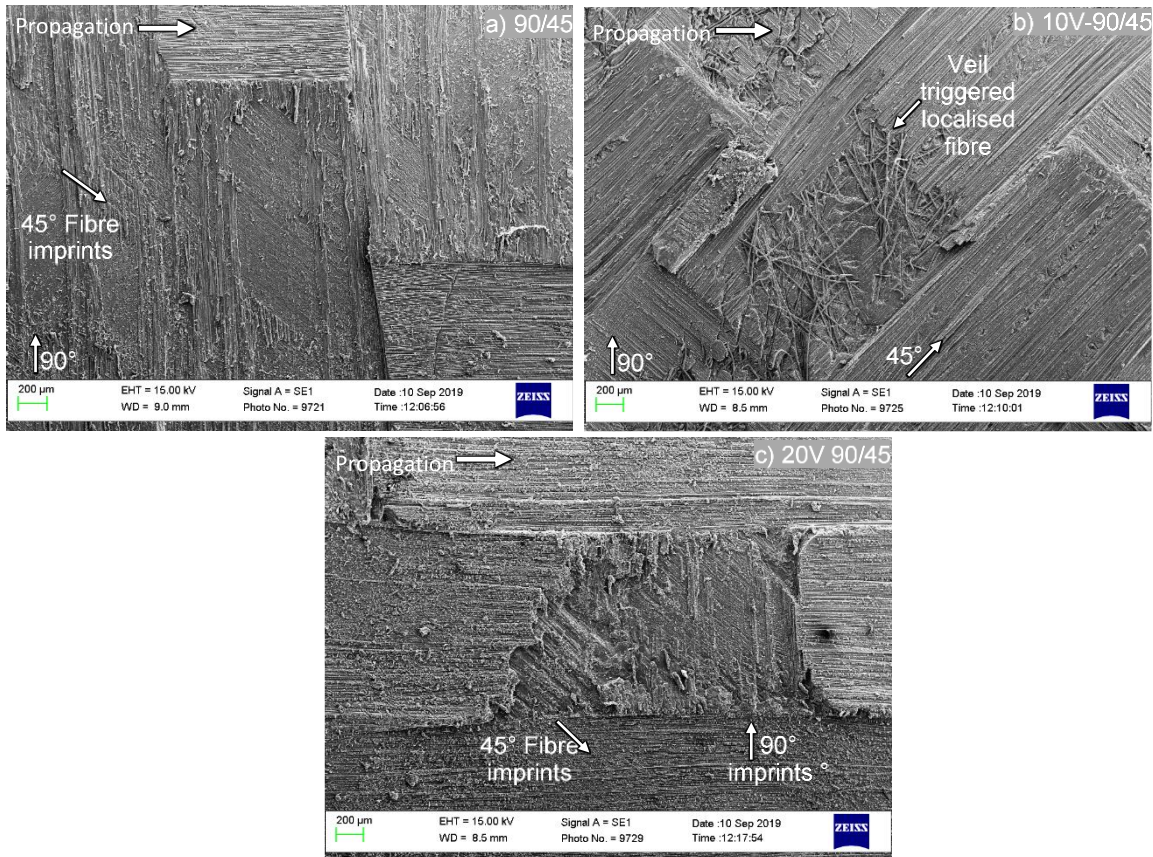


Fig.75 Fracture surface of 90/45 dominated mid-plane: (a) without veil; (b) 10gm⁻² veil; (c) 20gm⁻² veil.

Fracture surface features for the 0/45 dominated midplane (Fig.75) appear similar to those at the 90/45 mid-planes. The fracture surface morphology of the 10 gm⁻² interleaved specimens exhibits significant surface irregularities, observable fibre breakages, fibre-resin imprints and enhanced fibre bridging. The 20 gm⁻² interleaved specimens exhibit further indications of the cavernous fibre-matrix interactions, which demonstrate resistance to the fracture initiation and propagation. Interleaving the 0/45 dominated mid-planes formed a 25% enhancement using the 10 gm⁻² veil and 23% enhancement using 20 gm⁻² veil in the average G_{IIC} (Table.17). Compared to the G_{IIC} from 0/0 dominated mid-planes, the 0/45 dominated interfaces are 1.72 times greater for non-interleaved, 1.40 times greater for 10 gm⁻² interleaved and 1.37 times greater for 20 gm⁻² interleaved specimens. The source of the increase is a result of the off-axis 45° dominated interfacial ply against the 0° (directing) ply which triggers local tendencies for mixed-mode II/III cusp and imprint morphologies as shown in Fig.75b.

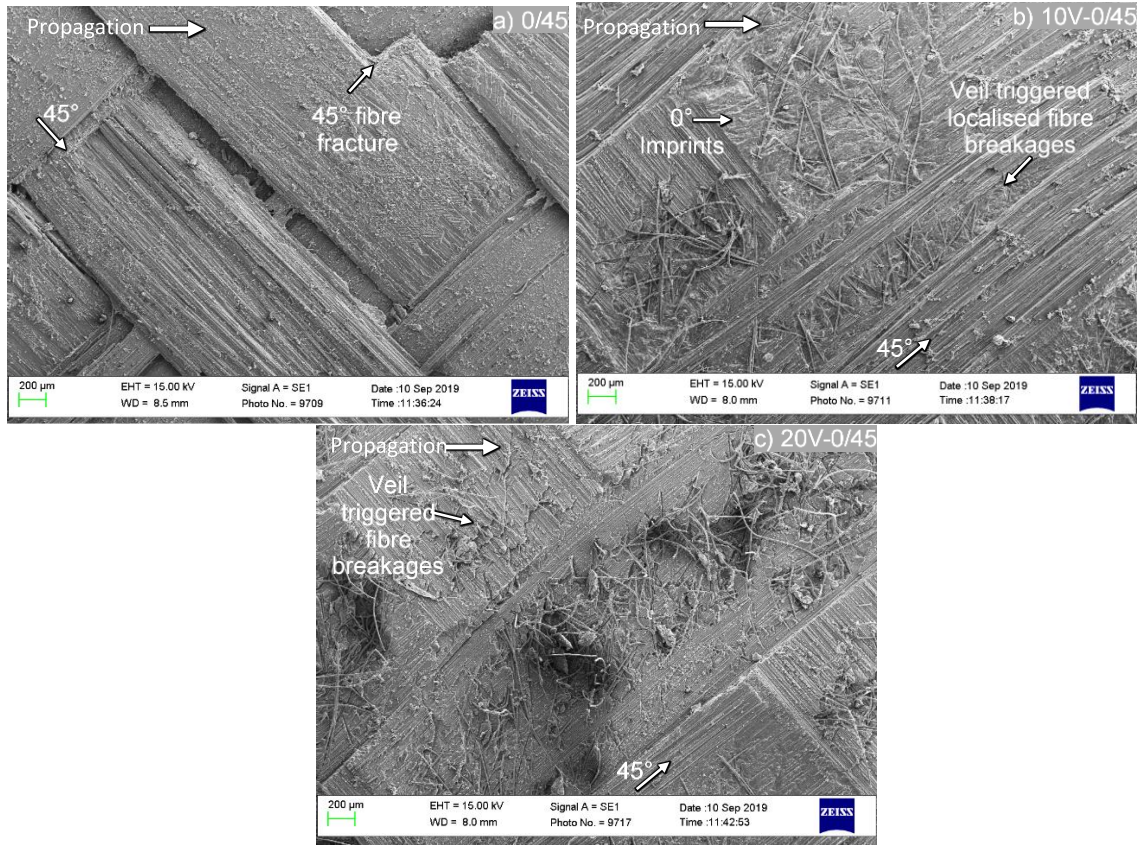


Fig.75 Fracture surface of 0/45 dominated mid-plane: (a) without veil; (b) 10gm⁻² veil; (c) 20gm⁻² veil.

3.4.6 Discussion

The delamination growth process and ILFT of 5HS weave CFRP laminates for multi-directional layups have been affected by many factors. Previous work relating to 5HS laminates with off-axis interfaces [143,205,223,306,307] revealed variances in the interlaminar fracture resistance. Whereas this work examines the combined influence of biased interfacial fibre orientation with veil interleaving for toughness considerations. The results emphasise the vast influence of dominant fibre orientation and demonstration of friction variations within woven laminates associated with fibre level interactions. The details and summary of enhancements on interlaminar fracture resistance are presented in Table.17-Table.18 and Fig.70 with clear applicability for equivalent G_{IIC} scaling predictions.

The SEM images from the delaminated surfaces of the specimens presented similar trends on upper and lower surfaces having identical fracture features. The fracture surfaces with veils exhibit coarse and irregular morphologies, whereas specimens without veils exhibit a smooth texture and suggest a reduction in fibre friction-related damages. The surface irregularity and interlaminar damage features are found to be highly dependent on the interfacial fibre orientation as presented in this work. The existence of shear hackles and fibre-matrix imprints were observed, which represent intense shear deformation due to shear stresses occurring locally. The distinct differences in shear triggered fractures and detailed fibre level interactions when using transversely dominant plies could provide reasoning for an enhanced ILFT. The fracture surface morphology of non-interleaved specimens for off-axis dominated plies (90° and 45°) revealed fibre matrix striations with increased fibre pull-out and localised fractures. In comparison to the 0° dominant interface (being a directing interface), off-axis interfaces can trigger localised mixed-mode II/III cusps and shear hackle features as demonstrated in the fracture morphology.

The results demonstrate that testing of non-interleaved and interleaved laminates following the prescribed 0° orientation as per D7905/D7905M – 14 [54] is inadequate to understand the detailed multidirectional interlaminar damage behaviours. Additionally, the need for woven fabric test standardization are apparent, to better understand the fracture mechanisms in multidirectional woven laminates with toughening interlayers. Additionally, damage modelling commonly disregards local migration tendencies, fibre related interactions and influence of interleaving, however, these are crucial to understanding the progressive damage in laminated structures. The fracture toughness values obtained from the multidirectional 3ENF tests can be suited for simulating laminated structures for the orientations and veil interleaving configurations considered in this study.

The understanding developed from the combined testing of interfacial fibre orientation and interleaved veil on the mode II fracture resistance can aid in revealing further improvements in woven 5HS laminae. Design flexibility and controllability of the G_{IIC} by manipulation of interfacial fibre orientation and by addition of PPS veils could be avenues of future exploration for CFRP applications. The industrial relevance to the damage tolerance design optimisation and modelling strategy of laminated woven composites can be subsequently explored based on the outcomes of the investigation.

3.4.7 Conclusions

The mode II interlaminar fracture behaviour, the influence of dominant fibre orientation, presence of PPS veil interleaves and weave structure on the delamination resistance of 5HS woven fabric CFRP laminates have been explored in this work. Specimen stacking configurations for fifteen mid-plane interfaces (Table.16) were investigated to study the combined effects with the following conclusions being drawn:

- Interlacing between the warp and weft yarns leads to a greater fracture resistance from the transverse (90°) interfaces which promote fibre-matrix de-bonding and localised fibre fractures resulting from their increased friction interactions.
- The fracture behaviour presented different growth behaviours where longitudinal (0°) dominant interfaces presented a sudden growth after the peak elastic deformation whereas transverse interfaces fractured gradually.
- For two-dimensional fracture growth, the interface anisotropy can result in elongated delamination profiles based on the differences in fracture resistance owing to the interfacial fibre orientations.
- Veil interlayers can increase the initiation ILFT ranging from 21% to 75% depending on the interfacial orientations.
- The veil areal density marginally influenced the ILFT. However, areal density related enhancements are demonstrated to be significant dependant on the opening mode and interfacial fibre orientation [25,26,223,308,309].
- A reduction in GIIC scatter when interleaved with the 20 gm^{-2} veil implies the fracture toughness of the PPS veil is recorded instead of the parent 5HS interface.

REFERENCES

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4 Performance of Carbon Fibre-reinforced plastics under LVI with Veil Interleaves

4.1 Influence of Veil Distribution on the Delamination of Multidirectional CFRP Laminates under Single and Repeated Low-Velocity Impact.

Pages 157-180

4.1

4.1 Influence of Veil Distribution on the Delamination of Multidirectional CFRP Laminates under Single and Repeated Low-Velocity Impact

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Abstract: This paper focuses on the effect of impact damage response of unidirectional carbon fibre reinforced polymer laminates interleaved with non-woven polyphenylene sulfide (PPS) veils under low-velocity impact (LVI). Tests were conducted under LVI on a baseline layup of (0/90)_{7s}, modified with various levels of veil interleaving at the middle and outside plies under single and repeated impact. Results demonstrate that veil interleaving can reduce and alter the damage from inter-laminar to intra-laminar for veils positioned in the outside plies. Repeated impact tests located at the same impact site under the same energy level demonstrate the ability for the veil to inhibit the fracture growth process where it is present. Furthermore, the interleaved configurations show a change in slope after the Delamination Threshold Load (DTL), implying the level of damage is lower than those with fewer quantities of the veil and positioned at the mid-level. Exciting changes in the impact force history is also observed where a smoothing effect of the DTL is achieved through interleaving and secondary impact. Fracture migration effects are also reduced where the veils are present, hence providing significant knowledge on the details of fracture migration and damage characteristics when using CFRP laminates with interleaved PPS veils under LVI.

Keywords: A-Carbon fibres; B-Delamination; B-Impact Behaviour; C-Interfacial Strength; C-Damage tolerance.

4.1.1 Introduction

Carbon Fibre Reinforced Polymer (CFRP) laminates have been increasing in use for lightweight structural applications. Their desired strength-to-weight proportions and electrochemical tolerance permit lightweight and robust structures, leading to reduced environmental impact and operating costs associated with fuel consumption. Conversely, their meagre impact tolerance is a vital problem for the future expansion and deployment, especially under Low-velocity Impact (LVI) [73], which are often difficult to detect, owing to the internal damage and impact sites being Barely Visible [44]. LVI can occur unexpectedly

during manufacturing, servicing and during maintenance of a component. The damage is mostly under the barely visible category if the impact energy is low [44,45]. In order to improve CFRP's performance and post-impact mechanical properties such as their in-plane compressive strength and interlaminar performance, toughening methods have been developed.

Toughening using thermoplastic additives to improve the interlaminar toughness and fibre-matrix enhancements exist; including fibre surface treatments, commingling with tougher fibres [13–19], interleaving [20–26] and through-thickness Z-pinning [27–30]. Using interleaving techniques, it is possible to introduce the characteristics of thermoplastics to modify the interlaminar properties at bespoke locations within a laminated structure. This work aims to primarily investigate the inclusion of PPS veil under single and repeated impact with various veil layup configurations. The laminate plates for this investigation are designed to consider the effects of interleaving at the outside and the mid interfaces in various quantities. The details of the layup are given in Table.20, where mid interfacial plies are present; the effect of shear loading will be considered near the neutral axis (NA). The inclusion of interleaves at outside interfaces takes into consideration direct and bending stress related effects during the impact. The novelty of this study compared to the work in literature is the identification of ideal locations for veil applicability and identify mechanisms involved for veil interleaving influencing the fracture migrations.

4.1.2 Materials and Experimental Procedures

The pre-impregnated CFRP material for the impact tests was supplied by SHD Composites, consisting of UD carbon 3K Toray T300 fibres, with MTC510 toughened epoxy system of 37% resin weight [290]. It has an areal density of 150gm^{-2} , a nominal ply cure thickness of 0.143mm with relevant mechanical properties listed in Table.19. To investigate the effect of veil influence under LVI, PPS veils were positioned at various interfaces as listed in Table.20 with a control non-interleaved configuration for comparison. The veils, Optiveil OP-49-48 [280] were supplied by Technical Fibre Products, with a nominal areal density of 10gm^{-2} and pre-cured thickness of 0.09mm. Veils were manufactured using a wet-lay process with chopped strand PPS fibres of $10\mu\text{m}$ diameter and length of 6mm in a styrene-acrylic binder.

Table.19 Mechanical properties of MTC510- UD150-HS-37% RW [290].

Property	Result	Standard
Fibre Density [g/cm ³]	1.8	N/A
Matrix Density [g/cm ³]	1.21	N/A
Fibre volume fraction [%]	53.4	N/A
Cured ply thickness [mm]	0.143	N/A
0° Tensile strength [MPa]	2112	BS EN ISO 527-5
0° Tensile modulus [GPa]	110	
0° Poisson's ratio	0.34	
90° Tensile strength [MPa]	54	
90° Tensile modulus [GPa]	8.2	
90° Poisson's ratio	0.01	EN 2850 Type B
0° Compression strength [MPa]	988	
0° Compression modulus [GPa]	105.1	
90° Compression strength [MPa]	200	
90° Compression modulus [GPa]	9.3	ASTM D3518
In-Plane shear strength $\pm 45^\circ$	99	
In-Plane shear modulus $\pm 45^\circ$	3.6	BS EN ISO 14130
Interlaminar shear strength	85	

Twenty-four single impact and nine repeated impact specimens were fabricated under identical manufacturing conditions for each layup as listed in Table.20. The LVI specimens were rectangular plates with the dimensions of 150(L)×100(W)×4.0mm(H). For the interleaved configurations, the Optiveil was positioned at various locations emanating from the midplane and others with veil towards the top and bottom surfaces. Twenty-eight UD CFRP layers were cold consolidated in groups of 4 plies under vacuum during the layup and subsequently placed on a flat steel plate with release film, caul plate, breather fabric and cured within a vacuum bag in the autoclave. The panels were cured under pressure at 620 KPa (90 psi), with a target bag vacuum of less than 5 KPa, and temperature setpoint of 100°C for 4 hours as prescribed for MTC510 [290]. The impact tests were assessed under the same principle as specified in ASTM-D7136M-15 [52], with essential deviations in layup configurations for studying the influence of veil distribution. Unlike the quasi-isotropic layup configuration of [45/0/-45/90]_{4S} as specified in [52], a single interfacial fibre arrangement of 0/90 repetitions, as shown in Table 2 were manufactured for the LVI specimens.

The layup configurations used in this study are designed to capture the damage response and impact force history while keeping the baseline CFRP layup as $[0/90]_{7s}$, thereby minimizing variations from the fibre alignment and orientation related reinforcement. The control (C) group are designed to benchmark the impact response and damage characteristics with no veil interleaves. M3 to M11 include interleaves propagating outward of the mid-plane with 3 to 11 veil interlayers to capture the interlaminar shear response. O6 to O12 have interleaves at the outside inter-layers having 6 to 12 interleaves respectively to capture direct and bending stress effects.

Table.20 Layup configurations for LVI and repeat LVI specimens.

Code	N° of veils	Stacking sequence (V= 10gm ⁻² veil insertion)
C*	0	$[0/90]_{7s}$
M3	3	$[0/90]_6/0/V/90/V/90/V/0/[90/0]_6$
M5*	5	$[0/90]_6/V/0/V/90/V/90/V/0/V/[90/0]_6$
M7	7	$[0/90]_5/0/V/90/V/0/V/90/V/90/V/0/V/90/V/0/[90/0]_5$
M9	9	$[0/90]_5/V/0/V/90/V/0/V/90/V/90/V/0/V/90/V/0/V/[90/0]_5$
M11	11	$[0/90]_4/0/V/90/V/0/V/90/V/0/V/90/V/90/V/0/V/90/V/0/V/90/V/0/[90/0]_4$
O6*	6	$0/V/90/V/0/V/90/[0/90]_5/[90/0]_5/90/V/0/V/90/V/0$
O8	8	$0/V/90/V/0/V/90/V/[0/90]_5/[90/0]_5/V/90/V/0/V/90/V/0$
O10	10	$0/V/90/V/0/V/90/V/0/V/90/[0/90]_4/[90/0]_4/90/V/0/V/90/V/0/V/90/V/0$
O12	12	$0/V/90/V/0/V/90/V/0/V/90/V/[0/90]_4/[90/0]_4/V/90/V/0/V/90/V/0/V/90/V/0$

Note: C=Control, M=Mid interfaces and O= Outside interfaces.

** Also considered for repeated impact.*

The CFRP laminates were subjected to LVI using an Imatek drop-weight tower (Fig.77) with a calibrated Kistler 9331B in-line load cell attached to a 25mm semi-hemispherical impactor tip with Kistler C3008 signal amplifier for data acquisition. The total mass of the striker and carriage assembly was 4.275 Kg, and striker mass was 0.345 Kg. The resulting average velocity of 3.35m/s for an impact-energy of 25J was determined. Test specimens were aligned centrally and clamped using rubber-tipped toggle clamps onto the impact fixture as recommended by ASTM-D7136 [52]. The drop-weight tower also incorporated an anti-rebound feature triggered by laser and pneumatic controls to prevent multiple rebound impacts. The laser gate sensor activates data acquisition when the failing carriage crosses the sensor only milliseconds before the impact.

For consideration of repeated impact nearby the impact site, a second and third impact were also performed to observe any further damage potential at the same energy level. Repeated impacts were conducted for the C, M5 and O6 layup configurations. For the repeated impacts, one condition was to impact the same location two times for under a same energy level and another with a location shift by 2mm in x and y-axis. A third condition was to impact three impacts at the same location and energy level for comparison of the impact force history and damage evolution. The impact load, duration, velocity and acceleration were obtained for further analysis in an attempt to gather significant and novel impact force history understanding when using veil interleaves at various interfaces.

To assess the damage and make a comparison, ultra-sonic-scanning was employed for all composite panels using a Sonatest veo 16:64© with a 64 element linear probe in a water tank. Ultrasonic scan images were post-processed and the damaged area calculated to identify the variations resulting from the interleaved veil placement and repeated impacts.

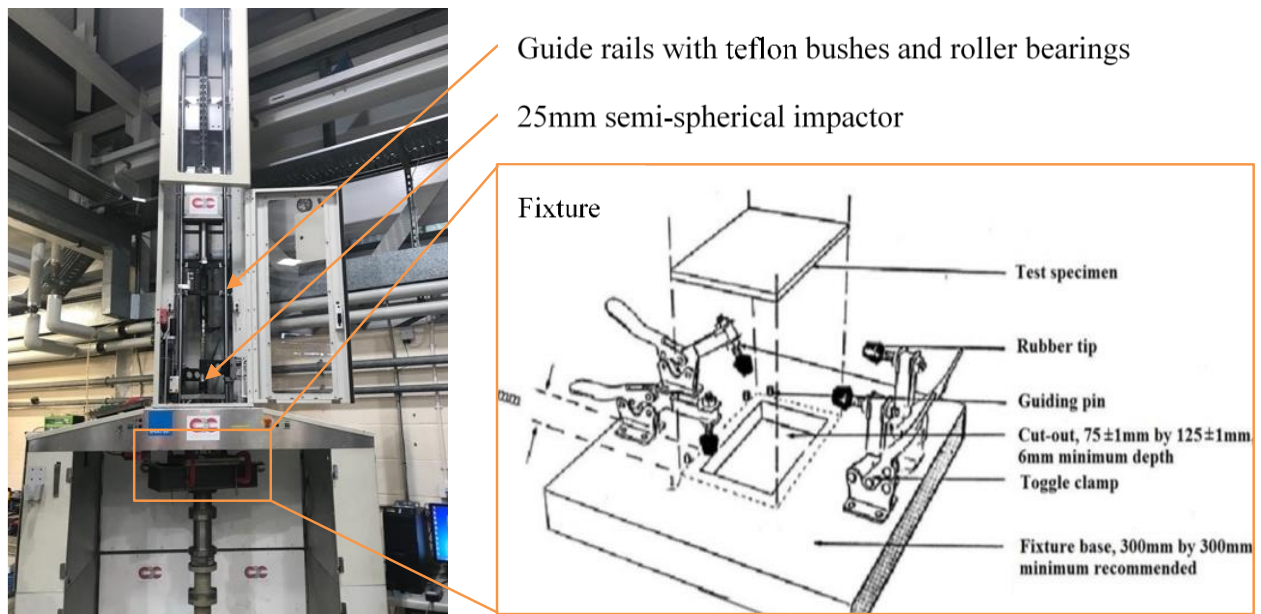


Fig.77 Drop weight tower set-up and clamping fixture.

The tested specimens after ultrasonic scanning were cut and wet polished up to 1200 grit at the impact site epicentre and observed using an optical microscope (Leica DM2700) and Scanning Electron Microscope (SEM) (TESCAN Lyra3). The SEM scanning was made possible with thin film deposition of Gold using the sputter coating process and observed using secondary beam electrons, and a backscatter capture technique.

4.1.3 Results

4.1.3.1 Impact force history – Single Impact

The following results present an overview of the loading history for all configurations considered. Fig.78(a&c) presents the impact force history for the middle and outside interleaved configurations, respectively. Whereas, Fig.78(b&d) presents the force-displacement during the LVI event for the middle and outside interleaved, respectively. Initial observations show that the impact responses are very consistent between repetitions and a clear difference exists for the various interleaved configurations.

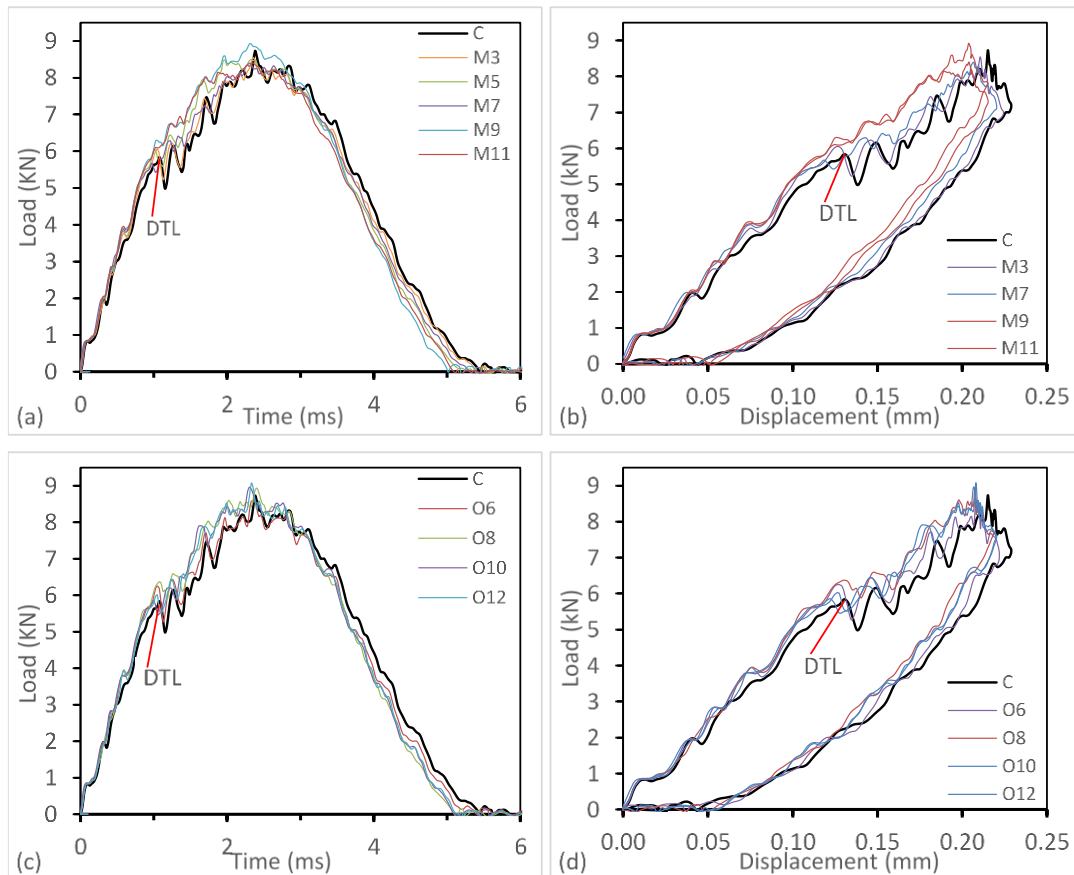


Fig.78 Impact force history (left) and load-displacement (right) for mid-interleaved (a-b) and outside-ply (c-d).

The first significant abrupt load drop during impact is recognized as the Delamination Threshold Load (DTL). It can be observed in Fig.78(a&c), and is followed by further progressive damage over the short impact duration. Increasing the veil count has reduced the severity of the abrupt load drops after the DTL point and almost completely smoothed for the M11 and O12 layup configurations. These observations are likely a result of stable damage propagation or a reduction in delamination magnitude.

The C-Scan analysis can help to understand the extent of the damage. A further method is to cut a strip from the impacted specimen at the impact site epicentre with a wet diamond circular saw and observe the cross-section under optical and electron microscopy. The DTL is approximately 5.92kN and occurs at the same instance for all configurations, albeit its local oscillatory magnitude is altered with the presence of veils. The summary of the key results is listed in Table.21, the results are consistent with those in literature, but the added influence of veil distribution is understated in terms of damage magnitude. The smoothing of the impact-force profile suggests otherwise; were cross-sectional microscopy (Section 4.1.5) highlights other significant factors brought about from the presence of veils; such as damage relocation and migration.

The load-displacement in Fig.78(b&d) also shows fascinating impact-force profiles, where the fluctuation in load-displacement during impact is reduced when using veils. The result, although not significantly apparent on the loading history, still demonstrated the capability of thermoplastic veil interleaving. The cross-section examination (Section 4.1.5) will subsequently show further details regarding the toughness enhancement with the presence of veils and their distributions.

Table.21 Summary of maximum results - Single Impact

Layup Code	Max Load kN	Velocity m/s	Acceleration g	Energy J	DTL Load kN	C-Scan Damage Area mm²
*C_Ave	8.75	3.36	203.43	24.60	5.84	687.0
M3_Ave	8.79	3.36	204.38	24.58	6.09	716.5
δ %	0.46%	-0.04%	0.47%	-0.08%	4.23%	4.29%
M5_Ave	8.64	3.36	200.96	24.65	5.99	736.5
δ %	-1.21%	0.11%	-1.21%	0.22%	2.43%	7.21%
M7_Ave	8.40	3.36	195.13	24.64	5.66	770.5
δ %	-4.06%	0.08%	-4.08%	0.16%	-3.21%	12.15%
M9_Ave	9.01	3.37	209.39	24.77	6.29	741.4
δ %	2.91%	0.34%	2.93%	0.68%	7.57%	7.86%
M11_Ave	8.58	3.36	199.39	24.66	Not	784.0
δ %	-1.97%	0.13%	-1.98%	0.26%	Visible	14.12%
O6_Ave	8.87	3.35	206.27	24.54	6.38	545.5
δ %	1.39%	-0.12%	1.40%	-0.24%	9.19%	-20.6%
O8_Ave	8.94	3.37	207.78	24.75	6.38	572.5
δ %	2.13%	0.30%	2.14%	0.61%	9.19%	-16.7%
O10_Ave	9.03	3.37	209.95	24.78	6.20	579.0
δ %	3.19%	0.37%	3.21%	0.75%	6.03%	-15.7%
O12_Ave	8.77	3.37	203.87	24.78	5.80	613.0
δ %	0.22%	0.37%	0.22%	0.74%	-0.82%	-10.8%

* δ %, With reference to C_Ave.

4.1.3.2 Impact force history –Repeated Impact

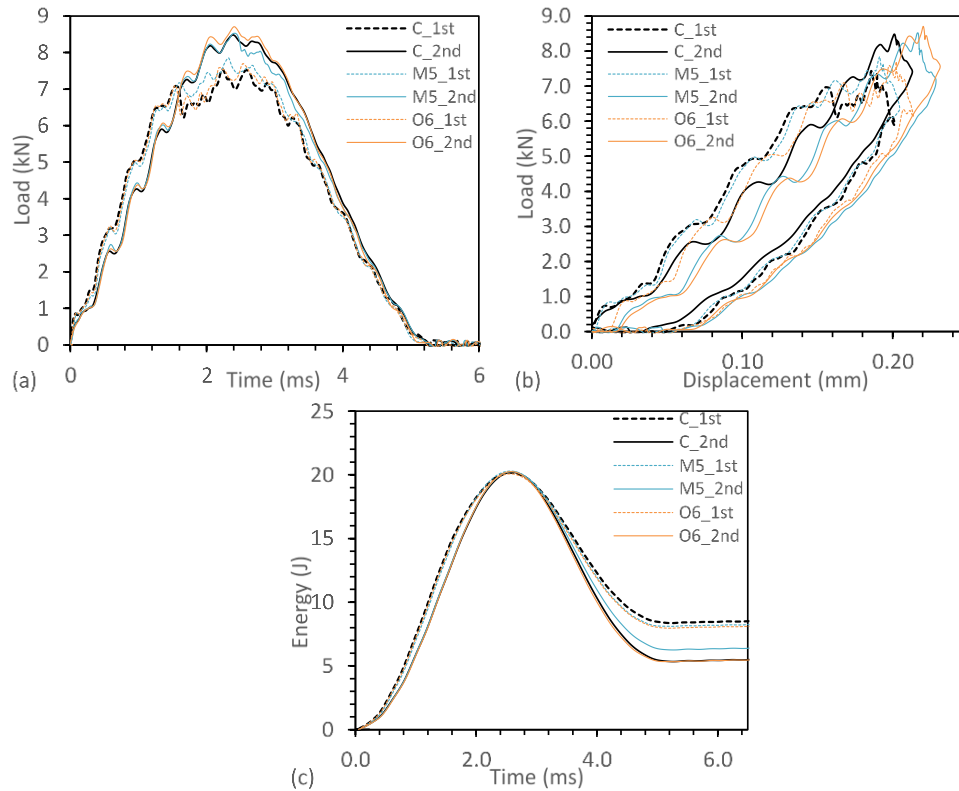


Fig.79 Repeated impact at the same location (a) Impact force history, (b) Load displacement, (c) Energy absorption.

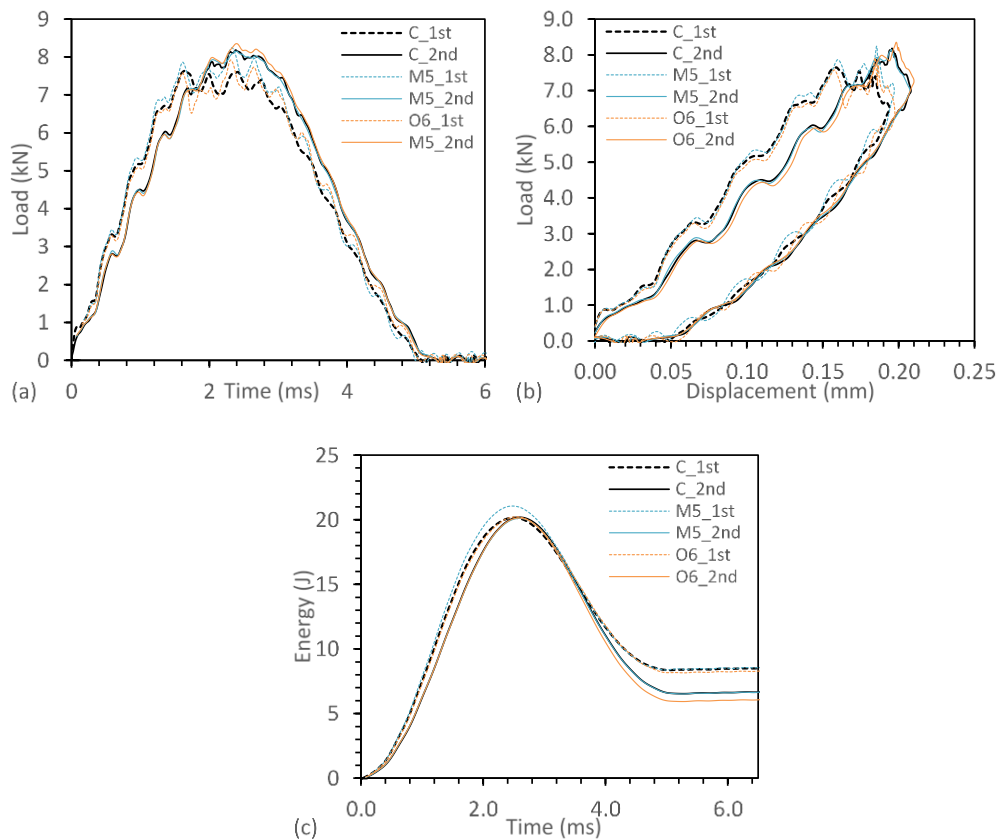


Fig.80 Repeated impact at 2mm x & y shift location (a) Impact force history, (b) Load displacement, (c) Energy absorption.

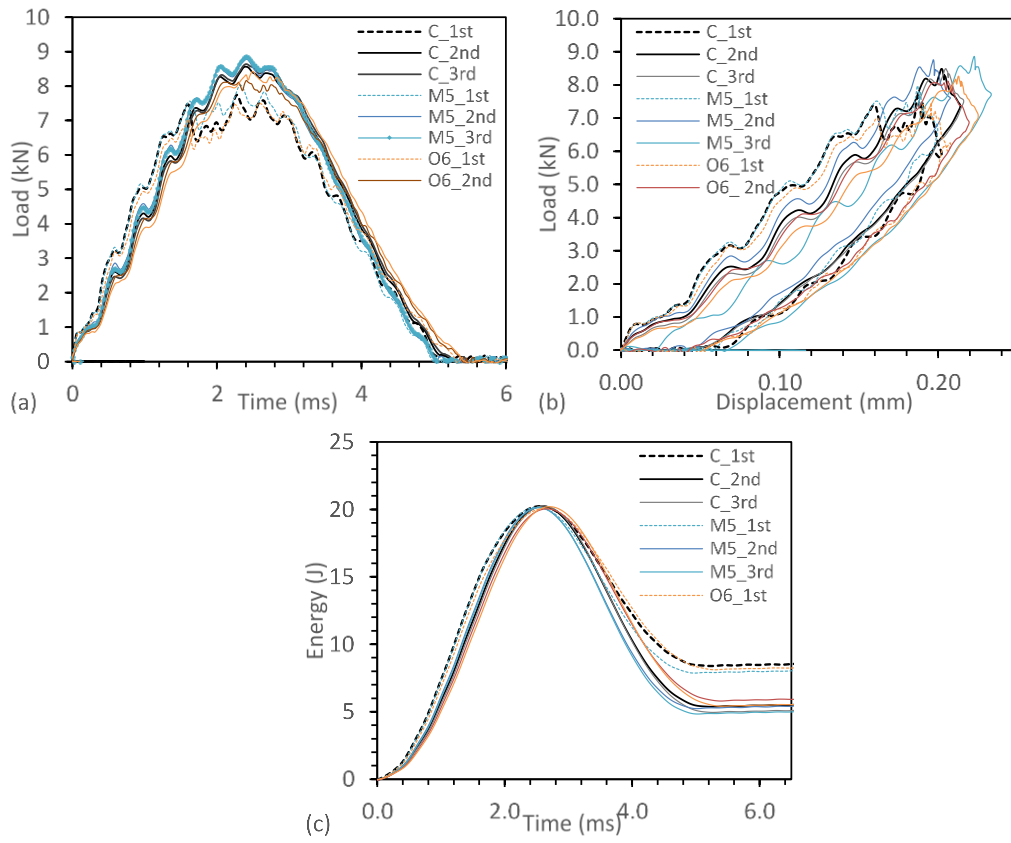


Fig.81 Repeated impact at the same location (a) Impact force history, (b) Load displacement, (c) Energy absorption.

Table.22 Summary of maximum results - Repeated Impact.

Repeated Impact	Layup Code	Max Load kN	Velocity m/s	Acceleration n g	Energy J	DTL Load kN	Damage Area mm ²
Double Impact Same Location	C-1	7.54	3.07	178.84	20.17	6.99	^a 731
	C-2	8.48	3.07	201.25	20.20	**	
	δ %	12.46%	0.08%	12.53%	0.17%	-	
	M5-1	7.84	3.08	186.07	20.30	**	849
	M5-2	8.52	3.08	202.32	20.25	**	
	δ %	8.69%	-0.14%	8.73%	-0.28%	-	
	O6-1	7.69	3.06	182.51	20.08	7.01	806
	O6-2	8.71	3.08	206.67	20.26	**	
	δ %	13.16%	0.45%	13.24%	0.91%	-	
Double Impact Shift Location 2mm in x & y	C-1	7.65	3.07	181.42	20.16	7.65	^b 765
	C-2	8.18	3.07	194.15	20.19	**	
	δ %	6.98%	0.07%	7.02%	0.15%	-	
	M5-1	8.24	3.14	195.65	21.07	7.81	875
	M5-2	8.15	3.07	193.29	20.11	**	
	δ %	-1.20%	-2.30%	-1.20%	-4.55%	-	
	O6-1	7.91	3.08	187.74	20.25	7.58	816
	O6-2	8.35	3.07	198.13	20.14	**	
	δ %	5.51%	-0.27%	5.53%	-0.54%	-	
Triple Impact Same Location	C-1	7.75	3.08	183.78	20.25	7.47	^c 822
	C-2	8.56	3.08	203.20	20.27	**	
	C-3	8.64	3.08	205.06	20.26	**	
	δ %	11.52%	0.04%	11.58%	0.07%	-	
	M5-1	8.03	3.07	190.53	20.20	7.52	894
	M5-2	8.78	3.07	208.36	20.17	**	
	M5-3	8.86	3.07	210.42	20.17	**	
	δ %	10.39%	0.10%	10.44%	0.19%	-	
	O6-1	7.51	3.07	178.18	20.11	7.01	859
	O6-2	8.16	3.07	193.68	20.08	**	
	O6-3	8.41	3.08	199.59	20.23	**	
	δ %	11.95%	0.36%	12.02%	0.72%	-	

* **δ %**, between impacts ** DTL Not Visible

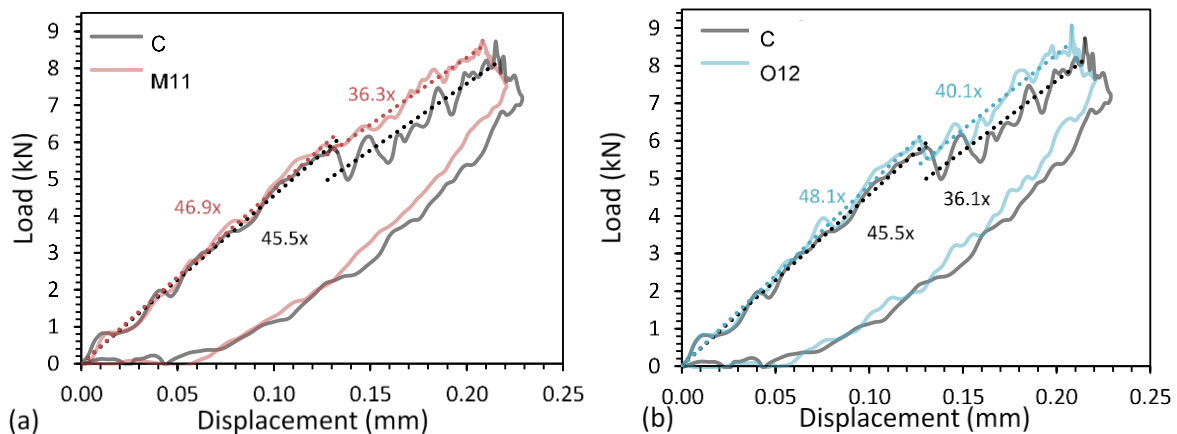


Fig.82 Load-Displacement comparison of maximum veils for mid-plies (a) and outside-plies (b).

The load-displacement curve for the 'C' configurations with the highest number of veil interleaves are presented in Fig.82. All curves present a change in slope between the initial loading and fracture loading after the DTL point of 5.92kN. It is also evident that veil interleaved specimens have a significant reduction at the DTL load-drop event compared to the 'C' specimens. Therefore, it is conclusive that interleaving contributes to the impact behaviour, although quantitatively the enhancements first appear minimal. The energy absorbed by the specimens is the area enclosed within each curve and can be quantitatively compared.

The post-DTL slope change for the 'C' configuration is from 45.5kN/mm to 36.1kN/mm and represents a 20.6% reduction. The change in slope represents a reduction in stiffness resulting from the initial matrix cracks and delamination during the progress of the damage during the impact. The veil interleaved configurations for the M11 configuration has a decrease of 22.6% whereas for O12 configuration the slope reduction is 16.6% after the DTL. The explanation for slope change when interleaves are positioned towards the outside interfaces are explained in Fig.97 where they are more effective on bending and direct loading. The differences appear insignificant at first; however, the presence of veils in the outside interfaces has limited the post-DTL stiffness change compared to mid-plane interleaved configurations. The slope before DTL appears higher in all instances where veils are present, indicating that the initial stiffness has increased slightly by interleaving with veils.

4.1.4 Damage size characterisation

Impacted specimens were assessed for damage as described previously using ultrasonic scanning techniques. Fig.84-Fig.85 presents the total through-thickness damage scans where any damage was observed. Variations exist for the extent of damage resulting from the veil positioning and veil quantity in comparison to the 'C' configuration. To link the extent of damage with the influence of veil distribution under LVI, ultrasonic scanning can aid in quantification for the internal damage.

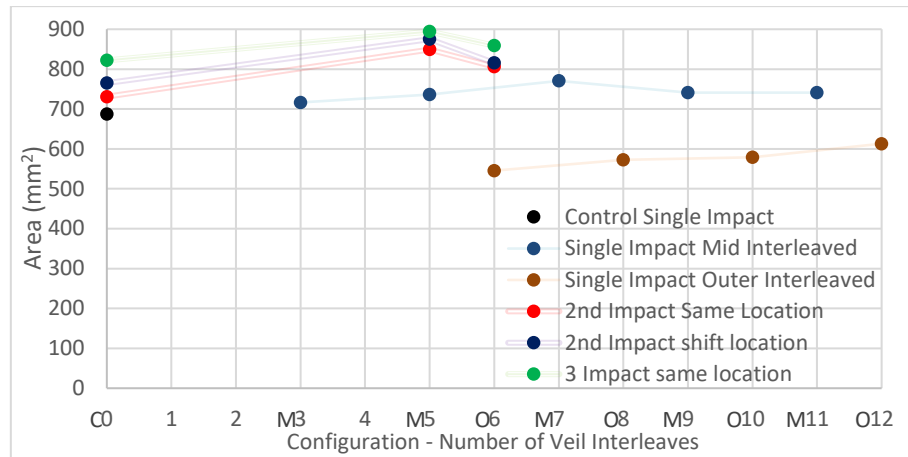


Fig.83 Damage area comparison against veil distribution.

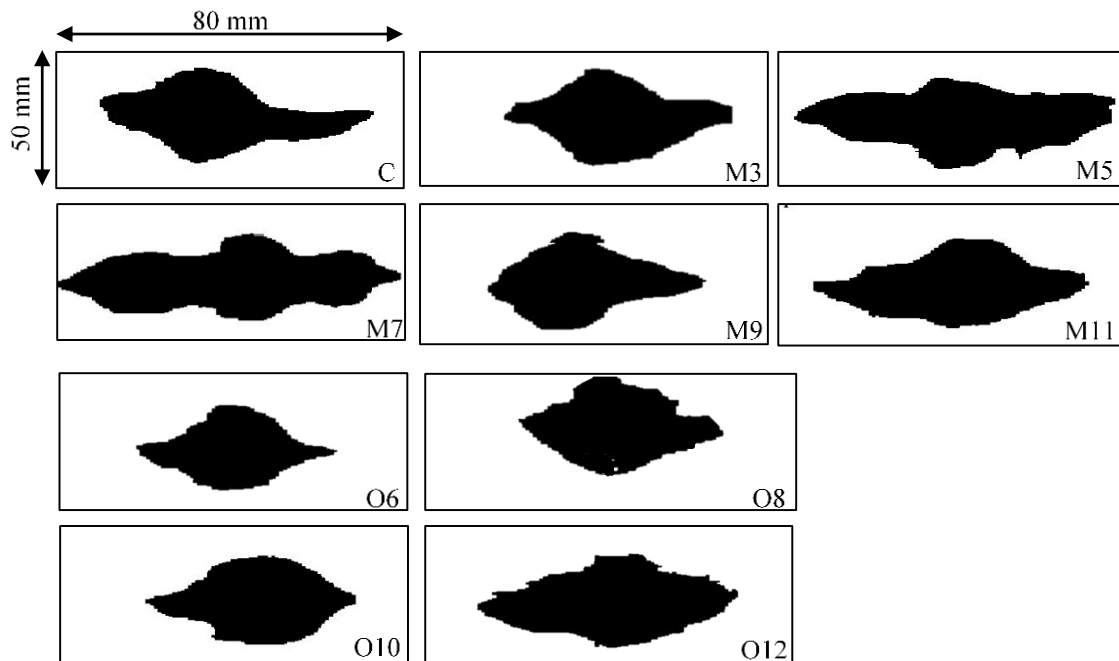


Fig.84 Typical damage area C-scan negatives for all single impact interleaved configurations.

The outside interleaved configurations have lower damage areas than that of the middle-interleaved configurations as observed in Fig.84. The shape of the extent of damage is significantly different, where outside interleaved configurations present oval-shaped damage and mid-interleaved having an elongated damage profile. The damaged area for the single impacts appear to be increasing with the increasing veil counts for outside interfaces and is contrary to the toughening

mechanisms commonly observed. The mid interfaces appear to have a fixed damage size compared to the outside interleaved configuration. The damaged area patterns are not, however entirely demonstrative of the impact damage extent when interleaved veils are considered. Differences in localised failure mechanism such as intra-laminar failure and migration cannot be captured with low-resolution ultrasonic scanning techniques. Therefore, care should be exercised when interpreting damage size due to the potential variance of damage modes. Additionally, the presence of strain concentration brought about from the interleaving may have a detrimental effect in causing intra-laminar failure. Further explanations supporting these observations are examined in the cross-sectional microscopy (Section 4.1.5).

Repeated impact specimens were also examined using ultrasonic C-scanning technique as presented in Fig.85 and summarised for damage area measurements in Fig.83. The impact damage resulting from repeated impacts at the same location appears to have increased the damaged area. The results suggest that repeated impact increases the extent of damage; however, the impact force history having a smoothed contour means the damage is less pronounced as if it were the first impact. For shifted location impacts, the extent of damage is also increased with similarities to the same location repeated impact.

For all repeated impacts, the mid-interleaved configuration appears to have more damage than that of the outside ply interleaving, suggesting that shear effects are more significant when veils are introduced towards the middle during the repeated impact. The veils have improved the damage extent when inserted at the outside plies, suggesting that direct and bending stress regions are more suited for veil interleaving.

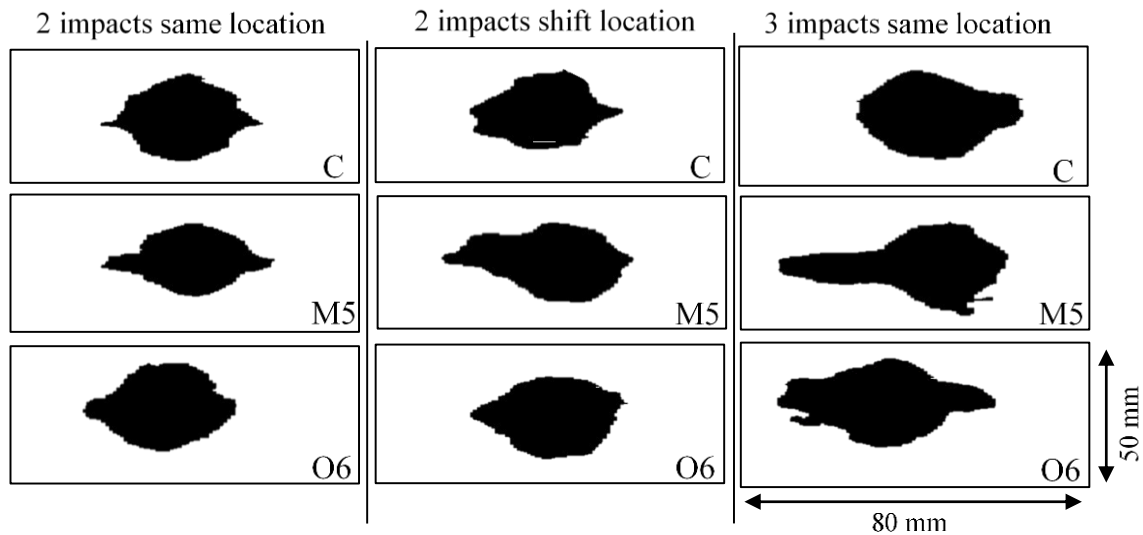


Fig.85 Typical damage area C-scan negatives for all repeated impact interleaved configurations.

4.1.5 Cross-sectional damage morphology

A Lyra Tescan SEM was employed to further study the internal damage cross-section for all configurations under single and repeated impact; to clarify the results of the ultrasonic scan measurements. The SEM also provides explanations for the impact force history response and identify which locations the veils are most suitable within the cross-ply layup configuration by considering localised fibre-matrix interactions.

4.1.5.1 Cross-section damage microscopy – Single Impact

The 'C' configuration on Fig.86 presents the delamination positions closest to the impact surface. The 'C' configuration has six levels of delamination where the most significant delamination is located at the 4th interface with a depth of 75 μ m. Fig.86b presents a detailed view of one of these positions where delamination occurred between a 90° and 0° ply. The SEM is consistent to the C-scan which presented elliptical damage profile along the 0° lengthwise fibre direction. The matrix cracks shown in Fig.86b, are propagating between the layers with the indication that they initiate as intra-laminar. This level of detail is significant as the veil interleaved fracture behaviour is different and shifted to purely intra-laminar.

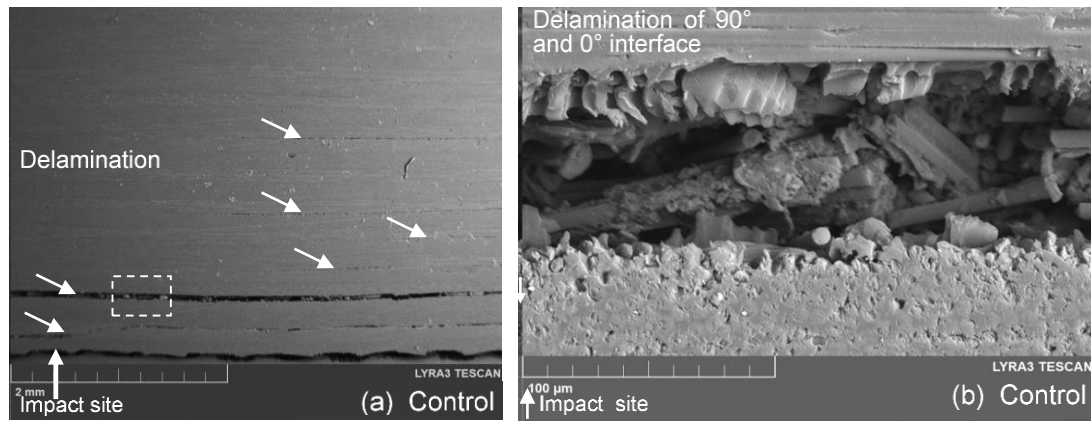


Fig.86 SEM of control configuration 'C'; (a) Overview, (b) magnified view.

The mid interleaved M3 layup configuration (Fig.87) has similar damage characteristics as the 'C' configuration, where primary failure is at the same depth from the impact site and mainly inter-laminar with some regions as intra-laminar. The damage location has not been affected in comparison to the 'C' configuration; however, the size has altered slightly for the interleaved configurations. The reason for the minimal change is that the veil interleaves are located towards the mid-plane near the neutral axis, hence only effective for the in-plane shear stress regions towards the mid-plane.

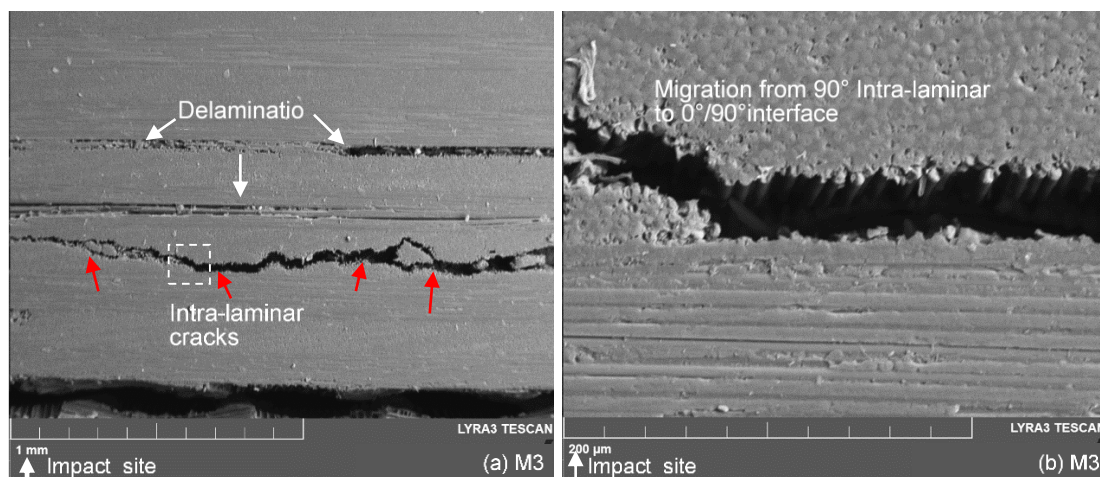


Fig.87 SEM of M3 configuration; (a) Overview, (b) magnified view of migration.

The cross-sectional fracture for the M5 layup configuration is presented in Fig.88. Fractures similar to those observed in M3 (Fig.87) are apparent, suggesting that veil interleaving is more likely to be effective at the outside interfaces for cross-ply layups as the tested conditions at 25J and a 25mm impactor.

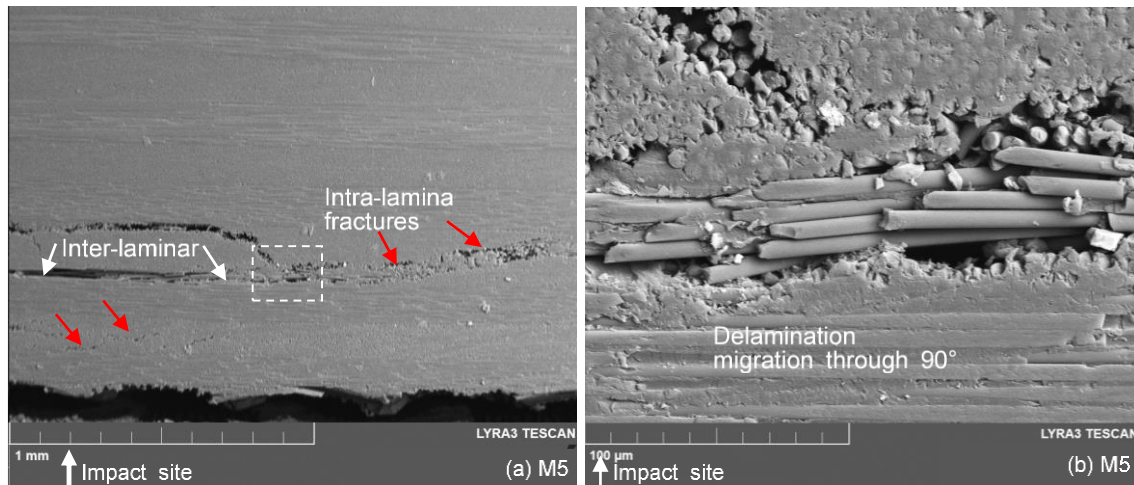


Fig.88 SEM of M5 configuration; (a) Overview, (b) magnified view of migration.

For the M7 configuration, Fig.89 presents an interesting finding where the number of failed regions have been reduced to three interfaces as inter-laminar. The depth of the largest fracture is less than the non-interleaved configuration and only 25 μm. The presence of large intra-laminar and migrating fractures is reduced and apparent in Fig.89a.

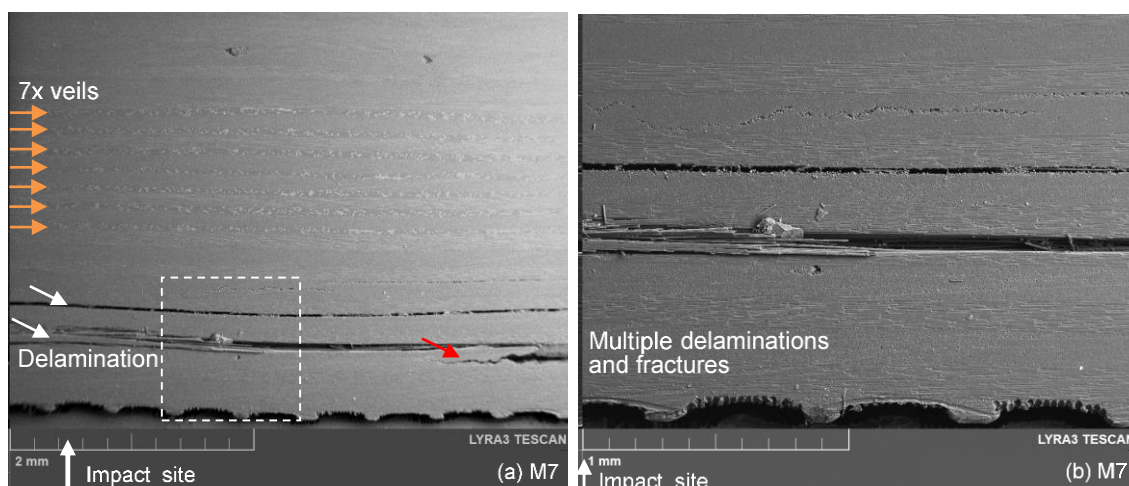


Fig.89 SEM of M7 configuration; (a) Overview, (b) magnified view of multiple delaminations.

For the M9 configuration, Fig.90 presents one of the two identical entirely delamination zones. No damage was found to be existent near the region of the veil interleaves. The largest delamination has a depth of 5 μm and is entirely an interlaminar failure as presented in Fig.90b. For the M11 configuration, similar fracture features as those found in M9 were observed.

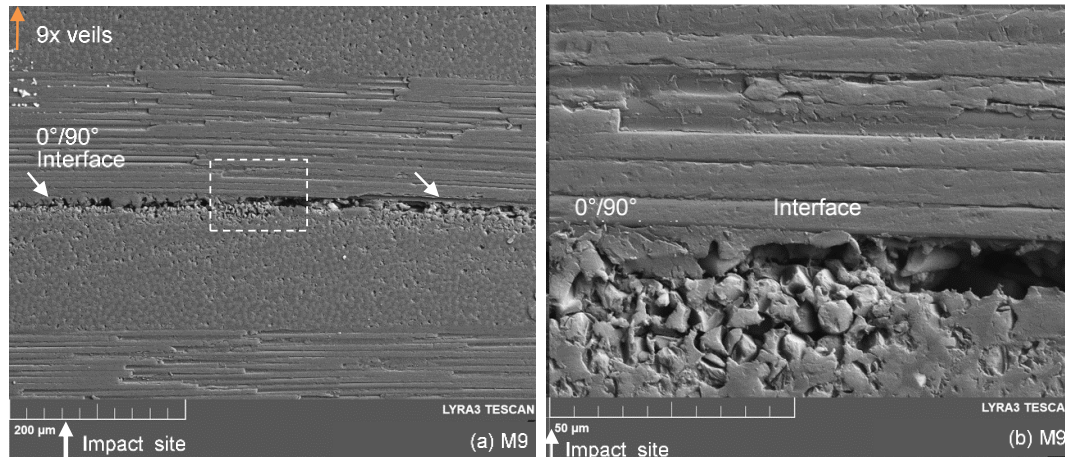


Fig.90 SEM of M9 configuration; (a) Overview, (b) magnified view of 0/90 interface delamination.

A different trend is observed for veils positioned at the outside interfaces where Fig.91 presents the O6 configuration. The exact location of the damage is within the 4th CF ply (90°) as predominantly intra-laminar failure below the impact site. The depth of this fracture being approximately 40 μm is considerably less than the 75 μm from the 'C' configuration (Fig.91). The damage was localised to this interface at this cross-section however the volume under the impact site may have other locations for three-dimensional consideration of the damage.

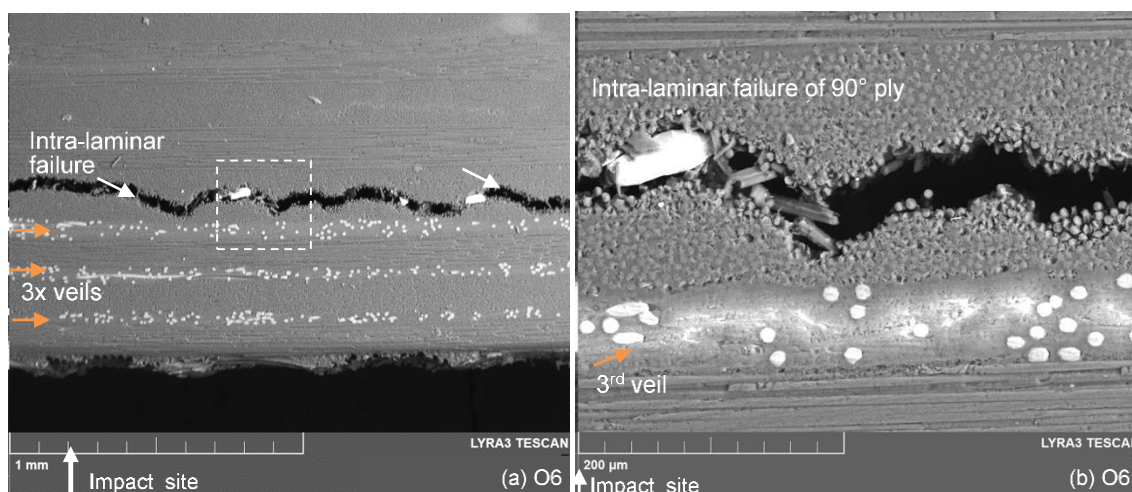


Fig.91 SEM of O6 configuration; (a) Overview, (b) magnified view of 90° intra-laminar failure.

The cross-section fracture observations for the O8 configuration is shown in Fig.92 where the fracture is intra-laminar within the 4th CF layer and surrounded by the PPS veils between the 3rd and 4th veil interleave. The fracture position has not moved in comparison to the 'C' configuration; nevertheless, interleaving appears to have distributed the fractures into many smaller matrix cracks. The other difference is that the fractures do not propagate through the veil interleaved interfaces and are unable to trigger interlaminar failure due to the PPS enhanced interface. Fig.92 showing the O8 configuration has intra-laminar fractures that are different to O6 in Fig.91; the critical difference being the strain concentration caused by the interleaved layers surrounding the failed layer on both faces of the 4th ply.

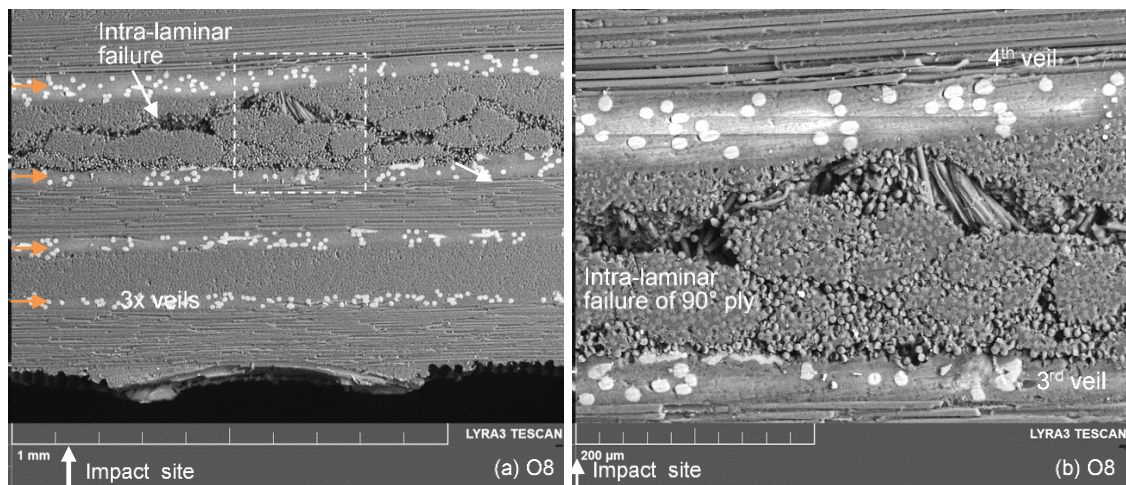


Fig.92 SEM of O8 configuration; (a) Overview, (b) magnified view of 90° intra-laminar failure.

The configurations O10 and O12 provide clarity into the number of interfaces affected by the impact being within the outside layers for the impact energy level of 25J using a 25mm semi-hemispherical impactor. For higher impact energy, the fractures may coalesce and form elsewhere, including the back face depending on the impact energy. Fig.93 shows the failure to be intra-laminar within the 4th ply and is consistent with that of the O6 and O8 configuration being intra-laminar dominant failure modes. The depth of the fracture is also reduced to 20μm which is a considerable difference to that observed at the 'C' and 'M' configurations.

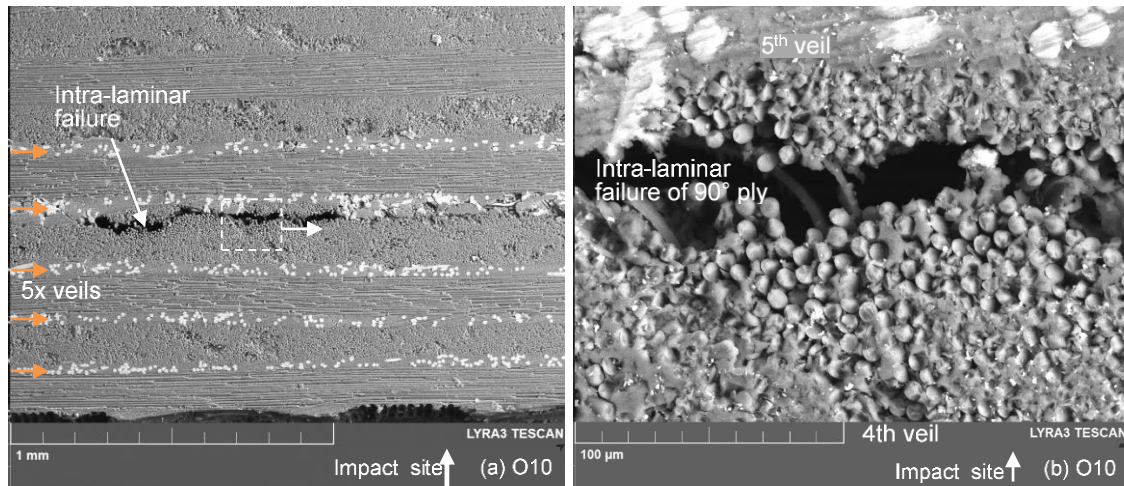


Fig.93 SEM of O10 configuration; (a) Overview, (b) magnified view of 90° intra-laminar failure.

4.1.5.2 Cross-section damage microscopy – Repeated Impact

As examined under the single impact, a Lyra3 Tescan SEM is also employed to examine the cross-sectioned impact damage for the repeated impact samples. For 'C' configurations, impacted twice at the same location (Fig.94(a&b)) the following differences are observed: two delamination positions near the surface exist alongside a deep matrix crack. Similar observations exist for the twice impacted specimens with a 2 mm planar shift in x and y (Fig.94(c&d)). The shifted position may have exaggerated the damage growth and is supported by the C-scan data (Fig.83 & Fig.85) along with the observations using the SEM. The repeated impact with a shifted location for the 'C' configuration has a few delamination zones with a large matrix crack at the 90° second layer, the damage depth and span is more pronounced than the previous test at the same location and against the original impact condition. For the triple impact condition at the same location for the 'C' configuration, the specimen appears to have similar fracture features to that impacted twice at the same location. The deep intra-laminar feature for the triple impact (Fig.94(e&f)) appears to be present at the 4th ply (90°) with an inter-laminar failure between the 8th and 9th ply.

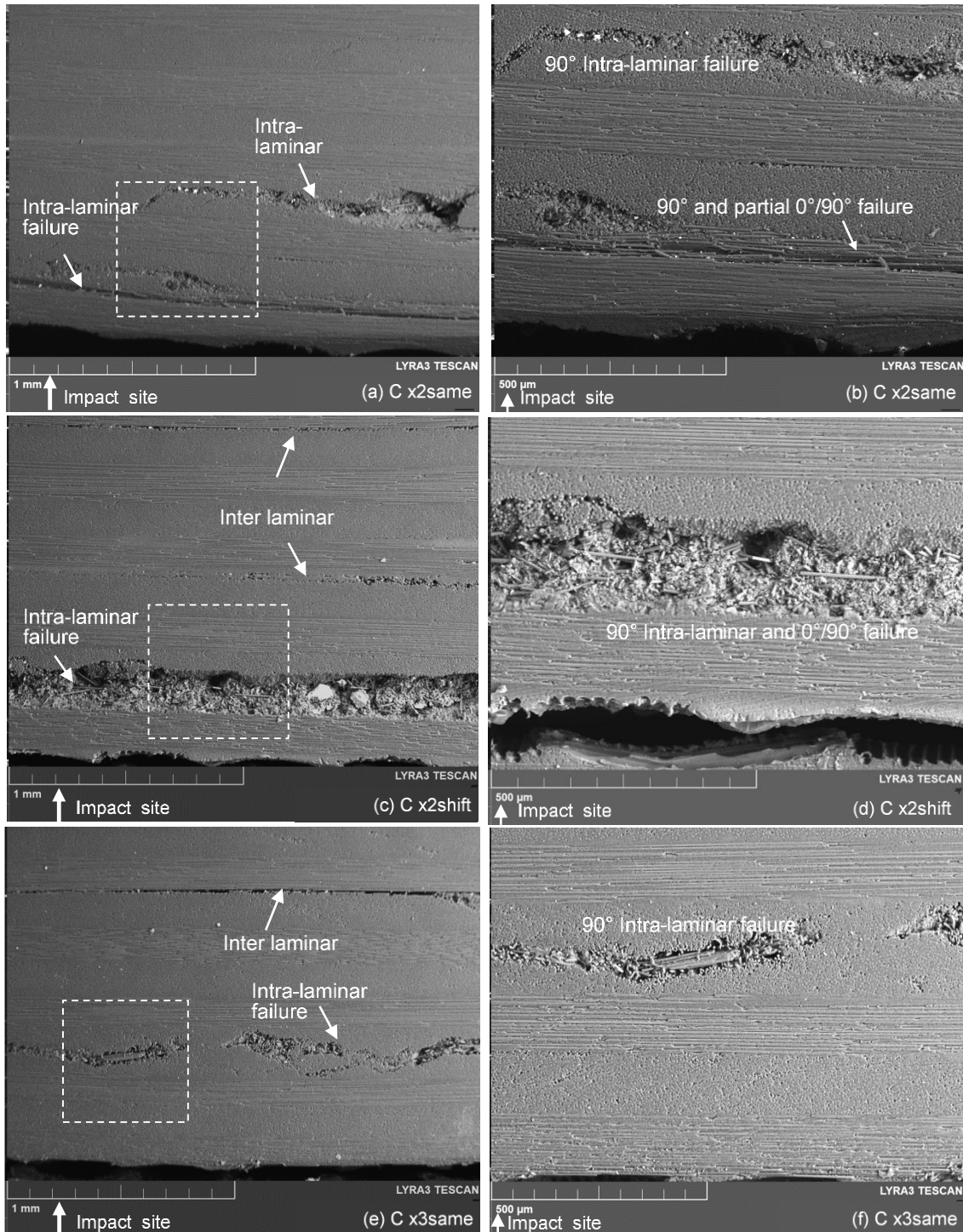


Fig.94 SEM of repeated impact 'C' configuration; (a-b) 2x same location, (c-d) 2x shift location, (e-f) 3x same location.

'M' configurations have similarities to 'C' configurations; where the M5 impacted two or three times have similar features as the 'C' group with fractures on the outside interfaces when impacted at the same location. However, for the shifted location test condition, inter-laminar damage can be observed closer to the mid-

interfaces for the first time. The fractures surround the veils on either side but do not penetrate the veil interlayer where fracture migration is triggered and arrested. This is an exciting finding where further propagation is hindered by the presence of the veil interleaved layers, as observed in Fig.95b. Although the crack inhibiting behaviours may be widespread throughout the volume under the impacted area, observations at this cross-section provide evidence of the veils enhancing the damage tolerance. Many crack inhibiting features similar to these are likely present throughout different cross-section positions and is the scope for future work.

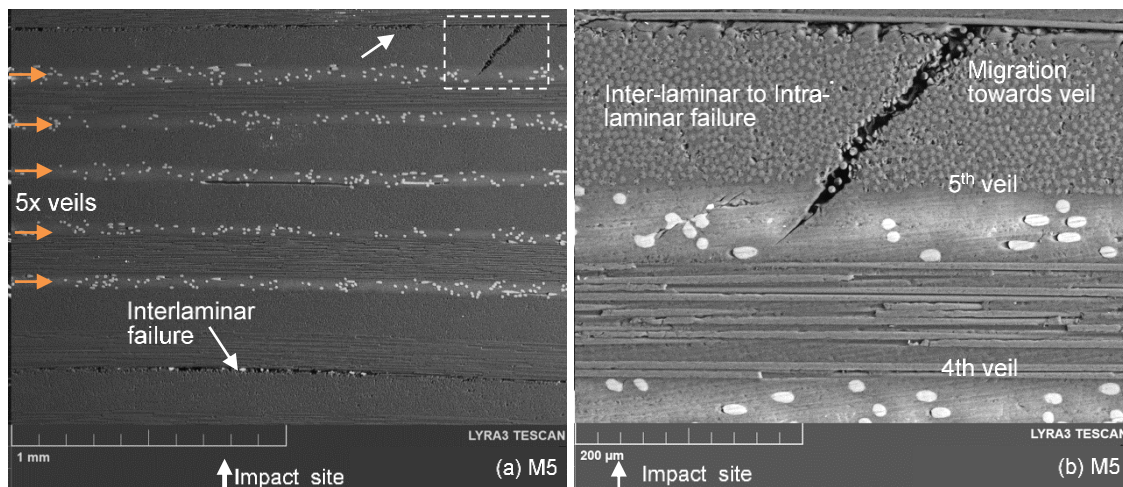


Fig.95 SEM of repeated impact for M5; (a) Overview, (b) magnification of migrating fracture.

For the O6 configuration under repeated impact, the damage is located once again closer to the impact surface (Fig.96). The exact location of the damage is within the 4th CF ply (90°) as a predominantly intra-laminar failure. The damage was localised to this interface at this cross-section; however, the volume under the impact site may have other locations for three-dimensional consideration of the damage.

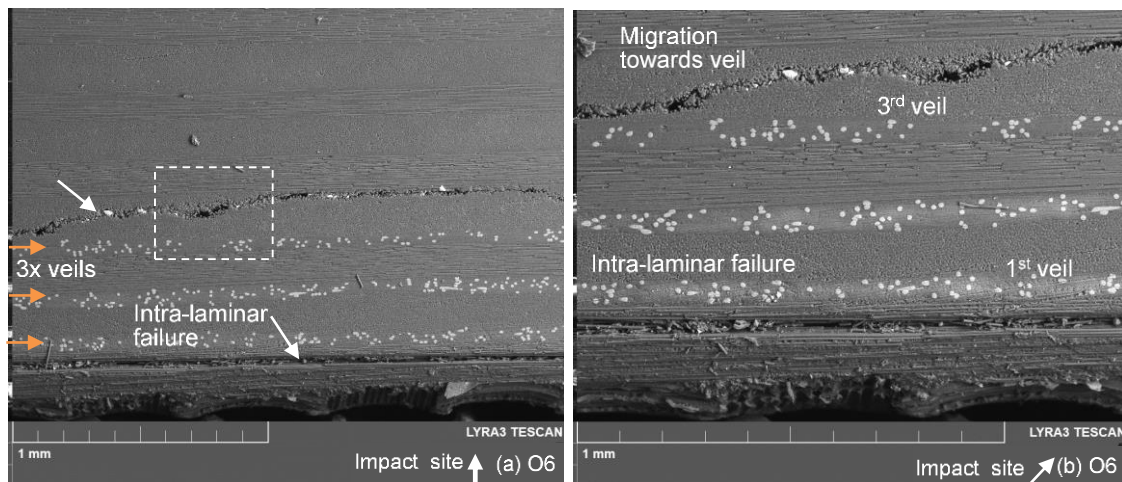


Fig.96 SEM of repeated impact for O6; (a) Overview, (b) magnified view of migrating fracture.

4.1.6 Discussion

Results from this investigation on LVI for cross-ply laminates interleaved with veils has generated exciting results with clear industrial relevance for laminates susceptible to impact damage. The tests show a clear identification of the DTL on the impact-force history of the 'C' configuration and interleaved configurations M3 to M9 and for O6 to O10. The DTL indicates the initiation of delamination within the laminate during the impact and is identified by the first significant load reduction event. Interestingly, the gradual fading of the DTL by increasing the number of veils is observed (Fig.78) where the severity of the load-reduction is lower when veil counts increase. A decreasing magnitude of the load during a repeat impact event would suggest smaller delamination being initiated within the laminate; conversely, the damage (Fig.84-Fig.85) does not appear reduced for the 'M' configurations.

The damaged area is an overlapped summation, with little through-thickness damage information; therefore it is plausible that damage areas are smaller, but the number of interfaces affected is more widespread, or the presence of multiple delaminations exist, making the area appear unchanged or greater. Upon examination of the fractured cross-sections under SEM, the amount of laminae undergoing fractures is lowered from 6 to 2 for 'O' configurations (outside interleaved) as veil quantities are increased. Additionally, the magnitude of matrix fractures and the delamination extent is reduced with higher veil quantities. However, this could be a result of a slightly increased stiffness as suggested by

the slope change in Fig.82. The change in initial and post-impact slope is apparent for all load-displacement plots (Fig.78-Fig.81) and demonstrate that although the thickness of the veil is merely 60-70 μm based on the SEM, the effect on the stiffness is still apparent (Fig.97). The crack depth reduces from 75 μm to a consistent 20 μm for configurations with 7 or more veils, also down to 5 μm for 9 to 12 veil interleave configurations.

The configurations with the veils close to the mid-plane do not affect the single impact responses based on the results, and the reasoning can be justified by being in proximity to the neutral axis (NA). For the outside ply interleaved configurations, the veils have changed the dominant failure mechanism from inter-laminar for low veil counts, into intra-laminar for higher veil counts. Under the repeated impact tests, the 'M' configurations appear to be inhibiting further growth through the layers (Fig.95b). The fracture propagation energy has been absorbed by the localised semi-elastic region resulting from the veil interleaving. For the 'O' configurations, identical failure mechanisms are apparent as with single impact, where the dominant failure mechanisms are intra-laminar (Fig.96). As the fractures cannot propagate the interleaved interfaces due to localised elasticity, a new path and propagation forms within the layers as intra-laminar matrix fractures. Additionally, the localised strain concentration resulting from the veils will contribute and cause fractures to migrate away from the interface under brittle fracture.

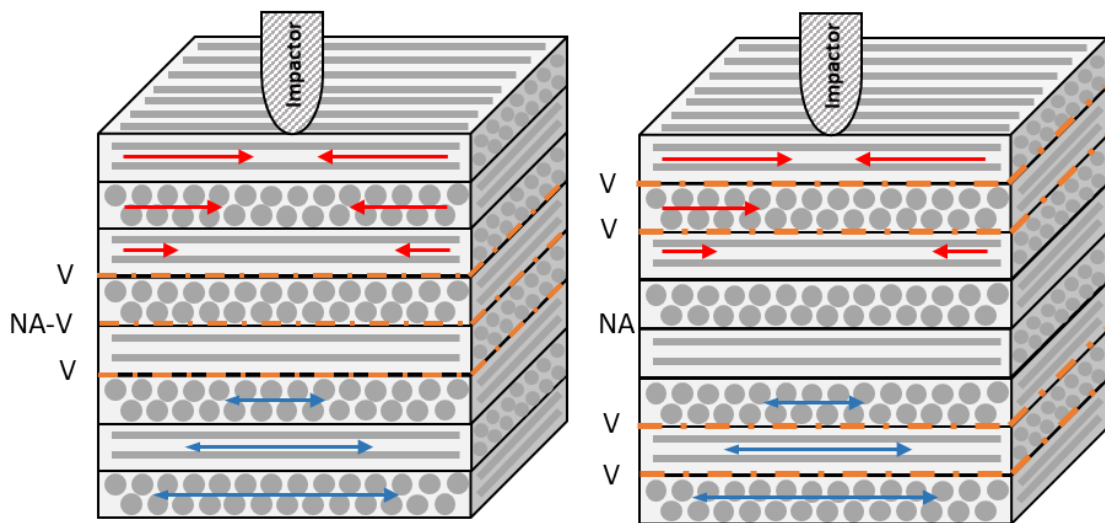


Fig.97 Stiffness illustration of loading conditions for: 'M' configurations (left) and 'O' configurations (right).

4.1.7 Conclusions

The influence of single and repeated low-velocity impact on veil interleaved CFRP laminates have been investigated. A summary of conclusions can be collated from the results of the study as follows:

- Veil interleaving is most effective when used towards the outside plies for the selected layup configurations, energy level and impactor size in this investigation.
- DTL increases up to 9%; however, DTL is not visible in heavily interleaved configurations. The repeated impacts do not show a further DTL for repeated impacts as expected. A completely shifted impact location on the laminate may indeed present a secondary DTL dependant on the shift distance, impact energy and impactor size.
- The fading of the DTL by increasing the veil count is observed where the amplitude of the load-drop is reduced as veil quantity is increased. The decreasing magnitude of the abrupt load-drop during an impact event means smaller delamination initiation within the laminate and DTL being less apparent.
- C-scans show the outside interleaved configurations have less damage than the middle-interleaved configurations. The shape of the extent of damage is significantly different, where outside interleaved configurations present oval-shaped damage and mid-interleaved having elongated damage.
- Veil interleaves in larger quantities can increase the stiffness of the laminate, resulting from the greater separation of the laminae and the increased thickness of the cured veil.
- Interleaving PPS veils improve the impact performance by restricting inter-laminar damage, however intra-laminar damage at a nearby layer is the by-product.
- Localised strain concentrations introduced by the veils can exist which allow the interleaved interfaces to focus the strain concentration to a position with higher modulus and fail under intra-laminar conditions.

- The PPS veil interleaves used in CFRP laminates can inhibit fracture propagation into the interface, as demonstrated within the repeated impact tests.

The applicability of the PPS veils in CFRP structures is now better understood where this work provides the potential variations which can be encountered during the introduction of veils at various positions within a laminate. The interlaminar enhancement exists by the addition of PPS interleaves; however, the shifting of damage position into intra-laminar matrix cracks means an ideal approach would be fibre-level toughening systems introduced earlier. Though, interleaving remains a primary toughening method owing to the ability to be applied where necessary within the laminate and do not require unique assembly methods within the manufacturing process.

REFERENCES

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5 Discussion

5.1 Research Methodology

The subject of damage tolerance enhancement in composites is a broad one. The need for damage-tolerant structures and their growing application requires successive adaptations and modifications to current material systems. To investigate the critical nature of low-velocity impact-related damage; the internal damage mechanisms must be evaluated. Various enhancement opportunities must be explored, which involve a study of other research, material selection, test standards, testing and verification of enhancements.

During this research, the subject of study is mostly experimental owing to the nature and current understandings of damage mechanisms. The process of research for the subject of damage tolerance enhancement uses a cross-sectional approach of study; meaning individual aspects of damage and enhancements are isolated and explored.

For the experimental aspect of damage tolerance enhancement, the root cause of the damage must be defined; therefore, the identification of low-velocity impact variables are essential. It is then crucial to consider qualitatively which material systems are affected and explore previous literature. Once a suitable approach has been established, the aspects must be discretized into various smaller tasks (Fig.98). Experimental work is sub-divided into; mode I & II, and Low-velocity impact for various ply orientations for unidirectional and 2D woven 5HS fabrics with various toughening levels.

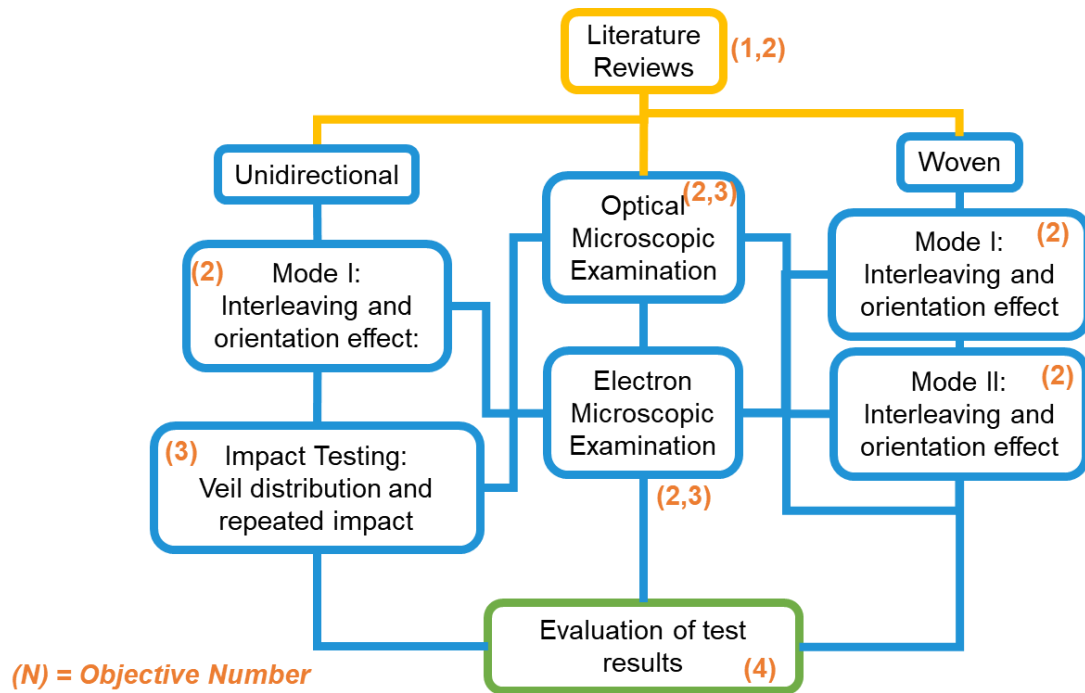


Fig.98 Research task dissemination.

Mode I fracture toughness (Sections.3.1-3.3): Thermoplastic veil interleaves were recognised from the literature to improve the fracture toughness of a composite; however, the level of enhancement was unclear and an intriguing topic for further study. Mode I fracture toughness testing was a primary test; to identify compatibility concerns for the material combinations and cure cycles. Improved fracture toughness is often due to the increased inter-ply thickness of the matrix interface, with reduced shear stress concentration owing from the stiffness and load mismatch. Therefore, low areal density for veil interleaves were selected to reduce the interface thickness effects of veil addition.

Mode II fracture toughness (Section.3.4): Investigations required a steady crack propagation; where numerous methods have been developed to measure the IFT, G_{IIC} in the past. Standard tests include using the ‘three-point’, and ‘four-point’ end notched flexure (3ENF and 4ENF) and end-loaded split (ELS) methods. 3ENF tests suffer from unstable crack propagation [93]; however, for this investigation controlled initial propagations are only required. As a result, controlled short distance fracture propagations using the 3ENF method was sufficient for accurate calculation of the G_{IIC} with interleaved veils and baseline virgin layups.

The interfacial bias of 5HS CFRP laminates (Sections.3.3-3.4): It was apparent from the UD CFRP mode I loading that veils could enhance the interlaminar bridging and reduce the directional bias sensitivity of woven 5HS CFRP laminae. To explore the effect of veils on the directional bias of 5HS weave fabrics at their inter-ply interface; various bias orientations were proposed and tested for mid-plane fracture tests under modes I & II. A novel finding from the study indicated that the directional bias between plies was reduced by using veils, were all except grouped biased 0° interfaces had similar fracture toughness values.

Low-velocity impact (Section.4.1): The drop-weight-tower was employed to assess the impact response of the composite laminate with and without interleaved materials. The variation of interleave quantity and placement allows studying damage positions to identify differences and observe mechanisms. Repeated impact tests were used to identify differences in further damage progression when impacted at the same location and shifted locations.

Common laminae orientations were tested, both with and without veils for various interfacial stacking configurations to enable a thorough investigation. To maintain the validity of the experimental comparisons, only interfacing plies where propagation occurred were included. The fracture toughness properties using veils can now be used as cohesive zone parameters to simulate alternative layups to conduct optimisation for impact damage reduction. Furthermore, numerical work will follow with veil enhanced cohesive zone interfaces obtained from the tests. Overall, the direction is to characterise the damage mechanisms of veils for enhanced impact damage tolerance while maintaining a minimal influence approach to the overall strength and stiffness of the laminate.

Achieving aims:

Develop knowledge for enhanced damage tolerance design of composite structures in the aerospace industry using interleaved veils.

To facilitate a review of the aims of the project, the following demonstration of objectives 1-4 are striking. Likewise, a complete summary of enhancements and observations from the research are listed in section 6.1.

Achieving Objectives:

1. *Review current materials and methods for toughness enhancement of laminates.*

This project contributes to the gap in knowledge for thermoplastic additives used for toughness enhancements in composite laminates, specifically for non-woven veil interleaves in combination by considering the reinforcement orientation effects.

2. *Understand toughening mechanisms by investigating the effect of interfacial fibre orientation and interleaving on delamination resistance.*

The rationalisation of toughening mechanisms using interleaves was demonstrated to improve fracture toughness for both UD and 5HS CFRPs under modes I & II, by additional fibre-matrix interactions, bridging, reduction of resin-rich regions, migration tendencies and the inclusion of ductile thermoplastic regions.

3. *Understand the impact performance of composite laminates by varying the through-thickness veil distribution.*

The experimental results achieved from the mode I & II tests have aided in the decisions for veil interleaving CFRPs under low-velocity impact. The qualitative observations from the fracture surface morphology have provided a depth of information to understand the fibre-matrix interactions under each mode separately. Therefore, armed with this qualitative understanding of variations of toughness enhancement based on interfacial orientations; the impact specimens were designed to remove the potential for migration tendencies by using a cross-ply layup for the 5HS CFRP. The impact results with veil interleaves showed that veils in the outer plies could reduce or remove the occurrence of a DTL in sufficient layers.

4. *Provide the industry with further knowledge for the enhanced damage tolerance design of laminated composite structures.*

The understanding of fracture toughness obtained from 0° oriented fibres in the crack direction is over-conservative for composite laminates. The current standards [53,54] must specify the fracture interfaces for multi-directional considerations. The methods do not specify the conditional validity requirements when using multi-directional laminates, where features such as migrations and multiple delaminations are present.

The fracture mechanics methods are standardised for the evaluation of the strain energy release rates of specimens, yet each must have inclusions for the use of tougheners, additives and conditional criteria for their validity. Material characterisation by introducing thermoplastic veil interleaves in CFRPs was demonstrated as added value to the knowledge. However, caution is recommended when using interleave materials for multi-directional laminates due to toughness enhancements being often over-estimated, especially for unidirectional fibre laminates. This is due to dominating fracture path migration failures if the growth is transverse to the interface fibres.

The qualitative results collected under mode I & II and can become semi-empirical inputs to subsequent simulation exercises. Where the performance under low-velocity impact was demonstrated to have redirected the damage position and form as intra-laminar fractures compared to dominated inter-laminar regions previously. The introduction of veils in various positions and quantities have demonstrated a reduction in DTL magnitude and eventual fading when using higher interleave counts towards the impact face of the panels. Nonetheless, the thickness sensitivity should also be considered for applicability of veil interleaves.

5.2 Discussion

Impact damaged composite laminates as established have detrimental in-plane characteristics. Although the current practice for the DTD of composites for applications prone to safety-critical failure is employing a 'no-growth' philosophy, a proposition of this work is to promote alternative techniques and a reconsideration of the current 'no-growth' approach.

The veil enhanced damage mechanisms resulting in improved fracture resistance, and enhanced damage tolerance was due to the inclusion of lower modulus PPS fibres. These discrete thermoplastic regions within the laminate provide a cushioning effect while increasing the localised ductility resulting in increased fibre-matrix interactions at the fracture process zone.

The concept of controlling the fracture propagation, which leads to delamination has been demonstrated; using fracture inhibiting PPS veil interleaves in combination with their effectiveness in multidirectional laminates. Under these notions, it was demonstrated that veil interleaving could influence the IFT and damage propagation during impact, but their benefits are grossly overestimated when considering only IFT tests based on 0° plies. The IFT for UD laminates is enhanced when using veils for all interfacial fibre orientations. However, the interface fibre orientation itself is the dominant factor affecting their inhibiting capability. It was shown that although the presence of veils under mode I loading can increase the apparent fracture toughness, migration triggering mechanisms are present as a result of off-axis plies. This demonstrates the need for further thermoplastic fibrous regions within the intra-laminar regions. The approach to achieve a hybrid laminate has been described under the topic of yarn intermingling (Section 2.10.3).

An identical study was conducted on 5HS CFRP weave laminae with the interfacial fibre orientation bias as a variable. It was discovered that interleaving the laminate with PPS veils had increased the IFT and reduced the orientation sensitivity of the biased interfaces under mode I loading. This meant that 5HS CFRP laminae could be oriented as per the component's stiffness requirements while using grouped and off-axis plies without detriment to the interfacial toughness drawbacks.

Under mode II it was established that a variation in interfacial fibre orientation of the 5HS CFRP laminae with the use of veils had a general improvement to the IFT. The fracture surface morphology revealed the increased fibre-matrix interactions as a result of the localised elasticity from the thermoplastic veil at the interface.

Furthermore, it was established that damage could be restricted or contained under LVI by the use of PPS veils. A change in fracture location was also observed under LVI, where the introduction of veils at outside ply-interfaces triggered intra-laminar failure compared to the dominant inter-laminar failure that was typically observed. This provided confidence that under LVI and two-dimensional propagation scenarios, the interleaving using veils were still effective despite the variations observed in the IFT tests.

It was also established that under repeated LVI, the impact force responses were significantly greater with evidence of minimal further damage under the same energy level. Additionally, the presence of veils has demonstrated the crack inhibiting characteristics by altering the crack path and arresting further propagation entirely.

The experimental studies within this research endeavour were designed to express proof of concept for the control of damage growth. Throughout the field of advanced composites, a comprehensive verification is desired, including additional cases under the fatigue-life study, strain rate dependency and searching for an optimum plateau for veil interleaving.

The research activities within the scope of damage initiation and propagation control; to ultimately enhance the damage tolerance of conventional thermoset laminates has been achieved. A parametric based exploration of damage initiation and propagation has demonstrated that the design of composites can be enhanced using thermoplastic modifiers. The outcomes from the further development and contribution to the topic can result in the reduction of damage sensitivity of composite structures. Damage tolerance can be realized by increasing the resistance of progressive damage, ensure reliability and safety by using mostly conventional certifiable materials and maintaining structural performance using a minimum change approach. Furthermore, the potential of damage growth manipulation by using embedded healing networks may facilitate the widespread development of self-healing and repair through capillary dispersion at crack locations, thereby prolonging the service life.

6 Conclusions

The outcomes from this research provided much-needed insight into the influence of nonwoven PPS interleaved veils and its influence on the ITF to enhance the damage tolerance capability of laminated composites. This chapter concludes the project into two sections, where the key results of the research activities are summarised in section 6.1, and the overall project conclusions are further elaborated in section 6.2 to determine if PPS veil interleaving is worth future research activities.

6.1 Key results

The following is a summary of original contributions to the knowledge regarding the development of toughening techniques. Considering veil interleaved laminates are low-cost, high-quality, can be implemented quicker, have improved impact and interlaminar enhancements, their development from this work by considering detailed dissemination of interface variables can be valuable. The examination efforts conducted throughout this research project which have been highlighted in the previous chapters proved to be beneficial to advance the field of toughening techniques with their enhancement capabilities. The following conclusions can be drawn for suitable toughening techniques used for improving the damage tolerance of laminates:

- The use of non-woven fibrous veil interleaves is a cost-efficient method of toughening, that can practically be introduced into the manufacturing processes using pre-impregnated lamina.
- The widespread implementation into aerospace structures is not fully developed but remains a viable toughening technique for low-cost manufacture.
- The interfacial fibre orientation has considerable influence on resistance to delamination initiation and propagation.
- Under Mode I:
 - UD specimens with off-axis fibres have further load carrying capacity beyond the initial elastic deformation. These also have higher apparent IFT than specimens with the 0°/0°.

- For UD materials, fracture path migration is the dominant failure mechanism for off-axis plies.
- The 0° plies achieve an apparent IFT increase up to 48% using 10gm⁻² PPS veil compared to the control.
- PPS veils at the interface improve the fracture resistance. However, the improvements are fibre-orientation-dependent.
- Veil interleaved toughening is over-projected if based on 0°/0° midplanes, therefore incorrectly prescribed for multi-directional laminates as intra-laminar failure and migration are not considered by many.
- Veil influence remains after fracture migration away from the veiled interface for UD plies, indicating minor enhancements to surrounding lamina.
- 5HS CFRP laminae interfacial fibre orientation bias is often overlooked.
- Orientation bias of 5HS CFRP laminae influences the IFT. 90° dominated interfaces have increases up to x2.9 that of 0° dominated interfaces without any veil enhancements.
- For 5HS CFRP tests, the areal density of the PPS veil improves the IFT; enhancements up to 111% achieved using 10gm⁻² and 136% using 20gm⁻².
- Reduced interface orientation bias sensitivity was achieved using 20gm⁻² PPS, meaning improved design and manufacturing capabilities for 5HS CFRP laminae.
- 5HS CFRP fracture surfaces with veils revealed the reduction of resin-rich zones, added fibre peeling, fibre fractures, increased surface roughness, fibre imprints and localised intra-laminar fracture tendencies.
- High areal density veils can form long and slender voids along the fibre tows; which, can contribute to a momentous decrease in compressive strength for laminates.

- Under Mode II:
 - Interlacing within 5HS CFRP laminae leads to a higher resistance from the transverse yarns and promotes fibre-matrix de-bonding and fractures.
 - Propagation for 5HS CFRP with 90° dominated interfaces required higher energy than along the 0° dominated interface due to the friction interactions of transverse yarns being greater.
 - For 5HS CFRP, 2D fracture growth could result in an elongated peanut-shaped delamination (Fig.7) due to the orientation bias.
 - Interleaving increased IFT for delamination initiation up to 75% against the control layups, depending on the veil areal density and interface orientation bias of the 5HS CFRP laminae.
 - For the 5HS CFRP, fracture resistance variances between 10gm⁻² and 20gm⁻² veil were marginal. However, under mode III and mixed-mode can be more significant.
- Under LVI:
 - For UD LVI, veil interleaving was most effective when used towards the outside plies for the selected layup configuration, energy level and impactor size considered.
 - UD LVI with veils causes the DTL to increase by 9%. However, DTL is not discernable in configurations with higher veil counts; therefore, no large scale delaminations were triggered, but intra-laminar damage was indeed observed.
 - UD LVI with veils causes the amplitude of the load-drop to reduce as veil quantity is increased, meaning only smaller fractures being initiated as the DTL becomes less apparent.
 - UD LVI with veils causes fading of the DTL by increasing the veil count. DTL is also wholly eliminated with higher veil counts.
 - Veil interleaves in larger quantities can increase laminate stiffness, resulting from the greater separation of laminae due to veil thickened inter-ply regions.
 - UD CFRP with interleaves shows improved impact performance by restricting inter-laminar damage, however, triggering intra-laminar damage at a nearby lamina.

- Veil induced localised strain concentrations can exist, allowing interleaved interfaces to focus strain-concentrations to a position with higher modulus and fail under intra-laminar conditions.
- Under repeated LVI:
 - For UD CFRP repeated LVI, tests do not show a secondary DTL for repeated impacts as expected. An entirely shifted impact location on the laminate may indeed present a DTL dependant on the shift distance, impact energy and impactor size.
 - UD CFRP repeated impacts show higher impact forces on the 2nd impact with noticeable smoothing of the loading profile, indicating less further damage.
 - A noticeable change in stiffness is apparent for the 2nd impact, suggesting a stiffness loss can be compared as a parametric study.
 - Enclosed areas of the force-displacement during the UD CFRP LVI tests show decreasing trends, suggesting less plastic deformation and fracture damage.
 - The PPS veil interleaves used in UD CFRP laminates can arrest fracture propagation into the interface as demonstrated within the repeated impact tests.
- Combining multidirectional layups and veil interleaving provided detailed insights of fracture behaviour and failure mechanisms of inter-laminar, intra-laminar and fracture plane migration.
- Advancement of failure mechanisms knowledge and their interactions are demonstrated to optimise the damage tolerance and performance of laminated composite structures.

6.2 Project Conclusions

Damage formation, initiating as matrix-cracks and delaminations within fibre-reinforced polymer composites is an inevitable downfall owing to lack of plastic deformation capability when external energy absorption is required. To facilitate the development of damage-tolerant composites, it is expected to characterise damage morphologies and understand progressive failures to design laminates that mitigate these eventualities. Historically, the methods employed to mitigate

damage eventualities have been under the 'no-growth' approach, optimising laminate stacking sequences or by developing stronger and tougher composite constituents. Alternatively, damage tolerance manipulation can be achieved by controlling the initiation and location of fractures, by utilising materials with added plasticity which are not unfavourable to the composite performance at a structural level.

An all-encompassing aim of this exploration was to advance the knowledge for enhancing the damage tolerance of composites. To suppress the damage growth, firstly the interlaminar crack growth and impact behaviour were explored and disseminated. The added value to the process of interleaving PPS veils in conventional thermoset CFRP composites is substantial; based on a deeper understanding of microscopic damage features and performance characteristics when used in multi-directional laminates.

Chapter 2 included a thorough literature review on the subject of interlaminar fracture arrest techniques and potential exploitation methods of damage tolerance enhancement. Various techniques were identified ensuing this review as suitable candidate solutions for damage suppression and crack manipulation within a composite material.

Chapter 3 experimentally explored two combinations of these techniques via mode I DCB and mode II 3-ENF testing. It was apparent from the initial investigations that interleaving and interface orientation variation had the most promising method for local damage manipulation without drastic changes to the current laminates. The fabrication of composite specimens with interleaved materials within conventional pre-impregnated CFRP was a flexible and straightforward solution for currently deployed designs. The veil interleaving was a relatively simple technique to be incorporated into the laminate whilst following guidelines from standard test methodologies. The veil was found to have a great influence on local fibre interactions in all tests alongside unveiling the effective toughening of the interface regions.

Chapter 3 also explored the suitability of interleaved veils within 2D woven laminates, namely 5HS CFRP with the inherent interfacial fibre orientation bias. It was established that interleaving veils had a beneficial impact on the orientation

sensitivity relating to the interfacial bias of the 5HS CFRP laminae. The toughening performance on UD and 5HS CFRPs using veils was established under a multitude of fibre orientations. It was discovered within Chapter 3, that the prospect of veil interleaving could trigger preferential changes to the fracture behaviour of composites. Contrary to expectation, the observations for UD laminates showed that veil interleaving enhancements are numerically over-projected in comparison to other literature owing from dominated intra-laminar fracture for orientations other than 0° . The focus was subsequently established on the effective fracture toughness and migration tendencies when using off-axis UD plies in their presence.

Chapter 4 explored the damage initiation and formation under LVI in addition to exploring interleaved veil on impact damage. The formation of impact damage under various impact energies within woven CFRPs was presented and found interesting behavioural trends in the absence of the DTL. Similarly, for the UD CFRP, interleaves were strategically positioned at the inner and outside interfaces to assess the interlaminar shear behaviour and bending performance under impact loading. Following the LVI testing, it was evident that veil interleaving undoubtedly influenced the fracture propagation and subsequently altered the fracture path and position within the laminate under LVI.

Under LVI, the levels of interleaving demonstrated that for 4mm thick laminates as tested, the most effective placement for veil interleave was at the outside ply interfaces. The overall damage extent did not appear to change, but significant differences within the impact-force history and load-displacement indicate a stiffened laminate when veils were introduced and hence affected the absorbed-energy based on the enclosed load-displacement areas.

The veil enhanced damage mechanisms resulting in improved fracture resistance and enhanced damage tolerance are due to the inclusion of lower modulus PPS fibres. These discrete thermoplastic regions within the laminate provide a cushioning effect while increasing the localised ductility resulting in increase fibre-matrix interactions at the fracture process zone.

Although composite laminates used in industry employs a 'no growth' DTD philosophy, the outcome of this work highlights a need to reconsider this approach. The occurrence of damage can indeed be controlled by crack arresting techniques as explored or using the other methods of toughening described in Chapter 2. The value added to the field necessitates further exploration of intermingled hybridisation based on the behaviours of veil interleaves changing the damage mechanism from inter-laminar towards intra-laminar dominated.

The applicability of the PPS veil interleaves in CFRP structures are now better understood where this research provides the potential variations which can be encountered during the introduction of veils at various positions within a laminate. The interlaminar enhancement exists by the addition of PPS interleaves, though the shifting of damage position into intra-laminar matrix cracks means a better approach would be alternative toughening systems such as yarn level commingled hybrids. Interleaving nonetheless remains an essential 'effective' toughening method; owing from their applicability within the laminate as they do not require unique assembly methods within existing manufacturing processes.

7 Future Work

This research provided much-needed insight concerning the influence of nonwoven interleaved veils and its influence on the ITF to improve the damage tolerance capability of laminated composites. As the rise of new analytical and experimental methods are proposed and developed, further advances in toughness enhancement techniques will inevitably be developed. Further topics of investigation are acknowledged in this chapter for continuing the development of composites beyond the scope of this project.

7.1 Interleaves

This research has added to the development of using crack arresting interleaves which use discrete layers for toughening. Interleaved materials are direct options for incorporation within laminated composites. Interleaves maintain the global strength and integrity of the parent laminate under various loading conditions and generally don't exhibit elevated complexity or manufacturing costs.

Like with most material selection options, there are a host of potentially suitable materials which could be implemented for interleaving. As explored in the literature, many efforts have been made to cover the variety of material selection options; however, there remain many unexploited material combinations which could provide far superior enhancements. For example, a combination of options such as veil interleaving with stitching as implemented by Tsotsis [271,272] can further improve the development and enhance the material combinations further for a range of veil fibres and density. Further development into interleaved and stitched combinations could improve the impact damage propagation resistance but also enhance the in-plane compression properties.

From the material selections explored, interleaved PPS veil was selected for its curing suitability and compatibility with the epoxy pre-preg matrix. When using the 20gm^{-2} veil interleave, the resin absorption could have been excessive, which would locally increase the fibre volume fraction at the interlayer. The downside to this possibility is the potential loss of peel strength of the surrounding reinforcement. Improving the adhesion strength could be to use appropriately 'sized' interleaves. Alternatively, by combinations of treatment methods namely; surface abrasion, chemical coating or surface binders could lead to a superior interleaved material.

7.2 Cohesive zone modelling with veil placement optimisation

Recent interest in cohesive zone models to study and optimise the use of interleaves has been made by Saghafi *et al.* [310] and Giuliese *et al.* [311]. The modelling strategy based on sizing assumptions from coupon tests is problematic to achieving accurate interleaving predictions. Owing from the numerous material combinations and manufacturing variations available; further work is required to accurately predict the effect of interleaving for optimisation and ultimately achieving damage tolerant structures.

7.3 Cohesive zone model calibration and genetic algorithm

For future research endeavours, an exciting subject worth considering is the development of optimised prediction solutions. From the material characterisation in this work, the simulations can be tuned and correlated to enhance predictions by implementing genetic algorithms to ultimately find optimum toughening solutions.

7.4 Combination of hybrid toughening and z-pinning

An additional prospect for the development of damage tolerant solutions is the combination of Z-pinning and thermoplastic interleaves which has never been previously investigated. The combination may provide interesting outcomes as the thermoplastic interleaves could reduce the propagation distance for a given crack opening, providing the Z-pins with an opportunity to improve the through-thickness behaviour. The philosophy behind the idea of combining the two methods arose from considering the crack length for mode I is relatively large when Z-fibres begin taking load [30]. The interleaving aspect for the combination was theorised from the outcomes of the research activities from Chapter 3, where the delamination length for a given opening displacement was found to be restricted when veils are present. The consideration of Z-pinning in combination with thermoplastic toughened laminates is an interesting topic due to their ability to be utilised where necessary in comparison to global additive toughening of the matrix. The prospect of combined z-pinning with interleaving is beyond the scope of this work; however, the combined effects are provocative.

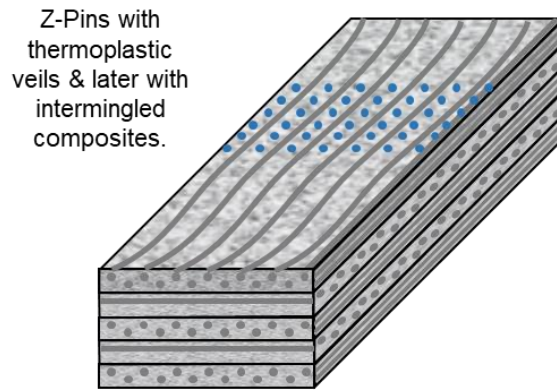


Fig.99 Combined veil interleaving with Z-pins concept.

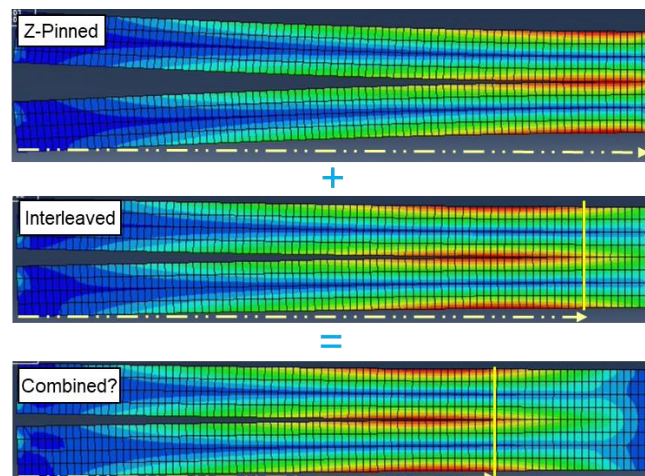


Fig.100 Potential outcome for combinations of interleaving and z-pinning.

7.5 Intermingled fibres

Laminates using films or discontinuous thermoplastic fibres can often trigger premature failure at an alternating location. The formation defined elastic regions are spread throughout the lamina and can cause premature delamination if the interleaving material is not suitably sized. Therefore, it is apparent that intermingling could be a preferred option as regions of local elasticity can hinder or provide crack propagation retardation effects. The benefits of using intermingled yarn reinforcements are due to the toughening fibres being closely blended with the parent material, rather than at perfectly defined interfaces. This allows intermingling to be far superior owing to the presence of toughening fibres in the desired locations and within fibre tows. The scope of comingled fibre yarns is one which could become excessively costly for immediate implementation into current laminates. However, the advancement in manufacturing and processing techniques today could assist in the future development of these hybrid yarn materials. As a result, a wealth of further study must be conducted for future applicability of intermingled yarn fabrics.

7.6 Healing networks

Self-healing materials are inevitably a next-generation material, owing to their applicability in high-performance structures. The concept of self-healing composites promises great potential when internal micro-cracks and fractures are driven towards a vascular network of repairing fluid. The self-healing vascular network can make it possible for a partial or even complete repair of the damage with the penetrating feature stemming from capillary movements of low-viscosity adhesive fluids.

As previously established, significant degradation of composite structures, particular laminated composites can occur due to external damage. Frequently these damaged regions are internal and barely visible, which can lead to catastrophic failure of the structure. Various concepts using extrinsic methods have been theorised; including encapsulation, vascular-fibre intermingling, and hollow fibre embedment. Conclusively, the self-healing composites allow the possibility of damage management within materials. Self-healing networks begin healing of damage at early stages and promise to recover the mechanical performance without detriment. The self-healing novelties may allow a dramatic increase in service life, helping to reduce environmental impact, maintenance costs and permit lighter and reliable structures. However, technological restrictions, economic dedication and a deeper understanding are required for their future implementation.

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