Analysis of Compression Failure in Multidirectional Laminates

by

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Abstract

Even though the attractive properties of fibre-reinforced polymer composites have led to them becoming essential materials for a wide variety of high performance applications, they pose several drawbacks such as low compressive strength, low delamination resistance and high sensitivity to defects. These drawbacks have led designers to adopt a damage tolerance approach, whereby damage growth is deemed as failure. However, this approach has resulted in heavy composite structures with conservative configurations, which has somewhat negated the significant weight saving potential that composites offer over traditional materials. A step change in the approach by which damage growth is tolerated could provide designers with the freedom to develop novel composite structures.

Despite an improvement in the understanding of composites failure, particularly in unidirectional laminates, compressive failure of multidirectional composites is still not fully understood. Therefore, the initial objective of this research project was to investigate the compressive failure processes multidirectional composites, leading to development of material-based approaches (i.e. introduction of a secondary material to the parent composite) which could offer compressive crack arrest/redirection. Such an approach would facilitate the adoption of a damage growth approach for composites design. An extensive experimental, fractographic, theoretical and numerical study on the compressive failure of multidirectional composites was conducted, resulting in the main failure mechanisms being identified and the sequence of events that lead to global fracture being deduced. The influence of the layup, specimen geometry (such as compact, plain and sandwich panel compression) and the proportion of shear loading on the compressive performance of multidirectional laminates were characterised. These observations were then used to validate numerical models, thus yielding more physically based predictions.

In the process of formulating novel crack arrest/diversion solutions in composite structures, various concepts ranging from hybridisation to carbon nanotubes and piezoelectric actuators, were investigated. However, after consideration of the relative maturity of these technologies and the time constraints, the latter two approaches were not pursued. Given the absence of an explanation of the hybrid effects observed in composites in the literature, an extensive study was carried out to investigate the effect of hybridisation on the compressive performance of multidirectional composite laminates. For this study, two systems of unidirectional pre-preg tapes with the same epoxy resin but different carbon fibre types and tow sizes were employed. It was identified in this study that hybridisation of selective ply interfaces influenced the location and severity of the fracture mechanisms. Finally, in a complementary study on delamination fracture toughness of hybrid composites, a significant improvement was observed in the delamination resistance (doubling in Mixed Mode I/II toughness) compared to the monolithic composites, indicating that the behaviour of the hybrid interfaces was critical for the compressive performance of the hybrid laminates.
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Nomenclature

Lower case Roman letters

\( \alpha \)  
\( \alpha \)  
\( \alpha \)  
\( \alpha_{ij} \)  
\( \alpha_0 \)  
\( b \)  
\( h \)  
\( l_x \)  
\( l_{x,\text{max}} \)  
\( t_s \)  
\( t_i \)  
\( t_t \)

Upper case Roman letters

\( A_i \)  
\( C \)  
\( E_{11} \)  
\( E_c \)  
\( E_{i} \)  
\( E_t \)
Nomenclature

\( F \) \hspace{1cm} \text{delamination fracture toughness correction factor (MBT)}

\( F_{IK} \) \hspace{1cm} \text{fibre kinking failure index}

\( F_M \) \hspace{1cm} \text{matrix failure index}

\( F_{SP} \) \hspace{1cm} \text{ply splitting failure index}

\( G_{xy} \) \hspace{1cm} \text{shear modulus in the } (x, y) \text{ plane}

\( G_C \) \hspace{1cm} \text{material fracture toughness}

\( G_{IC} \) \hspace{1cm} \text{Mode I delamination fracture toughness}

\( G_{II} \) \hspace{1cm} \text{Mode II delamination fracture toughness}

\( G_{ij} \) \hspace{1cm} \text{shear modulus in the } ij \text{ plane } (ij = 12, 23, 13)

\( G_t \) \hspace{1cm} \text{energy associated with the } t_{n} \text{ traction component}

\( G_s \) \hspace{1cm} \text{energy associated with the } t_{s} \text{ traction component}

\( G_l \) \hspace{1cm} \text{energy associated with the } t_{l} \text{ traction component}

\( G_T \) \hspace{1cm} \text{total fracture toughness}

\( K_{nn} \) \hspace{1cm} \text{stiffness associated with the } t_{n} \text{ traction component}

\( K_{ss} \) \hspace{1cm} \text{stiffness associated with the } t_{s} \text{ traction component}

\( K_{ll} \) \hspace{1cm} \text{stiffness associated with the } t_{l} \text{ traction component}

\( L \) \hspace{1cm} \text{DCB, ELS and MMB specimen length}

\( P \) \hspace{1cm} \text{applied load/force}

\( S^L \) \hspace{1cm} \text{longitudinal shear strength}

\( S^L_{is} \) \hspace{1cm} \text{in-situ longitudinal shear strength}

\( S^T \) \hspace{1cm} \text{transverse shear strength}

\( S^T_{is} \) \hspace{1cm} \text{in-situ transverse shear strength}
Nomenclature

\( V_f \) ................................................................. fibre volume fraction

\( V_m \) ................................................................. matrix volume fraction

\( X^C \) ................................................................. longitudinal compressive strength

\( X^T \) ................................................................. longitudinal tensile strength

\( Y^C \) ................................................................. transverse compressive strength

\( Y^T \) ................................................................. transverse tensile strength

\( Y_{is}^T \) ............................................................... in-situ transverse tensile strength

Lower case Greek letters

\( \beta \) ................................................................. kink band/inclination angle

\( \gamma_{x,y} \) ............................................................. shear strain in the xy plane

\( \delta \) ................................................................. load point displacement

\( \delta_{nf} \) ............................................................. displacement at total separation in the 3 direction (normal)

\( \delta_{sf} \) ............................................................. displacement at total separation in the 1 direction (shear)

\( \delta_{tf} \) ............................................................. displacement at total separation in the 2 direction (shear)

\( \delta_{n}^{o} \) ............................................................ displacement which corresponds to the \( t_n \) traction component at failure initiation

\( \delta_{s}^{o} \) ............................................................ displacement which corresponds to the \( t_s \) traction component at failure initiation

\( \delta_{t}^{o} \) ............................................................ displacement which corresponds to the \( t_t \) traction component at failure initiation

\( \varepsilon_0 \) ............................................................. strain at failure initiation

\( \varepsilon_{\text{max}} \) ...................................................... strain at which the material has failed (2dVUMAT)

\( \varepsilon_y \) ............................................................. strain in the y direction (parallel to the fibre direction)

\( \eta_T \) ................................................................. coefficient of transverse influence

\( \nu_{ij} \) ............................................................... Poisson’s ratio at the \( ij \) plane \((ij = 12, 23, 13)\)
Nomenclature

\( \sigma_{11} \) .................................................. longitudinal tensile stress

\( \sigma_{11}^m \) .................................................. longitudinal tensile stress in the misalignment coordinate

\( \sigma_{22} \) .................................................. transverse compressive stress

\( \sigma_{22}^m \) .................................................. transverse compressive stress in the misalignment coordinate

\( \sigma_e \) .................................................. applied compressive stress

\( \sigma_t \) .................................................. applied tensile stress

\( \sigma_N \) .................................................. normal component of the traction vector in a potential matrix fracture plane

\( \tau \) .................................................. applied shear stress

\( \tau_{12}^m \) .................................................. longitudinal shear stress in the misalignment coordinate

\( \tau_{23}^m \) .................................................. transverse shear stress in the misalignment coordinate

\( \varphi \) .................................................. additional fibre rotation

\( \bar{\varphi} \) .................................................. initial fibre rotation

\( \chi_I \) .................................................. correction factor for Mode I delamination fracture toughness

\( \chi_{II} \) .................................................. correction factor for Mode II delamination fracture toughness

\( \psi \) .................................................. rotation of kink band plane

Upper case Greek letters

\( \Delta \) .................................................. effective delamination extension to correct for rotation of DCB arms at delamination front
Chapter 1 – Introduction

1.1 Motivation
Currently advanced composite materials are used in many primary structural applications in the aerospace, automotive and maritime industry. Even though the properties they offer make them very attractive materials, composites pose some drawbacks which hinder them from replacing conventional metallic structures. Low compression strength, poor delamination and impact resistance, defect sensitivity and brittle/catastrophic fracture are some of the issues which composite scientists and designers are trying to improve to fully exploit the capabilities that composites could offer.

The complex and anisotropic nature of composites leads to a complicated response to loading and consequently multiple and interacting failure mechanisms. Isotropic materials usually exhibit fracture normal to the principal stress, whilst in composites fracture is usually a combination of different failure mechanisms. These failure mechanisms can be grouped into translaminar, interlaminar and intralaminar fracture[1,2]. Loading conditions, ply configuration, the respective properties of the constituents and their interface can greatly influence the failure mechanisms and their interactions.

Whilst composites generally offer superior mechanical properties compared to metallic alloys, their performance in longitudinal compression is poor. When a composite structure is subjected to compressive load, delaminations often occur and consequently lead to local weakening. The deterioration of out-of-plane support of the load-bearing fibres due to delamination influences fibre microbuckling which leads to the collapse of the composite structure. Since fibre microbuckling is a key mechanism which can lead to failure, hindering this failure mechanism may be the best route to avoid catastrophic failure and thus improve the compressive performance of fibre-reinforced composites.

The poor compressive performance of fibre-reinforced composites is also related to their sensitivity to defects. In particular, this sensitivity has led designers to deem crack
growth as failure[3]. In fact, this no-growth approach has led to heavy structures with conservative configurations which to some degree have negated the significant weight saving potential that composites offer over metallic structures.

Designs which utilise the concept of crack growth tolerance, i.e. permitting crack growth before deeming failure to have occurred could introduce novel composite material additions or structural features capable of inhibiting catastrophic failure. This approach which is widely accepted in metallic structures[4], can provide designers with the opportunity to introduce innovative structures that offer significant weight savings. Some damage tolerance approaches have used toughened zones[5] and structural features[6-8]. Although advanced composites offering high fracture toughness currently form integral parts of aircraft structures[9], their complex behaviour calls for further research.

The transition from the no-growth approach (damage resistance) to the crack growth tolerance approach can offer great advantages but there are some issues which need to be taken in account. As mentioned earlier, composites exhibit brittle fracture i.e. unstable crack propagation (with little scope for plasticity), which needs to be considered in the design of structural features. Thus the adoption of a crack growth tolerance approach would require a thorough consideration of the various factors such as defect sensitivity and residual stresses which can affect the in-service behaviour.

Essentially, a successful crack growth tolerance approach might require an effective interaction between material and structural features which can attenuate the failure. While the introduction of a second novel material to the parent material must offer optimal properties to inhibit crack initiation and growth (material features), the incorporation of special-purpose features must be also capable of enhancing the mechanical properties and when needed efficiently redirect or arrest the propagating crack (structural features). Both approaches individually play a vital role in the concept of crack arrest, but there is also a need to act in synergy so the composite structure can tolerate damage growth without
sacrificing overall performance (parasitic mass). In fact it is important to ensure that any fall in specific pristine performance does not fall below the enhanced crack growth tolerance.

To summarise, there are two approaches to design against failure in composite materials. The first is the damage resistance in which no damage is tolerated, i.e. the composite is so tough that it will not be detrimentally damaged when exposed to the most severe, once in a lifetime, threat. However, this conservative philosophy has led designers and materials suppliers to develop heavy structures and systems which offer high fracture toughness but modest compressive performance. The philosophy of crack growth tolerance approach, which is the approach taken in this project, suggests that damage growth is tolerated and can be arrested or redirected to a region where it will consequently be self-healed[10]. A crack growth tolerance approach can offer significant benefits such as lighter and complex yet safer structures which would be difficult to achieve by a no-growth approach. Such an approach can potentially provide the structure with the freedom to tolerate significant crack growth while in service. This would consequently lead to reduced need for inspection and repair. Moreover, the designers would be given the freedom and confidence to develop groundbreaking designs. Considering all these advantages, crack growth tolerance is an approach which can take full advantage of composite materials and revolutionise their utilisation in high performance applications.

This research project (Exploratory Crack Arrest Theme 1) is part of the CRack Arrest and Self-Healing in COMposite Structures (CRASHCOMPS) project funded by the EPSRC and Dstl[10]. CRASHCOMPS is a joint research program between Imperial College London and University of Bristol, led by Dr. Emile S. Greenhalgh (Imperial College London) and Professor Ian Bond (University of Bristol).

1.2 Aims
During the past two decades, research on compression crack arrest in composites has mainly focused on inhibiting or arresting interlaminar fracture. As for translaminar fracture,
research has mainly focused on structural arrest approaches for tensile failure. The dearth of studies on arrest of compressive translaminar cracks raised the need to thoroughly investigate the failure modes relating to this particular type of damage and propose ways to hinder translaminar compressive failure. Therefore the aims of this project was to undertake an in-depth investigation of the compressive failure mechanisms which occur in multidirectional composite laminates and consequently to suggest novel approaches for arresting/redirecting compressive damage. This work was focused on the material rather than the structural level, i.e. by introducing a second material to the parent in order to inhibit compressive failure. Crack arrest by the means of structural concepts was covered by another research project under CRASHCOMPS[10].

1.3 Novelty

In this research prior to developing ways to inhibit or arrest compressive translaminar cracks, the compressive failure of multidirectional composite laminates was thoroughly studied. This study constitutes the first attempt in the open literature to extensively study the compressive failure of multidirectional composite laminates using experimental, fractographic, theoretical as well as numerical tools. Moreover, factors which greatly influence the compressive behaviour such as loading conditions, layup configuration and specimen geometry were considered in order to suggest a failure sequence applicable to various types of multidirectional composite laminates. Once the key aspects of compressive failure were identified, novel crack arrest concepts were proposed and investigated. These crack arrest concepts were hybridisation, carbon nanotube reinforcement and piezoelectric actuation. While the carbon nanotube reinforcement and piezoelectric actuation were not pursued due to the immaturity of these concepts and time constraints, the experimental study on hybridisation was the first in the literature to thoroughly investigate the effect of hybridization (same matrix-different fibre types) on the compressive performance of multidirectional laminates. Finally, the outcomes of this work have been published and presented in European and International conferences as well as a journal paper [11-13].
Chapter 2 – Literature Review

2.1 Introduction
The purpose of this chapter is to provide an overview of the literature relating to compressive failure and crack arrest concepts amenable to compressive failure in composites. The first part of this chapter deals with compressive failure of continuous fibre-reinforced composites and covers the state-of-the-art studies on unidirectional composites and high performance multidirectional composites. This knowledge underpins the experimental study of compressive fracture on multidirectional composites, presented in Chapter 4 and Chapter 5.

The second part of this chapter reviews the crack arrest concepts which have been identified in composite materials (such as biocomposites) and could be used in high performance fibre-reinforced composites. Finally novel material systems, which might offer crack arrest capabilities, are presented and evaluated.

2.2 Compressive Failure of Fibre-Reinforced Composites
In general, compressive failure of fibre-reinforced composite laminates is the result of a combination of failure mechanisms often acting in synergy. Compressive failure is strongly related to fibre instability (microbuckling) and matrix yielding, in a region where an imperfection is present such as misaligned fibres or manufacturing defects. Failure mechanisms can be grouped into translaminar, interlaminar and intralaminar fracture [1,2].

![Illustration of translaminar, intralaminar and interlaminar fracture modes](image)

Figure 2-1 Illustration of translaminar, intralaminar and interlaminar fracture modes[1].
Unlike tensile failure, the processes by which compressive failure in fibre-reinforced composites occurs are complex with the fracture not normal to the loading direction and the fractured surfaces continue to carry load even after failure has occurred.

In real structures, multidirectional fibre-reinforced composites are used to achieve optimal performance. Even though multidirectional composites are used extensively, the $0^\circ$ plies carry most of the load in a composite laminate. Therefore, a thorough understanding of the failure mechanisms relating to unidirectional laminates is of great importance and acts as the basis for comprehending the compressive performance of multidirectional composites. In this chapter a brief overview of the state-of-the-art studies on compressive failure in unidirectional fibre-reinforced composites is given, although the main focus is the thorough comprehension of compressive failure of multidirectional composites in light of the pertinent literature.

### 2.2.1 Unidirectional Composites

The main failure mechanisms which may occur separately or in synergy during compressive failure in UD composites are summarised as follows:

**Fibre failure**

- *Elastic microbuckling* which is a term used to describe the shear deformation of the matrix, where there is no relative lateral displacement across the failure zone (Figure 2-2a)

- *Plastic microbuckling or Kinking* is the large non-linear deformation of the matrix which is followed by fibre fracture at two points and a relative displacement across the failure zone (Figure 2-2b)

- *Fibre shear failure* which occurs at the reinforcement usually due to internal defects or flaws at the surface of the fibres (Figure 2-2c)
Matrix failure or fibre/matrix interface failure

- *Ply splitting* which occurs when the matrix fractures parallel to the fibre direction (Figure 2-3a) or in the laminate plane (delamination - Figure 2-3b and Figure 2-3c)

Figure 2-2 Fibre Failure Modes: (a) Elastic microbuckling; (b) Kinking and (c) Fibre fracture.

Figure 2-3 Matrix and fibre/matrix modes; (a) Longitudinal ply splitting; (b) Ply splitting in the laminate plane; (d)Shear band formation.
Shear band formation, in which fracture occurs in a band at 45° with respect to the loading direction (Figure 2-3d)

2.2.1.1 Fibre Failure

Elastic Microbuckling

Microbuckling is a failure mechanism in which linear elastic deformation occurs[14] and is highly influenced by the shear properties of the matrix, due to the large shear deformation that the matrix undergoes during microbuckling. Apart from the low shear properties of the matrix, imperfections such as defects at the fibre/matrix interface, fibre misalignment, residual stresses, porosity and fibre waviness can have a significant effect on the compressive performance of the composite laminate by promoting microbuckling[15].

The current failure models for elastic microbuckling are based on Rosen’s model[16], who employed a simple two-dimensional model where fibres were represented as beams in an elastic matrix. Two possible modes can occur; the extension (out-of-phase fibre deformation) and the shear mode (in-phase fibre deformation). Even though Rosen’s model still stands as the most pioneering model for elastic microbuckling, it overestimates compressive strength by up to two orders of magnitude when compared to the reported experimental results[15]. Most of the studies[15] which followed Rosen’s, employed Greszczuk’s model[17-19]. In Greszczuk’s approach the incorporation of the energy released during bending of the fibres, provided results which were in good agreement with the experimental studies. Greszczuk also suggested that the modulus of the matrix could lead to different failure mechanisms, such as matrix cracking (intermediate modulus) or compressive fibre failure (high modulus).

In later studies, other concepts such as fibre misalignment and matrix non-linearity were incorporated in the models. In particular, it has been reported in the literature that even a small initial fibre misalignment (2-3°) could depress the compressive strength[20-22]. Furthermore, to accommodate the fact that most commercially available matrices did not
behave linearly as Rosen[16] had suggested, Jelf and Fleck[23] developed a composite material model (spaghetti or glass rods in a silicon elastomer matrix) where the matrix behaved in a non-linear manner. However, the predicted compressive strength was not in accordance with the experimental observations[15].

Although this failure mode has been thoroughly studied using finite element models[15, 24, 25], the predictions struggle to replicate the experimental observations. This has been attributed to the fact that these models were simplistic and factors greatly degrading the compressive performance of the first generation fibre-reinforced composites were omitted. Moreover, elastic microbuckling does not occur in high performance unidirectional fibre-reinforced composites, but instead another failure mode occurs, *kinking*, which is discussed below.

**Plastic microbuckling/Kinking**

The term *plastic microbuckling* (or *kinking* to which it is also referred throughout the literature) is used to describe the large deformation which the matrix experiences due to significant rotation of the fibres. As compressive loading increases, so does the severity of a kink band until fibre fracture ensues at two points[26,27]. The kink band formation may be followed by development of further kinking in the adjacent fibres or by other mechanisms mentioned later on in this section.

The duration of kinking is very short and in most cases is catastrophic because it is initiated by fibre fracture, localized fibre microbuckling or shearing of the matrix. The load is then suddenly redistributed around the defect; the matrix is overloaded and hence permanently deformed at the broken fibre sites. Figure 2-4 illustrates the main features and parameters used in modelling the kink band formation, where \( w \) is width of the kink band, \( \beta \) the inclination angle, \( \overline{\varphi} \) the initial fibre rotation and \( \varphi \) the additional fibre rotation.

Throughout the literature, there has been a debate as to whether kinking is a compressive failure mechanism in its own right or merely the irreversible stage of elastic
fibre microbuckling as was described in the previous section[15]. The majority of studies, as Argon[26] first proposed, suggest that kinking is an independent failure mode. In fact, if kinking was merely the irreversible stage of elastic fibre microbuckling, the boundary of the kink band should be in the plane of the highest fibre bending stresses and thus the boundaries should lie perpendicular to the loading axis. On the contrary, the edges of the kink band lie at an angle ranging from 30° to 45°[15, 28]. Kinking is associated with large deformations within the material and therefore it is expected that kink bands lie along planes of maximum shear stresses[27]. Hence the fibre misalignment angle $\varphi$ and the shear yield strength of the matrix are the key factors which control the compressive behaviour of composites.

![Figure 2-4 Geometry of a kink band, with width w, initial fibre misalignment $\varphi$, inclined at an angle $\beta$][29].

In their detailed review of compressive failure of unidirectional composites Schultheisz and Waas[15] suggest that kinking is triggered by manufacturing defects such as fibre misalignment and ply splitting (matrix cracking) and not by fibre fracture as Hahn et al.[30] and Chaudhuri[31] had suggested. Along the same lines, Pinho[27] proposed that ply splitting is the mechanism which initiates the kink band formation and not fibre microbuckling. According to Pinho, as a kink band propagates a zone of ply splits develops ahead of the kink band and as the kink band approaches the split density increases. Therefore the shear strength of the matrix and the fibre/matrix interface strength stand as the properties which dictate the compressive performance[27,32].
Even though in some studies\cite{33-35} parameters such as fibre diameter, fibre volume fraction and fibre/matrix interfacial toughness have been modelled and the kinking/splitting interaction mechanism explained, there has been no experimental validation and neither has the effect of the adjacent off-axis plies on the behaviour of the load-bearing plies been considered. Nevertheless, a series of recent studies\cite{36-38} have provided improved micromechanical models to investigate kink formation and its propagation, which provide good agreement with the experimental results\cite{39,40}.

A notable study which proposed a micromechanical model for kinking was suggested by Pimenta et al.\cite{41,42}, who investigated kinking experimentally as well as numerically. Pimenta et al. noted that the process of kink band formation could be divided into three events or domains (Figure 2-5). Upon loading, the imperfection (e.g. fibre misalignment) induced bending of the fibres and shearing of the matrix (elastic domain). The matrix then yielded until the peak load was reached (softening domain-peak load). At that point the rotation of the fibres increased and the deformation was localized in a narrow band, accompanied by further yielding of the matrix (softening domain-post peak). Finally, once the failure strength of the fibres was reached, the fibres started to fail from the outer plies moving inwards point at which the kink band had been fully defined and ultimate failure had occurred.

![Figure 2-5 Stages of fibre kinking (loaded), (i) elastic domain; (ii) softening domain; (iii) fibre failure domain\cite{41}.](image)
Gutkin et al. [41] investigated the longitudinal compressive failure in unidirectional composites and proposed a series of failure mechanisms which lead to the collapse of the laminate (Figure 2-6). With the aid of fractographic analysis it was shown that during the compressive failure of a notched composite there was a transition from shear driven compression to kink band formation, and that ply splitting played a significant role in the kink-band formation.

![Kink-band formations in a unidirectional specimen][43]

**Figure 2-6 Kink-band formations in a unidirectional specimen[43].**

It should be noted that there are two different types of microbuckling (in-plane and out-of-plane) which can occur in composite laminates. As mentioned previously, microbuckling failure is promoted by reduced lateral support of the load-bearing fibres. This drop in lateral support can arise due to ply splitting, delamination or due to features such as a notch or free edge.

In the case of in-plane microbuckling, the fibres buckle due to lack of lateral support in the vicinity of a ply split or at the tip of a notch creating a band of microbuckled fibres. In-plane microbuckling propagates in a stable manner and does not immediately lead to catastrophic failure. On the other hand, out-of-plane microbuckling usually occurs when a
delamination is present next to a load-bearing ply and is an unstable failure mechanism which can lead to catastrophic failure.

**Fibre Fracture**

Fibre failure is considered throughout the literature as an alternative compression failure mechanism or a mechanism which initiates the failure process as identified by Gutkin[43]. This failure mode generally occurs when the matrix is sufficiently stiff and strong and the fibre/matrix interface strong enough to prevent microbuckling and kinking. In this instance, failure will occur when the fibre strength is exceeded. Such a failure mode was first identified by Ewins and Ham[44]. Moreover, the presence of defects along the fibre and at the laminate surface deteriorates the strength of the fibres and can lead fibre fracture[45]. In general, fibre fracture is not considered as a dominant failure mode in high performance composites[15].

**Fibre Failure Criteria**

The inadequacy of the failure criteria to precisely predict fracture and the difficulty in fully comprehending the failure process of fibre reinforced composites, led composites experts to set up the World Wide Failure Exercise[39,40,46-50]. This exercise was initiated in order to test and evaluate all the existing models in unidirectional and multidirectional composites. Regarding compressive failure, Argon’s initial approach[26] for fibre kinking has been significantly improved, leading to a more precise prediction of kink formation. Davila et al.[36] proposed an improved criterion for fibre kinking using a Mohr-Coulomb criterion[46], and suggested an expression for fibre misalignment $\phi$ (Figure 2-7) in relation to the parameters that play an essential role during compressive failure.

![Figure 2-7 Imperfection in fibre alignment idealized as local region of waviness][36].
The criterion proposed by Davila and further developed by Pinho (LaRC05[51, 52]) for fibre kinking provides improved correlation with the experimental observations compared to Hashin[53,54] (which does not account for the contribution of the in-plane shear that can deteriorate the compressive strength of the ply) and other criteria used in WWFE. The widely accepted failure criterion LaRC05 for fibre failure is given as follows:

\[
F_{KINK} = F_{SPLIT} = \left( \frac{\tau_{23}^m}{S_T - \eta_T \sigma_2^m} \right)^2 + \left( \frac{\tau_{12}^m}{S_L - \eta_L \sigma_2^m} \right)^2 + \left( \frac{\sigma_2^m}{Y_T} \right)^2
\]  

Equation 2-1

where \( \psi \) is the rotation of the kink band plane and \( \varphi \) is the misalignment angle of the subsequent rotation:

\textit{Kinking will occur when:} \( \sigma_1 \leq -X^C / 2 \)

In Figure 2-8 and Figure 2-9, a schematic representation of the aforementioned approach is given, indicating the physical meaning of the various stress and traction components in the respective coordinate systems.
Literature Review

Regarding those which are not depicted, the strength components, $S^{\alpha}_{T}$, $S^{\alpha}_{L}$, and $Y^{\alpha}_{T}$, the so-called in-situ strengths are related to the in-situ effects, that arise in the tensile and shear strengths when a ply is constrained by plies of different fibre orientation. Finally $\eta$ is an experimentally determined constant which may be regarded as an internal material friction parameter[51].

Figure 2-9 Traction components acting on the matrix fracture plane[36].

2.2.1.2 Matrix and Fibre/Matrix Interface Failure
In the next two sections the matrix and fibre/matrix interface failure modes associated with compressive failure of composites are presented.

Ply Splitting
Matrix cracking or ply splitting (as it is commonly referred) is a fracture mode associated with crack development at the fibre/matrix interface[1,55,56]. In unidirectional composites, intralaminar failure develops due to tensile forces transverse to the fibres or shear forces parallel to the fibres. In fact, composites are prone to ply splitting because of the low fibre/matrix interface strength. The factors which dictate the development and the extent of the ply splitting are the interfacial fibre/matrix strength, the strength and stiffness of the matrix as well as the presence of other defects particularly delamination.
According to Sjögren[57,58] three different types of ply splitting can occur during failure(Figure 2-10). The extent to which each of these mechanisms occurs is dictated by the fibre/matrix interfacial strength as well as the fibre and matrix bulk strengths[1]. In particular, if the matrix strength is low, cohesive fracture of the matrix will occur (Figure 2-10a), whereas if the fibre/matrix interfacial strength is low fracture will develop at the interface (Figure 2-10b). Finally, in case of high fibre/matrix interfacial strength translaminar failure will occur (Figure 2-10c) instead, however, this mode is very rare in the toughened resins commonly used nowadays.

![Figure 2-10 Illustration of ply splitting micromechanisms[57].](image)

The strong interaction between ply splitting and other failure mechanisms is also evident in multidirectional laminates. For example, in impact damage, ply splitting induces delamination [59] or delaminations migrate through ply splits [1,60].

**Shear band formation**

Another failure mode which has been observed in fibre-reinforced composites with very low fibre volume fraction is shear band formation. According to Fried[61], fracture in glass-
reinforced composites occurs in a band that is angled 45° with respect to the loading
direction (Figure 2-3c), and is highly controlled by the point at which matrix yields. Similarly,
Jelf and Fleck [23] also observed this failure mode in composites made of spaghetti fibres in
brittle paraffin wax matrix and slender glass rods in Plaster of Paris (calcium sulphate
hemihydrate) matrix. Generally, this failure mechanism is not expected to occur in advanced
composites with conventional volume fractions (higher than 40%).

**Matrix Failure Criteria Associated with Compression**

Researchers have recognised the need to distinguish between the failure modes related to fibre and matrix. Although Hashin[53,54] was the first to suggest different failure criteria for fibre and matrix (further developed by Sun et al.[62]), these criteria do not accurately predict matrix compressive damage.

Puck and Schürmann[48,49], suggested that matrix compressive failure is dominated by in-plane shear and occurs perpendicular to the ply and parallel to the fibres, i.e. \( \alpha = 0^\circ \), where \( \alpha \) is the fracture plane angle, a key element in Puck and Schürmann’s hypothesis[36]. As the compression stress increases it reaches a point where \( \alpha = 40^\circ \) and afterwards \( \alpha = 53 \pm 2^\circ \) for pure transverse compression[27,48,49]. However, the most up-to-date matrix failure criterion LaRC05[51] provides a more in-depth mechanics analysis compared to the previous versions in the LaRC series[36-38,63-65]. In fact, in LaRC05 the micromechanics of the failure process at the microscopic and mesoscopic scale have been taken in account, thus allowing for solutions to be computed for laminae as well as laminates. The LaRC05 criterion is given by the following expressions, whilst the physical meaning of the stress and traction components is provided in Figure 2-8 and Figure 2-9:

\[
F_{M}^I = \left( \frac{\tau_T}{S_T - \eta_T \sigma_N} \right)^2 + \left( \frac{\tau_L}{S_L - \eta_L \sigma_N} \right)^2 + \left( \frac{\langle \sigma_N \rangle}{Y_T} \right)^2
\]

**Equation 2-2**

*Matrix failure initiates when* \( F_{M}^I \geq 1 \)
Finally considering Equation 2.1, the LaRC05 criterion suggests that splitting (or fibre splitting as it is referred in the criterion) will occur when:

$$\sigma_t \geq -\frac{X_c}{2}$$

### 2.2.2 Multidirectional Laminates

Building on the knowledge of compressive failure in unidirectional composites, the state-of-the-art studies on compressive failure of multidirectional composites are presented, starting with cross-ply laminates and proceeding to the more complex multidirectional laminates.

#### 2.2.2.1 Cross-ply Laminates

Compressive failure of cross-ply laminates was recently studied by Gutkin et al.[43] who used in-situ SEM to directly observe kink-band formation. According to this study the fracture in a notched cross-ply laminate initiated at the notch tip in the form of a shear driven fibre compressive failure which propagated at 45° with respect to the load-bearing fibres (Figure 2-11).

![Figure 2-11 Typical compressive fractures in a cross-ply specimen; (a) Overall view of the fracture process; (b) schematic of the failure process; (c) Definition of the three different patterns; (d) Close-up view of the transition region[43].](image-url)
This fracture consequently transformed into a kink-band which propagated in the plane at 25° with respect to the fibre orientation. Under increased compressive loading the faces of the shear crack slid over each other whilst the fibres ahead of the shear crack rotated. Due to the applied compressive load and the rotation of the fibres, a kink-band was subsequently formed. The friction generated between these faces led to the rotation of the fibres on the lower part of the crack and hence fibre fracture due to bending and compressive stresses (Figure 29). Finally, in this study it was suggested that the presence of the 90° plies inhibited the 0° ply splitting because it constrained the fibre rotation, effectively delaying the transition to kink-band formation.

2.2.2.2 Multidirectional Laminates
Early studies conducted on compressive failure of multidirectional composites coincided with the utilization of fractography for post-failure analysis[66-71]. Studies by Ewins[44] and Potter[68] were the first on fractography of carbon fibre reinforced composites which had failed in compression. Potter employed two multidirectional configurations ([45/0/-45/0]3S and ([45/-45/0]/45/-45/0)/45/-45/0]5) with a side notch to initiate compressive failure, and observed that extensive translaminar cracks (D) (Figure 2-12a) had formed in the vicinity of the notch in the majority of the plies (where the stress concentration was maximum) and were related to the induced delamination (C) (Figure 2-12a).

![Figure 2-12 (a) Fibre fracture in off-axis plies; (b) magnification of fracture next to the hole[68].]
Potter also observed that the off-axis plies affected the behaviour of the axial load-bearing plies and that the translaminar cracks in the axial plies led to in-plane shear fractures at the off-axis plies (B) (Figure 2-12a). However, delamination in this study was only observed as a result of translaminar failure and not as a primary failure mechanism which often occurs in laminated composites[1].

![Figure 2-13](a) Buckling failures in axial ply; (b) mixed compression and shear failure[68].

Along the same lines, a very interesting phenomenon was observed by both Potter[68] and Pinnell et al.[72]. The presence of an intralaminar crack at the 45° ply governed the fibre microbuckling at the axial plies and reduced the load carrying capability of the axial plies, i.e. the ply split acted as stress concentration on the 0° layers (Figure 2-13).

However, Pinnell et al.[72], who studied the compressive failure of multidirectional laminates in plain compression (to alleviate the influence of the notch), tested both monolithic laminates and sandwich panels and also observed a step-like pattern (Figure 2-14). Pinnell noted that the axial ply followed the fracture path of its adjacent angle ply translaminar failure and the fibres had failed at an angle with respect to the fibre axis at the edge of each step and perpendicular to the fibre axis within each step. This suggested that fibres failed under in-plane shear[1] (Figure 2-14).
Pinnell et al. also observed that the delaminations (Figure 2-15) were mainly located at 0°/45° and 45°/90° ply interfaces. Similar results were observed by Purslow[73]. To support these findings, a stress analysis was also conducted to obtain the interlaminar stresses near the free edges. Pinnell suggested the sequence of events which led to global fracture. In particular, the skins had failed by a combination of shear and compression which resulted in the fracture by global buckling of the one skins. With reference to Figure 2-14, the intralaminar fracture of the 45° ply (close to the surface) was followed by translaminar fracture in the adjacent 0° ply. Once the 45° ply had failed, the axial fibres began to shear, following the fracture line of the angle plies. Consequently the 0° ply sheared parallel to the 45° ply fibres until it reached the fracture path introduced by the angle ply. This process continued across the width of the specimen resulting in the observed step-like pattern (Figure 2-14).

Interestingly, according to these same studies the fracture of the axial ply had occurred prior or simultaneously with delamination. If the 0°/45° delamination had occurred first, the axial ply would have lost support, buckled and failed in a different way to that described above. Had the delamination occurred first, the delamination section would have buckled resulting in a high interlaminar stress concentration at the delamination tip. Increased loading would have caused buckling driven delamination resulting in global instability and catastrophic fracture. Although similar failure mechanisms were observed in
the two different configurations, in general monolithic laminates buckled, whereas the sandwich skins were restricted to ensure pure membrane (in-plane) failure. Although the study by Pinnell was the first to suggest the sequence of events which led to global failure, the fractographic analysis was mainly focused on the outer 0° load-bearing plies and there was no fractographic examination of the ply interfaces that had failed by delamination.

![Figure 2-15](image)

**Figure 2-15** (a) Longitudinal cross-section of (0/45/90/-45)s monolithic laminate (×6.5); (b) Longitudinal cross-section of sandwich skins 1 (up) and 2 (down) (×5.5) [72].

Shikhmanter *et al.* [71] verified the results obtained by Potter [68] and Pinnell *et al.* [72] but also quantified the step height of the step-like morphology which was found to be a multiple of 0.010-0.014 mm for the angle plies and 0.018-0.020 mm for the axial plies. This was attributed to the difference in the height of the steps i.e. in the angle plies it was lower due to the difference in orientation of the fibres with respect to the loading direction. In the same study, delaminations were also observed across the width of the specimen. In delaminated 0°/90° ply interfaces, dense narrow bands were observed; this was evidence of 90° ply splitting prior to delamination.

With regards to the quasi-isotropic laminates, delamination in the 45°/-45° ply interface was also characterized by resin-poor and resin-rich zones due to manufacturing. Finally evidence of local formation of both interlaminar peel (Mode I) and shear (Mode II) fracture was observed. Similar fracture features were observed by the study of Cina [66]. However, none of these studies suggested how delamination had interacted with ply splitting and translaminar fracture.
Usually in structural components the surface layers are angle plies for damage tolerance reasons. The compressive failure of such structures reveals some important features relating to the series of events which lead to global fracture. A novel technique to investigate the source and the sequence of failure mechanisms during compressive loading was suggested by Greenhalgh and Cox [74]. In this study, multidirectional side notched specimens of different outer plies configurations were investigated.

The arrowhead technique proposed by the authors was based on the concept that the surface ply split features could provide valuable information about the failure initiation site and propagation. Figure 2-16 is representative of the suggested technique and shows the observed surface split distribution. According to the same study, translaminar compressive cracks were found to have grown from the notch. Diagonal surface splits (parallel to the ply direction) were observed below the cracks with less surfaces splitting above the cracks (Figure 2-16).

![Figure 2-16 Coupon with 45°/-45° surface lay-up (side-notched) with splitting][74].

Although there is literature on compressive failure of laminates containing notches and open holes [75,76] and most notably the work of Soutis [77-83], until recently none have attempted to characterise the influence of delamination in the failure process with the aid of fractographic analysis. However, Suemasu et al. [84] investigated the compressive failure
mechanisms of multidirectional laminates with an open hole where the importance of delaminations in the failure process was highlighted. In this study it was observed that in a laminate with high interlaminar toughness, in-plane microbuckling of the load-bearing fibres triggered the failure process, whereas with low interlaminar toughness, delaminations constituted initial failure and were accompanied by sudden overall failure.

Finally in a recent study, Prabhakar and Waas[85] investigated the interaction of compressive failure modes in a multidirectional unnotched laminate. Based on the Pagano and Pipes[86] and Martin et al.[87] theoretical formulations, as well as experimental results, the weakest ply interface (45°/45°) in a (-45/45/90/0)₅ was determined and used to build a numerical model. The outcomes of this model were in good agreement with the experimental results.

Prabhakar and Waas also tested thicker laminates, however no significant influence of scaling of laminae in the laminate was observed. It was highlighted that delamination was the first failure to occur which then induced fibre kinking, while there was no evidence of whether this failure sequence was representative of the particular layup or how the suggested failure sequence was influenced by scaling of the laminate. Moreover, in this study there was no comment on the failure propagation after the kinking had occurred as

Figure 2-17 Damage progresses during open hole compression of a T800H/3633 (45/0/-45/90)ₛ₅ specimen[84].
well as on the influence of the observed delaminations in other ply interfaces (90°/0° and 45°/90°) on the failure propagation.

### 2.2.3 Summary of Literature in Compressive Failure

Several studies have been conducted to investigate the failure mechanisms in multidirectional CFRPs under compressive loading. Even though different specimen configurations have been used, common failure mechanisms have been noted. Notched specimens have been used throughout the literature, hence failure initiated around the notch/hole, mainly as longitudinal ply splitting and propagated across the specimen. In addition, studies which employed unnotched specimens in plain compression to alleviate the effects of the stress raiser were also presented.

However, there seems to be a discrepancy as to which mechanism accompanies ply splitting. Delamination in most cases was induced by ply splits but could also occur after or simultaneously with kinking. Essentially, delamination separates the laminate into two or more sub-laminates which can buckle out-of-plane and consequently lead to multiple delaminations across the width of the specimen. Usually, depending on the load application, multiple delaminations lead to global fracture and collapse of the laminate.

Another possible series of events which can occur under certain conditions should also be noted. Upon compressive loading, delaminations are likely to occur before ply splitting. This can occur when the laminate buckles globally leading to flexural loads in addition to the in-plane compressive loading inducing large interlaminar strains. Even though delamination reduces the load bearing capability of the axial plies, little or no fibre fracture occurs at this failure sequence since delaminations dominate across the specimen width. This failure mode is often referred to as “green stick” failure[1] and is illustrated in Figure 2-18. Although the failure mechanisms which lead to fracture of composites material are not entirely understood especially under compression, in light of the literature reviewed above three main failure modes can be distinguished (as illustrated in Figure 2-1):
- Fibre Fracture – Translaminar Failure
- Ply Splitting (Matrix Cracking) – Intralaminar Failure
- Delamination – Interlaminar Failure

The first two processes were discussed in the previous section (2.2.1), but in multidirectional laminates the additional failure mode of delamination plays an important role.

![Figure 2-18 Compression failure in multidirectional laminate with (a) limited delamination, (b) delamination prior to failure](image)

Delamination[1,88-92] occurs in laminated composites due to excessive interlaminar stresses at the interfaces between adjacent plies. This failure mode is generally associated with defects induced during the manufacturing process, out-of-plane stresses introduced by initial fibre misalignment or impact damage. Delaminations strongly interact with translaminar damage causing a reduction in the lateral support of load-bearing plies leading to further damage growth and even premature failure. Although in the literature there is a huge body of work on delamination, mainly focussed on analysing unidirectional coupons[93], there is also
considerable work on delamination in multidirectional ply interfaces[1,88]. Even though delamination is a dominant compressive failure mode in multidirectional laminates, as it will be shown in the experimental studies (Chapter 4 and Chapter 5), attempting to cover this failure mode is beyond the scope of this literature review. Further information about this failure mode can be found in the literature[1,93].

2.3.4 Compression Specimen Geometry Selection
The complex nature of composite materials and their sensitivity to defects and stress concentrations has been essential in the determination of the optimal specimen geometry for compression testing. Specimen size and shape, stress concentrations from notches/holes, global buckling and load application are only a few of the factors which need to be taken into account when a compression test fixture is developed[15].

Compressive failure of composites has been studied since the late 1960s, but it was only recently that compression tests became standardised (D3410[94], D6641[95], D5467[96]) by ASTM. In the open literature many testing configurations and specimen geometries[15] have been suggested, mainly incorporating a combined compressive and shear end loading, however there is no widely accepted standard. This is because the tests which have been developed to study the compressive performance of composite laminates have been specially tailored to investigate particular failure mechanisms such as fibre microbuckling. A thorough historic overview of the early compressive test fixtures can be found in the review by Shultheisz[15]. However, compression testing of sharp or circular notched specimens was not discussed in this review.

Although many specimen geometries have been suggested, such as untabbed rectangular specimens (using the Wyoming Combined Loading Compression (CLC) Test Method[97]) rectangular with end tabs (Celanese-IITRI) and sandwich composites (four-point bending)[15], specimen configurations which incorporated a stress raiser (notch or hole) have been also been extensively used to study the compressive failure of composite
laminates [43,44,74,78-84,98,99]. In fact, specimens containing notches (sharp or round) or holes, provide a more controlled compressive behaviour if membrane load can be ensured and global buckling is avoided. After thorough consideration of the literature and the objectives of this study, as explained in detail in Chapter 3, composite specimens containing notches were selected for this investigation. However, the high sensitivity of the compressive failure on the notch size and sharpness was well acknowledged. To encompass the effect of the notch size and sharpness, in this research, notches of various geometries and sharpness were utilised.

2.3 Crack Arrest

2.3.1 Crack Arrest in Composites

Composites offer a good combination of properties such as strength and stiffness compared to metals, nevertheless they are more brittle and thus more prone to fracture. Upon exceeding a critical stress, the crack propagates rapidly in an unstable manner rather than failure developing in a more benign manner such as via plasticity like in metals. Therefore, it is very difficult to arrest a rapidly propagating crack.

The driving force behind crack propagation is the stored strain energy in the material. Hence to prevent a crack from propagating, this released strain energy must somehow be absorbed. In the literature several attempts have been made to investigate the potential of crack arrest mainly of translaminar failure in tension or interlaminar failure, where the majority involves introduction of structural features within the component such as buffer strips and z-pins. Although delamination is an important failure mechanism for compressive failure of composite laminates, it is beyond the scope of this literature review to cover such concepts. However, more information can be found in the relevant literature [100-103].

The philosophy of this project is to suggest solutions capable of arresting or redirecting translaminar compressive cracks by employing novel material approaches. There is a dearth of studies in the literature, especially for compression, in this field. Therefore the
aim of this section is to provide a brief overview of crack arrest concepts which have been suggested in the literature. These mainly focus on natural composites which often perform and fail under compression. Finally, a brief overview of the concepts which have been considered as potential crack arrest candidates are presented, whilst those that were chosen for this project are discussed.

2.3.2 Crack Arrest Concepts in Natural Composites
In this section an overview of the failure processes in natural composites and the way nature arrests compression cracks will be provided. In light of this knowledge, such concepts could potentially be implemented in synthetic fibre-reinforced composites materials. Albeit four groups of natural composites (namely bone, teeth, wood and nacre) have been studied, no literature pertinent to compressive failure arrest in bone and teeth is available according to the author's knowledge. Therefore, only the concepts on how compressive failure is arrested in wood and nacreous shells will be presented in this section.

2.3.2.1 Wood
Wood has a complex hierarchical architecture with four levels of structure: molecular, fibrillar, cellular and macroscopic. The most important of all these levels is cellular which contributes to the high mechanical properties. Cellulose, the main component of the cellular structure, consists of microfibrils with amorphous and crystalline regions. Macroscopically wood is composed of cellulose macrofibrils in a matrix of lignin, hemicellulose and other compounds.

Wood, as a composite, exhibits high anisotropy in mechanical properties. As mentioned above, cellulose fibrils are the most important component and being the reinforcement greatly influences the mechanical properties of wood. According to Gibson and Ashby[104], wood exhibits three orthogonal planes of symmetry (radial, tangential, axial) and the stiffness and strength are much higher in the axial direction (by a factor between 2 and 20 respectively). When wood is subjected to compressive loading (common loading condition), the performance in the three directions is quite different. The failure mode
depends on the loading direction as well as the direction of reinforcement. As Figure 2-19 illustrates, six failure modes can occur under compression in the axial direction; crushing, wedge splitting, shearing, splitting, crushing and splitting, and finally, brooming and end rolling.

![Failure types of non-buckling clear wood in compression parallel to grain](image)

**Figure 2-19** Failure types of non-buckling clear wood in compression parallel to grain: (a) crushing; (b) wedge splitting; (c) shearing; (d) splitting; (e) crushing and splitting; (f) brooming and end rolling (redrawn from Bodig[105]).

![Failure types in clear wood in compression at an angle with respect to the grain](image)

**Figure 2-20** Failure types in clear wood in compression at an angle with respect to the grain: (a) crushing of an earlywood zone; (b) shearing along a growth ring (redrawn from Bodig[105]).

It should be noted that each of these failure modes can only occur if ply splitting does not occur first. In the radial direction, compressive load causes uniform bending/plastic collapse which consequently induces crushing failure in the earlywood zone (Figure 2-20a).
Finally, in the tangential direction compressive loading leads to off-axis splitting (Figure 2-20b). As in synthetic fibre-reinforced composites, cracks in wood propagate more easily parallel (splitting) rather than perpendicular to fibres. Therefore arrest of propagating cracks parallel to the grain is more difficult and usually occurs in the latewood region where the cells are smaller thus inducing a complicated path for the crack to propagate. On the contrary, cracks in the perpendicular direction tend to be arrested in the early wood region where the cells are larger and the walls are thinner. However, in the arrest of such cracks the remarkably high (approximately 10 times greater than synthetic fibre-reinforced composites) fracture toughness also plays a significant role.

2.3.2.2 Nacreous shells

Nacre is an organic-inorganic bio-composite which is the main constituent of the inner layer of abalone and pearl oyster shells. Both the outer (Periostracum) and inner layers (Epithelium, Prismatic Calcite and Nacreous Aragonite) of the shells are made of calcium carbonate CaCO$_3$, however the microstructure of these layers differs. Concerning the outer layer this is made of calcite, the rhombohedral form of CaCO$_3$, whereas nacre is composed of aragonite, the orthorhombic form of calcium carbonate, as Figure 2-21 illustrates.

![Figure 2-21 Structure of typical abalone shell][106].

Nacreous shells exhibit special performance in compression and crack arrest capabilities. In particular, nacre consists of a tiled structure of crystalline aragonite which plays a critical role in the mechanical properties and acts as crack deflector. Under compression loading, the microstructure of nacre fails by fibre kinking similar to that in
CFRPs (Figure 2-22). Moreover the angle $\alpha$ in nacre is relatively similar to the angle usually observed in synthetic fibre-reinforced composites (approximately $35^\circ$). Although the anisotropy in the mechanical properties is high both perpendicular and parallel to the nacre tiles, there is a remarkable phenomenon which nacre exhibits. The compressive strength is higher (approximately double) perpendicular to the reinforcement compared to that parallel to the tiles due to ply splitting and kinking occurring parallel to the reinforcement[107].

![Figure 2-22](image1.png)

**Figure 2-22** Mechanisms of damage accumulation in nacreous region of abalone through plastic microbuckling[108].

![Figure 2-23](image2.png)

**Figure 2-23** (a) Cross-section of abalone shell showing how a crack, starting at the left is deflected by viscoplastic layer between calcium carbonate lamellae; (b) Schematic drawing showing arrangement of calcium carbonate in nacre, forming a miniature “brick and mortar” structure[109].
Considering the toughening mechanisms occurring in the nacreous biocomposites, nacre exhibits a sophisticated means of arresting cracks. In this mechanism the organic matrix and the reinforcement/matrix play an essential role. The interaction between the tiles and consequently the layers due to the viscoelastic organic interactions, caused by the organic matrix, make the propagation of a crack very tortuous. As Figure 2-23 illustrates, the compression crack needs to propagate through a complicated path to reach the other end of the nacre structure.

2.3.2.3 Summary of crack arrest in natural composites
To summarise, biological and natural composites exhibit crack arrest capabilities, mainly by creating a complicated path for the crack to propagate, through which most of the released strain energy is dissipated. Although such crack arrest concepts seem very attractive, they cannot be directly applied to synthetic fibre-reinforced composites as far as this study is concerned. This is due to the fact that modifying the microstructure at such a level was not practical within the constraints of this project. However the effect of the special structure of nacreous shells on the compressive failure could be exploited in a non-material approach which would make compressive cracks tortuous. In light of this knowledge, other concepts which could offer crack arrest capabilities have been studied and are presented in the following section.

2.3.3 Material Candidate Concepts for Crack Arrest in Composites
Several novel materials have been considered as crack arrest/redirect solutions such as carbon nanotube reinforcing, piezoelectric actuation, hybridization, shear thickening fluids, electrostrictive polymers and shape memory polymers.

Carbon nanotubes is an allotropic phase of carbon with cylindrical nanostructure[110]. Due to their low density, superior mechanical, electrical and thermal properties compared to conventional materials, these materials have been considered as excellent candidates for reinforcing high performance composites[111]. Regarding crack
arrest, carbon nanotubes would provide local stiffening of the matrix around the reinforcement and hence inhibit fibre microbuckling.

Piezoelectric materials exhibit a special property; when they are subjected to mechanical load they generate electric field and vice versa[112]. Their ability to operate as actuators and sensors has made these materials very attractive especially for health monitoring and shape control of composite structures[113]. As a potential crack arrest concept, piezoelectric actuation would induce a tensile stress field ahead of the crack tip to decelerate, redirect or halt a propagating compressive crack.

In composites, hybridisation is the term which is used to describe the process of incorporating different types of reinforcements or resins within the same composite material. The main purpose of hybridisation is to enhance particular properties and lower the overall manufacturing cost of composite laminates[114]. Hybridization would focus upon the enhancement of the lateral support of the load-bearing plies which consequently would inhibit propagation of translaminar cracks. Although the first step is to assess the effect of hybridisation on the compressive performance, at a later stage discrete areas of hybrid material or architecture would be placed in strategic areas such as along a propagating compressive crack.

Shear thickening fluids (STFs) are colloidal suspensions which upon rapid loading exhibit significant stiffening[115]. These materials exhibit a non-Newtonian flow behaviour, which is characterized by a discontinuous increase in viscosity with increasing shear stress. Under normal conditions they behave as a liquid, but when sudden mechanical stress is applied, the particles form a network in a few milliseconds, which leads to a solid-like behaviour. In fact it has been reported that STFs improved the ballistic properties of aramid fabrics and the damping properties[115-117]. Even though no studies are available in the literature on the effect of such materials on the compressive performance, it is thought that shear thickening fluids would induce local stiffening at key points (prone to crack initiation)
by their ability to yield rapid stiffening upon abrupt load application (non-Newtonian flow behaviour).

Electrostrictive polymers is another type of smart materials which exhibits the same performance as piezoelectric materials, i.e. they induce a strain field upon electrical actuation[118,119]. The magnitude of deformation is proportional to the square of applied voltage and inversely proportional to the thickness of electrostrictive polymer[118]. Nevertheless, although there is evidence that electrostrictive polymers can induce strains two orders of magnitude greater than that in piezoelectric materials, the slow response and the complex algorithms which are required for actuation hinder these materials from replacing traditional materials such as piezoelectric, thermomechanical and electrostatic.

Finally, shape memory polymers (SMPs) are materials which are capable of returning to their original shape after they have been deformed with the application of external stimulus[120-122]. Although this ability has been originally observed in shape memory alloys (SMAs), the mechanism is more complex in polymers[122]. Moreover, in shape memory polymers the stimulation can be achieved apart from thermal load with visible and IR light irradiation, electric field or even immersion in water. Even though shape memory polymers have been employed in composites mainly with carbon nanotubes and short carbon fibres, there is no literature on the actuating capabilities of these materials. As crack arrest concepts, shape memory polymers would be utilized in a pre-deformed state and upon thermal load application they would return to the original state by inducing simultaneously tensile stresses at strategic points.

In light of the literature presented above, Table 2-1 briefly presents the key advantages and disadvantages of each of these concepts considered in this project. After consideration, only one of those concepts was chosen for this project namely hybridization, taking in account the maturity of each concept and the time boundaries of this work.
<table>
<thead>
<tr>
<th>Materials</th>
<th>Advantages</th>
<th>Disadvantages</th>
<th>References</th>
</tr>
</thead>
<tbody>
<tr>
<td><strong>Hierarchical Composites</strong>&lt;br&gt;(CNT reinforcement)</td>
<td>Superior strength and stiffness, chemical compatibility, high matrix stiffening locally</td>
<td>Large scale manufacturing still not practical, relatively low volume fraction of grafted CNTs (approximately 15%), deterioration of tensile properties</td>
<td>[110,111,123-127]</td>
</tr>
<tr>
<td><strong>Piezoelectric actuation</strong></td>
<td>High local stiffening, large strain energy release, delamination suppression</td>
<td>Slow response to actuation, brittleness, chemical incompatibility, complex programming required</td>
<td>[112,113,128-131]</td>
</tr>
<tr>
<td><strong>Hybridization</strong></td>
<td>Local stiffening, delamination and impact resistance improvement, improvement of through-thickness performance, lower cost</td>
<td>Modest improvement of strength and stiffness, need for chemical compatibility</td>
<td>[114,132-140]</td>
</tr>
<tr>
<td><strong>Shear thickening fluids</strong></td>
<td>Rapid local stiffening, superior compressive and shear performance, low cost</td>
<td>Manufacturing limitations (consolidation), limited to woven fabrics and flexible structures (i.e. body armour)</td>
<td>[115,116,141]</td>
</tr>
<tr>
<td><strong>Electrostrictive polymers</strong></td>
<td>Chemical compatibility, less complex programming is required in comparison to piezoelectrics, resilient</td>
<td>Modest deformation and strain energy release, moisture sensitivity, poor heat resistance and performance at high voltages, slow response to actuation</td>
<td>[118,119]</td>
</tr>
<tr>
<td><strong>Shape memory polymers</strong></td>
<td>Pre-stressing, shape memory effect can be achieved by tailoring of the molecular structure</td>
<td>Manufacturing limitations (consolidation), large residual stresses, moisture sensitivity</td>
<td>[118-122,142]</td>
</tr>
</tbody>
</table>

*Table 2-1 Crack arrest concept evaluation.*
The class of hybrid composites which has attracted most attention is the sandwich composite laminates[132]. These hybrid composites are made by incorporating an interlayer (composite or not) between adjacent carbon and other fibre-reinforced laminae. Some common examples of the latter are the sandwich panels made of fibre-reinforced laminae and honeycomb core, CARE® (carbon reinforced aluminium laminate), GLARE® (glass reinforced aluminium laminate) and ARALL® (aramid reinforced aluminium laminate)[132]. These hybrid composites have attracted attention due to the higher fatigue and impact resistance[138, 143-146] they offer compared to the conventional composites used in the aircraft industry and hence they have been utilized in primary structural applications, e.g. fuselage(GLARE®). Although these materials are considered hybrids, their aim is to change the geometry and not the intrinsic material behaviour. Therefore it is beyond the scope of this review to cover in detail this class of hybrids. More information can be found in the relevant literature[9,132].

Alternative hybridisation approaches have also been suggested in the literature such as hybrid commingled yarns [139,147-149]. However, these approaches have focused on the improvement of delamination fracture toughness. In particular, Thanomsilp and Hogg observed that the introduction of polymeric fibres (PP, PA) into the glass fabric improved the Mode I delamination fracture resistance (due to fibre bridging) but had a moderate effect on Mode II delamination fracture toughness. Nonetheless, the introduction of mPET fibres, which dissolved in the matrix had little effect on Mode I delamination fracture toughness but enhanced Mode II delamination fracture toughness (due to the significant plastic deformation of the thermoplastics[149]. Thanomsilp and Hogg also studied the effect of hybridisation on the penetration impact resistance[139]. Ditto, the introduction of PP and PA fibres into the glass fabric improved the impact resistance, whereas the introduction of mPET fibres had no effect whatsoever. In fact, in the fractographic analysis it was observed that the thermoplastic fibres deformed plastically and that the plastic deformation was characterised by drawing of the fibres in comparison to the clean fracture of the glass fibres.
To summarise, hybrid composites offer significant improvement in particular properties in comparison to traditional CFRP composites. Even though the main attention has been paid to impact resistance, it has been observed that hybridisation offer enhanced delamination resistance mainly due to fibre bridging. As it was reported in this chapter, translammar compressive fracture is greatly influenced by delamination. Therefore a way to improve the compressive performance of a composite laminate may be to enhance the lateral support of the load bearing plies which is one of the reasons of employing hybridization as a crack arrest/redirection solution. The majority of studies have used different materials or architectures to hybridize composite laminates to improve some key properties. However, according to the author’s knowledge there are no studies which have attempted hybridizing a composite laminate using different types of carbon fibres. Therefore this approach was taken in the current study and is presented in detail in Chapter 6.

2.4 Summary of Literature Review

In this chapter, the mechanisms which lead to compressive failure of fibre-reinforced composites were presented with the aid of the state-of-the-art studies and theories found in the literature. In addition, gaps in the literature, especially on compressive failure of multidirectional laminates, were identified. It was noted that no studies suggest clearly the sequence of failure events that lead to global failure of multidirectional composites and explain the role of delamination on the compressive failure process. In light of this knowledge, the inhibition and arrest of translammar compressive failure was discussed and a brief overview of the concept which could potentially provide crack arrest capabilities in fibre-reinforced composites was provided, namely hybridisation.

The gaps in the literature on compressive failure of multidirectional composite laminates and the suggested crack arrest concepts underpin the experimental studies presented in Chapter 4, Chapter 5 and Chapter 6. In particular, in Chapter 4 an extensive study on the compressive failure of multidirectional composite laminates is presented. In Chapter 5, the effect of hybridisation on the compressive performance is investigated and in
Chapter 6 a complementary study on the delamination fracture toughness of hybrid interfaces is also provided. An overall discussion of the results is given in Chapter 7 and conclusions drawn in Chapter 8, whereas in Chapter 9 the implications of this research project and recommendations for future work are provided. However, prior to that in Chapter 3, the experimental procedure and the analytical and numerical tools which were used in this work are documented.
Chapter 3 – Methodology

3.1 Chapter Introduction

The aim of this chapter is to present and justify the different techniques utilised throughout this work. In particular, a detailed presentation of the experimental procedures, post-failure examination, theoretical analysis and numerical modelling is provided.

3.2 Experimental Studies

In this section, the details of the experimental studies on compression testing of monolithic and hybrid multidirectional composites as well as the delamination fracture toughness testing of hybrid composites are presented.

3.2.1 Compressive Testing

3.2.1.1 Multidirectional Composites

Materials

For the compressive failure of multidirectional composites study, Hexcel IM7/8552 unidirectional pre-preg tape was used with a nominal ply thickness of 0.125 mm (0.123 ± 0.002 mm, experimentally calculated – see Appendix A) and nominal fibre volume fraction of 60% (59.7 ± 0.6%, experimentally calculated – see Appendix B). This high performance system is widely used in the aerospace industry. According to the manufacturer[150] and the literature[52,151], the mechanical properties of the IM7/8552 lamina are: longitudinal compression strength (-1690/-1590 MPa) and Young’s modulus (150/142 GPa), longitudinal tensile strength (2724/2560 MPa) and modulus (164/165 GPa), in-plane shear strength (120/90 MPa) and modulus (5.17/5.6 GPa). A detailed presentation of the reported values of the IM7/8552 lamina mechanical properties is given later in this chapter.
Manufacturing and Specimen Configuration

Panels with dimensions 430 mm × 300 mm were manufactured according to the supplier’s recommendations. Each panel comprised 32 plies with different stacking sequence, \((0/90)_{8S}\), \((90/0)_{8S}\), \((0/90/45/-45)_{4S}\) and \((-45/45/0/90)_{4S}\). These cross-ply and multidirectional configurations have been extensively used for research in numerous publications as well as studied by industry. For the cross-ply and multidirectional layups the relative position of the axial and off-axis plies provided the opportunity to investigate the contribution of these load-bearing and off-axis plies on the compressive behaviour of the multidirectional laminates. It should be noted that the \((90/0)_{8S}\) and \((0/90/45/-45)_{4S}\) configurations were deemed as the cross-ply and multidirectional baseline configurations respectively.

The specimen geometry used in this study was similar to that employed by Pinho et al.[152] to measure the translaminar fracture toughness associated with kink-band formation (Figure 3-1). The compact compression (CC) configuration was chosen due to the reported stable crack growth and the absence of global buckling during compression loading, since no other compression specimen geometry was found in the literature which fully matched those characteristics.

To manufacture the CC specimens, the panels were cut with a wet saw to dimensions 60 mm × 65 mm. Holes for loading pins were then drilled and a semi-circular notch was introduced using a 4 mm wide diamond-coated circular saw, extending 31 mm from the free edge. The notch edges were widened to avoid contact between opposing faces during loading. To ensure that the notch shape and the widening of the notch edges were identical in all specimens, the desired dimensions were sketched in detail on each specimen and a wooden template was employed. Finally, a speckle pattern was painted on the surface to facilitate Digital Image Correlation (DIC). The surface was initially painted white and a fine black speckle pattern (5-15 µm diameter) was introduced on top using an airbrush (Figure 3-2). More information about the DIC calibration can be found in Appendix C.
Compact Compression Testing

The compact compression testing was conducted in a 10-ton servo-hydraulic Instron machine equipped with a 10 kN load cell. Five specimens per configuration were tested and Linear Variable Differential Transducers (LVDT) were attached to the loading pins to record opening displacement during loading.
Compressive Failure of Multidirectional Composites Study

After installing the specimen onto the test machine, a pair of Digital Image Correlation cameras (Schneider Kreuznach Componon S-2.8/50 mm) were placed approximately 10 cm from the specimen surface which was illuminated using two halogen lights (40 Watts). The DIC system employed for the surface deformation measurement was GOM Aramis (v5.7). Digital Image Correlation is an optical method which allows accurate two and three dimensional measurement of changes in digital images. This method uses images of the surface of a structure with a painted stochastic speckle pattern and allocates coordinates to the image pixels. Initially, a reference image represents the undeformed state of the structure. As soon as load or moment is applied onto the structure, the optical system incrementally registers the deformed states of the structure. Then the speckle patterns of these images are compared with the reference image and the displacement and deformation of the structure is calculated.

To identify the location of each object point, the optical system uses a correlation algorithm, which is based on the tracking of the grey value pattern $G(x, y)$ in locally neighbouring facets (square or rectangular image details). Regarding the accuracy of the optical system, GOM Aramis (v5.7) was capable of a 0.04 pixel matching accuracy and 0.02% strain accuracy (Appendix C). More information regarding the background of the Digital Image Correlation method can be found in the literature[153-156].

Finally, all the specimens were loaded in displacement control at a rate of 1mm/min. The data acquisition step for load-displacement was one second and DIC frames were taken every two seconds, covering the specimen area highlighted in Figure 3-3. Loading was halted prior to catastrophic failure or contact between opposing pre-crack faces.

3.2.1.2 Multidirectional Hybrid Composites

For the hybrid compression failure study four configurations with an identical layup, $(0/90/45/-45)_2S$, were used. This stacking sequence was the baseline multidirectional
configuration which was used in the study described previously. However, in this instance
the material had to be changed because of supply problems.

Materials

The materials used in this study were HTS/MTM44-1 and IMS/MTM44-1 unidirectional pre-
preg tapes with a nominal ply thickness of 0.250 mm (0.246 ± 0.011 mm and
0.249 ± 0.012 mm experimentally calculated – see Appendix D) and nominal fibre volume
fraction of approximately 60% (60.2 ± 0.7% and 59.5 ± 0.9% experimentally calculated – see
Appendix E), supplied by CYTEC[157]. However, the tow size was 12K for HTS/MTM44-1
and 24K for IMS/MTM44-1.

The mechanical properties of the unidirectional pre-preg laminae are given in
Table 3-1. The configurations used in this study were: monolithic HTS (HTS), hybrid
HTS/IMS with 0° and 90° HTS plies and ±45° IMS plies (HTS_IMS_A), hybrid HTS/IMS with
0° and 90° IMS plies and ±45° HTS plies (HTS_IMS_O) and finally monolithic IMS (IMS).
Note that in this study onwards, for the configurations made of pure HTS and IMS
respectively (i.e. the non-hybrid), the term monolithic will be used.
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<table>
<thead>
<tr>
<th>Property</th>
<th>HTS/MTM44-1</th>
<th>IMS/MTM44-1</th>
</tr>
</thead>
<tbody>
<tr>
<td><strong>Strength (MPa)</strong></td>
<td></td>
<td></td>
</tr>
<tr>
<td>Longitudinal Compression</td>
<td>-1330</td>
<td>-1459</td>
</tr>
<tr>
<td>Longitudinal Tension</td>
<td>2159</td>
<td>2738</td>
</tr>
<tr>
<td>In-plane Shear</td>
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<tr>
<td><strong>Modulus (GPa)</strong></td>
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<td></td>
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<tr>
<td>Longitudinal Compression</td>
<td>123.2</td>
<td>147.2</td>
</tr>
<tr>
<td>Longitudinal Tension</td>
<td>128.9</td>
<td>174.6</td>
</tr>
<tr>
<td>In-plane Shear</td>
<td>4.11</td>
<td>3.60</td>
</tr>
</tbody>
</table>

Table 3-1 Mechanical Properties of HTS/MTM44-1 and IMS/MTM44-1 laminae[157].

As will be presented in the following sections three different types of compression test have been employed to study the compressive performance of hybrid laminates; compact compression, plain compression and sandwich panel compression. The compact and plain compression tests were employed for coupon-sized specimens while the sandwich panel compression was used for element-sized components. Concerning the coupon-sized specimens, the two compression tests exhibited different fracture propagation behaviour. In particular, the fracture in compact compression was more progressive (i.e. stable) whereas in plain compression the fracture was dynamic.

Moreover, for these three specimen configurations, Digital Image Correlation was used to record surface deformations. In addition to this, these three different compression tests provided the opportunity to conduct a geometry sensitivity study. Finally, the sandwich panel compression test was utilized to study the performance of such hybrid laminates in larger size components and thus narrow the gap between coupons and structural components.
Manufacturing and Specimen Configuration

Following the manufacturing process discussed in section 3.2.2.1, four panels with dimensions 430 mm x 300 mm were manufactured. Each panel had identical layup \((0/90/45/-45)_{2S}\) but different material configurations, \(HTS\), \(HTS\_IMS\_A\), \(HTS\_IMS\_O\) and \(IMS\) respectively, as noted in Figure 3-1.

The specimen geometry used in this study was nominally identical to that in the previous section (Figure 3-1) and thus the same procedure was followed to fabricate the CC specimens. Five specimens were tested per configuration. Regarding the Digital Image Correlation, for the hybrid configuration testing GOM Aramis (v6.2) was utilized instead, capable of 0.04 pixel matching accuracy and 0.01% strain accuracy. The area covered by the DIC in this case was similar to that illustrated in Figure 3-3.

Plain compression (PC) specimens were manufactured and cut following the same procedure as that described for CC specimens. However, the specimen geometry in this instance (adopted from Greenhalgh et al.[158]) was rectangular with dimensions 132 mm x 50 mm and a 6 mm cutter was used for producing the notch (Figure 3-4a). For this case five specimens per configuration were tested. Similar to the CC case, GOM Aramis (v6.2) was employed on a specimen area which is illustrated in Figure 3-4b (Appendix C).

To obtain the hybrid skins for the sandwich panels, eight panels with dimensions 330 mm x 300 mm were manufactured. Although the same multidirectional stacking sequence was used, in this instance each panel comprised eight double thickness unidirectional plies, \((0/90/45/-45)_S\). The material configurations used for this study were identical to the configurations used for the compact and plain compression specimens, \(HTS\), \(HTS\_IMS\_A\), \(HTS\_IMS\_O\) and \(IMS\).

The panels were then cut to dimensions 270 mm x 200 mm for the sandwich panel skins and a side notch (24 mm diameter) was cut only in half skins as shown in Figure 3-5. Moreover, Hexcel 7.9-1/4-40(5052)T[159] aluminium honeycomb with 40 mm thickness and
Hexcel Redux 312[160] film adhesive with an areal weight of 0.73 kg/m² were used to form the sandwich panel. Once the honeycomb was bonded using the adhesive film onto the skins, the sandwich panels were then cast into aluminium end blocks using bendalloy 70. Bendalloy (also known as Wood’s metal or Lipowitz’s alloy[161]) is a eutectic and fusible alloy with a melting point of 70°C, made of 50% Bismuth (Bi), 26.7% Lead (Pb), 13.3% Tin (Sn) and 10% Cadmium (Cd). The reason for using this particular alloy to cast the sandwich panel into the end blocks was due to the lower cost compared to the normally used fibre-reinforced potting resin and its reusability once it is heated over 70°C. The aluminium end blocks were afterwards machined parallel to ensure uniform load distribution during testing (± 0.1 mm).

![Figure 3-4](image-url)

Figure 3-4 (a) Plain compression specimen configuration - all dimensions in mm (adopted from Greenhalgh et al.[158]), (b) area employed for DIC in a PC specimen.

Eventually, to facilitate Digital Image Correlation (GOM Aramis (v6.2)), the front skin was painted with a black speckle pattern on a white background. However, due to the larger dimensions of the panels compared to the compact and plain compression a coarser speckle pattern (20-35 μm) was used (in comparison to that in CC and PC specimens – 5-15 μm –
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see Appendix C). The area used for the Digital Image Correlation is highlighted in Figure 3-5 by the dotted red line.

![Figure 3-5 Geometry of sandwich panel with hybrid skins.](image)

**Compact Compression Testing**

For the compact compression testing of hybrid composites a 10-ton servo-hydraulic Instron machine equipped with a 10 kN load cell was employed. In the same manner, as for the compressive testing of multidirectional composites (Section 3.2.1.1), five specimens of each layup were tested. Linear Variable Differential Transducers (LVDT) were attached to the loading pins to record closing displacement during loading. Although the same DIC system was employed for this case (as in section 3.2.1.1), different cameras (Schneider Kreuznach Componon S-1.4/100 mm), illumination (Schneider Kreuznach LED lights – 40 Watts) and software version (GOM Aramis - v6.2) were utilised.
Plain Compression Testing

For the plain compression test a 50-ton Zwick 1488 machine with hydraulic grips equipped with a 200 kN load cell and a fixture shown in Figure 3-6 were utilised. This consisted of end clamps and an antibuckling guide (Figure 3-7). Upon attaching the specimen onto the clamps, extra care was taken to ensure an equal gap between each clamp and the antibuckling rig as well as that the specimen was positioned in the middle of the antibuckling rig (Figure 3-6).

![Figure 3-6 Plain Compression fixture[158].](image)

![Figure 3-7 Detail of the antibuckling rig geometry[158].](image)
For these tests the use of Linear Variable Differential Transducers was not possible due to the complex geometry of the testing rig. DIC was also performed on the plain compression test through the window in the antibuckling guide. Finally, all the specimens were loaded in displacement control at a rate of 2 mm/min, load-displacement data were recorded every second and DIC pictures were taken every three seconds. In this instance loading could not be halted prior to catastrophic failure of the specimens.

**Sandwich Panel Compression Testing**

Sandwich panel were tested in compression testing using a 250-ton servo-hydraulic compression machine. Per contra to compact and plain compression, strain gauges were utilized to record the deformation of the panel skins. Two strain gauges were attached at the front skin and two at the back skin in locally abraded sites. In this test uniaxial strain gauges (FLA-2.11) with 2 mm length, 120 ± 0.3 Ohm resistance and 2.11 ± 1% gauge factor supplied by Tokyo Sokki Kenkyuio were utilised.

In addition to the machine load-displacement and strain gauge readings, Digital Image Correlation was performed to measure the surface deformation during compressive loading using the cameras described earlier (Schneider Kreuznach Componon S-2.8/50 mm). In this instance, the cameras were placed approximately 1.5 m from the specimen surface and at 25° with respect to each other. Once the specimen was placed onto the 250-ton machine, a Linear Variable Differential Transducer was attached onto the lower loading plate to record the machine displacement.

For the recording of the machine load-displacement, the voltage from the strain gauges and the LVDT an Instron controller was used. The sandwich panels were tested until failure in displacement control at a rate of 0.5 mm/min while DIC images were taken every three seconds. Finally, a Phantom V12.1 high-speed camera equipped with a Zeiss ZF2/100 lens was used to enable crack speed measurement, at 60,000 fps with a resolution of 256 x 128 pixels.
3.2.2 Delamination Resistance of Hybrid Composites

The aim of this study was to assess the delamination resistance of hybrid composite interfaces and compare it with the delamination resistance of the constituent materials. Considering the absence of available delamination resistance values (peel and shear mode) in the open literature for the two systems used in this study, HTS/MTM44-1 and IMS/MTM44-1 (supplied by CYTEC), the characterization of the two systems was essential. It was also necessary to generate these data for the compressive failure study. For this study three mode mixities were employed; 100% Mode I, 75% Mode I and 0% Mode I.

3.2.2.1 Manufacturing

Three panels with dimensions 430 mm × 300 mm were manufactured according to the supplier’s recommendations. Each panel was composed of 16 unidirectional plies with a non-adhesive polytetrafluoroethylene (PTFE) 15 μm thick film, placed along one edge extending 50 mm into the laminate at the midplane. The panels were then cut using a dry saw to rectangular specimens with dimensions 150 mm × 20 mm, making sure the film insert was at least 50 mm from the free edge. The measured specimen thicknesses of the three configurations, monolithic HTS/MTM44-1 monolithic IMS/MTM44-1 and their hybrid, were 4.09 ± 0.12 mm, 4.14 ± 0.07 mm, 4.12 ± 0.17 mm respectively.

The same specimen configuration was used for all the specimens tested in Mode I, Mode II and Mixed Mode I/II. In the case of the hybrid specimens, the unidirectional layup utilised was (IMS/HTS/IMS/HTS/IMS/HTS) to ensure almost identical bending stiffness $D_{ij}$ for the upper and lower arm (0.85% difference – see Appendix F) as well as an overall coupling stiffness matrix $B_{ij}$ equal to zero. These requirements were essential in order to avoid stiffness coupling and dissimilar bending of the arms, which would lead to discrepancies in the results. In particular, arms with five, six and seven plies were investigated, however, only the (IMS/HTS/IMS/HTS/IMS/HTS) configuration
could meet both criteria. For the determination of the optimal layup with the required characteristics, as given by the Laminate Plate Theory[162], LAP software was employed.

3.2.2.2 Double Cantilever Beam (Mode I)

For the Mode I interlaminar fracture toughness ($G_{IC}$) tests the double cantilever beam test (DCB), the most common Mode I delamination resistance test, was employed (Figure 3-8) according to the D 5528 ASTM standard [163].

The DCB specimen, as shown in Figure 3-8, was a rectangular composite specimen with dimensions 150 mm x 20 mm ($L \times b$) and uniform thickness of 4 mm ($h$) which was composed of 0° unidirectional plies and contained a non-adhesive film insert at the midplane that acted as a delamination initiator ($a$), 50 mm from the free edge (Teflon film – Figure 3-8). The initial delamination $a_0$ was 40 mm (Figure 3-8). Loading blocks were bonded on both arms as shown in Figure 3-8. To apply the loading blocks, the end of the two arms where lightly grit blasted and then wiped clean with acetone to remove any contamination. Finally, after the surface preparation, a two-part epoxy (Araldite® 2012) was applied to bond the arms to the loading blocks and it was cured at room temperature for 24 hours.

![Figure 3-8 Double Cantilever Bending fixture[164].](image)

To facilitate visual detection of the delamination front, both sides of the specimens were coated with correction fluid. The first 5 mm ahead of the insert were then marked with vertical lines every 1 mm and the remaining 20 mm were marked with vertical lines every
5 mm. The specimens were afterwards mounted in a 1-ton 4504 Instron machine equipped with a 1 kN loading cell. To monitor the delamination front, a camera with a controlled crosshead mounted on a track was employed. This camera was capable of distinguishing the delamination front with an accuracy of ±0.05 mm.

Subsequently, the specimens were loaded in two phases. During initial loading at a displacement rate of 1 mm/min, load-displacement data were recorded and the delamination length was registered every 1 mm for the first 5 mm. As soon as the delamination front reached 5 mm the load was halted, the specimen was unloaded with a displacement rate of 5 mm/min and the length of the pre-crack was noted. Then the specimens were reloaded at a displacement rate of 1 mm/min until the final delamination length was reached or unstable crack growth had caused total separation of the two arms.

For the calculation of the Mode I interlaminar fracture toughness \([163]\), \(G_{IC}\), three data reduction methods were used, namely modified beam theory (MBT)\([165]\), compliance calibration method (CC)\([166]\) and modified compliance calibration method (MCC)\([167]\). The expressions for these three data reduction methods are:

\[
G_{IC} = \frac{3P\delta}{2b(a + \chi h)} \quad \text{(MBT)} \quad \text{Equation 3-1}
\]

\[
G_{IC} = \frac{nP\delta}{2ba} \quad \text{(CC)} \quad \text{Equation 3-2}
\]

\[
G_{IC} = \frac{3P^2C^{2/3}}{2A_{bh}} \quad \text{(MCC)} \quad \text{Equation 3-3}
\]

where \(P\) was the load, \(\delta\) was the load point displacement, \(b\) was the specimen width, \(a\) was the delamination length, \(\chi\delta\) (or \(\Delta\)) a factor which could be determined experimentally by generating a least squares plot of the cube root of compliance against delamination length, \(C\) was the measured compliance of the specimen and \(n\) was the slope of the plot of \(\log C\) versus \(\log a\). Note that the modified beam theory (MBT)\([165]\), also allowed for the
determination of the modulus of elasticity in the fibre direction measured in flexure from the following expression:

\[ E_y = \frac{64(a + |\Delta|)^3 P}{\delta bh^3} \]  

Equation 3-4

3.2.2.3 End Loaded Split (Mode II)

While for Mode I delamination fracture toughness characterisation the double cantilever test is universally accepted, in Mode II different tests have been used to measure the delamination fracture toughness. Some of those test methods are the End Notched Flexure (ENF)[168-170], the End Loaded Split (ELS)[168, 171, 172], the Centre Notched Flexure (CNF)[172] and the Cantilever Bend End Notched (CBEN)[172].

ENF and ELS are generally preferred, whilst CNF and CBEN are rarely used [88]. Although ENF and ELS yield similar results, in general ELS is more stable and generates less scatter[173]. For this reason, the ELS method was chosen for the measurement of Mode II delamination fracture toughness. However, there is still no ASTM standard for ELS Mode II testing.

The specimen used for the ELS test was nominally identical to the specimen used in the DCB test. Nevertheless, the length \( L \) shown in Figure 3-9 was approximately 95 mm and the film insert (50 mm). The initial delamination \( a_0 \) was 40 mm. The uncracked end of the specimen was restrained from vertical movement by rollers whilst the cracked end was loaded downward so stable growth could be achieved. Prior to loading, a loading block was applied on the upper arm (Figure 3-9). The loading block was bonded on the specimen in the same manner as for the DCB specimen. Finally, the process of data monitoring and recording was also similar as that in the previous section.
Several data reduction methods are available for measuring Mode II delamination fracture toughness such as modified beam theory (MBT), compliance calibration (CC) and shear deformable plate theory (SDPT). In this study the modified beam theory was employed because it accounts for the support flexibility correction at the clamped end. The expression for the delamination fracture toughness is the following[174]:

\[
G_{IIc} = \frac{9}{4} F \left( \frac{P^2 (a + \sqrt{a})^3}{b^3 h^3 E_{11}} \right)
\]

Equation 3-5

where \( E_{11} \) is the axial elastic modulus of the laminate and \( F \) is the correction factor which is related to the crack length shortening due to the large displacements involved in the specimen. Yet in the case where the arms are relatively stiff, as in UD specimens, this factor approximates unity[165].

3.2.2.4 Mixed-mode Bending (Mode I/II)

For the Mixed Mode I/II delamination fracture toughness \( (G_T) \) measurement of hybrid composites, the mixed-mode bending test (MMB) was employed (Figure 3-10) according to the D 6671 ASTM standard[175]. The use of this test allowed the determination of the delamination fracture toughness at various mode mixities (in this instance 75% Mode I).
In this test the load was applied via end blocks which were placed in the delaminated section and through rollers that bore against the specimen at the non-delaminated section. This combined load was achieved by the use of a lever. While the lever was applied vertically halfway between the base roller and the end blocks (introducing Mode II), the specimen was held stationary by the base of the apparatus at the lower end block and at the other end of the specimen with a roller. Hence as the lever was pushing down the specimen, the upper arm was pulled up via the end block (introducing Mode I). The length of the lever, \( c \), is very important for the MMB test. In particular, if the length of the lever is varied the ratio of the load pulling on the end block to the load bearing through the roller is varied as well. Effectively the mode mixity of the test is varied. Thus by varying the lever length appropriately the required mode mixity can be achieved. In this study the chosen mode mixity was 75% Mode I with a lever length of 78.4 mm.

The specimens had nominally identical dimensions to that of DCB and the initial delamination length was 40mm. The loading blocks were applied onto the upper and lower arms and the horizontal sides were coated with the same procedure as described in Section 3.2.3.3. For this study a calibrated 10-ton 4504 Intron equipped with a 10 kN load cell was used. Finally, the same procedure for loading, data monitoring and recording described for DCB testing was followed in this instance as well.
For the calculation of the total mixed mode delamination fracture toughness the values of the fracture toughness in the two modes were calculated, bearing in mind the corrections mentioned in the two previous sections and the MMB apparatus configuration. The expressions for the delamination fracture toughness in the two modes[176-178] are:

\[
G_I = \frac{3P^2(a + \chi_I h)}{b^2 E_1 h^3}\left[1 - \frac{c + b}{L}\right] - \frac{c}{b}^2
\]

Equation 3-6

\[
G_{II} = \frac{9P^2(a + \chi_{II} h)}{4b^2 E_1 h^3}\left[1 - \frac{c + b}{L}\right] + \frac{c}{b}^2
\]

Equation 3-7

where \(\chi_{II}\) is the correction factor for Mode II delamination fracture toughness and is considered equal to 0.42\(\chi_I\) [177] and the total delamination fracture toughness is:

\[
G_T = G_I + G_{II}
\]

Equation 3-8

### 3.3 Fractographic Analysis

This section provides an overview of the techniques used for fractographic analysis during this project.

#### 3.3.1 Optical Microscopy

Optical microscopy [1,173,179-181] is a valuable tool for failure analysis and quality control of composites. Optical microscopy was employed for fractographic analysis for the compact and plain compression specimens as well as for the examination of the interface between HTS and IMS plies at the delamination fracture toughness specimens.

#### 3.3.2.2 Specimen Preparation

For the compact compression specimens, rectangular sections of 45 mm x 20 mm were carefully cut as shown in Figure 3-11a and Figure 3-11b. In this instance the one edge of the section coincided with the notch and optical inspection was conducted on two sections.
(at the notch and 15.5 mm away from the notch; at the midpoint between the notch and the free edge (Figure 3-11c).

Figure 3-11 Optical microscopy specimen preparation.

The rectangular section was mounted in potting resin (Araldite® LY 5052 (resin) + Araldur® 5052 (hardener) with a 100:38 mixing ratio), then ground and polished to achieve a smooth polished surface (up to 1 μm grit), using a Struers Sapphire 350E polisher/grinder. For the plain compression specimens, the same procedure was followed to obtain polished specimens. Nevertheless, since the specimen was larger, the rectangular section had to be cut in half and then put in separate pots. For the delamination fracture toughness specimens, cross-sections of the three different configurations perpendicular to the fibre direction were obtained to compare the hybrid ply interfaces (HTS_IMS) with the interfaces of the monolithic configurations (HTS and IMS). To examine the polished surfaces an Olympus BHM incident optical microscope was employed, equipped with an NK eyepiece (2×), an SPLAN objective lens (5×/0.13) and a Q-Imaging MicroPublisher 5.0RTV camera.

3.3.2 Scanning Electron Microscopy

Scanning Electron Microscopy[1,182,183] (SEM) is a widely used method to examine and characterize fracture surfaces due to the high magnification and depth of field.
3.3.1.1 Specimen Preparation

Selected specimens (Figure 3-12a) were dissected as shown in Figure 3-12b, using a dry saw to produce a rectangular section 45 mm × 20 mm enclosing the damage. Since in some specimens the fracture did not propagate all the way to the free edge, mechanical load had to be applied to separate the section into two halves (Figure 3-12c). This may have induced some additional damage to the fractured surfaces.

![Figure 3-12 SEM specimen preparation.](image)

For the fractographic analysis two representative specimens from each configuration were chosen. The dissected components of each specimen were bonded on 25 mm diameter aluminium stubs (Figure 3-12d) using a two-part epoxy (Araldite® 2012). To avoid dust and loose debris obscuring the fracture surfaces, compressed air was blown onto the fracture surfaces. Subsequently, the surfaces were sputter-coated with gold and marked with silver dag to ensure electrical conductivity. Electron microscopy analysis was then conducted using a Hitachi S-3400N microscope at an acceleration voltage of 15 kV and magnifications up to 10,000. It should be noted that for the plain compression specimens a similar procedure was followed. However, for the DCB, ELS and MMB specimen the
sections mounted onto the stubs were different. Due to the specimen configuration and the nature of the testing, rectangular sections from selected specimens were obtained. In particular, nominally identical sections from the upper and lower arm of each specimen, 40 mm × 20 mm, were mounted onto aluminium stubs next to each other to ease the comparison. These rectangular sections included 5 mm from the area where the Teflon film insert has been placed.

3.3.3 X-ray radiography
In this study, X-ray radiography was utilised for fractographic analysis because it was best suited for looking at localised translaminar damage, intralaminar damage especially at the notch as well as the extent of the interlaminar damage.

3.3.3.1 Specimen Preparation
Throughout this study, dibromomethane (C₂H₄Br₂) was used as penetrant, supplied by Sigma-Aldrich Chemistry. This was because an organic solvent could not leave a residue on the fractured surfaces[181]. Albeit general instructions for X-Ray specimen preparation have been reported in the literature[184], these instructions may not apply to different composite geometries and thicknesses. For this reason, a parametric study was conducted. After a series of tests on relatively small specimens of 4 mm thickness an optimum combination of soaking and drying time was obtained. It was found that a soaking time of 2 to 5 minutes was adequate for the penetrant to reach the full extent of the damaged area and a drying time of 15 to 20 minutes produced a satisfactory contrast between the pristine composite and the damage.

Once the specimen was dry, it was placed inside the X-Ray cabinet on top of a radiation sensitive plate. The X-ray system was a 43804N Faxitron Series supplied by Hewlett-Packard (120 kV/4 mA). The tube voltage for the composite specimens was 25 kV, the tube current was 4mA and the exposure time was 120 seconds. After the scanning procedure the image plate was taken to an HD-CR 35 NDT X-ray scanner, supplied by
DÜRR[185], to produce the digital image according to the supplier’s instruction. It should be noted that for the X-Ray radiographic inspection of the compact and plain compression the entire specimens were scanned while for the sandwich panels only a section of approximately 100 mm × 50 mm of the entire front skin was scanned (in the vicinity of the notch).

3.4 Theoretical Analysis

To assess the elastic behaviour of the laminate and provide an estimation of the stress distribution in the laminate, the Classical Laminate Theory (CLT) was utilized[162, 186, 187]. In particular, LAP code (supplied by Anaglyph Ltd.) was used to obtain the stress distribution across the laminate width (including residual stresses) of the various configurations which have been tested throughout this work[188] as well as the compliances, although for the latter no loading or specimen configuration is taken in account. LAP was employed for both monolithic (Chapter 4) and hybrid multidirectional laminates (Chapter 5) and two cases were studied, pure compression (-0.1% strain) and combined shear and compression (0.1% strain each). The latter was conducted to assess the effect of the combined load on the stress distribution since it is thought that the compact compression specimen experiences a combined shear and compression load due to its complex geometry (Figure 3-1).

The stress distribution both under pure compression and combined compression-shear load was determined, as well as the unnotched compressive strength[81, 188, 189] and the Budiansky-Fleck-Soutis notched compressive strength (BFS)[83,188]. Finally, the stress of the 0° load-bearing plies at failure for the three compression tests were also obtained by LAP and compared to investigate how hybridisation influenced the behaviour of the 0° load-bearing plies.

Albeit the determination of the axial mechanical and residual stresses was conducted using the LAP code, the interlaminar stresses could not be determined since LAP is based on the Classical Laminate Theory (CLT) which does not take in account interlaminar
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stresses. To overcome this, ABAQUS was employed to obtain the interlaminar stresses using the cohesive zone elements (see next section). To achieve this, a path was created across the thickness of the CC specimen 2 elements away from the notch tip to alleviate any effect from the notch. In the cohesive zone elements, the $S_{33}$ stress component corresponded to the Mode I interlaminar stresses, while the Mode II interlaminar stresses were given by $\sqrt{(S_{13})^2 + (S_{23})^2}$ as suggested by Pinho et al.[190].

3.5 Numerical Analysis

The classical analytical expressions are not able to accurately predict the performance of complex geometries such as laminates which contain notches or holes, particularly when non-linear material behaviour is considered. This has led to the use of the Finite Element Method (FEM).

3.5.1 Compressive Failure of Multidirectional Composites

Finite Element Analysis (FEA) was employed to predict the compressive performance of multidirectional composite CC specimens and to identify the dominant failure modes at initiation. Only in-plane damage (intralaminar and translaminar) was initially considered by modelling a laminate with shell elements and a ply-level continuum damage material model. Delamination does, however, significantly affect compressive failure and is important to capture in a model. Cohesive elements were then added to the model such that every ply interface could potentially delaminate while in-plane damage was still considered as before. Each ply was then represented by a shell element.

The generic Hashin based damage initiation criteria were tried as well as a user-defined model which is described in more detail in Section 3.4.1.2. In the following sections the details of the finite element models which were built to predict the compressive performance of multidirectional composites are provided, whereas the results are presented in Chapter 4 and Chapter 5.
3.5.1.1 Numerical Model Details

For the numerical analysis of the compressive failure of multidirectional composites (CC), Simulia ABAQUS/Explicit v6.10[191] dynamic explicit commercial solver was used in preference of the static ABAQUS/Implicit to better cope with the severe material non-linearity occurring during compressive failure as well as the crushing of the plies following that. While a static implicit solver is more efficient for mildly non-linear problems, it was found necessary to use the dynamic explicit solver for modelling more realistically progressive compressive failure in multidirectional composite laminates. To do so, a half thickness model of the compact compression specimen was built for each different layup, (0/90)_8S, (90/0)_8S, (0/90/45/-45)_4S and (-45/45/0/90)_4S.

Elements, Boundary Conditions and Outputs

The initial model comprised 2590 linear hexahedral continuum shell elements of type SC8R with 11526 degrees of freedom. This type of three-dimensional elements accounted for finite membrane strains, large rotations and allowed the shell thickness to change with element deformation. Load was introduced to the model laminate via rigid pins to avoid premature damage (Figure 3-13a) and at each pin an acceleration (~2 m/s ^2) was applied.

![Figure 3-13 (a) CC specimen finite element model; (b) notch detail.](image)

Regarding the boundary conditions, horizontal rigid body movement in the plane was prevented by setting ~A_1 = 0 on the pins whilst one laminate surface was constrained against...
out-of-plane displacement to represent symmetry plane (\( A_3 = 0 \)), since only half the actual thickness was modelled. For every model, field and history outputs were recorded. The field output variables were: fibre compressive damage (DAMAGEFC), fibre tensile damage (DAMAGEFT), tensile matrix damage (DAMAGEMT), compressive matrix damage (DAMAGEMC), shear damage (DAMAGESHR), logarithmic strain components (LE), stress components and invariants (S), scalar stiffness degradation (SDEG), and the history outputs were reaction force from the pin (RF) and displacement (U)[191].

**Material Data Input**

The CC specimen contained thirty-two IM7/8552 plies of thickness 0.122 mm and fibre volume fraction of 59.68\% (experimentally determined – see Appendix A and Appendix B), of which sixteen were modelled. The mechanical properties of the lamina and the fracture initiation energies (Hashin criterion[53,54]) used for this model are given in Table 3-2 and Table 3-3. As mentioned earlier in this section, Hexcel IM7/8552 is a widely used carbon fibre/epoxy material system and numerous studies in the literature and technical reports have been devoted to obtaining the mechanical properties of this system [52,150, 151,192-197]. Nevertheless, considering the values suggested in the literature there is a wide range and only limited studies provide the full set of properties required for the finite element analysis (Table 3-2).

To try to characterise and obtain all the mechanical properties needed for the numerical analysis for this material system would require an extensive and time consuming testing scheme and this was outside the scope of this work since the mere comparison between different layups does not necessitate highly accurate values. According to the author’s knowledge, in the literature there is only one source which can provide the full set of properties for the IM7/8552 with high confidence and credibility. This is the WWFE II in which the mechanical properties of this system were obtained using a round robin test scheme. Therefore the set of values provided by the WWFE II was used for the numerical analysis of the compact compression of multidirectional IM7/8552 specimens. Note that for the fibre
volume fraction, across the literature 60% fibre volume fraction and 0.125 mm lamina thickness are assumed, however, in this numerical analysis the values obtained experimentally in this work was used instead (i.e. 0.123 mm and 59.7% respectively – see Appendix A and Appendix B).

<table>
<thead>
<tr>
<th>Property</th>
<th>Hexcel</th>
<th>EDAVCOS</th>
<th>WWFE II</th>
<th>Literature</th>
</tr>
</thead>
<tbody>
<tr>
<td>$E_1$</td>
<td>164 GPa</td>
<td>165 GPa</td>
<td>165 GPa</td>
<td>150–171.42 GPa</td>
</tr>
<tr>
<td>$E_2$</td>
<td>11.38 GPa</td>
<td>9.4 GPa</td>
<td>8.4 GPa</td>
<td>9.08–11 GPa</td>
</tr>
<tr>
<td>$E_3$</td>
<td>11.38 GPa</td>
<td>10.5 GPa</td>
<td>10.5 GPa</td>
<td>9.08 GPa</td>
</tr>
<tr>
<td>$G_{12}$</td>
<td>5.17 GPa</td>
<td>4.5 GPa</td>
<td>5.6 GPa</td>
<td>4.6–5.9 GPa</td>
</tr>
<tr>
<td>$G_{13}$</td>
<td>5.17 GPa</td>
<td>4.29 GPa</td>
<td>5.6 GPa</td>
<td>5.29 GPa</td>
</tr>
<tr>
<td>$G_{23}$</td>
<td>3.9 GPa</td>
<td>3.19 GPa</td>
<td>2.8 GPa</td>
<td>3.973 GPa</td>
</tr>
<tr>
<td>$X^T$</td>
<td>2724 MPa</td>
<td>2600 MPa</td>
<td>2560 MPa</td>
<td>1531–1690 MPa</td>
</tr>
<tr>
<td>$X^C$</td>
<td>-1690 MPa</td>
<td>-1500 MPa</td>
<td>-1590 MPa</td>
<td>-(1415–1690) MPa</td>
</tr>
<tr>
<td>$Y^T$</td>
<td>110 MPa</td>
<td>60 MPa</td>
<td>73 MPa</td>
<td>100 MPa</td>
</tr>
<tr>
<td>$Y^C$</td>
<td>-250 MPa</td>
<td>-290 MPa</td>
<td>-185 MPa</td>
<td>-(105–120) MPa</td>
</tr>
<tr>
<td>$S^L$</td>
<td>120 MPa</td>
<td>90 MPa</td>
<td>90 MPa</td>
<td>59.2–120 MPa</td>
</tr>
<tr>
<td>$S^T$</td>
<td>50 MPa</td>
<td>-</td>
<td>69.7 MPa</td>
<td>-</td>
</tr>
<tr>
<td>$V_{12}$</td>
<td>0.32</td>
<td>0.30</td>
<td>0.34</td>
<td>(0.3–0.362)</td>
</tr>
<tr>
<td>$V_{13}$</td>
<td>0.32</td>
<td>0.31</td>
<td>0.34</td>
<td>(0.32)</td>
</tr>
<tr>
<td>$V_{23}$</td>
<td>0.44</td>
<td>0.487</td>
<td>0.5</td>
<td>(0.5)</td>
</tr>
</tbody>
</table>

Table 3-2 Suggested IM7/8552 lamina properties reported in the literature[52,63,150,151,192-196].

Longitudinal critical fracture energy values were taken from Pinho et al.[152]. These values were measured for T300/913 and according to the author’s knowledge this study is the only study to date which reports critical energy release rate for fibre fracture in the 0°
layers decoupled from the matrix crack propagation in the 90° layers as previously reported in the literature using cross-ply laminates[80,83,198-201]. Although T300/913 is a system with different properties than IM7/8552, assuming that the value obtained by Pinho et al. could be applicable in this case too, was thought to be more well-founded than guessing a value for the IM7/8552. Transverse critical fracture energies were estimated with the tensile value based on delamination toughness and Bazant’s crack band saturation[202]. The energy associated with transverse crushing has purely been estimated. More rigorous methods for determining these values do not exist at the moment.

<table>
<thead>
<tr>
<th>Fracture Energies (kJ/m²)</th>
</tr>
</thead>
<tbody>
<tr>
<td>Longitudinal tensile</td>
</tr>
<tr>
<td>Longitudinal compressive</td>
</tr>
<tr>
<td>Transverse tensile</td>
</tr>
<tr>
<td>Transverse compressive</td>
</tr>
</tbody>
</table>

<table>
<thead>
<tr>
<th>Elastic (GPa)</th>
</tr>
</thead>
<tbody>
<tr>
<td>$E_{nn}$</td>
</tr>
<tr>
<td>$E_{ss}$</td>
</tr>
<tr>
<td>$E_{tt}$</td>
</tr>
</tbody>
</table>

<table>
<thead>
<tr>
<th>Quads Damage (MPa)</th>
</tr>
</thead>
<tbody>
<tr>
<td>$t_{nn}$</td>
</tr>
<tr>
<td>$t_{ss}$</td>
</tr>
<tr>
<td>$t_{tt}$</td>
</tr>
</tbody>
</table>

<table>
<thead>
<tr>
<th>Damage Evolution (J/m²)</th>
</tr>
</thead>
<tbody>
<tr>
<td>$G_{nn}$</td>
</tr>
<tr>
<td>$G_{ss}$</td>
</tr>
<tr>
<td>$G_{tt}$</td>
</tr>
</tbody>
</table>

Table 3-3 Fracture energy and cohesive zone properties for IM7/8552.

Mesh Refinement and Cohesive Elements
Subsequently, a series of improvements were made to the initial model described previously. Firstly, the model was re-meshed and the mesh in the vicinity of the notch was refined to achieve better accuracy for the prediction of fracture propagation (Figure 3-15). Secondly,
the interfaces between the plies were modelled using cohesive elements to model interlaminar failure. In this instance 64914 elements with 211212 degrees of freedom were used for the entire model. Each ply was modelled by a layer of SC8R continuum shell elements with cohesive elements (COH3D8) inserted at all ply interfaces[191]. The mechanical properties were identical to those shown in Table 3-2. The properties of the cohesive elements are shown in Table 3-3.

![Figure 3-14](image)

**Figure 3-14** (a) Modified CC specimen model; (b) notch detail.

**Mesh Sensitivity**

Finally, to check for mesh convergence two models were built in addition to the model described above. The first model had a coarser mesh especially around the notch and for that purpose 41044 elements (21184 SC8R/19860 COH3D8) with 136332 degrees of freedom were used. The finer model had a finer mesh around the notch and 158844 elements (81984 SC8R/76860 COH3D8) with 503628 degrees of freedom were used. The meshes of the three models are shown in Figure 3-15.

![Figure 3-15](image)

**Figure 3-15** (a) Coarse mesh; (b) Intermediate mesh; (c) Fine mesh.
Note that the models depicted in Figure 3-15b and Figure 3-15c were run using High Performance Computing (HPC) due to the high computational power required.

3.5.1.3 Failure Criteria
The only composites failure criterion which is implemented in the standard ABAQUS code to date (v6.11) is the Hashin failure criterion[53]. According to this criterion, failure can be initiated due to fracture in the matrix or the fibres. In particular, four different failure mechanisms which can initiate failure are suggested: matrix tension, matrix compression, fibre tension and fibre compression, while in-plane shear damage is taken as the maximum of the four direct damage components. More information about this failure criterion is given in Appendix A.

2dVUMAT
As has been suggested in the literature, Hashin’s failure criterion cannot always consistently predict the failure of fibre-reinforced composite laminates, especially in compression[36]. ABAQUS/Explicit has an interface which allows the user to implement general constitutive equations as a user subroutine VUMAT[191]. Via this interface constitutive models of arbitrary complexity can be defined for any ABAQUS/Explicit element type. Hence failure criteria more complex and more accurate than Hashin’s can be implemented in the code to potentially provide more realistic prediction of the failure process.

In this project, a failure model was utilised with the ABAQUS/Explicit code, the 2dVUMAT[203] which was built by Professor Lorenzo Iannucci and Dr. Jesper Ankersen at the Aeronautics Department, Imperial College London with further details given in Appendix B. Similar to the ABAQUS Hashin implementation, the 2dVUMAT code is for modelling damage and failure in unidirectional composites where through-thickness effects are neglected. In contrast to Hashin’s failure criterion[53], 2dVUMAT incorporates a non-linear softening shear response with rate dependence. For fibre dominated failure the maximum stress criterion is used, whilst for matrix dominated failure the Mohr-Coulomb[49] formulation is employed. Finally, in this code there is an output variable, namely SDV16,
which corresponds to the $l_x/l_{x,max}$ ratio, where $l_x$ is the characteristic element length which, in ABAQUS, is the square root of the element area. The maximum element size is given by material properties as per Equation 3-8. In the 2dVUMAT code, this length is related to the strain at which the material has failed by Equation 3-9:

$$l_{x,max} = \frac{2G_C}{\sigma_C^2}$$  \hspace{1cm} \text{Equation 3-9}$$

$$\varepsilon_{max} = \frac{2G_C}{\sigma_C l_x}$$  \hspace{1cm} \text{Equation 3-10}$$

The 2dVUMAT code checks whether the chosen element size can ensure the correct $G_C$ dissipation. By plotting the SDV16 it can be assessed whether the element size is adequate and thus the correct $G_C$ is dissipated. If the ratio is less than one a correct mesh has been chosen. In the case where the mesh is too coarse, 2dVUMAT sets $\varepsilon_{max} = 1.05\varepsilon_0$ to prevent snap back in the material model[203].

### 3.5.1.4 Cohesive Zone

Interlaminar failure can lead to great loss of laminate stiffness or even premature failure. Hence to realistically model interlaminar fracture the interface between the plies has to be modelled. For this purpose specially formulated elements which used a cohesive zone framework (cohesive elements) were employed to model the ply interfaces. In this cohesive zone framework, fracture is regarded as a gradual process in which debonding takes place across a cohesive zone and is resisted by cohesive tractions. Therefore this cohesive zone (cohesive elements) represents the cohesive forces which occur due to debonding rather than a physical material. When such damage occurs at the interface between two plies, or in general at an interface, the cohesive elements open to simulate damage initiation or growth. As the crack propagates, the crack path follows these cohesive elements and thus the propagation of an interfacial crack depends on the presence of cohesive elements. To
describe the behaviour of the interfacial damage, traction-separation laws are employed[204]. Essentially, as the cohesive elements open and fracture propagates, the traction initially increases until a maximum value (initiation) and then it decreases to zero which leads to a total local separation (Figure 3-16).

![Figure 3-16 Typical traction-separation response[191].](image)

In the dynamic explicit solver (ABAQUS/Explicit), a stable time increment for time marching the solution is determined by the dilatational wave speed across the smallest element in the model. The cohesive elements must have a small and finite thickness which often sets the time increment size. Hence, the addition of cohesive elements to a dynamic explicit model often increases the analysis cost considerably. Further details about the cohesive zone elements are provided in Appendix I.

3.5.2 Compressive Failure of Hybrid Composites

3.5.2.1 Numerical Model Details

Analysis of hybrid laminates was performed with a model similar to that described in section 3.4.1 (Figure 3-14). However since the plies used in the testing had double thickness, the model was composed of 8 plies again modelling effectively half of the specimen thickness. The hybrid composite CC model comprised 64914 elements with 211212 degrees and had one continuum shell element (SC8R) per ply with cohesive elements (COH3D8) at all ply interfaces. Note that for the two material systems used in the hybridisation study (HTS/MTM44-1 and IMS/MTM44-1), according to the author’s knowledge there are no studies in the literature reporting laminate mechanical properties. To characterise these
Compressive Failure of Multidirectional Composites Study

systems in order to obtain the full set of properties (needed in ABAQUS), would require an extensive testing scheme which would be beyond the scope of this work. In addition, taking into account the comparative nature of this study absolute values are considered not to be crucial. Therefore, the laminate properties of the two materials used in this model (shown in Table 3-4), are those reported by CYTEC[157].

<table>
<thead>
<tr>
<th>Property</th>
<th>HTS/MTM44-1</th>
<th>IMS/MTM44-1</th>
</tr>
</thead>
<tbody>
<tr>
<td><strong>Elastic Modulus (GPa)</strong></td>
<td></td>
<td></td>
</tr>
<tr>
<td>$E_1$</td>
<td>123.2</td>
<td>174.2</td>
</tr>
<tr>
<td>$E_2$</td>
<td>88</td>
<td>147</td>
</tr>
<tr>
<td>$E_3$</td>
<td>88</td>
<td>147</td>
</tr>
<tr>
<td>$G_{12}$</td>
<td>4.1</td>
<td>3.6</td>
</tr>
<tr>
<td>$G_{13}$</td>
<td>3.9</td>
<td>3</td>
</tr>
<tr>
<td>$G_{23}$</td>
<td>3.9</td>
<td>3</td>
</tr>
<tr>
<td><strong>Poisson's ratio</strong></td>
<td></td>
<td></td>
</tr>
<tr>
<td>$v_{12}$</td>
<td>0.3</td>
<td>0.3</td>
</tr>
<tr>
<td>$v_{13}$</td>
<td>0.2</td>
<td>0.2</td>
</tr>
<tr>
<td>$v_{23}$</td>
<td>0.2</td>
<td>0.2</td>
</tr>
<tr>
<td><strong>Strength (MPa)</strong></td>
<td></td>
<td></td>
</tr>
<tr>
<td>$X^T$</td>
<td>2159</td>
<td>2738</td>
</tr>
<tr>
<td>$X^C$</td>
<td>-1330</td>
<td>-1459</td>
</tr>
<tr>
<td>$Y^T$</td>
<td>66</td>
<td>111</td>
</tr>
<tr>
<td>$Y^C$</td>
<td>-230</td>
<td>-250</td>
</tr>
<tr>
<td>$S^L$</td>
<td>105</td>
<td>76</td>
</tr>
<tr>
<td>$S^T$</td>
<td>105</td>
<td>50</td>
</tr>
</tbody>
</table>

Table 3-4 Lamina Mechanical properties of HTS/MTM44-1 and IMS/MTM44-1 used for hybrid composites modelling[157].

The layup of half the laminate was (0/90/45/-45)$_S$ with lamina thickness and fibre volume fraction of 0.246 mm and 60.23% respectively for the HTS/MTM44-1 and 0.249 mm and
59.51% for the IMS/MTM44-1 (experimentally determined – see Appendix D and Appendix E). Regarding the delamination fracture toughness values for the two systems, no values were available in the literature and those reported by the supplier had been obtained by vacuum-only consolidation and not by autoclave[203]. Hence, for the numerical analysis the values which were obtained by testing in Mode I, Mode II and Mixed-Mode I/II (Chapter 6) were fed into the models (Table 3-5). Regarding the fracture energies, the application of the values tabulated below is in accordance to the reasoning described in Section 3.5.1.1.

<table>
<thead>
<tr>
<th>Property</th>
<th>HTS/MTM44-1</th>
<th>IMS/MTM44-1</th>
</tr>
</thead>
<tbody>
<tr>
<td>Fracture Energies (kJ/m²)</td>
<td></td>
<td></td>
</tr>
<tr>
<td>Longitudinal tensile</td>
<td>90</td>
<td>90</td>
</tr>
<tr>
<td>Longitudinal compressive</td>
<td>80</td>
<td>80</td>
</tr>
<tr>
<td>Transverse tensile</td>
<td>1</td>
<td>1</td>
</tr>
<tr>
<td>Transverse compressive</td>
<td>10</td>
<td>10</td>
</tr>
<tr>
<td>Elastic (GPa)</td>
<td></td>
<td></td>
</tr>
<tr>
<td>$E_{nn}$</td>
<td>9</td>
<td>9</td>
</tr>
<tr>
<td>$E_{ss}$</td>
<td>5</td>
<td>5</td>
</tr>
<tr>
<td>$E_{tt}$</td>
<td>5</td>
<td>5</td>
</tr>
<tr>
<td>Damage Evolution (kJ/m²)</td>
<td></td>
<td></td>
</tr>
<tr>
<td>$G_n$</td>
<td>0.36</td>
<td>0.38</td>
</tr>
<tr>
<td>$G_s$</td>
<td>0.93</td>
<td>1.14</td>
</tr>
<tr>
<td>$G_t$</td>
<td>0.93</td>
<td>1.14</td>
</tr>
<tr>
<td>Quads Damage (MPa)</td>
<td></td>
<td></td>
</tr>
<tr>
<td>$t_n$</td>
<td>60</td>
<td>60</td>
</tr>
<tr>
<td>$t_s$</td>
<td>80</td>
<td>80</td>
</tr>
<tr>
<td>$t_t$</td>
<td>80</td>
<td>80</td>
</tr>
</tbody>
</table>

Table 3-5 Properties used for the cohesive zone of hybrid composites modelling.
Chapter 4 – Compressive Failure of Multidirectional Composites Study

4.1 Chapter Introduction
In this chapter the results from the study on compressive failure of multidirectional composite laminates are presented; these include the outcomes from testing, fractographic evaluation, theoretical and numerical analysis.

4.2 Compressive Failure of Multidirectional Composites

4.2.1 Mechanical Testing
The results of the compact compression tests are summarized in Table 4-1 based on five specimens per layup, while typical force-displacement curves of the four configurations are shown in Figure 4-1. Moreover, the experimentally obtained in-situ ply thickness for the various layups is also given in Table 4-1 (Appendix A).

<table>
<thead>
<tr>
<th>Layup</th>
<th>Peak Load (kN)</th>
<th>Compliance (mm kN⁻¹)</th>
<th>In-situ Lamina Thickness (mm)</th>
<th>Measured Specimen Thickness (mm)</th>
</tr>
</thead>
<tbody>
<tr>
<td>(90/0)₈s</td>
<td>-5.02 ± 0.09</td>
<td>0.021 ± 0.005</td>
<td>0.121 ± 0.09</td>
<td>3.96 ± 0.05</td>
</tr>
<tr>
<td>(0/90)₈s</td>
<td>-4.73 ± 0.07</td>
<td>0.019 ± 0.001</td>
<td>0.124 ± 0.07</td>
<td>3.98 ± 0.04</td>
</tr>
<tr>
<td>(0/90/45/-45)₄s</td>
<td>-6.11 ± 0.05</td>
<td>0.017 ± 0.001</td>
<td>0.123 ± 0.09</td>
<td>4.04 ± 0.02</td>
</tr>
<tr>
<td>(-45/45/0/90)₄s</td>
<td>-5.99 ± 0.05</td>
<td>0.016 ± 0.003</td>
<td>0.123 ± 0.05</td>
<td>4.01 ± 0.04</td>
</tr>
</tbody>
</table>

Table 4-1 Compact Compression test results.

As Figure 4-1 illustrates, the cross-ply and multidirectional configurations behaved in a different manner during compression loading. In particular, the multidirectional configurations ((0/90/45/-45)₄s and (-45/45/0/90)₄s) exhibited a stiffer elastic response than the cross-ply configurations ((90/0)₈s and (0/90)₈s – contrary to the CLT prediction (Table 4-2)), although for a given configuration (cross-ply or multidirectional) the elastic response was almost independent of the layup. The compliance and the fracture initiation
were similar between the two cross-ply configurations and between the two multidirectional configurations, albeit the damage propagation following failure initiation differed significantly.

![Figure 4-1 Representative compressive testing results of CC configurations.](image)

<table>
<thead>
<tr>
<th>Layup</th>
<th>Compliance (mm kN⁻¹)</th>
<th>Theoretical Compliance (mm kN⁻¹)</th>
<th>Theoretical Shear Modulus (GPa)</th>
</tr>
</thead>
<tbody>
<tr>
<td>(90/0)₈s</td>
<td>0.021 ± 0.005</td>
<td>0.013</td>
<td>5.6</td>
</tr>
<tr>
<td>(0/90)₈s</td>
<td>0.019 ± 0.001</td>
<td>0.013</td>
<td>5.6</td>
</tr>
<tr>
<td>(0/90/45/-45)₄s</td>
<td>0.017 ± 0.001</td>
<td>0.018</td>
<td>23.9</td>
</tr>
<tr>
<td>(-45/45/0/90)₄s</td>
<td>0.016 ± 0.003</td>
<td>0.018</td>
<td>23.9</td>
</tr>
</tbody>
</table>

Table 4-2 Comparison between experimental and theoretical compliances for the various configurations (theoretical compliances don’t account for loading).

Such an improvement in the compressive performance should be mainly attributed to the incorporation of the angle plies, which essentially enhanced the effective shear stiffness
Compressive Failure of Multidirectional Composites Study

(Table 4-2). Considering the CC specimen geometry (Figure 3-1) such an improvement in the compressive performance should be expected since the load is applied via the pins and therefore a moment is acting at the crack tip.

In addition, in the two multidirectional configurations the initial load drop was more gradual compared to that in the crossply configurations (Figure 4-1). Upon failure initiation, the damage propagation in the (0/90)$_{8S}$ and (0/90/45/-45)$_{4S}$ configurations was more progressive in comparison to the (90/0)$_{8S}$ and (-45/45/0/90)$_{4S}$ configurations. The difference in the compressive performance between the four different configurations indicated that the layup had a significant effect at the failure process.

4.2.1.1 Nominally identical (90/0)$_{8S}$ baseline specimens

Nominally identical specimens were also compared to investigate the inherent variability in the compressive performance. The behaviour of nominally identical specimens was quite similar with similar elastic regions as the small scatter indicates (Figure 4-2), whilst the large drop in the load after initial failure was observed in both nominally identical specimens.

![Figure 4-2 Representative load-displacement curves for nominally identical (90/0)$_{8S}$ cross-ply configurations.](image)
4.2.1.2 Cross-ply specimens – (90/0)$_{8S}$ and (0/90)$_{8S}$

The initial elastic response of the two cross-ply configurations was similar, as shown in Figure 4-1. However, the baseline (90/0)$_{8S}$ cross-ply configuration was more compliant and approximately 7% stronger than the (0/90)$_{8S}$ cross-ply configuration. Once the failure load had been reached, the response of the two cross-ply configurations differed. In particular, the fracture propagation at the (90/0)$_{8S}$ configuration was less progressive in comparison to the (0/90)$_{8S}$ configuration (Figure 4-1) while similar scatter was observed in both configurations (Table 4-1).

4.2.1.3 Multidirectional specimens – (0/90/45/-45)$_{4S}$ and (-45/45/0/90)$_{4S}$

In contrast to the cross-ply configurations, the behaviour of the multidirectional specimens was more complicated. The incorporation of angle plies enhanced the apparent compressive stiffness and strength of the CC specimens as it can be seen from the force displacement plots in Figure 4-1. The failure load of the multidirectional configurations was approximately 20% higher than the failure load of the cross-ply configurations (Figure 4-1). This improvement was attributed to the in-plane shear enhancement imbued on the laminate by the ±45° angle plies (Table 4-2). Moreover, albeit the (-45/45/0/90)$_{4S}$ configuration exhibited a stiffer response, its overall compressive behaviour was inferior to the (0/90/45/-45)$_{4S}$ baseline configuration, both prior and after failure initiation.

4.2.2 Digital Image Correlation (DIC)

DIC was then employed to characterise the deformation of the CC specimens during compressive loading, particularly just prior to failure. The distribution maps, shown in the following figures, illustrate the surface strain distribution ($\varepsilon$) in the direction of the load-bearing fibres (0°) and the in-plane shear strain ($\gamma_{xy}$) distribution across the surface. In fact, the latter could be only plotted as shear angle ($\alpha$ - (deg)) by the Aramis software[155], with a relation to the shear strain as follows: $\gamma_{xy} = \tan(\alpha)$. All the strain distribution maps have been generated at the same scale to aid comparison and they represent the strain
distributions just prior to failure in each representative specimen. Note that DIC was principally employed to qualitatively compare the different layups rather than provide quantitative measurements.

**4.2.2.1 Nominally identical cross-ply \((90/0)_{8S}\) configurations**

To gauge the inherent variability in the DIC data, comparison was initially made between nominally identical specimens. As Figure 4-3 illustrates, the strain distributions of the two nominally identical specimens were similar especially around the notch (red dotted line). In addition, in Figure 4-4, the distribution of the \(\varepsilon_y\) versus the section length (the distance from the centre of the notch to the free edge) is shown.

![Figure 4-3](image.png)

*Figure 4-3 Representative DIC strain distribution \((\varepsilon_y)\) in nominally identical \((90/0)_{8S}\) cross-ply configurations (a), (b); shear angle \((\alpha)\) in \((90/0)_{8S}\) in nominally identical \((90/0)_{8S}\) cross-ply configurations (c), (d), just prior to the crack initiation in two nominally identical specimens (-4.99 kN and -4.93 kN respectively). Red dotted semicircles represent the notch locations.*
As it was anticipated, the strains were high in the vicinity of the notch (red dotted line) and gradually diminished extending from the notch. Note that the curves shown in Figure 4-4 correspond to the same specimens as those shown in Figure 4-2. Although the distributions were similar, the strains at the notches differed. Nevertheless, the details close to the notch should be ignored since the strain distribution locally was greatly influenced by the detail of the notch shape, albeit it has been ensured that all notches were nominally identical (Chapter 3).

### 4.2.2.2 Cross-ply configurations – (90/0)₈S and (0/90)₈S

With regards to the cross-ply configurations, as it can be seen in Figure 4-5a, the axial strains ($\varepsilon_y$) just prior to failure in the baseline configuration (90/0)$_{8S}$ were slightly lower than the strains in the (0/90)$_{8S}$ configuration (Figure 4-5b). Moreover, the tensile strains at the free edge were higher in the baseline configuration (90/0)$_{8S}$. As for the shear strains (Figure 4-5c and Figure 4-5d) a difference in the distribution was observed, in particular the strains in the vicinity of the notch, as well as the scatter in the baseline configuration, were lower.

![Representative axial strain distribution ($\varepsilon_y$) versus the distance from the notch for two nominally identical (90/0)$_{8S}$ cross-ply configurations.](image)
4.2.2.3 Multidirectional configurations – (0/90/45/-45)_{4S} and (-45/45/0/90)_{4S}

Typical axial strain ($\varepsilon_y$) distributions of the two multidirectional configurations are shown in Figure 4-6a and Figure 4-6b. Just prior to failure the surface strain in the (-45/45/0/90)_{4S} configuration was higher along the specimen length, implying that the presence of the angle plies at the surface yielded higher strains. The different strain distribution around the notch suggested that the position (or depth) of the angle plies and thus the different layup led to different failure mechanisms. A comparison of the axial strain distribution versus the section length of the four configurations is given in Figure 4-7. Clearly, the strains at the notch were higher in the cross-ply configurations than those in the multidirectional configurations.
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Figure 4-6 DIC strain distribution ($\varepsilon_y$) in (a) QI-baseline (0/90/45/-45)$_{4S}$, (b) (-45/45/0/90)$_{4S}$ multidirectional configurations; shear strain distribution ($\alpha$), in (a) QI-baseline (0/90/45/-45)$_{4S}$, (b) (-45/45/0/90)$_{4S}$ multidirectional configurations just prior to the crack initiation (-6.13 kN and -5.96 kN respectively). Equivalent scale in terms of strain ranges from -0.009% to 0.017%.

Figure 4-7 Representative axial strain distribution ($\varepsilon_y$) versus the distance from the notch of the (0/90)$_{8S}$, (0/90)$_{8S}$, (0/90/45/-45)$_{4S}$ and (-45/45/0/90)$_{4S}$ configurations.
Even though away from the notch the strains converged, the multidirectional configurations exhibited higher tensile strains at the specimen free edge. What is more, the tensile stresses especially close to the free edge were significantly higher due to the presence of the 0° load-bearing plies at the surface. Although there was a difference in the strain along the loading direction, the shear strain distribution was relatively similar for the two multidirectional configurations (Figure 4-6c and Figure 4-6d).

4.2.3 Fractographic Analysis

In the following sections, the results from the fractographic analysis are presented. In particular, the dominant failure mechanisms were identified and the sequence of failure events which led to global failure of the various configurations is suggested.

4.2.3.1 Optical Microscopy

In this section, the results from the optical microscopic examination of specimens with different layups are presented. In particular, representative fracture morphologies acquired from nominally identical specimens at the notch and 15.5 mm away from the notch are illustrated and a graphical representation of each respective failure sequence which indicates the dominant failure modes is also given.

The characterisation of the fracture at these two locations, which has not been reported elsewhere in the literature, was considered essential for the interpretation of the failure propagation along the specimen length as well as the determination of the effect of post-failure damage on the fracture morphologies. In light of the knowledge on compressive failure mechanics presented in Chapter 2, the suggested interpretation of the failure process provided in this section for the various cross-ply and multidirectional layups, is based on evidence from numerous nominally identical specimens of each layup (cross-ply and multidirectional) at the notch and 15.5 mm away from the notch and therefore considered representative of the respective layup. To ease the reader, micrographs at higher magnifications indicating the fracture propagation, are also provided.
Cross-ply configuration – \((90/0)_{8S}\) (baseline)

Firstly, regarding the baseline configuration, \((90/0)_{8S}\) (Figure 4-8), the most dominant and first failure mechanism to have occurred was delamination (B) at the 22/23 ply interface (Figure 4-8b). This was apparent because the adjacent translaminar in-plane shear fracture (A) was discontinuous across this delamination. Should the in-plane shear fracture had occurred prior to delamination it would have propagated all the way to the surface. Therefore, the formation of the interlaminar fracture prior to the in-plane shear fracture implies that the critical strain release energy rate was exceeded before the in-plane shear strength. The interaction between the in-plane shear fracture and the delamination is illustrated at higher magnification both at the notch and 15.5 mm away from the notch in Figure 4-9 and Figure 4-10.

![Figure 4-8](image)

**Figure 4-8** (a) Observation faces position on the CC specimen; (b) schematic showing the sequence of the failure events; (c) optical microscopy pictures (x10) illustrating the fracture propagation at the notch (d) and 15.5 mm away from the notch in a typical \((90/0)_{8S}\) cross-ply configuration with ply numbers shown.
As soon as the delamination (B) at the 22/23 ply interface occurred, the laminate was
split into two sub-laminates and consequently the load was shed and redistributed. Upon
increased load the two sub-laminates behaved independently. The local instability caused by
the delamination (B), triggered the through-thickness shear fracture (A) in the left sub-
laminate, which propagated diagonally up to the surface of the specimen. This shear fracture
(A) angled at 53 ± 2° with respect to the loading direction (Figure 4-8c).

The failure mechanism which followed the 23/24 delamination (B) and the in-plane
shear fracture (A) was a secondary translaminar shear fracture (D) in the right sub-laminate.
Although the delamination at the 24/25 ply interface (C) was quite distinct, it could not have
occurred prior to the secondary translaminar fracture. The reason is that it had not
propagated beyond the shear fracture but appeared to have been induced by this
translaminar fracture. The delamination at the 30/31 ply interface (E) was most likely formed
after or simultaneously with the 24/25 delamination (C), induced as well by the shear fracture
(A). Note that the selective interaction of the dominant delamination (B) with the shear
fracture in one of the two sub-laminates indicates that the stress state in the two sub-
laminates differed.

![Figure 4-9 In-plane shear fracture-delamination interaction; (a) at the notch and (b) 15.5 mm
away from the notch (x20).](image)

The fracture morphology 15.5 mm away from the notch did not greatly differ compared to
that at the notch (Figure 4-8d ). However, the secondary in-plane shear fracture (D) had
propagated in a more complex manner. The propagation of the crack had followed a step-like path to the surface indicating that along the specimen further interlaminar fractures had occurred in that sub-laminate (Figure 4-10b).

![Figure 4-10 Fracture morphology of secondary in-plane shear fracture (a) at the notch and (b) 15.5 mm away from the notch (x20).](image)

It should be noted that the fracture morphology of the secondary in-plane shear fracture at the notch had been obscured by post-failure damage (Figure 4-8 and Figure 4-10). The presence of larger amount of post-failure damage at the notch was expected. This is due to the fact that the fracture surfaces at the notch kept sliding over each other upon increased loading even after failure had occurred. For this reason the fracture at the notch is described as “older” in comparison to that away from the notch[1].

**Cross-ply configuration – (0/90)$_{8}$s**

With regards to the second cross-ply configuration (Figure 4-11), the fracture was more severe than the fracture of the (90/0)$_{8}$s baseline configuration. Although a through-thickness shear fracture (A) had formed at 53 ± 2° with respect to the loading direction (Figure 4-11b), as in the previous case, the series of events which had led to catastrophic failure differed.

In particular, the fracture morphology in this instance was more discrete propagating across the width of the laminate (Figure 4-11d and Figure 4-12b). In the light of the morphologies at the notch shown in Figure 4-11c, it is clear that the translaminar shear fracture (A) had occurred prior to any interlaminar fracture. That is, the delaminations at the
8/9 (B) and 28/29 (E) ply interfaces had formed after the in-plane shear fracture (Figure 4-13), implying that the in-plane shear strength was exceeded prior to the critical strain release energy rate. The fracture morphology at the notch indicated that the shear fracture (A) had initiated at the left section of the laminate and propagated diagonally from left to right (Figure 4-11b and Figure 4-12a). This was more evident away from the notch, since at the notch the post-failure damage caused a “misalignment” of the propagation plane (Figure 4-11).

![Figure 4-11](image)

Figure 4-11 (a) Observation faces position on the CC specimen; (b) schematic showing the sequence of the failure events; (c) optical microscopy pictures (x10) illustrating the fracture propagation at the notch (d) and 15.5 mm away from the notch in a typical (0/90)$_{8S}$ cross-ply configuration with ply numbers shown.

As the fracture propagated along the specimen length, additional failure mechanisms developed (Figure 4-11d). For instance, the delaminations at the 15/16 and 23/24 ply interface (C and D respectively) were not present at the notch, which implies that the interlaminar fractures observed at the notch, 8/9 (B) and 28/29 (E), may have occurred due
to post-failure damage or there was migration of the delamination to a different ply interface through ply splitting.

![Figure 4-12 Fracture morphology of in-plane shear fracture (a) at the notch and (b) 15.5 mm away from the notch (x20).](image)

Nonetheless, considering the severity and extent of the interlaminar fracture at the 15/16 ply interface (C), it is clear that it occurred prior to the delamination at the 23/24 (D) ply interface and caused a significant local instability which led to the separation of the laminate into two sub-laminates (Figure 4-11 and Figure 4-14). That coupled with the slight misalignment in the propagation plane of the shear fracture, indicates that the delamination at the 15/16 ply interface (C) have led to the global instability (out-of-plane) of the specimen.
Figure 4-14 Fracture morphologies of the delaminations 15.5 mm away the notch (a) at the 15/16 ply interface (C) and (b) at the 23/24 ply interface (D) (x20).

*Multidirectional configurations* – *(0/90/45/-45)_{4S} (Baseline multidirectional)*

Although the incorporation of off-axis plies improved the compressive strength of multidirectional laminates (Section 4.2.1), the resulting fracture morphologies were more complicated (Figure 4-15).

Figure 4-15 (a) Observation faces position on the CC specimen; (b) schematic showing the sequence of the failure events; (c) optical microscopy pictures (x10) illustrating the fracture propagation at the notch (d) and 15.5 mm away from the notch in a typical *(0/90/45/-45)_{4S} multidirectional configuration with ply numbers shown.*
This was due to the higher post-failure damage and the higher amount of interlaminar fractures in comparison to the cross-ply configuration. The latter was anticipated since in multidirectional configurations both Poisson mismatch and shear-extension coupling would have been present between the plies and thus higher interlaminar stresses could have been induced [1]. Considering Figure 4-15c, a translaminar in-plane shear fracture similar to the fracture observed in the cross-ply laminates was also evident, however, the formation and propagation was more complicated and no delaminations were continuous across any shear fracture. Judging from the severity of the individual fractures, the failure appeared to have initiated due to through-thickness shear fractures (B and C) which started from the mid-plane and propagated diagonally through the thickness at 53 ± 2° with respect to the loading direction (Figure 4-16). These fractures led to the initial instability of the laminate (Figure 4-15b).

![Figure 4-16 Fracture morphology of the primary in-plane shear fracture 15.5 mm away from the notch (a) B and (b) C (x20).](image)

Upon increased load, the formation of shear fracture changed the stress state in the material which led to the formation of secondary through-thickness shear fractures (A and E) as shown in Figure 4-15b. The in-plane shear fracture on the left, of lower severity, propagated in two directions inducing a secondary shear fracture and a delamination at the 7/8 ply interface (H). Consequently, these secondary shear fractures induced the interlaminar fracture at the 15/16 ply interface (G) (Figure 4-18).
Figure 4-17 Fracture morphology of the secondary in-plane shear fracture at the notch (a) A and (b) E (x20).

As the fracture propagated there was a change in the fracture processes, as Figure 4-15d illustrates. Comparing the two fracture morphologies, a shift in the locations of the major damages was observed. In particular, the delamination at the 20/21 ply interface (F) occurred prior to the shear fractures and led to the separation of the laminate into two sub-laminates, which consequently behaved independently. The local instability caused by this interlaminar fracture triggered the formation of the shear fractures. Once those in-plane shear fractures had occurred, damage propagated via the delaminations at the 5/6 (I) and 24/25 ply interfaces (D) respectively.

Figure 4-18 Fracture morphologies of the delaminations 15.5 mm away the notch (a) at the 7/8 ply interface (H) and (b) at the 15/16 ply interface (G) (x20).

Eventually, the laminate was not able to withstand any further load and the fracture propagated to the surface which led to the global fracture of the laminate (out-of-plane).
Similar to the (0/90)\textsubscript{8S} cross-ply configuration, delamination occurred after the shear fracture suggesting that the shear strength was exceeded before the critical strain release energy rate. Even though a higher amount of delamination was observed in this case, as described in this section delaminations seem to have formed in a later stage of the fracture process. This coupled with the evidence of higher amount of post-failure damage (Figure 4-15), implies that some of the delaminations may have been post-failure damage.

**Multidirectional configurations – (-45/45/0/90)\textsubscript{4S}**

The fracture of the (-45/45/0/90)\textsubscript{4S} multidirectional configuration, as Figure 4-19c and Figure 4-19d illustrates, was more progressive and severe than the fracture of the baseline configuration (0/90/45/-45)\textsubscript{4S}.

Figure 4-19 (a) Observation faces position on the CC specimen; (b) schematic showing the sequence of the failure events; (c) optical microscopy pictures (x10) illustrating the fracture propagation at the notch (d) and 15.5 mm away from the notch in a typical (-45/45/0/90)\textsubscript{4S} multidirectional configuration with ply numbers shown.
This might have contributed to the lower failure load (Figure 4-1). The main difference between the two multidirectional configurations was the severity of the delaminations as well as the greater amount of post-failure damage at the notch.

In this instance, the fracture initiated due to interlaminar fracture in contrast to the fracture processes of the (0/90/45/-45)$_{4S}$ configuration, suggesting that the critical strain release energy rate was exceeded prior to the in-plane shear strength. In particular, interlaminar fractures occurred at the 12/13 (A) and 21/22 ply (D) interfaces (Figure 4-19b). Although it was very difficult to distinguish which of these two fractures occurred first, judging from their severity and the overall fracture morphology, the delamination at the 12/13 (A) ply interface was more likely to have occurred prior to the delamination at the 21/22 (D) ply interface (Figure 4-20).

The formation of these two interlaminar fractures (12/13 and 21/22 ply interfaces) effectively separated the laminate into three sub-laminates. Upon increased loading these fractures induced the translaminar shear fracture (C) and consequently two secondary shear fractures formed (B and E) at the sub-laminates. Although further shear fracture and delaminations were identified, it is most likely that most of these fractures were post-failure events (Figure 4-21).

![Figure 4-20 Fracture morphologies of the delaminations at the notch (a) at the 12/13 ply interface (A) and (b) at the 21/22 ply interface (D) (×20).]
Considering Figure 4-19d, in which the fracture morphology was “younger” than the morphology at the notch (Figure 4-19c), the translaminar shear fracture was angled at $53 \pm 2^\circ$ with respect to the loading direction, whilst at the notch was larger. This discrepancy in the angle was attributed to the large amount of post-failure damage induced in the laminate (sliding of the sub-laminates) and due to bending of the specimen after the failure had occurred. Finally, what optical microscopy revealed, in contrast to the baseline multidirectional configuration $(0/90/45/-45)_{4S}$ was that there was no shift in the location of the major fractures between the fracture morphology at the notch and 15.5 mm away from it.

![Figure 4-21 Fracture morphology of the secondary in-plane shear fracture 15.5 mm away from the notch (a) B and (b) E (x20).](image)

4.2.3.2 Scanning Electron Microscopy

In this section, the outcomes from the fractographic analysis using scanning electron microscopy are presented. In particular, the dominant failure modes which occurred in the various cross-ply and multidirectional configurations are identified and their interactions are highlighted. Prior to comparing the fracture in the different configurations, the fracture process in two nominally identical $(90/0)_{2S}$ cross-ply specimens is presented. This is to deduce the inherent variability and thus glean what is significant when comparing different configurations.
Inherent variability in fracture morphology between nominally identical cross-ply \((90/0)_{8S}\) configurations

The difference in the strain distribution around the notch and the failure load between the two nominally identical configurations, noted by the testing and DIC results, was also reflected in the local fractures observed by electron microscopy. Figure 4-22 and Figure 4-23 illustrate the difference in the fracture surfaces between the two nominally identical cross-ply specimens. Concerning the first configuration, fracture had initiated due to a \(90^\circ/0^\circ\) ply interface delamination supporting the optical microscopy findings. However, in the second configuration interlaminar fracture was not so extensive (Figure 4-22b). From the fracture morphology, as it can be clearly seen in Figure 4-22a, the fracture propagation was more progressive. At this point it should be noted that the post-failure damage in the two specimens was different.

Considering Figure 4-23a and Figure 4-23b, there was clearly a difference in the extent and severity of the interlaminar fractures between the two nominally identical specimens. The fracture morphology of the baseline specimen (Figure 4-23a) exhibited a greater amount of interlaminar fracture and less in-plane shear fracture. For the second \((90/0)_{8S}\) configuration, as the fracture morphology illustrates, the interlaminar fracture was not so extensive. Therefore, based on these observations it can be noted that no variability in the dominant failure modes and the failure sequence was observed. However, the severity of these failures differed, implying that these two nominally identical configurations exhibited different post failure damage.

Cross-ply configurations – \((90/0)_{8S}\) and \((0/90)_{8S}\)

The fractographic examination of the two cross-ply configurations also supported the observed difference in compressive response (Figure 4-1). Figure 4-24 and Figure 4-25 show the failure mechanisms which occurred during compression in the two different cross-ply configurations at the notch. As mentioned in the previous section, the fracture in the \((90/0)_{8S}\) configuration (Figure 4-24a and Figure 4-25a) initiated by an interlaminar fracture in the \(90^\circ/0^\circ\) ply interface which caused the two sub-laminates to slide over each other,
triggering secondary failure modes that led to global instability and eventually to catastrophic fracture.

On the contrary, in the (0/90)$_{8}$ configuration as Figure 4-24b and Figure 4-25b illustrate, delamination seem to have occurred after the shear fracture and the delamination extent was very limited. The discrete fracture morphology was evidence that another failure mechanism had occurred prior to interlaminar fracture and led to fracture of this configuration. Indeed evidence of 0° in-plane shear fracture was observed (Figure 4-25b) which verified the optical microscopy findings (Figure 4-15).

**Multidirectional configurations – (0/90/45/-45)$_{4S}$ and (-45/45/ 0/90)$_{4S}$**

Albeit the presence of angle plies in the laminate contributed to the improvement in the compressive response as shown in Section 4.1, it led to more complex fracture surfaces in comparison to the cross-ply configurations. In fact, the location of the angle plies had a significant effect on the failure mechanisms and the propagation of the fracture (Figure 4-26 and Figure 4-27).

Concerning the (0/90/45/-45)$_{4S}$ baseline multidirectional configuration, the failure was triggered by a shear fracture adjacent to the mid-plane and evidence is shown in Figure 4-26a. The deep delamination at the 45°/-45° ply interface which was observed in the optical microscopy analysis (Figure 4-26a) was also evident in this instance. Moreover, the step-like fracture morphology indicated that the translaminar shear fractures had formed and interacted with the off-axis intralaminar fractures[1].

In the second multidirectional configuration, (-45/45/0/90)$_{4S}$ (Figure 4-26b), the fracture morphology revealed the relative severity of the damage with evidence of shear fractures. However, the delaminations were deeper indicating that they had occurred prior to any in-plane shear fractures and more post-failure damage was evident. In particular, the delamination at the 90°/90° ply interface which was observed in the optical microscopy was evident in this instance as well.
Figure 4-22 (a) and (b) Inherent variability of the fracture morphologies in nominally identical cross-ply (90/0)\textsubscript{s} configurations at the notch (×50). and (c) and (d) 10.8 mm away from the notch (×100) respectively.
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Figure 4-23 (a) and (b) Inherent variability of the fracture morphologies in nominally identical cross-ply \((90/0)_\text{as}\) configurations 10.8 mm away from the notch \((\times 100)\).
Figure 4-24 Micrographs for (a) typical $(90/0)_{ss}$ and (b) typical $(0/90)_{ss}$ cross-ply configurations at the notch ($\times50$).
Figure 4-25 Micrographs for (a) typical (90/0)$_{ss}$ and (b) typical (0/90)$_{ss}$ cross-ply configurations at 10.8 mm from notch ($\times 100$).
Figure 4-26 Micrographs for (a) typical (0/90/45/-45)$_{4S}$ and (b) typical (-45/45/0/90)$_{4S}$ multidirectional configurations at the notch (x50).
Figure 4-27 Micrographs for (a) typical (0/90/45/-45)$_{4S}$ and (b) typical (-45/45/0/90)$_{4S}$ multidirectional configurations 10.8 mm from the notch (x100).
The delamination at the 45°/0° ply interface shown in Figure 4-26b was probably post-failure damage. As the fracture propagated along the specimen, similar failure mechanisms were observed (Figure 4-27a and Figure 4-26b). However, upon increased load the number of delaminations increased which led to global instability of the laminate. The delaminations in the surface plies were clear indicating that the fracture had propagated all the way to the surface.

To summarise, with the aid of fractographic analysis, the key failure mechanisms of the four configurations were identified and the sequence of the events which led to catastrophic failure was proposed. Optical and scanning electron microscopy revealed that the layup greatly influenced the failure process. In all four configurations, delamination and shear fracture were the dominant failure mechanisms, however, their interaction differed across the four configurations. In particular, in the (90/0)₈₅ cross-ply and (-45/45/0/90)₄₅ multidirectional configuration delamination triggered failure (at 90°/0° and 90°/45° ply interfaces respectively) which consequently induced translaminar shear fracture. On the contrary, in the (0/90)₈₅ cross-ply and (0/90/45/-45)₄₅ in-plane shear fracture triggered the failure which induced delamination (at 0°/90° and 45°/-45° ply interfaces respectively).

4.2.4 Theoretical Analysis
Using the properties tabulated in Table 3-2 and the thermal properties reported in the literature[205] for the IM7/8552 material system[150], LAP was used to calculate the mechanical stress distributions in the four configurations under pure compression (0.1% strain) and combined compression-shear (0.1% strain each). In addition to mechanical stresses, stresses due to curing and thermal (residual) effects were also considered by using curing temperature 180°C and room temperature 20°C. Nevertheless, no stresses due to hygroscopic effects were considered in this study because the laminates were tested soon after fabrication. In this theoretical analysis the complex geometries (CC) were not taken in account and thus the increased shear stresses induced in such a geometry (CC) were not considered in that sense. Note that the theoretical analysis is purely based on linear elastic
analysis and thus is comparative between stacking sequences in terms of stresses and should not be related to the experimental strength and failure presented in the previous sections.

The total applied stresses (mechanical and residual) in each ply are illustrated in Figure 4-28, parallel to the loading direction in half the laminate due to symmetry. In Figure 4-28, it can be seen that the stresses were much higher in the 0° plies, as it was expected since these plies bore most of the compressive load. Moreover, the stresses in the 0° plies on the two multidirectional configurations were higher than those in the cross-ply configurations, whilst the stresses in the 90° were slightly lower.

![Figure 4-28 Representative stress distribution in (a) (0/90)_{8S}, (b)(90/0)_{8S}, (c) (0/90/45/-45)_{4S}, (d) (-45/45/0/90)_{4S} for -0.1% applied strain.](image)

A more in depth illustration of the mechanical and residual stresses is provided in Figure 4-29, where a comparison between the magnitude of the mechanical and the non-mechanical stresses for (90/0)_{8S} cross-ply and (0/90/45/-45)_{4S} multidirectional configurations is made. In particular, for -0.1% applied strain the magnitude of the residual stresses was...
found to be considerable compared to the mechanical stresses as expected, especially in the 90° plies.

![Figure 4-29 Comparison of the stress distribution (residual-mechanical) in (90/0)_{3S} and (0/90/45/-45)_{4S} configurations for -0.1% applied strain.](image)

In addition to the theoretical analysis presented above, LAP was also used to obtain the applied stresses in the four configurations under a combined shear-compression load (0.1% applied strain each (Figure 4-30 and Figure 4-31)). This analysis was conducted in order to assess the effect the introduction of shear load had on the total stress distribution, since in the DIC analysis there was evidence of applied shear strain on all compact compression configurations (higher in multidirectional).

Considering Figure 4-28, Figure 4-29, Figure 4-30 and Figure 4-31, the introduction of shear load did not have great effect on the 0° load-bearing plies. The stress of the 0° load-bearing plies in the cross-ply configurations did not change but slightly dropped (approximately 2%) in the two multidirectional configurations. While there was no change in the 90° plies, the stress drop in the 0° load-bearing plies was accompanied by a large
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increase in the stress on the -45° plies and a large decrease in the stress of the +45° plies (both by approximately 100%).

Figure 4-30 Representative stress distribution in (a) (0/90)$_{8S}$, (b)(90/0)$_{8S}$, (c) (0/90/45/-45)$_{4S}$, (d) (-45/45/0/90)$_{4S}$ for a combined shear (0.1%) and compression strain (-0.1%).

Figure 4-31 Comparison of the stress distribution (residual-mechanical) in (90/0)$_{8S}$ and (0/90/45/-45)$_{4S}$ configurations a combined shear (0.1%) and compression strain (-0.1%).
Since LAP is based on the Laminate Plate Theory, the interlaminar stresses could not be calculated in order to determine which ply interfaces were more prone to delamination. However, large differences in the axial stresses (Figure 4-28) would imply that high interlaminar stresses should be expected in the respective ply interfaces. Therefore considering Figure 4-28, LAP indicated that 0°/90° ply interfaces were more prone to delamination than 0°/±45° or 90°/±45° ply interfaces.

Finally, LAP yielded the unnotched and notched compressive strengths for the four configurations [98,99,188,189,206,207], both for pure compression and combined compression-shear load. For this case, given the configuration of the CC specimen (Figure 3-1), a notch size of 4mm was used. The formulations used by LAP predicted higher compressive strength (both notched and unnotched) for the cross-ply configurations than in the multidirectional configurations (Table 4-3). Albeit the multidirectional and cross-ply configurations had different layup, LAP predicted identical unnotched[188,189] and BFS (Budiansky-Fleck-Soutis) notched compressive strength[83,188].

<table>
<thead>
<tr>
<th>Layup</th>
<th>Unnotched Compressive Strength (MPa) – Pure Compression</th>
<th>Unnotched Compressive Strength (MPa) – Combined Compression-Shear</th>
<th>BFS Notched Compressive Strength (MPa) – Pure Compression</th>
<th>BFS Notched Compressive Strength (MPa) – Combined Compression-Shear</th>
</tr>
</thead>
<tbody>
<tr>
<td>(0/90)₁ₛ</td>
<td>-784.0</td>
<td>-525.3</td>
<td>-225.3</td>
<td>-187.5</td>
</tr>
<tr>
<td>(90/0)₁ₛ</td>
<td>-784.0</td>
<td>-525.3</td>
<td>-225.3</td>
<td>-187.5</td>
</tr>
<tr>
<td>(0/90/45/-45)₁ₛ</td>
<td>-558.2</td>
<td>-373.5</td>
<td>-216.4</td>
<td>-166.3</td>
</tr>
<tr>
<td>(-45/45/0/90)₁ₛ</td>
<td>-558.2</td>
<td>-373.5</td>
<td>-216.4</td>
<td>-166.3</td>
</tr>
</tbody>
</table>

On the contrary, this was not observed in the mechanical testing (Figure 4-1). The introduction of shear load led to a drop in both unnotched and notched compressive strength. In fact, the unnotched strength in the cross ply configurations decreased by
approximately 30% while in the multidirectional configurations the compressive strength dropped by approximately 50%. The strength reduction of the notched compressive strength was approximately 16% and 25% for the cross-ply and multidirectional configurations respectively.

4.2.5 Numerical Analysis

In this section, the results from the numerical analysis of compressive failure of compact compression specimens are presented. As noted in Chapter 3, the purpose of this study was to assess the ability of the available built-in failure criteria in ABAQUS to model compressive failure of multidirectional composite laminates and compare against the experimental results. For this reason, the numerical study was focused on initiation of compressive failure, i.e. the first failure event to occur in various configurations with compact compression specimen geometry.

As it was described in Chapter 3, three different types of models were built and utilised in this study. It should be noted that in all three models the compliance was modelled as well. Initially a simple model of the CC specimen was built which used the Hashin failure criteria and delamination was ignored. Due to the overestimation of the compressive performance, the model was refined and each ply interface was modelled with cohesive zone elements to account for delamination while the Hashin failure criteria were used for the ply fracture.

Finally, the 2dVUMAT failure criterion was employed via the VUMAT subroutine and the cohesive elements were still included. All four configurations (cross-ply and multidirectional) were modelled in accordance to the testing procedure described in the last section. However, a comparison will be conducted only for the two baseline configurations, (90/0)$_{8S}$ and (0/90/45/-45)$_{4S}$. This is because the 2dVUMAT model was run only for the two baseline configurations due to time constraints. The predicted failure for all configurations is given in Table 4-4. Figure 4-32, and Figure 4-34 illustrate the comparison between the three models and the testing results for (90/0)$_{8S}$ and (0/90/45/-45)$_{4S}$ respectively.
Figure 4-32 Comparative force-displacement curves of the three models against test results for the (90/0)$_{8S}$ compact compression cross-ply baseline specimen.

<table>
<thead>
<tr>
<th>Layup</th>
<th>Without cohesive zone-Hashin</th>
<th>With cohesive zone-Hashin First Failure</th>
<th>With cohesive zone-Hashin Second Failure</th>
<th>With cohesive zone-2dVUMAT</th>
<th>Experimental observation</th>
</tr>
</thead>
<tbody>
<tr>
<td>(0/90)$_{8S}$</td>
<td>Fibre damage</td>
<td>Fibre damage</td>
<td>Delamination</td>
<td>N/A</td>
<td>Fibre damage</td>
</tr>
<tr>
<td>(90/0)$_{8S}$</td>
<td>Fibre damage</td>
<td>Fibre damage</td>
<td>Delamination</td>
<td>In-plane shear damage</td>
<td>Delamination</td>
</tr>
<tr>
<td>(0/90/45/-45)$_{4S}$</td>
<td>Fibre damage</td>
<td>Fibre damage</td>
<td>Delamination</td>
<td>Fibre damage</td>
<td>Fibre damage</td>
</tr>
<tr>
<td>(-45/45/0/90)$_{4S}$</td>
<td>Fibre damage</td>
<td>Fibre damage</td>
<td>Delamination</td>
<td>N/A</td>
<td>Delamination</td>
</tr>
</tbody>
</table>

Table 4-4 Failure prediction of the three models for the four different configurations.

As it can be seen in Figure 4-32, the results from the three models are plotted against the test result for the (90/0)$_{8S}$ configuration (Test). The first simple model (W_out), over predicted the compressive performance of the CC specimen and in terms of failure load
Compressive Failure of Multidirectional Composites Study

by approximately 40%. This overestimation was mainly attributed to the fact that the ply interfaces were not modelled and hence delamination was not taken in account. For this model, the first failure which occurred was fibre fracture (DAMAGEFC) accompanied by shear damage (DAMAGESHR). As soon as the ply interfaces were modelled (W_del), the predicted compressive performance was improved by approximately 30% and approximated the test result. This model also predicted that fibre failure (DAMAGEFC) occurred first, while delamination (SDEG) occurred subsequently.

Finally, the 2dVUMAT failure criterion predicted that in-plane shear damage (SDV3) triggered the failure process and was accompanied by compressive fibre damage (SDV4). Delamination in this final model occurred later. Considering the four force-displacement curves, it can be noted that the modelling of delamination by cohesive elements improved the prediction of the compressive performance significantly, while the improved models approximated the failure load and displacement, none of the two could provide an accurate prediction of both the failure load and displacement.

To assess the sensitivity of the model to the input mechanical properties, the set of data reported in LaRC05 [52,151] were used in addition to the dataset provided by Hexcel[150]. In Figure 4-33 the test results are compared against the results from the ABAQUS model using the data set provided by Hexcel (W_del_Hex) [150] and that obtained by a round-robin test as reported in WWFE II (W_del_LaRC05) [52,191] (see Table 3-2). Clearly, the use of the mechanical properties suggested in LaRC05 did not provide with a better prediction neither of the stiffness not the failure load. Note that no change in the dominant failure mode was observed with the application of the LaRC05 values (Table 4-4).

With regards to the baseline multidirectional configuration, in the first model which did not account for delamination the compressive performance was overestimated and particularly both the failure load (approximately 15%) and deformation. Fibre fracture (DAMAGEFC) was the first failure to occur. Consequently the incorporation of cohesive zone elements led to an underestimation of the compressive performance.
Figure 4-33 Comparative force-displacement curves between test and two different IM7/8552 material data sets for the (90/0)_{8S} compact compression cross-ply baseline specimen.

Figure 4-34 Comparative force-displacement curves of the three models against test results for the (0/90/45/-45)_{4S} compact compression multidirectional baseline specimen.
Both models which employed cohesive zone elements predicted that fibre damage (DAMAGEFC) triggered the failure process that was then accompanied by delamination (SDEG). Another interesting point can be made considering the force displacement curves shown in Figure 4-34. Although these two models utilised different failure criteria, they predicted a large stiffness drop upon the failure load having been reached. On the contrary the test showed that the specimen was able to withstand further load after the failure load had been reached. Similar to the cross-ply configuration, the mechanical properties reported in WWFE II (W_del_LaRC05) for the IM7/8552 [52,151], were also used in this instance to compare against the test results and the model which used the data set suggested by Hexcel (W_del_Hex) [150]. The comparison between the test result and the prediction of the two models suggest that while a slight change in the failure load was observed for the model using WWFE II values, both models failed to predict the experimental compression behaviour. Apart from the deviation in the failure load by approximately 10% both models did not also predict the behaviour once the failure load had been reached (Figure 4-35).

Figure 4-35 Comparative force-displacement curves between test and two different IM7/8552 material data sets for the (0/90/45/-45)_{4S} compact compression multidirectional baseline specimen.
As it was noted in Section 4.2.4, the interlaminar stresses in the four configurations could not be obtained by the Laminate Plate Theory. However, in ABAQUS with the aid of the cohesive zone elements, the interlaminar stresses were determined. In particular, the stress component corresponded to the Mode I interlaminar stresses (Figure 4-36), while the Mode II interlaminar stresses (Figure 4-37) were given by $\sqrt{(S_{13})^2 + (S_{23})^2}$ as suggested by Pinho et al.[190]. It should be noted that only half the ply interfaces are shown in Figure 4-36 and Figure 4-37 since only half the plies were modelled, due to symmetry. The interlaminar stresses shown in Figure 4-36 and Figure 4-37 were taken at -3.5 kN, before the first failure had occurred or these stresses had caused delamination.

![Mode I interlaminar stress distribution of the four different configurations obtained by ABAQUS at -3.5 kN.](image)

According to Figure 4-36 and Figure 4-37, some ply interfaces exhibited higher interlaminar stresses than others, both Mode I and Mode II, and thus they were more prone to delamination. In particular, the $(90/0)_{BS}$ exhibited higher interlaminar stresses than those of the $(0/90)_{BS}$ cross-ply configuration and the $(-45/45/0/90)_{4S}$ higher than those of the $(0/90/45/-45)_{4S}$ multidirectional configuration. The results agree with the observations from the fractographic analysis (Section 4.2.3), i.e. that in the $(90/0)_{BS}$ and $(-45/45/0/90)_{4S}$
delamination triggered the failure process. It should be noted that ABAQUS predicted higher Mode II interlaminar stresses for all four configurations than those in Mode I, and that higher interlaminar stresses occurred in -45°/45° or 0°/±45° or 90°/±45° ply interfaces rather than 0°/90° ply interfaces. Note that the magnitude of the interlaminar stresses predicted by ABAQUS for particular ply interfaces, differed significantly implying that the layup greatly influenced the distribution of both Mode I and Mode II interlaminar stresses and thus the development of delamination.

Figure 4-37 Mode II interlaminar stress distribution of the four different configurations obtained by ABAQUS at -3.5 kN.

As it was described in Chapter 3, in order to check for mesh convergence two modified versions of the model shown in Figure 4-34 (W_del) were employed, one with a coarser and one with a finer mesh (Figure 3-15) which used Hashin’s failure criterion. This model (FE-Intermediate) corresponds to the one used to obtain the interlaminar stresses shown in Figure 4-36 and Figure 4-37. As it can be seen in Figure 4-38, the performance of the three models is also compared with the representative compressive performance obtained in the experimental analysis to assess the prediction of each model.
Initially, what can be noted is that all three FE models could not accurately approximate the experimental stiffness and particularly predicted a more compliant behaviour. Concerning the coarse model, apart from the underestimation of the stiffness, it predicted very large displacements before the failure load had occurred, even though the failure load was predicted fairly accurately. With respect to the intermediate model (W_del), although the failure strain was closer to that observed in the experimental study, the stiffness and the failure load were underestimated. Subsequently, the introduction of a finer mesh led to a more accurate prediction of the failure load, however the stiffness was as well underestimated. Note that the load-displacement curve was interrupted earlier than the other two curves due to the high process time that was required.

Moreover, apart from the underestimation of the stiffness and the failure load, both models (intermediate and fine) predicted a large drop in stiffness after the failure load had been reached, which was not the case in the experimental study where the (0/90/45/-45)_4S specimen bore higher load for larger displacements. It should also be noted that in the model
with the fine mesh the predicted first and second failure did not differ in comparison to those presented in Table 4-4. The mesh sensitivity was also assessed for the models which were fed with material properties reported in WWFE II, in the same manner as described above. The load-displacement curves predicted by the three FE models with different meshes are compared against the test results in Figure 4-39. Along the same lines with the results shown in Figure 4-38, while the refinement of the mesh improved the prediction of the failure load, the displacement and post-failure behaviour were not in accordance with the experimental observation. The computational time for the model with the coarse mesh was 53 hours and 8 CPUs, 32 hours and 16CPUs for the model with the intermediate mesh and 80 hours and 16CPUs for the model with the fine mesh, while for all three models 4800 MB of RAM was utilised.

![Figure 4-39 Finite element models of a (0/90/45/-45)_4S employed to check for mesh convergence using material properties reported in WWFE II [52,191].](image-url)

Finally, concerning the 2dVUMAT model, as described in Chapter 3, there is an output variable (SDV16) which corresponds to the $l_x/l_{x,\text{max}}$ ratio (where $l_x$ is the characteristic element length) and assesses whether the chosen mesh is correct and the
$G_C$ can be properly dissipated. In particular, if this ratio is lower than unity then the chosen mesh is correct. In Figure 4-40 as the contour plot illustrates this ratio is well below one across the specimen and therefore it can be suggested that the chosen mesh was adequate in terms of in-plane fracture energy dissipation.

![SDV16 contour plot of SDV16 which corresponds to the $l_x/l_{x,\text{max}}$ ratio.](image)

To summarise, in this chapter the study of the compressive behaviour of multidirectional composite laminates using compact compression specimens was reported. In particular, the results from the experimental analysis (testing and Digital Image Correlation) and the consequent fractographic analysis (Optical and Scanning Electron Microscopy) were presented. Moreover, analytical tools were employed for the stress analysis and FEA (ABAQUS) was also performed to assess the capability of the currently available in the code failure models to capture the compressive failure of a complex fixture, such as compact compression specimen with different layups. This study highlighted that compressive failure is very sensitive to layup (be it cross-ply or multidirectional), in the sense that the layup influences the stress distribution, the dominant failure mechanisms as well as their interaction. This knowledge underpins the experimental study presented in the following chapter where the effect that the introduction of a new material in the stacking sequence (hybridisation) has on the compressive performance, failure mechanisms and sequence of a $(0/90/45/-45)_{2S}$ multidirectional laminate is investigated.
Chapter 5 – Compressive Failure of Hybrid Multidirectional Composites Study

5.1 Chapter Introduction
In this chapter the outcomes of the study on hybridization of composite laminates are presented for compression (compact, plain and sandwich panel) covering mechanical testing, fractographic, theoretical as well as numerical analysis.

5.2 Compressive Failure of Hybrid Multidirectional Composites

5.2.1 Compact Compression

5.2.1.1 Mechanical Testing
The compact compression results of the (0/90/45/-45)\textsubscript{2S} monolithic and hybrid laminates are summarized in Table 5-1 while typical force-displacement curves, based on five specimens per configuration, are shown in Figure 5-1. The chosen multidirectional layup, (0/90/45/-45)\textsubscript{2S}, is the same as that used in the study presented in Chapter 4. However, the materials used for the hybridisation study were HTS/MTM44-1 and IMS/MTM44-1, the latter with 40% higher $E_{11}$ [157]).

<table>
<thead>
<tr>
<th>Layup</th>
<th>Peak Load (kN)</th>
<th>Compliance (mm kN\textsuperscript{-1})</th>
<th>In-situ Lamina Thickness (mm)</th>
<th>Measured Specimen Thickness (mm)</th>
</tr>
</thead>
<tbody>
<tr>
<td>HTS</td>
<td>-4.96 ± 0.10</td>
<td>0.018 ± 0.001</td>
<td>0.246 ± 0.011</td>
<td>3.96 ± 0.03</td>
</tr>
<tr>
<td>HTS_IMS_A</td>
<td>-5.35 ± 0.07</td>
<td>0.014 ± 0.001</td>
<td>0.248 ± 0.005</td>
<td>4.01 ± 0.05</td>
</tr>
<tr>
<td>HTS_IMS_O</td>
<td>-5.86 ± 0.09</td>
<td>0.015 ± 0.001</td>
<td>0.248 ± 0.008</td>
<td>3.99 ± 0.03</td>
</tr>
<tr>
<td>IMS</td>
<td>-5.63 ± 0.08</td>
<td>0.016 ± 0.001</td>
<td>0.249 ± 0.012</td>
<td>4.02 ± 0.02</td>
</tr>
</tbody>
</table>

Table 5-1 Representative compact compression results of monolithic and hybrid configurations.

These materials differed from IM7/8552 as used in the previous study (Chapter 4) with the change being due to supply issues. In Table 5-1, the experimentally determined in-situ ply
thickness for the various configurations is also given (Appendix D). Considering the force-displacement curves shown in Figure 5-1, the monolithic and hybrid configurations exhibited different compressive performance. In particular, the HTS.IMS_O hybrid configuration exhibited the highest failure load followed by IMS, HTS.IMS_A and HTS. Even though the compliances of the four configurations were similar, the damage propagation which followed failure initiation varied, suggesting that the failure process had also been different. The test results shown in Figure 5-1 indicate the hybridisation of the laminate enhanced the compressive performance (with respect to the baseline HTS). The replacement of the HTS angle plies (±45°) with IMS enhanced the compressive performance by approximately 8% (HTS.IMS_A). In fact, the improvement of the overall stiffness was in accordance to the prediction of the rule of mixtures (within statistical error), however, it was quite lower than the CLT prediction (Table 5-2 - [162]).

![Figure 5-1 Compact Compression results of typical monolithic and hybrid (0/90/45/-45)2s configurations.](image)

The improvement in the compressive performance was approximately 18% when HTS 0° and 90° plies were replaced by IMS plies (HTS.IMS_O) and the compliance was in
Compressive Failure of Hybrid Multidirectional Composites Study

accordance to the prediction of CLT and the rule of mixtures (within statistical error – Table 5-2). The improvement of the compressive performance in the HTS_IMS_A configuration was mainly attributed to the increase of the overall effective shear stiffness, whereas in the HTS_IMS_O configuration was attributed to the increase of the effective shear stiffness and the enhancement of the compressive performance due to the higher compressive strength of the IMS plies (Table 3-1).

<table>
<thead>
<tr>
<th>Layup</th>
<th>Experimental Compliance (mm kN(^{-1}))</th>
<th>Theoretical Compliance (mm kN(^{-1}))</th>
<th>Rule of Mixtures (mm kN(^{-1}))</th>
</tr>
</thead>
<tbody>
<tr>
<td>HTS</td>
<td>0.018 ± 0.001</td>
<td>0.014</td>
<td>N/A</td>
</tr>
<tr>
<td>HTS_IMS_A</td>
<td>0.014 ± 0.001</td>
<td>0.009</td>
<td>0.013</td>
</tr>
<tr>
<td>HTS_IMS_O</td>
<td>0.015 ± 0.001</td>
<td>0.013</td>
<td>0.013</td>
</tr>
<tr>
<td>IMS</td>
<td>0.016 ± 0.001</td>
<td>0.010</td>
<td>N/A</td>
</tr>
</tbody>
</table>

Table 5-2 Comparison between experimental and theoretical compliances for the various CC monolithic and hybrid configurations (theoretical compliances don’t account for loading).

5.2.1.2 Digital Image Correlation (DIC)

Figure 5-2, Figure 5-3 and Figure 5-4 illustrate the surface strain distribution (\(\varepsilon_y\)) in the direction of the load-bearing fibres (0°) and the in-plane shear strain (\(\gamma_{xy}\), where \(\gamma_{xy} = \tan(\alpha)\)) distribution across the surface, with a 0.02% accuracy. The strain distribution maps have been generated at the same scale to aid comparison and their principal aim is to provide a qualitative compassion between the various monolithic and hybrid configurations rather than a quantitative comparison.

Considering the strain distribution of the four configurations shown in Figure 5-2, the axial strains (\(\varepsilon_y\)) just before failure in the vicinity of the notch (HTS) were much higher in the baseline configuration than the other three configurations, indicating that different failure processes influenced the failure initiation. The axial strains (\(\varepsilon_y\)) around the notch in the
HTS.IMS.O which exhibited the highest failure load, were lower than all the other configurations, however, in this configuration the highest tensile stresses at the free edge were observed. Moreover, the strain distribution of the IMS and HTS.IMS_A was similar across the surface which implied that similar failure mechanisms may have occurred (Figure 5-1). A comparative graphical illustration of the axial strain distribution of the four configurations is given in Figure 5-3. As it can be clearly seen the strain distributions at the notch and around the notch are in accordance with the strain distribution maps shown in Figure 5-2 but started to converge close to the free edge, albeit the two laminates with IMS appear to have been consistently lower.

![Figure 5-2](image.png)

Figure 5-2 Representative DIC strain distribution ($\varepsilon_y$) in (a) HTS (0/90/45/-45)$_{2S}$; (b) HTS.IMS_A (0/90/45/-45)$_{2S}$; (c) HTS.IMS.O (0/90/45/-45)$_{2S}$ and (d) IMS (0/90/45/-45)$_{2S}$ configurations, just prior to the crack initiation (at -5.01kN, -5.29kN, -5.78kN and -5.59kN respectively).

In addition to the strains along the loading direction ($\varepsilon_y$) the shear strains were also recorded (as shear angle) shown in Figure 5-4 at the individual failure loads of the four configurations.
Figure 5-3 Representative axial strain distribution ($\varepsilon_y$) versus section length of the HTS (0/90/45/-45)$_{2S}$, HTS_IMS_A (0/90/45/-45)$_{2S}$, HTS_IMS_O (0/90/45/-45)$_{2S}$ and IMS (0/90/45/-45)$_{2S}$ configurations, prior to failure.

Figure 5-4 Representative DIC shear angle distribution ($\alpha$) in (a) HTS (0/90/45/-45)$_{2S}$; (b) HTS_IMS_A (0/90/45/-45)$_{2S}$; (c) HTS_IMS_O (0/90/45/-45)$_{2S}$ and (d) IMS (0/90/45/-45)$_{2S}$ configurations, just prior to the crack initiation (at -5.01kN, -5.29kN, -5.78kN and -5.59kN respectively). Equivalent scale in terms of strain ranges from -0.009% to 0.017%.
As the shear distribution maps illustrate in Figure 5-4, the shear strains in the HTS_IMS_O configuration were lower in comparison to the other configurations, especially the HTS_IMS_A and IMS which indicated that the presence of the tougher IMS on the angle plies led to higher shear strains.

5.2.1.3 Fractographic Analysis

X-Ray radiography

Upon failure, the compact compression specimens were examined using X-Ray radiography. Representative fracture morphologies of the four configurations used in this study are presented in Figure 5-5. In Figure 5-5 the notch is located on the left hand side of each specimen, and the X-Ray radiographs shown in the following figure correspond to the same representative specimens used for the optical microscopy described in the following section. As Figure 5-5 illustrates X-Ray radiography provided valuable information about the failure mechanisms and the failure process.

Even though delaminations, ply splits and the crack propagation path could be identified, the exact depth of those failure modes could not be assessed with accuracy. This was due to the inherent inability of the conventional X-Ray radiography to provide through thickness details. Furthermore, during this study there was no access to stereoscopic facility which could have produced a three-dimensional representation of the fracture. The most important features provided by X-Ray radiography, were the crack propagation path and the step-like fracture morphology which was presented in Chapter 2, since these fractures are related to the in-plane shear fractures occurring in multidirectional laminated composites. In particular, these step-like fractures initiated from the notch due to 0° and ±45° ply splitting (perpendicular and diagonal red-dotted lines) and propagated across the specimen length. The main delaminations observed throughout the four configurations occurred at the 0°/90°, 45°/90° and 45°/-45° ply interfaces (Figure 5-5).
Figure 5-5 Typical X-Ray radiographs of (a) HTS (0/90/45/-45)_2S; (b) HTS_IMS_A (0/90/45/-45)_2S; (c) HTS_IMS_O (0/90/45/-45)_2S and (d) IMS (0/90/45/-45)_2S configurations.
As noted above the location of those interlaminar fractures could neither be identified accurately nor could it be deduced whether those failure modes were primary or secondary. For that reason optical and scanning electron microscopy were employed.

**Optical Microscopy**

The fracture morphologies of the HTS baseline configuration indicated that delamination and in-plane shear fracture were the dominant failure modes (Figure 5-6).

![Figure 5-6](image)

Figure 5-6 (a) Observation faces position on the CC specimen; (b) schematic showing the sequence of the failure events; (c) optical microscopy pictures (x10) illustrating the fracture propagation at the notch (d) and 15.5 mm away from the notch in a typical HTS (0/90/45/-45)\textsubscript{2S} hybrid configuration with ply numbers shown.

In particular, the mechanism which had occurred first and triggered the failure process was delamination (A) at the 3/4 ply interface. This delamination separated the laminate into two sub-laminates which consequently acted independently. With regards to the sub-laminate to the right of the (A) delamination, the failure was caused by the delamination propagating upwards and via an in-plane shear fracture (B) in ply 3 (45°), where it migrated to the 2/3 ply
interface (Figure 5-7a). Upon increased load, the failure reached the surface. Failure in the sub-laminate to the right of the initial delamination (A) was triggered by in-plane shear fracture (C) of ply 12 (Figure 5-7b). This failure mode consequently induced secondary fracture. As it can be seen in Figure 5-7b, this in-plane shear fracture propagated in two directions, inducing delamination (D) at the 13/14 ply interface and the in-plane shear failure (E).

![Fracture morphology of (a) primary failure mechanisms and (b) secondary failure mechanisms 15.5 mm away from the notch (×50).](image)

The latter changed the stress state in the material and triggered the delamination (F) at the 9/10 ply interface (Figure 5-6). The remaining fractures shown in the fracture morphology in Figure 5-6 were triggered by the secondary failure mechanisms described above and thus were not critical for the early stages of the failure process.

The fracture morphology of the HTS.IMS.A hybrid laminate (Figure 5-8) indicated that delamination and in-plane shear fracture were also the dominant failure mechanisms. Considering the severity of the failure modes observed in this particular fracture morphology, the delamination (A) at the 14/15 ply interface triggered the failure process. Consequently, the initial delamination separated the laminate into two sub-laminates which failed independently (Figure 5-9a). After the loss of stiffness due to delamination (A) the sub-laminate to the left of delamination (A) essentially carried most of the load. As for the failure in the left sub-laminate, delaminations (B) and (C) at the 7/8 and 2/3 ply interfaces triggered the failure (Figure 5-9b). However, considering the secondary damages these delaminations
induced and their magnitude, delamination (B) at the 7/8 ply interface seems to have occurred first, and consequently via in-plane shear fracture (E), induced a longitudinal split (ply 5).

![Diagram](image)

**Figure 5-8** (a) Observation faces position on the CC specimen (b) schematic showing the sequence of the failure events; (c) optical microscopy pictures (×10) illustrating the fracture propagation at the notch (d) and 15.5 mm away from the notch in a typical HTS_IMS_A (0/90/45/-45)$_{2S}$ hybrid configuration with ply numbers shown.

![Images](image)

**Figure 5-9** Fracture morphology of (a) primary failure mechanisms and (b) secondary failure mechanisms 15.5 mm away from the notch (×50).
Finally, the delamination (C) at the 2/3 ply interface also induced an in-plane shear fracture (D). At this point it should be noted that delaminations (A) and (C) occurred at a hybrid 90°/45° interface (Figure 5-5b).

The fracture morphology of the second hybrid configuration, HTS.IMS.O, is shown in Figure 5-10.

![Fracture Morphology](image)

Delamination and in-plane shear fracture were again the dominant failure mechanisms. The delaminations (A) and (F) at the 15/16 and 3/4 ply interface respectively (Figure 5-10d) seem to have occurred prior to other failure modes. As for the failure modes related to delaminations (A) and (F), delamination (A) caused significantly more damage and hence...
greater loss of stiffness. Initially, this delamination induced an in-plane shear fracture (B) (Figure 5-11b). Consequently, the fracture propagated in two directions, causing a delamination (C) at the 14/15 ply interface and an in-plane shear fracture (D). Delamination (F) at the 3/4 ply interface propagated in two directions inducing two shear fractures (E) and (G) (Figure 5-11b). Although evidence of other failure mechanisms can be observed in Figure 5-10d such as longitudinal splitting (ply 5) and delaminations at 7/8 and 10/11 ply interfaces, these failure modes were not critical for the failure process since the laminate had already failed (Figure 5-11b).

![Fracture morphology of (a) primary failure mechanisms and (b) secondary failure mechanisms 15.5 mm away from the notch (×50).](image)

The fracture morphology of the second monolithic configuration, IMS, is shown in Figure 5-12. Although mainly delaminations and in-plane shear fractures were observed, the fracture morphology was more complex than the fracture morphologies of the two hybrids shown in Figure 5-8 and Figure 5-10.

In the fracture morphology shown in Figure 5-12d, three main failure mechanisms were observed: delamination (A) at the 14/15 ply interface, in-plane shear fracture (E) and delamination (I) at the 5/6 interface. Considering delamination (A) at the 14/15 ply interface (Figure 5-13a) and the related failure modes, delamination (A) occurred prior to (E) and (I) and the redistribution of the stresses in the material caused the first stiffness loss of the
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laminate. Delamination (A) consequently induced in-plane shear fracture (B) via which migrated to the 12/13 ply interface (C).

![Diagram](image)

Figure 5-12 (a) Observation faces position on the CC specimen; (b) schematic showing the sequence of the failure events; (c) optical microscopy pictures (x10) illustrating the fracture propagation at the notch (d) and 15.5 mm away from the notch in a typical IMS (0/90/45/-45)_{2S} hybrid configuration with ply numbers shown.

![Images](image)

Figure 5-13 Fracture morphology of (a) primary failure mechanisms and (b) secondary failure mechanisms 15.5 mm away from the notch (x50).
As for the in-plane shear fracture (E), it propagated in two directions inducing the delaminations (F) and (D) at the 7/8 and 9/10 ply interfaces respectively (Figure 5-13b). The formation of the delamination (F) caused a change in the stress state in the material which consequently induced in-plane shear fracture (G) and delamination (H) at the 6/7 ply interface (Figure 5-12d and Figure 5-13b). Finally, delamination (I) at the 5/6 ply interface occurred independently and did not cause significant secondary damage. Nevertheless, considering Figure 5-12d, this delamination seems to have coalesced with the in-plane shear fracture (E) but it may also be due to post-failure damage, induced by the sliding of the fractured surfaces over each other upon increased load.

**Scanning Electron Microscopy**

Scanning Electron Microscopy was employed to confirm the findings of the X-Ray radiography, Optical Microscopy and to provide an insight to the interaction between the failure mechanisms in the four configurations. In Figure 5-14 to Figure 5-19, the fracture morphologies of the four configurations at the notch are shown and the dominant failure mechanisms are highlighted.

Electron microscopy revealed similar dominant failure mechanisms to those observed from X-Ray radiography and optical microscopy. Both longitudinal and off-axis ply splitting can be clearly seen in Figure 5-14a and Figure 5-15a. These longitudinal ply splits occurred due to a sharp change in the direct stress across the notch which led to a rapid rise in the shear stress along the 0° load-bearing plies and acted as a site where ply splitting formed in the adjacent off-axis plies. As well as the intralaminar fractures, the dominant failure modes (delamination and in-plane shear fracture) which were identified both in the X-Ray radiography and optical microscopy were also observed.

In the HTS configuration, the delaminations at the -45°/45° and 90°/-45° ply interfaces, which were deemed as the dominant failure mechanisms by optical microscopy, were also evident (Figure 5-14a).
Figure 5-14 Typical micrographs of (a) HTS (0/90/45/-45)_{2S}; (b) HTS.IMS.A (0/90/45/-45)_{2S} configurations at the notch (x32).
Figure 5-15 Typical micrographs of (a) HTS_IMS_O (0/90/45/-45)_{2S}; (b) IMS (0/90/45/-45)_{2S} configurations at the notch (x32).
Figure 5-16 Typical Micrographs of (a) HTS (0/90/45/-45)$_{2S}$, (b) HTS_IMS_A (0/90/45/-45)$_{2S}$; (c) HTS_IMS_O (0/90/45/-45)$_{2S}$ configurations 15.5mm away from the notch (x50).
Figure 5-17 Typical Micrographs of (a) HTS.IMS.O (0/90/45/-45)$_{2S}$, (b) IMS (0/90/45/-45)$_{2S}$; (c) IMS (0/90/45/-45)$_{2S}$ configurations 15.5mm away from the notch (×50).
Figure 5-18 Typical Micrographs of (a) HTS (0/90/45/-45)_2S; (b) HTS_IMS_A (0/90/45/-45)_2S configurations 15.5mm away from the notch (×100).
Figure 5-19 Typical Micrographs of (a) HTS.IMS.O (0/90/45/-45)\textsubscript{2S}; (b) IMS (0/90/45/-45)\textsubscript{2S} configurations 15.5mm away from the notch (×100).
Moreover, \(0^\circ/90^\circ\) delamination (Figure 5-16a) on the surface was also observed and at the mid-plane evidence of in-plane shear fracture and the associated delaminations (Figure 5-16a). Electron microscopy confirmed the dominant failure mechanisms, i.e. delamination at the \(-45^\circ/45^\circ\) and \(90^\circ/-45^\circ\) ply interfaces (Figure 5-14b, Figure 5-16b and Figure 5-18b) in the HTS/IMS\_A hybrid configurations. In addition, evidence of both longitudinal and off-axis ply splitting at the notch can be seen in Figure 5-14b. In the second hybrid configuration, HTS/IMS\_O, electron microscopy also confirmed the findings from X-Ray radiography and optical microscopy. In particular in Figure 5-15a, Figure 5-17a and Figure 5-19a the dominant delaminations at the \(-45^\circ/45^\circ\) and \(0^\circ/90^\circ\) ply interfaces can be seen, at the same positions as optical microscopy revealed. Moreover, evidence of ply splitting at the notch was also observed. Finally, in the second monolithic configuration, IMS, evidence of the dominant failure mechanisms is shown in (Figure 5-15b, Figure 5-17b, Figure 5-19b). In particular the main delaminations at the \(-45^\circ/45^\circ\), \(90^\circ/45^\circ\) and \(0^\circ/90^\circ\) which are shown in Figure 5-12 were also observed. In addition to the main interlaminar fractures longitudinal ply splitting can be clearly seen as well as the related splitting of the adjacent off-axis plies (Figure 5-17b).

To summarise, the fractographic analysis highlighted the dominant failure mechanisms in the four representative configurations as well as the suggested sequence of events which led to catastrophic failure. The observations in the fractographic analysis suggested that delamination triggered the failure process, contrary to those in Chapter 4 \(((0/90/45/-45)_4S\) layup) where in-plane shear failure was observed to have triggered the failure and consequently induced delamination. The amount and severity of delaminations were much higher in this study. In the monolithic configurations (HTS and IMS), the dominant delaminations occurred at the \(45^\circ/-45^\circ\) and \(90^\circ/45^\circ\) ply interfaces respectively, the failure in the HTS/IMS\_A was characterised by delaminations mainly at hybrid ply interfaces \((90^\circ/45^\circ)\) whereas the failure in the HTS/IMS\_O was characterised by delaminations at non-hybrid interfaces \((0^\circ/90^\circ\) and \(45^\circ/-45^\circ)\).
5.2.1.4 Theoretical Analysis

For the theoretical analysis of the hybrid laminates, LAP was employed to determine the mechanical stress distribution as shown in Figure 5-20 (for half a laminate due to symmetry). The theoretical analysis presented here is purely based on linear elastic analysis and provides a comparison between stacking sequences in terms of stresses and has nothing to do with the actual strength and failure. In addition to the mechanical stresses, thermal curing stresses were also considered assuming a 180°C curing temperature and a room temperature of 20°C. Note that stresses due to hygroscopic effects were not taken in account since the laminates were tested soon after fabrication.

Moreover, LAP was used to identify the interfaces which exhibited large stress differences and hence would be more likely to delaminate, since LAP could not directly determine any interlaminar stresses. For this purpose the mechanical properties of the HTS-MTM44-1 and IMS/MTM44-1 were used (Table 3-1).

As it can be seen from the stress distribution of the monolithic configurations (Figure 5-20), the IMS configuration was at a higher overall stress state than the HTS configuration. The incorporation of the IMS plies (0° and 90°) increased the overall stress state of the hybrid configuration. This improvement was mainly due to the stiffer IMS fibres (approximately 40%). Contrary to the large difference in the load-bearing plies, the stresses in the off-axis plies did not differ much.

Concerning the mechanical and non-mechanical (residual) stresses, across the width of the four multidirectional configurations, representative distributions are shown in Figure 5-21 and Figure 5-22 where a comparison between the magnitudes of these stresses is made. As Figure 5-21 and Figure 5-22 illustrate, under 0.1% applied strain the incorporation of IMS 0° and 90° plies increased the residual stresses (by approximately 50%) while the incorporation of IMS ±45° plies had a moderate effect the residual stresses.
Figure 5-20 Typical total stress distribution in the four configurations for -0.1% applied strain.

Figure 5-21 Comparison of the stress distribution (residual-mechanical) in HTS (0/90/45/-45)_{2S} and HTS_{IMS,O} (0/90/45/-45)_{2S} configurations for -0.1% applied strain.
In addition to the theoretical analysis presented above, LAP was also used to obtain the applied stresses in the four configurations under a combined shear-compression load (0.1% applied strain each (Figure 5-23, Figure 5-26 and Figure 5-27)). In this instance, the application of additional shear load did not affect the stresses on the 0° and 90° plies, however it decreased the stresses on the 45° and increased the stresses on the -45° plies by approximately 10%. No notable change was observed in the magnitude of the residual stresses.

The introduction of shear load also led to larger differences in the stress at the 90°/45° and 45°/-45° ply interfaces implying that the additional load made these particular interfaces more susceptible to delamination (Figure 5-20 and Figure 5-23). Nevertheless, there was no change in the stress difference in the 0°/90° ply interface across the four configurations.
Figure 5-23 Typical total stress distribution in the four configurations for a combined shear (0.1%) and compression strain (-0.1%).

Figure 5-24 Comparison of the stress distribution (residual-mechanical) in HTS (0/90/45/-45)_{2s} and HTS_IMS_O (0/90/45/-45)_{2s} configurations for a combined shear(0.1%) and compression strain (-0.1%).
Moreover, the unnotched and notched compressive strength for the four configurations were also calculated (Table 5-3) for both pure compression load and combined compression and shear load. Contrary to the results shown in Table 4-3, LAP predicted different unnotched and notched compressive strength for the four (0/90/45/-45)_{2S} configurations. As it can be seen in Table 5-3, initially the introduction of shear load degraded the unnotched compressive strength by approximately 30%.

The introduction of a notch decreased the compressive strength by approximately 40% for all configurations, whereas the application of shear load degraded the compressive strength further (by approximately 50%), indicating that the shear load played a significant role in the compressive behaviour of both monolithic and hybrid configurations. Considering Table 5-1 and Table 5-3, it can be seen that although the prediction of the notched compressive under pure compression was not in agreement with the experimental results,
the incorporation of shear load led to a more accurate prediction of the ranking highlighting the importance of the shear load contribution.

<table>
<thead>
<tr>
<th>Layup</th>
<th>Unnotched Compressive Strength (MPa) – Pure Compression</th>
<th>Unnotched Compressive Strength (MPa) – Combined Shear-Compression</th>
<th>BFS Notched Compressive Strength (MPa) – Pure Compression</th>
<th>BFS Notched Compressive Strength (MPa) – Combined Shear-Compression</th>
</tr>
</thead>
<tbody>
<tr>
<td>HTS</td>
<td>-820</td>
<td>-497</td>
<td>-553</td>
<td>-367</td>
</tr>
<tr>
<td>HTS_IMS_A</td>
<td>-971</td>
<td>-588</td>
<td>-673</td>
<td>-441</td>
</tr>
<tr>
<td>HTS_IMS_O</td>
<td>-804</td>
<td>-557</td>
<td>-578</td>
<td>-421</td>
</tr>
<tr>
<td>IMS</td>
<td>-914</td>
<td>-634</td>
<td>-619</td>
<td>-461</td>
</tr>
</tbody>
</table>

Table 5-3 Unnotched and Notched Compressive Strength of the four hybrid multidirectional configurations from LAP analysis.

5.2.1.5 Numerical Analysis

In this section the results from the numerical analysis on the compressive failure of hybrid compact compression specimens is presented. It should be noted that the delamination fracture toughness values used in this study were obtained by the study presented in Chapter 6. Figure 5-26 illustrates the behaviour of the four different configurations, two monolithic (HTS and IMS) and two hybrid (HTS_IMS_A and HTS_IMS_O) as predicted by the numerical analysis.

All four models utilised the Hashin failure criteria built in ABAQUS and cohesive elements were used to model ply interfaces (based on experimentally obtained delamination fracture toughness values – see Chapter 6). The performance of the four configurations was similar however different stiffness and failure propagation was predicted. Considering the compressive performance of the four configurations from the experimental testing (Figure 5-1), the numerical analysis did not predict accurately the compressive performance of the four multidirectional configurations.
In particular, the numerical analysis predicted that the two monolithic configurations were stiffer and achieved higher failure loads. However, the experimental study showed that the HTS IMS O hybrid configuration exhibited superior compressive performance than all the other configurations. Moreover the numerical analysis predicted an abrupt drop in the stiffness as the failure load was reached whereas in the test results such drop was not observed, but instead all four configurations were able to bear further load for larger displacements (Figure 5-1).

A comparison between the test results and the two models (incorporating Hashin and 2dVUMAT failure criteria respectively) for the HTS IMS_A hybrid configuration is given in Figure 5-27. The predicted compressive performance of both models differed from the test result. In particular, both models predicted stiffer behaviour for the HTS IMS_A configuration, however the failure load was predicted accurately. Both models (W_del and 2dVUMAT) predicted a drop in the stiffness as soon as the failure load was reached, however, the failure propagation differed significantly due to the different failure criteria.
Finally, in Table 5-4 the failure modes as predicted by the models, are presented in detail. Clearly ABAQUS predicted the shift in the failure modes, suggested in the experimental study, however different failure modes were predicted by Hashin and 2dVUMAT failure criteria.

![Comparative force displacement curves of the two numerical models against test results for the (0/90/45/-45)_{2s} hybrid HTS_IMS_A compact compression specimen.](image)

As in Chapter 4, ABAQUS was also employed to obtain the interlaminar stresses in the four multidirectional configurations and assess the effect of hybridization on delamination. The interlaminar stress distribution, both for Mode I and Mode II, is shown in Figure 5-28 and Figure 5-29. Clearly, Mode II interlaminar stresses were higher than Mode I interlaminar stresses which may be attributed to the increased amount of shear induced by the load application via the loading pins. The magnitude of the interlaminar stresses predicted by ABAQUS was in agreement with the findings from the fractographic analysis. In particular, ABAQUS predicted that in the HTS configuration, the highest interlaminar stresses occurred in the 45°/-45° and 90°/45° ply interfaces, interfaces where the dominant delaminations occurred (Figure 5-6). Regarding the HTS_IMS_A hybrid interface, ABAQUS
predicted that the 90°/45° and 45°/-45° ply interfaces were more prone to delamination, as it was observed in the fractographic analysis (Figure 5-8).

<table>
<thead>
<tr>
<th>Layup</th>
<th>With cohesive zone-Hashin First Failure</th>
<th>With cohesive zone-Hashin Second Failure</th>
<th>With cohesive zone-2dVUMAT</th>
</tr>
</thead>
<tbody>
<tr>
<td>HTS</td>
<td>Fibre damage</td>
<td>Matrix damage</td>
<td>N/A</td>
</tr>
<tr>
<td>HTS_IMS_A</td>
<td>Fibre damage</td>
<td>Matrix damage</td>
<td>In-plane shear damage &amp; ply splitting</td>
</tr>
<tr>
<td>HTS/IMS_O</td>
<td>Fibre damage</td>
<td>Delamination</td>
<td>N/A</td>
</tr>
<tr>
<td>IMS</td>
<td>Fibre damage</td>
<td>Delamination</td>
<td>N/A</td>
</tr>
</tbody>
</table>

Table 5-4 Failure prediction of the three models for the four different configurations.

Figure 5-28 Mode I interlaminar stress distribution of the four different configurations obtained by ABAQUS.

Similarly, for the IMS configuration ABAQUS predicted higher interlaminar stresses for the 90°/45° and 0°/90° ply interfaces, which was in accordance with the dominant delaminations observed in the fractographic analysis. However, the prediction for the HTS_IMS_O configuration did not agree with the findings from the fractographic analysis which suggested
that the dominant delamination was observed in the $0^\circ/90^\circ$ ply interface. Despite that, the delaminations at $90^\circ/45^\circ$ and $45^\circ/-45^\circ$ ply interfaces were predicted accurately.

![Figure 5-29 Mode II interlaminar stress distribution of the four different configurations obtained by ABAQUS.]

5.2.2 Plain Compression

5.2.2.1 Mechanical Testing

The plain compression testing results of the four multidirectional laminates (monolithic and hybrid) are summarized in Table 5-5 while typical force-displacement curves, based on five specimens per configuration, are shown in Figure 5-30. In addition, in Table 5-5 the in-situ ply thickness measurements of the various configurations are also provided (Appendix D). As it can be seen from the load-displacement curves shown in Figure 5-30, monolithic and hybrid configurations exhibited different performance during compressive loading. Even though the load-displacement curves followed the same pattern (abrupt load drop upon failure initiation), the failure load and displacement differed implying different failure processes.
Table 5-5 Plain Compression results of monolithic and hybrid configurations.

<table>
<thead>
<tr>
<th>Layup</th>
<th>Peak Load (kN)</th>
<th>Compliance (mm kN⁻¹)</th>
<th>In-situ Lamina Thickness (mm)</th>
<th>Measured Specimen Thickness (mm)</th>
</tr>
</thead>
<tbody>
<tr>
<td>HTS</td>
<td>-54.45 ± 3.16</td>
<td>0.017 ± 0.001</td>
<td>0.246 ± 0.011</td>
<td>3.96 ± 0.03</td>
</tr>
<tr>
<td>HTS_IMS_A</td>
<td>-53.06 ± 1.78</td>
<td>0.016 ± 0.001</td>
<td>0.248 ± 0.005</td>
<td>4.01 ± 0.05</td>
</tr>
<tr>
<td>HTS_IMS_O</td>
<td>-52.25 ± 2.18</td>
<td>0.014 ± 0.001</td>
<td>0.248 ± 0.008</td>
<td>3.99 ± 0.03</td>
</tr>
<tr>
<td>IMS</td>
<td>-48.83 ± 0.88</td>
<td>0.013 ± 0.001</td>
<td>0.249 ± 0.012</td>
<td>4.02 ± 0.02</td>
</tr>
</tbody>
</table>

Figure 5-30 Representative Plain Compression testing results of monolithic and hybrid (0/90/45/-45)₂s configurations.

Considering the test results in Figure 5-30 show that the hybridization of the laminates both enhanced and degraded the compressive performance. In particular, the replacement of ±45° HTS plies with IMS plies deteriorated the compressive performance by approximately 8% (due to the decrease of the effective shear stiffness) while the decrease was approximately 18% when HTS 0° and 90° plies were replaced by IMS plies even though
the IMS fibres were much stiffer (Table 3-1). Although the HTS exhibited the highest failure load, its compliance was the highest which can be attributed to the lower elastic modulus of the HTS fibres. The compliances obtained in plain compression differed from those reported in compact compression for all configurations apart from the HTS (Table 5-6). In this instance the IMS was found to be the stiffest of all four configurations. Regarding the hybrid configurations, both HTS_IMS_A and HTS_IMS_O were in agreement with the CLT and rule of mixtures prediction, within statistical error (Table 5-6).

<table>
<thead>
<tr>
<th>Layup</th>
<th>Experimental Compliance (mm kN⁻¹)</th>
<th>Theoretical Compliance (mm kN⁻¹)</th>
<th>Rule of Mixtures (mm kN⁻¹)</th>
</tr>
</thead>
<tbody>
<tr>
<td>HTS</td>
<td>0.017 ± 0.001</td>
<td>0.014</td>
<td>N/A</td>
</tr>
<tr>
<td>HTS_IMS_A</td>
<td>0.016 ± 0.001</td>
<td>0.009</td>
<td>0.013</td>
</tr>
<tr>
<td>HTS_IMS_O</td>
<td>0.014 ± 0.001</td>
<td>0.013</td>
<td>0.013</td>
</tr>
<tr>
<td>IMS</td>
<td>0.013 ± 0.001</td>
<td>0.010</td>
<td>N/A</td>
</tr>
</tbody>
</table>

Table 5-6 Comparison between experimental and theoretical compliances for the various configurations.

5.2.2.2 Digital Image Correlation (DIC)

Figure 5-31 and Figure 5-32 illustrate the surface strain distribution (\( \varepsilon_{ij} \)) in the direction of the load-bearing fibres (0°) and the in-plane shear strain (\( \gamma_{xy} \)) distribution across the surface, with 0.02% accuracy. The following strain distribution maps have been generated at the same scale to aid comparison. Note that these strain distribution maps have been obtained in order to provide a qualitative comparison between the various configurations rather than a quantitative comparison. As it was expected, all the configurations exhibited higher compressive strain states and no tensile strains were observed compared to the compact compression results. This was attributed to the different geometry of the plain compression specimen.
Figure 5-31 Typical DIC strain distribution ($\varepsilon_y$) in (a) HTS (0/90/45/-45)$_{2S}$; (b) HTS_IMS_A (0/90/45/-45)$_{2S}$; (c) HTS_IMS_O (0/90/45/-45)$_{2S}$; (d) IMS (0/90/45/-45)$_{2S}$ configurations, just prior to the failure initiation (at -53.67 kN, -51.08 kN, -52.16 kN and -48.52 kN respectively).

Equivalent scale in terms of strain ranges from -0.009% to 0.017%.

Figure 5-32 Typical DIC shear angle distribution ($\alpha$) in (a) HTS (0/90/45/-45)$_{2S}$; (b) HTS_IMS_A (0/90/45/-45)$_{2S}$; (c) HTS_IMS_O (0/90/45/-45)$_{2S}$; (d) IMS (0/90/45/-45)$_{2S}$ configurations, just prior to the failure initiation (at -53.67 kN, -51.08 kN, -52.16 kN and -48.52 kN respectively).
Considering the strain distribution of the four configurations shown in Figure 5-31, the axial strains \((\varepsilon_y)\) just before failure in the four configurations followed the results which are shown in Figure 5-30. In particular, the axial strains were higher in the baseline configuration \((HTS)\) which failed at the highest compressive load and was followed by \(HTS\_IMS\_A\), \(HTS\_IMS\_O\) and \(IMS\). These results are also shown in the comparative graphical illustration of the axial strain distribution against the section length in the four configurations in Figure 5-33.

![Graph showing typical axial strain distribution \((\varepsilon_y)\) versus the distance from the notch of the HTS \((0/90/45/-45)_{2S}\), HTS\_IMS\_A \((0/90/45/-45)_{2S}\), HTS\_IMS\_O \((0/90/45/-45)_{2S}\) and IMS \((0/90/45/-45)_{2S}\) configurations, prior to failure.](image_url)

In addition to the strains along the loading direction \(\varepsilon_y\), the shear strains \(\gamma_{xy}\) were also recorded as shown in Figure 5-32. As the shear distribution maps show in all configurations the shear strains were similar. These results indicate that no significant shear strains occurred during testing and thus it can be suggested that nearly pure compressive stress was applied on the plain compression specimens and the use of the antibuckling guide was successful. However, a greater scatter in the shear strains can be observed in the
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HTS and HTS_IMS_A where the surface plies were made of HTS which implied that these configurations experienced higher strains than the other two configurations (HTS_IMS_O and IMS). Finally, the axial strains at the notch differed significantly, implying high notch sensitivity, whereas along the section length (distance from the notch) the difference in the strain state was moderate (Figure 5-33).

5.2.2.3 Fractographic Analysis

X-Ray radiography
Representative fracture morphologies of the four plain compression configurations are shown in Figure 5-34. It should be noted that these morphologies correspond to the representative specimens presented in the following sections. X-Ray radiography provided essential information about the dominant failure modes such as delaminations and ply splits. The step-like fracture which was observed in the compact compression specimens was also evident in this instance and the mechanism that induced these morphologies was also identical, i.e. 0° ply splitting at the notch (Figure 5-34). Nevertheless, these ply splits were shorter, since the notch in the plain compression specimen was less sharp than that in the compact compression specimen[1]. In addition, the dominant delaminations which occurred in the four configurations are also shown in Figure 5-34. Although delaminations at identical ply interfaces were observed such as 0°/90°, 45°/-45° and 0°/45°, the exact location of these delaminations could not be identified as well as whether these delaminations were primary or secondary failure modes.

Optical Microscopy
The fracture morphologies of the plain compression specimens were different from the fracture morphologies observed in the compact compression specimens. The main differences were the severity of the failure modes and the presence of significant post-failure damage.
Figure 5-34 Typical X-Ray radiographs of (a) HTS (0/90/45/-45)$_{2S}$, (b) HTS.IMS_A (0/90/45/-45)$_{2S}$, (c) HTS.IMS_O (0/90/45/-45)$_{2S}$ and (d) IMS (0/90/45/-45)$_{2S}$ configurations.
This was attributed to the nature of the plain compression test, which was more dynamic than the compact compression leading less progressive failure processes. The presence of multiple delaminations and a large in-plane shear fracture were the main features of the fracture morphology (Figure 5-35, Figure 5-37, Figure 5-39, Figure 5-41). Nevertheless, as mentioned above, post-failure damage even away from the notch was evident. The sequences of failure events described below for the four configurations are based on observations on numerous nominally identical specimens of each configuration and thus suggested as typical for the particular configuration under the loading conditions described in Chapter 3.

Figure 5-35 (a) Observation face position on the PC specimen; (b) schematic showing the sequence of the failure events; (c) optical microscopy pictures (×10) illustrating the fracture propagation at the notch (d) and 15.5 mm away from the notch, in an typical HTS (0/90/45/-45)_{2S} hybrid configuration with ply numbers shown.
Regarding the failure process of the first monolithic configuration, HTS, the mechanism which triggered the fracture was the delamination (A) at the 5/6 ply interface that caused loss of stiffness and separated the laminate into two sub-laminates (Figure 5-35 and Figure 5-36a). These two sub-laminates then failed independently. The sub-laminate to the left of this delamination (A), failed mainly due to delamination (B) at the 1/2 interface which consequently migrated via multiple in-plane shear fractures (Figure 5-36a). In-plane shear fracture (D) triggered the fracture in the right sub-laminate. In particular, once the load bearing ply (12) failed due to kinking, the in-plane shear fracture propagated in two directions. As the in-plane shear fracture propagated, the plies started to slide over each other inducing multiple delaminations, such as (C) in the 9/10 ply interface (Figure 5-36b).

![Fracture morphology of primary failure mechanisms and secondary failure mechanisms](image)

**Figure 5-36 Fracture morphology of (a) primary failure mechanisms and (b) secondary failure mechanisms 15.5 mm away from the notch (×50), in the HTS configuration.**

The fracture morphology of the first hybrid configuration *HTS.IMS.A* is shown in Figure 5-37. In this instance the fracture morphology was mainly characterised by multiple delaminations. The most important delaminations in this fracture morphology were (A), (B), (C), (D) at the 13/14, 9/10, 1/2 and 5/6 ply interfaces respectively (Figure 5-37). Considering the magnitude of the delaminations and the related failure modes, delamination (A) in the 13/14 ply interface occurred first and caused the first drop in the stiffness of the laminate and separated the laminates into two sub-laminates (Figure 5-38a). Upon increased loading, this
The failure process in the left sub-laminate was triggered by the delamination (D) at the 5/6 ply interface which separated further the sub-laminate into two sub-laminates (Figure 5-38b). Consequently, these two further sub-laminates failed due to two delaminations (C) and (B) at the 1/2 and 9/10 ply interfaces. As it can be seen in Figure 5-37c further delaminations were induced due to post-failure damage.
Figure 5-38 Fracture morphology of (a) primary failure mechanisms and (b) secondary failure mechanisms 15.5 mm away from the notch (x50), in the HTS_IMS_A configuration.

The fracture morphology of the second hybrid configuration, HTS_IMS_O is illustrated in Figure 5-39.

Figure 5-39 (a) Observation faces position on the PC specimen; (b) schematic showing the sequence of the failure events; (c) optical microscopy pictures (x10) illustrating the fracture propagation at the notch and (d) 15.5 mm away from the notch, in a typical HTS_IMS_O (0/90/45/-45)_2S hybrid configuration with ply numbers shown.
Similar to the previously discussed configurations, multiple delaminations were evident. The failure process was triggered by delamination (A) at the 4/5 ply interface (Figure 5-40a). The two sub-laminates which were formed, consequently failed in different manners. On the one hand, the left sub-laminate failed due to the delamination (B) at the 1/2 interface (Figure 5-40a). On the other hand, the failure process in the right sub-laminate, where more plies were present, was more complex. In this instance, the failure was triggered by an in-plane shear fracture (E) which propagated in two directions inducing the delaminations (D) and (F) at the 7/8 and 9/10 ply interfaces respectively. Consequently, the latter (F) caused local loss of stiffness and induced an in-plane shear fracture which propagated all the way to the surface (Figure 5-40b).

![Fracture morphology of (a) primary failure mechanisms and (b) secondary failure mechanisms 15.5 mm away from the notch (×50), in the HTS_IMS_O configuration.](image)

Finally, the fracture morphology of the IMS configuration is shown in Figure 5-41. Multiple delaminations and in-plane shear fractures were also evident in this configuration (Figure 5-41). The failure process of this configuration was triggered by delamination (A) at the 11/12 ply interface which formed two sub-laminates. The right sub-laminate failed due to delamination (D) at the 13/14 ply interface whereas the failure process of the left sub-laminate was more progressive (Figure 5-42a). As it can be seen in Figure 5-41d, delamination (A) at the 11/12 ply interface caused the loss of the lateral support which then led to formation of in-plane shear fracture (C).
Figure 5-41 (a) Observation faces position on the PC specimen; (b) schematic showing the sequence of the failure events; (c) optical microscopy pictures (×10) illustrating the fracture propagation at the notch and (d) 15.5 mm away from the notch, in a typical IMS (0/90/45/-45)$_{2S}$ hybrid configuration with ply numbers shown.

Figure 5-42 Fracture morphology of (a) primary failure mechanisms and (b) secondary failure mechanisms 15.5 mm away from the notch (×50), in the IMS configuration.
Upon increased loading this in-plane shear fracture propagated across the width of the sub-laminate inducing multiple delaminations such as (F) and (G) at the 9/10 and 5/6 ply interface respectively (Figure 5-42b).

**Scanning Electron Microscopy**

To compare against the findings from the X-Ray radiography and Optical Microscopy, Scanning Electron Microscopy was employed. The fracture morphologies of the four configurations at the notch and 15.5mm away from the notch are shown in Figure 5-43 to Figure 5-48 illustrating the dominant failure mechanisms and their interaction. Considering the fracture morphology of the HTS configuration at the notch (Figure 5-43a) and away from the notch (Figure 5-45a and Figure 5-47a), Scanning Electron Microscopy confirmed the presence of delaminations at the 0°/90° and 45°/-45° ply interfaces in accordance to the observations made in optical microscopy and X-Ray radiography (Figure 5-34a). Moreover, the delamination at the 0°/90° ply interface (Figure 5-43a) on the surface and evidence of in-plane shear fracture at the mid-plane were also observed (Figure 5-45a). In Figure 5-45a the red dotted line illustrates the off-axis ply splitting at a -45° ply which was induced by the ply splitting of the adjacent load-bearing ply. Note that the off-axis splits, as clearly observed by X-Ray radiography (Figure 5-34a), tended to initiate from the longitudinal split and tangential to the round notch.

With regards to the HTS_IMS_A configuration, the fracture morphology of is shown at the notch and 15.5mm away from the notch in shown in Figure 5-43b, Figure 5-45b and Figure 5-47b. The dominant delaminations 0°/90° and 45°/-45° ply interfaces which triggered the fracture process according to optical microscopy were also evident (Figure 5-43b and Figure 5-45b). Moreover, the characteristic longitudinal and off-axis ply splits at the notch can be seen in Figure 5-45b and as they propagated forming a step-like morphology (red-dotted elbow line).
Figure 5-43 Typical micrographs of (a) HTS (0/90/−45)_{2S} and (b) HTS_IMS_A (0/90/45/−45)_{2S} configurations at the notch (×32).
Figure 5-44 Typical micrographs of (a) HTS_IMS_O (0/90/45/-45)\(_{2s}\) and (b) IMS (0/90/45/-45)\(_{2s}\) configurations at the notch (×32).
Figure 5-45 Typical micrographs of (a) HTS (0/90/45/-45)\textsubscript{2S} and (b) HTS\_IMS\_A (0/90/45/-45)\textsubscript{2S} configurations 15.5 mm away from the notch (×50).
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Figure 5-46 Typical micrographs of (a) HTS_IMS_O (0/90/45/-45)$_{2S}$ and (b) IMS (0/90/45/-45)$_{2S}$ configurations 15.5 mm away from the notch (x50).
Figure 5-47 Typical micrographs of (a) HTS (0/90/45/-45)$_{2S}$; (b) HTS_IMS_A (0/90/45/-45)$_{2S}$ configurations 15.5 mm away from the notch (×100).
Figure 5-48 Typical micrographs of (a) HTS_IMS_O (0/90/45/-45)_{2S} and (b) IMS (0/90/45/-45)_{2S} configurations 15.5 mm away from the notch (×100).
The fracture morphology of the $HTS_{IMS\_O}$ configuration ($0^\circ$ and $90^\circ$ plies replaces with IMS) is shown at the notch in Figure 5-44a and 15.5mm away from the notch in Figure 5-44a, Figure 5-46a and Figure 5-48a. Delamination at the $0^\circ$/$45^\circ$ and $0^\circ$/$90^\circ$ ply interfaces was found to be the dominant failure mode both at the notch and away from the notch (Figure 5-44a, Figure 5-46a and Figure 5-48a). This confirmed the optical microscopy findings (Figure 5-40). Little evidence of off-axis ply splitting was observed (Figure 5-44a).

Finally, the fracture morphology of the second monolithic configuration, IMS, is shown at the notch (Figure 5-44b) and away from the notch (Figure 5-46b and Figure 5-48b). Similar to the observations from optical microscopy and X-Ray radiography, delaminations at the $0^\circ$/$90^\circ$ and $45^\circ$/$-45^\circ$ ply interfaces were evident, as well as secondary delaminations such as $0^\circ$/$-45^\circ$ delamination. Additionally, off-axis ply splitting (red-dotted elbow line) was also observed to have been formed by the adjacent load-bearing ply splitting (Figure 5-48b). However, the delamination which was also evident at the $0^\circ$/$-45^\circ$ ply interface was probably due to post-failure damage which is in accordance to the observations from the optical microscopy (Figure 5-41).

To summarise, with the aid of optical and scanning microscopy as well as X-Ray radiography, the dominant failure mechanisms were identified and the sequence of events which led to global failure of the four multidirectional configurations was suggested. Albeit the fractographic analysis noted that delamination was the dominant failure mode across the four configurations, the ply interfaces at which these delaminations occurred were different, suggesting that the hybridization influenced the formation of those delaminations. While in the monolithic configurations ($HTS$ and $IMS$) the dominant delamination occurred at the $0^\circ$/$90^\circ$ ply interface, hybridization seems to have influenced the failure process in the hybrid configurations ($HTS_{IMS\_O}$ and $HTS_{IMS\_A}$). In particular, in the $HTS_{IMS\_A}$ configurations which exhibited much higher failure load, delamination occurred at a non-hybrid ply interface ($45^\circ$/$-45^\circ$) whereas in the $HTS_{IMS\_O}$ configuration the delamination
which triggered the failure occurred at a hybrid interface (-45°/0°), which implies that the fracture toughness of the hybrid ply interfaces may have played a significant role in the compressive performance.

5.2.3 Sandwich Panel Compression

5.2.3.1 Mechanical testing

The results from the sandwich panel compression testing are summarised in Table 5-7 and the force-displacement curves are shown in Figure 5-49, based in one panel per configuration, with (0/90/45/-45)_{2S} skins. Considering the results shown in Table 5-7, the HTS_IMS_A configuration failed at the highest load followed by HTS, HTS_IMS_O and IMS. Even though the HTS_IMS_A exhibited the highest failure load, its compliance was the lowest among all four configurations, which was not the case in the plain compression. In fact, the ranking of the four configurations in terms of the compliances was in agreement to that observed in compact compression and not in plain compression.

<table>
<thead>
<tr>
<th>Layup</th>
<th>Peak Load (kN)</th>
<th>Compliance (mm kN⁻¹)</th>
</tr>
</thead>
<tbody>
<tr>
<td>HTS</td>
<td>-205.2</td>
<td>0.0092</td>
</tr>
<tr>
<td>HTS_IMS_A</td>
<td>-215.1</td>
<td>0.0054</td>
</tr>
<tr>
<td>HTS_IMS_O</td>
<td>-185.2</td>
<td>0.0068</td>
</tr>
<tr>
<td>IMS</td>
<td>-171.9</td>
<td>0.0063</td>
</tr>
</tbody>
</table>

Table 5-7 Sandwich panel compression testing results.

Moreover, in the HTS_IMS_A and IMS configurations, a small drop in stiffness was observed indicating that failure had initiated earlier than in the HTS and HTS_IMS_O configurations (Figure 5-49). A similar “neck” was observed in the IMS configuration which should be attributed to the premature failure in the front skin since no failure in the back skin was observed. Note that the failure of the HTS_IMS_A configuration was characterized by the
failure of the back skin, which failed shortly after the failure in the front skin had initiated (Figure 5-51).

Figure 5-49 Sandwich panel compression testing results of monolithic and hybrid (0/90/45/-45)_2s configurations.

In addition to the displacement of the machine, the strains from four strain gauges were recorded (Chapter 3). In Figure 5-50 the recorded strain from the four gauges against the machine displacement is shown. SG1 and SG2 correspond to the gauges positioned on the front skin whereas SG3 and SG4 correspond to the strain gauges placed on the back skin. As it can be clearly seen for the strain gauges located on the back skin there was no stiffness drop since they did not fail. Thus the use of strain gauges on both skins was essential to monitor and compare the behaviour of the two skins. It should be noted that the strain-displacement curves shown in Figure 5-50 correspond to the sandwich panel with HTS (0/90/45/-45)_2s skins.

The behaviour of the other configurations was similar apart from the HTS_IMS_A in which the back skin had failed. The front skin surface of the four configurations after failure is illustrated in Figure 5-51. The crack propagation in the HTS, HTS_IMS_O and IMS
configurations was similar, i.e. initiated from the stress raiser and propagated all the way to the free edge. On the contrary, in the HTS.IMS.A configuration the crack propagation was more complicated. In particular, a big jump occurred which interrupted the propagation of the crack in the front skin. This jump appeared to have occurred due to the failure of the back skin which caused a large stiffness drop (Figure 5-51).

![Figure 5-50 Representative behaviour of strain gauges on the HTS monolithic sandwich panel compression testing.](image)

5.2.3.2 Digital Image Correlation (DIC)

Direct surface strain in the loading direction (\( \varepsilon_y \)) is shown in Figure 5-52 and in-plane shear strain (\( \gamma_{xy} \)) is shown in Figure 5-53, with an accuracy of 0.02%. Plots just prior to failure are shown for each of the four laminate configurations using the same strain scale to aid comparison. It should be borne that these strain distribution maps have been recorded to provide a qualitative comparison between the various configurations rather than a quantitative comparison.
Figure 5-51 Front skin surface of (a) HTS (0/90/45/-45)$_{2S}$; (b) HTS_IMS_A (0/90/45/-45)$_{2S}$, (c) HTS_IMS_O (0/90/45/-45)$_{2S}$; (d) IMS (0/90/45/-45)$_{2S}$ sandwich panel configurations.
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With regards to the axial strain distribution of the four configurations, the HTS configuration exhibited the highest axial strains and was followed by the HTS_IMS_A, HTS_IMS_O and IMS. This trend can be seen in Figure 5-54 where the axial strain against the distance from the notch for the four configurations is given. Considering Figure 5-52c, the HTS_IMS_O configuration experienced higher axial strains at the notch compared to the other configurations. Since the strain distribution of the HTS_IMS_O distribution was not as high across the section length as at the notch, it indicates that this configuration was very sensitive to the notch geometry. Moreover, the scale in the IMS configuration (both axial and shear) is not the same as in the other three configurations. This is due to the lower overall strain state of the IMS configuration, which in the same scale could not have been properly presented and compared against the other three configurations.

![Figure 5-52 DIC strain distribution (ε_y) in sandwich panels with (a) HTS (0/90/45/-45)_{2S}; (b) HTS_IMS_A (0/90/45/-45)_{2S}, (c) HTS_IMS_O (0/90/45/-45)_{2S}; (d) IMS (0/90/45/-45)_{2S} skin configurations, just prior to the crack initiation (at -203.89 kN, -214.07 kN, -183.96 kN and -170.22 kN respectively).](image)

Regarding the shear strains, the distribution maps shown in Figure 5-53 indicate that the four configurations experienced similar shear strains. In particular the shear strains away...
from the notch were fairly similar approximating zero which implies that nearly pure compressive load was applied in the four configurations. Compared to the shear strains on the compact and plain compression specimens the shear strains the sandwich panels experienced were much lower indicating that almost pure compressive load was applied onto the panels.

![DIC shear angle distribution](image)

Figure 5-53 DIC shear angle distribution ($\alpha$) in sandwich panels with (a) HTS (0/90/45/-45)$_{2S}$; (b) HTS.IMS.A (0/90/45/-45)$_{2S}$, (c) HTS.IMS.O (0/90/45/-45)$_{2S}$; (d) IMS (0/90/45/-45)$_{2S}$ skin configurations, just prior to the crack initiation (at -203.89 kN, -214.07 kN, -183.96 kN and -170.22 kN respectively). Equivalent scale in terms of strain ranges from -0.003% to 0.003%.

With respect to the axial strain versus the sandwich panel section length shown in Figure 5-54, the difference in the strain state of the four configurations was clear. In this plot two main features can be seen, the significantly lower strain state of the IMS configuration and the high strains at the notch in the HTS.IMS.O configurations. As for the latter, taking in account the strains away from the notch, it seems that the high strains in the vicinity of the notch should be attributed to the notch sensitivity which is not representative of the overall strain state of that particular configuration.
5.2.3.3 High Speed Crack Propagation

High speed footage was recorded to enable crack tip tracking during the highly dynamic fracture. In Figure 5-56 and Figure 5-57 the propagation of the crack in each configuration is shown as a series of snapshots, each taken at 16 μs intervals. The development of the crack tip is depicted as a straight red line (assuming constant speed) and the measured crack speed is also noted. It should be noted that for the HTS_IMS_A the crack speed was measured up to the point the back skin had failed and induced the crack jump. Beyond that point (Figure 5-51) the crack propagation could not be measured and is therefore not shown in Figure 5-56.

![Axial strain distribution](image)

**Figure 5-54** Axial strain distribution ($\varepsilon_y$) versus section length of the HTS (0/90/45/-45)$_{2S}$, HTS_IMS_A (0/90/45/-45)$_{2S}$, HTS_IMS_O (0/90/45/-45)$_{2S}$ and IMS (0/90/45/-45)$_{2S}$ configurations, just prior to failure.

It has been reported in the literature that crack speed in dynamic fracture is linearly proportional to the strain energy at fracture, for uniformly loaded tensile strips[5]. In this case for uniformly compression loading, the results were plotted as crack propagation speed against failure load here (Figure 5-55). The failure load is squared to be directly proportional to strain energy and the trend is linear. Unfortunately, HTS_IMS_A exhibited premature back
skin failure, leaving the crack speed observation questionable. However, the remaining data points do suggest this linear relationship. Such an observation requires further test data to be fully confirmed.

X-Ray radiographs of the front skins of the four sandwich panels are shown in Figure 5-58. The section length of the front skin panels represents half the length of the actual skin length (100mm). Although as it can be seen the fracture morphology away from the notch was obscured by the aluminium honeycomb which was difficult to remove. Even though the loading conditions and the specimen configurations were different in comparison to the plain compression testing longitudinal and off-axis ply splits were also observed initiating from the notch.

![Graph showing Crack Speed Propagation versus Force for HTS (0/90/45/-45)_{2S}; HTS_IMS_A (0/90/45/-45)_{2S}; HTS_IMS_O (0/90/45/-45)_{2S} and IMS (0/90/45/-45)_{2S} skin configurations.](image)

Figure 5-55 Crack Speed Propagation versus Force for HTS (0/90/45/-45)_{2S}; HTS_IMS_A (0/90/45/-45)_{2S}; HTS_IMS_O (0/90/45/-45)_{2S} and IMS (0/90/45/-45)_{2S} skin configurations.

The most important finding of the X-Ray radiography in these sandwich panels was the considerably less evidence of interlaminar fracture in the skins, in comparison to the plain compression specimens, even though the load application was similar. In fact, the extent of interlaminar fracture seems to have been roughly equal to the notch diameter.
Figure 5-56 Crack propagation speed in sandwich panels with (a) HTS (0/90/45/-45)$_{28}$ and (b) HTS_IMS_A (0/90/45/-45)$_{28}$ skin configurations.
Figure 5-57 Crack propagation speed in sandwich panels with (a) HTS_IMS_O (0/90/45/-45)$_{2S}$ and (b) IMS (0/90/45/-45)$_{2S}$ skin configurations.
Figure 5-58 X-Ray radiographs of the four configurations.
This discrepancy can be mainly attributed to the constraint that the aluminium honeycomb offered to the skins, preventing that way the excessive out-of-plane deformation of the front skins. Therefore the fracture of these sandwich panels can be mainly attributed to translaminar and intralaminar fractures.

To sum up, in this chapter the effect of hybridisation of multidirectional composite laminates on the compressive performance and failure process was investigated. In particular, the approach was to compare the compressive performance of two monolithic carbon fibre/epoxy systems (CYTEC HTS/MTM44-1 and IMS/MTM44-1) against that of their respective hybrids in three different test and specimen configurations, compact compression, plain compression and sandwich panel compression. This study highlighted that the replacement of particular plies by others of a second material had both positive and negative effect in the compressive performance. Moreover, load application as well as specimen and notch geometry also influenced the effect of hybridisation, since no consistency in compression strength rankings was observed across the three compression tests. In addition, observations that hybridisation had also affected the location of the key delaminations were made. In particular, it was observed that delaminations at hybrid interfaces were more detrimental with respect to the compressive performance than that at a monolithic interface, which suggested that hybrid interfaces may have exhibited higher delamination fracture toughness. To confirm this hypothesis, a thorough study was carried out on delamination fracture toughness at hybrid and monolithic ply interfaces. The results from the experimental procedure (Mode I, Mode II and Mixed Mode I/II) as well as the consequent fractographic analysis are presented in the following chapter.
Chapter 6 - Delamination Fracture Toughness of Hybrid Composites

In this chapter, the results from the Mode I, Mode II and Mixed Mode I/II (75% Mode I) delamination fracture toughness tests are presented. For this study, three unidirectional configurations were used, two pure modes and one hybrid, (HTS)$_{16}$, (IMS)$_{16}$ and (IMS/HTS$_{4}$/IMS/HTS/IMS$_{2}$/IMS/HTS) respectively. Note that in the hybrid configuration the IMS was the uppermost ply. Finally it should also be noted that the manufacturing and testing of these delamination fracture toughness specimens were conducted in conjunction with an MSc student (Miss Chuan Li) while the planning, data reduction, fractographic analysis and interpretation of the results were carried out exclusively by the author.

6.1 Mode I – DCB

6.1.1 Results
The results from Mode I delamination fracture toughness testing for the three configurations HTS, HTS_IMS and IMS are shown in Figure 6-1, Figure 6-2 and Figure 6-3 respectively for all five specimens per configuration, while representative delamination fracture toughness against crack length curves are shown in Figure 6-4. From these figures it can be seen that the scatter was relatively low (<7%) for both initiation and steady state growth and thus the results can be deemed accurate and the testing procedure successful.

Mode I delamination fracture toughnesses for the three configurations are given in Table 6-1 for both initiation and steady state, as well as the percentage difference between the initiation and steady state growth, which relates to the fibre/matrix interface strength[1]. Moreover, the elastic moduli of the three configurations as determined by Equation 3-4 are also presented. Mode I fracture toughness of HTS was the lowest of the three configurations while the hybrid specimens yielded the highest delamination resistance. In comparison to the tougher monolithic IMS configuration, the enhancement was approximately 21% for initiation and 12% for steady state while in comparison to the HTS configuration the improvement was approximately 25% for initiation and 28% for steady state.
(\( \alpha \approx 65 \text{mm approximately} \)) respectively. The R curve for the hybrid configuration was very similar to that of the IMS configuration which may indicate a strong contribution of the IMS plies to the delamination fracture toughness of the hybrid configurations (Figure 6-4).

**Figure 6-1** Mode I delamination fracture toughness versus crack length curves for HTS DCB specimens.

**Figure 6-2** Mode I delamination fracture toughness versus crack length curves for HTS_IMS Hybrid DCB specimens.
According to the supplier (CYTEC), the delamination fracture toughness values for HTS/MTM44-1 and IMS/MTM44-1 are 310 J/m² and 340 J/m² respectively [208]. However, these fracture toughness values have been obtained using specimens which have been manufactured by vacuum consolidation and not autoclave. According to the author’s knowledge, no delamination fracture toughness data for these materials are available in the literature. In addition, considering the experimentally obtained elastic moduli values (Table 6-1) and those reported by the CYTEC (Table 3-1), a good agreement was noted for both systems since the supplier’s values fall within the experimentally obtained values range.
Regarding the elastic moduli of the hybrid configurations, it exceeded the elastic moduli of the IMS/MTM44-1 and also deviated from what a simple rule of mixtures would suggest.

Figure 6-4 Representative Mode I delamination fracture toughness curves for HTS, IMS and HTS_IMS Hybrid DCB specimens.

6.2 Mode II – ELS

6.2.1 Results

The results from Mode II delamination fracture toughness testing of the three configurations HTS, HTS_IMS and IMS are shown in Figure 6-5, Figure 6-6 and Figure 6-7 respectively for all five specimens per configuration. The scatter in all three configurations was relatively low (<8%) for both initiation and steady state growth \( (a \approx 65 \text{ mm}) \), which implies that the experimental procedure was successful and the obtained fracture toughnesses were accurate. The Mode II delamination fracture toughnesses for the three configurations are given in Table 6-2 for both initiation and steady state while representative delamination fracture toughness against crack length curves are shown in Figure 6-4. The hybrid configuration exhibited the highest delamination resistance and was followed by IMS and
Delamination Fracture Toughness of Hybrid Composites

HTS. Hence the results from the Mode II testing followed the trend which was observed in Mode I testing.

**Figure 6-5** Mode II delamination fracture toughness versus crack length curves for HTS ELS specimens.

**Figure 6-6** Mode II delamination fracture toughness versus crack length curves for HTS_IMS Hybrid ELS specimens.
Delamination Fracture Toughness of Hybrid Composites

Figure 6-7 Mode II delamination fracture toughness versus crack length curves for IMS ELS specimens.

<table>
<thead>
<tr>
<th>Layup</th>
<th>( G_{IC} (J/m^2) )-Initiation</th>
<th>( G_{IC} (J/m^2) )-Steady State</th>
<th>Percentage difference (%)</th>
</tr>
</thead>
<tbody>
<tr>
<td>HTS</td>
<td>926 ± 47</td>
<td>940 ± 58</td>
<td>7 ± 6</td>
</tr>
<tr>
<td>HTS_IMS</td>
<td>1302 ± 87</td>
<td>1303 ± 94</td>
<td>8 ± 7</td>
</tr>
<tr>
<td>IMS</td>
<td>1140 ± 64</td>
<td>1152 ± 91</td>
<td>8 ± 7</td>
</tr>
</tbody>
</table>

Table 6-2 Mode II test results.

The enhancement in the hybrid configuration was approximately 14% for initiation and 13% for steady state over monolithic IMS whilst in comparison to the HTS configuration the improvement was approximately 40% for initiation and 39% for steady state \( (a \approx 65 \text{ mm approximately}) \) respectively. A graphical representation of the apparent difference in the delamination fracture toughness values is shown in Figure 6-8. Interestingly, the R curve of the hybrid configuration was closer to the R curve of the IMS configurations. This, in addition to the fact that IMS was the uppermost ply, indicates that IMS plies may have influenced the delamination fracture toughness of the hybrid configuration more than the HTS plies.
6.3 Mixed-Mode – MMB

6.3.1 Results
The results from the Mixed-Mode I/II (75% Mode I) delamination fracture toughness testing of the three configurations HTS, HTS_IMS and IMS are shown in Figure 6-9, Figure 6-10 and Figure 6-11 respectively for all five specimens per configuration. These are plotted in terms of total delamination fracture toughness versus crack length. Similar to Mode I and Mode II, the scatter was relatively low (<8%) both for initiation and steady state especially for the HTS and the hybrid configurations. This indicates that the obtained fracture toughnesses can be considered accurate. Out of the three configurations, the hybrid configuration exhibited the highest total delamination resistance. The total delamination fracture toughness, both at initiation and at the steady state, was approximately double compared to the two monolithic configurations (Figure 6-12). The enhancement of the mixed-mode delamination fracture toughness was the highest recorded among the three fracture toughness tests.
Figure 6-9 Mixed Mode I/II delamination fracture toughness versus crack length curves for HTS MMB specimens.

Figure 6-10 Mixed Mode I/II delamination fracture toughness versus crack length curves for Hybrid MMB specimens.
Delamination Fracture Toughness of Hybrid Composites

Figure 6-11 Mixed Mode I/II delamination fracture toughness versus crack length curves for IMS MMB specimens.

<table>
<thead>
<tr>
<th>Layup</th>
<th>$G_T$ (J/m²)-Initiation</th>
<th>$G_T$ (J/m²)-Steady State</th>
<th>Percentage difference (%)</th>
</tr>
</thead>
<tbody>
<tr>
<td>HTS</td>
<td>537 ± 26</td>
<td>550 ± 22</td>
<td>7 ± 3</td>
</tr>
<tr>
<td>HTS_IMS</td>
<td>1001 ± 90</td>
<td>1223 ± 69</td>
<td>34 ± 8</td>
</tr>
<tr>
<td>IMS</td>
<td>513 ± 90</td>
<td>670 ± 20</td>
<td>58 ± 5</td>
</tr>
</tbody>
</table>

Table 6-3 Mixed Mode I/II test results.

The delamination fracture toughness values for the three configurations are tabulated below (Table 6-3) for both initiation and steady state loci (approximately 65mm), while representative delamination fracture toughness against crack length curves are shown in Figure 6-12. With a closer look at Table 6-3 it is clear that both Mode I and Mode II components in the hybrid configuration were approximately double in comparison to both monolithic configurations.
The relation between total delamination fracture toughness with the Mode I and Mode II components for each specimen is shown in Figure 6-13 and Figure 6-14 respectively. Note that since the total delamination fracture toughness was influenced mainly by the Mode I component, these plots would look similar if the total delamination fracture toughness had been plotted against the Mode II component.

![Graph showing fracture toughness curves](image)

**Figure 6-12** Representative Mixed Mode I/II delamination fracture toughness curves HTS, IMS and HTS/IMS Hybrid MMB specimens.

In Figure 6-13, which corresponds to the initiation locus, the difference between the hybrid and the monolithic configurations can be clearly seen even though there was a relatively high scatter in the hybrid configuration. Comparing the two monolithic configurations, a shift in the behaviour of the IMS specimens can be noted. That is, at the initiation locus, the IMS exhibited lower fracture toughness than the HTS configuration both in Mode I and Mode II (Figure 6-13) while at the steady state locus both components increased (Figure 6-14). Finally, at the steady state locus, the scatter in the hybrid configurations decreased in comparison the scatter at the steady state locus.
Figure 6-13 $G_T$ versus $G_{IC}$ component curves for HTS, IMS and HTS/IMS Hybrid MMB specimens at the crack initiation locus.

Figure 6-14 $G_T$ versus $G_{IIc}$ component curves for HTS, IMS and HTS/IMS Hybrid MMB specimens at the steady state locus.
6.4 Fractographic Analysis

While in the previous section the results from the DCB, ELS and MMB tests were presented suggesting that hybridization significantly improved Mode I, Mode II and Mixed Mode I/II delamination resistance, in this section the observations from the fractographic analysis are presented in order to provide an explanation of the interesting experimental results. In particular, the interfaces of the monolithic specimens (HTS and IMS) are compared with their respective interfaces in the hybrid specimens, in terms of morphology and evidence of features related to the particular delamination test mode.

6.4.1 Mode I

In this section the comparative study on the fracture morphologies of the three different configurations conducted on Mode I delamination fracture toughness specimens is presented. In Figure 6-15 to Figure 6-22 representative fracture morphologies of HTS (lower arm), IMS (lower arm) and HTS_IMS hybrid configuration (HTS-lower arm and IMS-upper arm) at the initiation and steady-state loci in various magnifications are shown. The fracture morphologies at initiation are shown in Figure 6-15 and Figure 6-16, where the Mode I characteristic dark morphology with rough texture is illustrated[1]. Regarding Figure 6-17, and Figure 6-18, the characteristic features found in Mode I specimens such as fibre bridging and textured microflow were observed in all configurations, both at the initiation and steady state area[1].

The most important finding of the fractographic analysis of the Mode I monolithic and hybrid configurations was the evidence of Mixed-Mode fracture features found in the hybrid interface, i.e. cusps (Figure 6-19 and Figure 6-20). Although at higher magnifications some evidence of cusps was also observed in the monolithic configurations, however, the hybrid configurations exhibited a significantly higher amount of cusps. The observed cusps are shown at a higher magnification in Figure 6-21 and Figure 6-22. To obtain these images the specimens were tilted 59°.
Figure 6-15 Typical Mode I fracture surface from (a) HTS and (b) IMS DCB specimens at initiation locus (x20) – (white arrow indicates the growth direction).
Figure 6-16 Typical Mode I fracture surface from (a) Hybrid (lower arm HTS) and (b) Hybrid (upper arm IMS) DCB specimens at initiation locus (x20) – (white arrow indicates the growth direction).
Figure 6-17 Typical Mode I fracture surface from (a) HTS; (b) IMS DCB specimens at steady-state locus (~65 mm – ×200) – (white arrow indicates the growth direction).
Figure 6-18 Typical Mode I fracture surface from (a) Hybrid (lower arm HTS) and (b) Hybrid (upper arm IMS) DCB specimens at steady-state locus (~65 mm – x200) – (white arrow indicates the growth direction).
Figure 6-19 Typical Mode I fracture surface from (a) HTS and (b) IMSDCB specimens at steady-state locus (~65 mm – x800) – (white arrow indicates the growth direction).
Figure 6-20 Typical Mode I fracture surface from (a) Hybrid (lower arm HTS) and (b) Hybrid (upper arm IMS) DCB specimens at steady-state locus (~65 mm – x800) – (white arrow indicates the growth direction).
Figure 6-21 Mode I fracture surface from (a) HTS and (b) IMS DCB specimens at steady-state locus (~65 mm – x2k) – (white arrow indicates the growth direction).
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Figure 6-22 Mode I fracture surface from (a) Hybrid (lower arm HTS) and (b) Hybrid (upper arm IMS) DCB specimens at steady-state locus (~65 mm – x2k) – (white arrow indicates the growth direction).
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As it can be clearly seen, the size, frequency and angle of the cusps observed in the hybrid configuration were higher. In fact this might have been a reason why the Mode I delamination fracture toughness in the hybrid configurations was much higher.

6.4.2 Mode II
A procedure similar to that used for Mode I specimens, was followed for Mode II with the aim to provide an explanation for the increase in the delamination fracture toughness observed in the hybrid configuration in comparison to the two monolithic configurations. Typical fracture morphologies of the interface in the monolithic and hybrid configurations at initiation loci are shown in Figure 6-23 and Figure 6-24, whilst Figure 6-25 to Figure 6-30 illustrate typical fracture morphologies at steady state loci.

At initiation, the characteristic dull and smooth surface morphology of Mode II was observed in all configurations (Figure 6-23 and Figure 6-24). Moreover cusps, the characteristic features of Mode II, were evident throughout the three configurations. As it can be seen in Figure 6-26, Figure 6-27 and Figure 6-28, the amount and size of cusps was higher in the hybrid configuration in comparison to the two monolithic configurations which implies larger fibre spacing and thicker interply resin layer than that in the monolithic configurations[1]. Another feature, which was highlighted in the fractographic analysis, was the irregular arrangement of the fibres in the hybrid configuration. This was mainly attributed to the difference in tow size of the two systems which induced fibre nesting (HTS/MTM44-1 and IMS/MTM44-1). The effect of this phenomenon, on the delamination fracture toughness, will be discussed in the following section. Regarding Figure 6-29 and Figure 6-30 which illustrates the cusps at a higher magnification (x2k), the difference in the density and their size can be clearly seen. The greater amount and larger size of the cusps, which were observed in the hybrid configurations, was due to the higher plastic deformation of the thicker resin layer between the plies, and can be deemed as one of the main reasons the hybrid configurations exhibited higher delamination fracture toughness.
Figure 6-23 Typical Mode II fracture surface from (a) HTS and (b) IMS ELS specimens at initiation area (x20) – (white arrow indicates the growth direction).
Figure 6-24 Typical Mode II fracture surface from (a) Hybrid (lower arm HTS) and (b) Hybrid (upper arm IMS) ELS specimens at initiation area (x20) – (white arrow indicates the growth direction).
Figure 6-25 Typical Mode II fracture surface from (a) HTS and (b) IMS ELS specimens at steady-state area (~65 mm × 200) – (white arrow indicates the growth direction).
Figure 6-26 Typical Mode II fracture surface from (a) Hybrid (lower arm HTS) and (b) Hybrid (upper arm IMS) ELS specimens at steady-state area (~65 mm × 200) – (white arrow indicates the growth direction).
Figure 6-27 Typical Mode II fracture surface from a) HTS and (b) IMS ELS specimens at steady-state area (~65 mm × 800) – (white arrow indicates the growth direction).
Figure 6-28 Typical Mode II fracture surface from (a) Hybrid (lower arm HTS) and (b) Hybrid (upper arm IMS) ELS specimens at steady-state area (~65 mm – x800) – (white arrow indicates the growth direction).
Figure 6-29 Mode II fracture surface from (a) HTS and (b) IMS ELS specimens at steady-state area (~65 mm – x2k) – (white arrow indicates the growth direction).
Figure 6-30 Mode II fracture surface from (a) Hybrid (lower arm HTS) and (b) Hybrid (upper arm IMS) ELS specimens at steady-state area (~65 mm – x2k) – (white arrow indicates the growth direction).
6.4.3 Mixed-Mode I/II (75% I)

In order to explain the large increase in the delamination fracture toughness in the hybrid configurations, Scanning Electron Microscopy was employed to examine the fracture morphologies of the interfaces in the monolithic and hybrid configurations. The fracture morphologies of the three configurations were investigated both at initiation (Figure 6-31 and Figure 6-32) and at the steady state sites (Figure 6-33 to Figure 6-38).

The fracture morphologies of the three different configurations at initiation were, as expected, characterised by a combination of Mode I and Mode II features. That is, a combination of Mode I features (limited fibre bridging) and Mode II features (shallow cusps) was observed. Similar features were observed in the three configurations, although it was clear that the fracture surface in the hybrid configuration was less smooth and contained more fibre bridging. This may justify the higher Mode I and Mode II components observed in the hybrid configuration results. At higher magnifications and at steady state locus, the larger amount of Mode I (fibre bridging) and Mode II features (cusps) supported the enhanced toughness over the monolithic materials (Figure 6-34 and Figure 6-36).

Finally, in Figure 6-37 and Figure 6-38 the shape of the cusps, which were observed in the three different configurations, is shown at a larger magnification. Clearly the density of the cusps in the hybrid interfaces was higher compared to both monolithic configurations. However, the cusps which were observed in the two monolithic configurations were of a different shape. Although in the HTS configurations the cusps were larger than the cusps observed in the IMS configurations, the density was significantly lower. This was mainly dictated by the spacing which was different due to the fibre tow size difference between the two materials. The variation in the density and thickness of the cusps which was observed in the three different configurations supported the observed difference in the Mode II components shown in the results section.
Figure 6-31 Typical Mixed-Mode I/II fracture surface from (a) HTS and (b) IMS MMB specimens at initiation area (×20).
Figure 6-32 Typical Mixed-Mode I/II fracture surface from (a) Hybrid (lower arm HTS) and (b) Hybrid (upper arm IMS) MMB specimens at initiation area (x20).
Figure 6-33 Typical Mixed-Mode I/II fracture surface from (a) HTS and (b) IMS MMB specimens at steady-state area (~65 mm – x200).
Figure 6-34 Typical Mixed-Mode I/II fracture surface from (a) Hybrid (lower arm HTS) and (b) Hybrid (upper arm IMS) MMB specimens at steady-state area (~65 mm – x200).
Figure 6-35 Typical Mixed-Mode I/II fracture surface from (a) HTS and (b) IMS MMB specimens at steady-state area (~65 mm – x800).
Figure 6-36 Typical Mixed-Mode I/II fracture surface from (a) Hybrid (lower arm HTS) and (b) Hybrid (upper arm IMS) MMB specimens at steady-state area (≈65 mm – x800).
Figure 6-37 Typical Mixed-Mode I/II fracture surface from (a) HTS and (b) IMS MMB specimens at steady-state area (~65 mm – x2k).
Figure 6-38 Typical Mixed-Mode I/II fracture surface from (a) Hybrid (lower arm HTS) and (b) Hybrid (upper arm IMS) MMB specimens at steady-state area (~65 mm – x2k).
6.5 Hybrid Interface Characterisation

The large improvement of the delamination fracture toughness, which was observed in the hybrid configurations in the previous section, was mainly attributed to the higher amount of fibre bridging and the formation of cusps in Mode I, the larger shear cusps in Mode II and the higher amount of fibre bridging and shear cusps in Mixed-Mode I/II specimens. However, it was thought that these may have not been the only reason for such an improvement.

In the previous section, high residual stresses were reported in the hybrid laminates (Figure 5-21), which were much higher than the residual stresses observed in the HTS and IMS configurations (Figure 5-22). This, in addition to the different tow size of the two systems (HTS/MTM44-1 and IMS/MTM44-1), may have also contributed to the large increase of the fracture toughness in the hybrid configurations. To assess the contribution of the residual stresses and the arrangement of the plies in the hybrid configurations, cross-sections from fracture toughness specimens were obtained and compared to the monolithic configurations as shown in Figure 6-39.

In the hybrid configuration (IMS/HTS₄/IMS//HTS/IMS/HTS₂/IMS/HTS) the individual plies (HTS and IMS) could be distinguished whereas in the two monolithic configurations it is very difficult to distinguish individual plies. However as it was expected, an increased amount of fibre nesting between nominally identical plies (for a given monolithic configuration) was observed as well as a large number of resin-rich pockets (Figure 6-40).

A resin rich layer was observed across the hybrid interface which might be one of the reasons (Figure 6-40) for the high fracture toughness of the hybrid configurations rather than increased fibre nesting. In Figure 6-39b it can be seen that the hybrid interfaces were not flat as but exhibited an undulated shape across the plies width. On the contrary, monolithic interfaces within the hybrid configurations were more flat such as the two interfaces in the upper arm between nominally identical HTS plies. The undulated shape of the hybrid
interfaces implied that the actual length of the interface might have been larger than the width of the specimens (25mm).

![Image of polished sections](image1)

**Figure 6-39** Representative polished sections perpendicular to the fibre direction of (a) HTS, (b) HTS/IMS Hybrid and (c) IMS delamination fracture toughness specimens (×10).

![Image of ply interfaces](image2)

**Figure 6-40** Ply interfaces at (a) HTS, (b) HTS/IMS Hybrid and (c) IMS configurations, perpendicular to the fibre direction (×10).

To assess the actual length of the hybrid interfaces and compare it against the interfaces of the monolithic configurations, Engauge Digitiser v4.1 was employed, supplied by SourceForge[241]. This software provided the ability to digitize the cross-section as shown in Figure 6-41 and transform it to a series of x and y points in a Cartesian coordinate system. However, to establish that the software was able to provide reliable values, the force-displacement curve of HTS.IMS_A configuration was used as trial data. As it can be
seen in Figure 6-42 the curve which was obtained by the software was quite similar compared to the actual curve. Hence the software was deemed overall reliable for this study.

![Figure 6-41 Digitisation of HTS/IMS Hybrid interface and data acquisition.](image)

To obtain the actual length of the hybrid and any other interface the interface was discretised as shown in Figure 6-43. Each of the red circular points represented each of the (x,y) points generated by the software (Figure 6-41), while the length between two points was treated as linear and constant.

![Figure 6-42 Benchmark establishment for the utilisation of Engauge Digitiser.](image)

The results from the estimated values of the interfaces in the three different configurations are shown in Table 6-4 where values for the interfaces at the midplane and elsewhere across the specimen thickness are tabulated. The estimated length of the hybrid
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interfaces and therefore the width of the plies within the hybrid configurations was higher than the width of the monolithic configurations by approximately 5%. Moreover, the actual width of the plies, in both monolithic and hybrid configurations, were slightly higher than the total the specimen width, 25 mm, which was the value that was used across this chapter to calculate the fracture toughnesses from Mode I, Mode II and Mixed-Mode I/II tests.

Figure 6-43 Interface discretisation and length estimation.

For the estimation of the actual length of any interface the following expressions were used:

\[
\Delta x_i = x_{i+1} - x_i \quad \text{Equation 6-1}
\]

\[
\Delta x_i = l_i \cos \theta \quad \text{Equation 6-2}
\]

\[
\theta_i = a \tan \left( \frac{y_{i+1} - y_i}{\Delta x_i} \right) \quad \text{Equation 6-3}
\]

\[
L = \sum_{i=1}^{N-1} \frac{\Delta x_i}{\cos \left( a \tan \left( \frac{y_{i+1} - y_i}{\Delta x_i} \right) \right)} \quad \text{Equation 6-4}
\]

Given that the specimen width \( b \) is inversely proportional to the delamination fracture toughness in all expressions presented in Section 3.2.2, the actual delamination
fracture toughnesses in Mode I, Mode II and Mixed-Mode I/II should be lower than those obtained by testing. Such an observation provides an explanation of the large increase of the delamination fracture toughness in Mode I, Mode II and Mixed Mode I/II. Since the hybrid interface was undulated it suggests that the crack plane may not have been aligned with local principal axes, which would effectively induce shear. This may have been the reason for the formation of cusps in the Mode I hybrid specimens and the increased amount of cusps in Mode II and Mixed Mode I/II in comparison to the monolithic configurations. Finally, it should be noted that such an undulated shape was not observed in the longitudinal direction, i.e. parallel to the fibre direction.

<table>
<thead>
<tr>
<th>Interface</th>
<th>Midplane (mm)</th>
<th>Elsewhere (mm)</th>
</tr>
</thead>
<tbody>
<tr>
<td>HTS/HTS</td>
<td>25.1 ± 0.1</td>
<td>25.1 ± 0.1</td>
</tr>
<tr>
<td>HTS/IMS</td>
<td>26.4 ± 0.2</td>
<td>26.3 ± 0.1</td>
</tr>
<tr>
<td>IMS/IMS</td>
<td>25.2 ± 0.1</td>
<td>25.2 ± 0.1</td>
</tr>
</tbody>
</table>

Table 6-4 Estimated values of interface length.
Chapter 7 – Discussion

7.1 Compressive Failure of Multidirectional Composites
The first of the studies conducted throughout this work was described in Chapter 4 where results from the compact compression testing of cross-ply and multidirectional laminates were presented. Subsequently, the dominant failure mechanisms were identified and the sequence of failure events in the various configurations were suggested in the light of the literature review (Chapter 2) and based on the observations from optical and scanning electron microscopy conducted on several nominally identical specimens. The outcomes from Chapter 4 demonstrated that the layup had greatly influenced the compressive performance of the four different cross-ply and multidirectional configurations, made of IM7/8552 prepreg tape and tested in compact compression. In fact, it was observed that the layup also influenced the sequence of failure mechanisms which had led to catastrophic failure of the different laminates. Such an observation provides scope for layup optimisation to enhance the compressive performance of multidirectional composite laminates.

Prior to the comparison between different configurations, nominally identical cross-ply configurations were tested to investigate the inherent variability. The experimental study indicated that the two nominally identical configurations behaved in a similar manner with a minor difference in the failure load (approximately 3% - Figure 4-2) and experienced similar axial and shear strains just prior to failure (Figure 4-3). Albeit similar failure mechanisms were identified in both nominally identical specimens the fracture morphologies differed (Figure 4-22 and Figure 4-23) implying that the failure process may have been different too. Nevertheless, a consistency in the locations of the key failure mechanisms (delamination and in-plane shear fracture) was observed. Furthermore, the severity of the post-failure damage was different in the two $(90/0)_{ss}$ cross-ply configurations, especially at the notch where the fracture had been “older”. Such observations in nominally identical cross-ply specimens are indicative of the highly complex nature of the compression failure process, which is subject to several factors such as specimen and notch geometry as well as
loading conditions, to name a few. In fact, even though a similar pattern is observed in terms of type and location of dominant failure modes, a difference in the extent of the particular failure modes and the amount of post-failure damage across the various specimens should be expected.

Regarding the cross-ply configurations, \((90/0)_{8S}\) and \((0/90)_{8S}\) exhibited different compressive performance. In particular, the baseline configuration \((90/0)_{8S}\) was less compliant and the failure load was approximately 6% higher than that of the \((0/90)_{8S}\) configuration. Even though the DIC data did not reveal any significant differences in the axial strains, the shear strains in the baseline \((90/0)_{8S}\) configuration were higher indicating that the two configurations experienced different shear strain fields just prior to failure (Figure 4-5). This of course was not captured by the CLT, which predicted identical shear moduli (Table 4-2). With regards to the failure process, as fractographic analysis revealed the failure initiation was associated with different failure modes in the two configurations. In the baseline cross-ply configuration \((90/0)_{8S}\), interlaminar fracture occurred first (Figure 4-8 and Figure 4-9) which consequently changed the stress state (shed and redistributed the stresses) in the material inducing two in-plane shear fractures (Figure 4-10). Although the theoretical analysis did not highlight any significant difference in the stress distribution (Figure 4-28), larger interlaminar stresses (both Mode I and Mode II) than those of the \((0/90)_{8S}\) configuration were predicted by ABAQUS model (cohesive zone), implying that the \((90/0)_{8S}\) configuration was more prone to delamination (Figure 4-36 and Figure 4-37). In fact, the fractographic analysis suggested that delamination had occurred prior to any in-plane shear fracture and therefore confirmed the outcomes of the numerical analysis.

On the contrary, in the \((0/90)_{8S}\) configuration in-plane shear fracture occurred prior to delamination (Figure 4-11 and Figure 4-12). This observation suggests that the interlaminar stresses were not high enough to cause delamination but instead the shear strength had been exceeded and induced in-plane shear fracture instead (Figure 4-13 and Figure 4-14). The numerical analysis (Figure 4-36 and Figure 4-37), confirmed that the magnitude of the
interlaminar stresses (both Mode I and Mode II) were much lower than those of the $(90/0)_8$ configuration for a given compressive load (-3.5 kN). Moreover, in this configuration the surface plies were composed of $0^\circ$ load-bearing fibres which lacked lateral support on the surface and at the notch. This, in addition to the fact that $0^\circ$ plies bore most of the compressive load, suggests that longitudinal splitting (most likely at the notch) had occurred first, which had induced in-plane shear fracture (and consequently delamination). Even though ply splitting may have formed prior to in-plane shear fracture, it was in-plane shear fracture which had triggered the failure process since ply splitting cannot directly cause unstable fracture\[1\]. That is, ply splitting promotes in-plane microbuckling from the notch which is a stable process, whilst delamination induces out-of-plane buckling which is an unstable process. Furthermore, it was observed that in both cross-ply configurations the fracture was characterised by the formation of an in-plane shear fracture which was angled at $53 \pm 2^\circ$ with respect to the loading direction, confirming the observations in the literature even though these being for uniaxial compression\[48,49,52\]. The findings from the optical and scanning electron microscopy in the two cross-ply configurations indicate the importance of the fractographic analysis in the process of identifying the sequence of failure events, which would otherwise be difficult given the limitations of the theoretical predictions, as described previously (Section 4.2.4).

With respect to the multidirectional configurations, albeit the laminates had 25% less $0^\circ$ plies and thus less load-bearing fibres parallel to the loading direction, the failure loads exhibited by the $(0/90/45/-45)_4$ and $(-45/45/0/90)_4$ multidirectional configurations were approximately 20% higher compared to those of the cross-ply configurations (Figure 4-1). This improvement in the in-plane shear performance is thought to have been due to the incorporation of the angle plies, which enhanced the effective shear stiffness (Table 4-2). It is also possible that this improvement is also related to the specimen geometry since the specimen was loaded via the pins and therefore a moment was applied at the crack tip (Figure 3-1). In fact, the theoretical analysis showed that the incorporation of shear load in
addition to the compressive load led to a large increase in the stresses in the off-axis plies (Figure 4-30), while ABAQUS models had indicated that the multidirectional configurations experienced higher shear stresses. Regarding failure, in the multidirectional configurations more complex failure processes were observed than those of the cross-ply configurations which made the interpretation of the damage propagation more arduous. Indeed the post-failure analysis indicated that the effect of the interlaminar fractures on the compressive behaviour was even more critical than in the cross-ply configurations. This was anticipated since in multidirectional configurations both Poisson mismatch and shear-extension coupling would have been present between the plies and thus higher interlaminar stresses could have been induced (Figure 4-36 and Figure 4-37). On the contrary, the residual stresses were much lower in the multidirectional configurations (Figure 4-29) and therefore their contribution to the laminate stress state had been limited. The same applies to the in-situ ply thickness and specimen thickness, which was found to be consistent across the four configurations (within statistical error) and therefore should not have influenced the failure processes (Table 4-1, Appendix A and Appendix B).

The two multidirectional configurations, (0/90/45/-45)_{4S} and (-45/45/0/90)_{4S}, exhibited relatively similar compressive responses (Figure 4-1). Nevertheless, the location of the angle plies in the two configurations yielded different strain distributions, indicating that the layup could have influenced the strain fields, which was also highlighted by the theoretical analysis (Figure 4-28 and Figure 4-30). The fracture morphology in both multidirectional laminates was different, however the through-thickness translaminar fractures were consistently angled at 53 ± 2° as observed in the cross-ply configurations [48,49,52] (Figure 4-15 and Figure 4-19). In the baseline multidirectional configuration (0/90/45/-45)_{4S}, in-plane shear fracture had occurred prior to any other failure mechanism. This consequently induced delaminations (45°/-45° ply interface) as Prabhakar and Waas[85] had suggested and further in-plane shear fractures which propagated in the reverse direction. The formation of these secondary damages along with further interlaminar damage and ply splits led to the global fracture of
this multidirectional configuration (Figure 4-15, Figure 4-17 and Figure 4-18). At the 0°/±45° ply interfaces which had not delaminated, a characteristic step-like morphology was observed[1]. This step-like fracture was the result of the interaction between the ply splitting of the off-axis plies with the translaminar fracture of the 0° load-bearing plies, as Potter[68] and Pinnell[72] had noted. Moreover, in the surface plies, longitudinal splits were observed implying that matrix cracking had formed early due to the lack of the support of the surface plies and the high stress concentration at the notch. However, judging from the location of the primary in-plane shear fracture and the fact that ply split cannot cause unstable out-of-plane microbuckling fracture, it is thought that although longitudinal splitting had occurred very early, it did not contribute as much to the fracture propagation as delamination. The formation of the in-plane shear fracture prior to any interlaminar failure, implies that the shear strength had been exceeded before the interlaminar strength (Figure 4-36 and Figure 4-37).

In the (-45/45/0/90)_4S multidirectional configuration the fracture initiated due to delamination and the damage was more severe which made the identification of the initial fracture event difficult but interlaminar fracture had occurred prior to any through-thickness translaminar fracture. These delaminations separated the laminate into three sub-laminates which performed independently and led to a large loss of stability. In fact, the fractographic analysis noted that the dominant delamination had occurred at the 90°/-45° ply interface as Pinnell[72] and Purslow[73] had suggested rather than at the 45°/-45° ply interface (Prabhakar and Waas[85]). The formation of delamination prior to any in-plane shear fracture was confirmed by the numerical analysis which noted that the interlaminar stresses were much higher than those of the (0/90/45/-45)_4S multidirectional configuration where in-plane shear fracture had occurred prior to delamination. Finally, as both electron and optical microscopy revealed, the post-failure damage was significant in both cases, especially in the vicinity of the notch where the fracture had been “older” (Figure 4-19 and Figure 4-26).
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In addition, in this study numerical analysis was employed to compare against the experimental results and assess the effectiveness of ABAQUS to model the compressive failure of multidirectional CC specimens. In the initial approach, where the ply interfaces had not been modelled, the compressive strength of both cross-ply and multidirectional laminates was over predicted by approximately 40% (Figure 4-34). Once the ply interfaces were modelled using cohesive zone elements and delamination was taken in account, the models approximated the experimental results, particularly for the (90/0)8S cross-ply laminate. In these models two failure criteria were employed, Hashin (the only failure criterion integrated in ABAQUS v6.10) and 2dVUMAT that was incorporated in the code via the VUMAT subroutine[203].

Even though in the cross-ply configuration the approximation was acceptable and the large drop in the stiffness was predicted (Figure 4-32), the predicted drop by both Hashin and 2dVUMAT was much larger than that observed in the experiment (Figure 4-1) indicating that the numerical analysis predicted greater damage in the load-bearing plies. In fact, in the models this large drop in stiffness coincided with the dominant delamination. On the contrary, in the experiment after the key delamination had occurred the resulting sub-laminates were still able to bear compressive load until they had failed. Moreover, to assess the sensitivity of the model to the input mechanical properties, the set of data reported in WWFE II [52,151] were used in addition to the dataset provided by Hexcel[150]. Clearly, the use of the mechanical properties suggested in WWFE II did not improve the prediction neither for the stiffness not the failure load compared to the model which was fed by the data set supplied by Hexcel (Figure 4-33). What is more is that no discrepancy in the dominant failure modes (type and location) was observed in the two models.

With regards to the multidirectional laminate (0/90/45/-45)4s the 2dVUMAT model provided a more realistic prediction of the compressive performance. Even though it did not accurately predict the failure load (neither did the Hashin failure criterion), the predicted force-displacement curve was very similar to the experimental curve (Figure 4-34). However
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after delamination had occurred a large drop in the stiffness was observed. On the contrary, in the experiment the multidirectional laminate was able to withstand further compressive load. Moreover, it should also be noted that also in this case the use of values obtained by a robin-round test and reported in WWFE II [52,151], did not improve the prediction of both stiffness and failure load compared to the compression behaviour predicted using the data set supplied by Hexcel[150]. Therefore, several issues were identified from the numerical analysis. Initially, these models could not accurately predict the compressive strength, which could be attributed to either the inability of the failure criteria used in the analysis to realistically model the failure process or more likely the use of mechanical properties which did not correspond to the actual values of IM7/8552. Additionally, the post-failure behaviour predicted by the models differed significantly; essentially the models (irrespective the failure criterion used) consistently predicted a large drop in the stiffness. As it was shown in the experimental study, particularly in multidirectional laminates, this was not the case. However it should be noted that in the numerical study the main focus was the failure initiation rather than the propagation, given that only half the laminate had been modelled. This could certainly account for the collapse of the numerical model, since the conditions at the midplane would have been violated.

In this study the dependence of the predicted compressive performance on the chosen mesh was also investigated. In particular, three different meshes were employed for the model which utilised the Hashin failure criterion (Figure 4-38). From the comparison of the ability of these models to predict the compressive performance of a (0/90/45/-45)AS, it was noted that a very coarse mesh cannot accurately predict either the failure load or the post-failure behaviour. The replacement of the data set (Hexcel) with that reported in WWFE II did not improve prediction neither of the stiffness but approximated better the failure load (Figure 4-39). As soon as the mesh was refined, especially in the vicinity of the notch (area of interest), a better approximation of the compressive strength was achieved. However, all three models could not approximate either the stiffness obtained by the experimental study
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or the post-failure analysis. Although more accurate stress and displacements were obtained in the vicinity of the notch, the computational time increased by approximately 150%. Finally, in the case of the 2dVUMAT model, the output variable SDV16 which corresponded to the $I_x/I_{x,\text{max}}$ (where $I_x$ is the characteristic element length) indicated that the mesh which had been chosen was correct (value way below unity) and thus no refinement of the mesh was conducted (Figure 4-40).

To summarise, in this study some very important observations were made on the compressive failure of multidirectional laminates. It was noted that the layup influenced the compressive performance and particularly the dominant failure modes and the sequence of events which led to global failure. Such observations indicate that the layup greatly influences the conditions under which failure occurs and particularly the magnitude of the various stresses (such as interlaminar and shear) that compete under a given compression load and specimen configuration.

In light of the knowledge acquired in this study (especially the fractographic analysis observations), a generic sequence of events which is likely to lead to catastrophic fracture is suggested for the widely used ($\pm 45/0/90)_s$ family of laminates utilised in high performance applications. Albeit this sequence may be applicable to other layups, further studies are required. The flow chart presented in Figure 7-1 illustrates the suggested sequence of the failure mechanisms which are likely to cause global failure of a ($\pm 45/0/90)_s$ multidirectional composite laminate in different scenarios. These scenarios depend on key factors such as the compressive load application, the premature formation of delamination and the presence of a stress raiser.

Prior to describing the suggested failure process for the various scenarios, a preliminary criterion which is related to whether pure compressive load can be applied onto the multidirectional composite laminate is used. Such a criterion is considered important since if pure compressive load cannot be ensured, multiple delaminations occur and the
laminate buckles out-of-plane. Although this scenario is not very likely to occur in coupon-sized specimens like the CC specimens, it can possibly occur in thin-walled composite elements and structures where antibuckling guides have not been utilised[1].

To aid the visualisation of the suggested compressive failure process as this is presented in Figure 7-1 and ease the reader, the two most likely to occur scenarios are shown in Figure 7-2 and Figure 7-3, as a series of images using a generic side-notch rectangular specimen and a (±45/0/90)_s multidirectional layup. In particular, Figure 7-2 illustrates the scenario where in-plane shear fracture occurs prior to delamination. Upon compressive loading, at approximately 60% of the compressive strength[1], longitudinal splits form at the notch, while as the in-plane shear stresses increase, off-axis ply splits also occur at the notch (Figure 7-2a).

As the load further increases, longitudinal splits will form (Figure 7-2b) and at the site of the off-axis splits the load-bearing fibres will buckle in-plane (Figure 7-2c). The load is then shed and redistributed and further longitudinal ply splitting is induced (Figure 7-2d). This will cause further microbuckling which will lead to the formation of the saw-tooth fracture morphology (Figure 7-2e). As the load-bearing fibres continue to fail, the compression crack propagates along the notch (Figure 7-2f). Consequently, the load is shed onto the adjacent off-axis plies and off-axis ply splitting will occur (Figure 7-2g). These off-axis plies will then fail by in-plane shear, inducing the saw-tooth fracture morphology (Figure 7-2h). Upon increased load, delamination is likely to occur at a ply interface adjacent to a load-bearing ply (Figure 7-2i), which will cause unstable failure, i.e. out-of-plane microbuckling (Figure 7-2j). Eventually, the compressive cracking will reach the free edge and the specimen will fail globally. This will cause extensive additional delamination and therefore post failure damage (Figure 7-2k and Figure 7-2l).
Figure 7-1 Suggested sequence of events during compressive failure of multidirectional fibre-reinforced composite laminates.
Figure 7-2 Suggested failure sequence where in-plane fracture occurs prior to delamination.
Figure 7-3 Suggested failure sequence where delamination occurs prior to in-plane fracture.
The second scenario, shown in Figure 7-3, illustrates the sequence of failure events in the case where delamination occurs first. As in the previous scenario, tangentially to the stress raiser longitudinal ply splits will occur due to the high stress concentration at the notch and poor transverse strength and off-axis ply splitting due to high in-plane shear stresses (Figure 7-3a and Figure 7-3b). Upon increased load, the critical strain release energy rate is likely to be exceeded prior to the in-plane shear strength and hence to cause delamination to occur most likely at a 45°/45° ply interface (Figure 7-3c). The delamination will split the laminate into two sub-laminates which will perform independently. As the load is shed and redistributed (Figure 7-3d) the two sub-laminates can fail by in-plane shear or by delamination, depending on the stress state in each sub-laminate. From Figure 7-3e to Figure 7-3k, the sequence of events which lead to the failure of one of the sub-laminates is shown (as described in the previous scenario), whilst for simplicity the failure process of the other sub-laminate is not shown. Note that it is the delamination which leads to the global failure (causing out-of-plane microbuckling).

The ability to predict the first failure event for a given multidirectional configuration using the knowledge acquired in this study is essential in order to engineer improved compressive performance. In fact, knowing in advance under what conditions delamination and in-plane shear fracture occur gives the ability to choose the appropriate layup and material system which would yield the optimum compressive performance. This can also be beneficial for the numerical analysis of multidirectional composite laminates since for a given layup and loading conditions, the analysis could be focused on modelling more realistically the dominant failure mechanism and thus make it less laborious and save considerable computational time.

Furthermore, to engineer the performance of composites under particular loading conditions, apart from the effect of the layup and material selection, notch and specimen geometry as well as load application, the effect of laminate size and scale effects on the compressive strength should also be taken into account, since the design of large-scale
composite structures is usually based on data obtained from coupon-sized specimens. The ability to predict the effect of laminate size and scale on the compressive performance would be a leap forward in the direction of reducing the cost of large-scale experiments and expanding the use of fibre-reinforced in primary aerospace structures. Even though there are several studies in the literature which have investigate the effect of size and scale on the compressive strength of fibre-reinforced composites[209], only a few focus on how the particular failure mechanisms as well as failure process are affected in compression[78,210-214]. In the most notable studies, those of Soutis[78,211,212] and Wisnom[213,214], different effects were noted due to specimen geometry (notched or unnotched), thickness scaling (ply level or sub-laminate), and notch geometry (hole diameter). Moreover, it has been highlighted that the different failure modes occurring in compression, fibre microbuckling, ply splitting and delamination, are affected by scaling in different ways.

Scaling of compressive strength in fibre-reinforced composites is essential and it should be taken in account when designing composite structures. However, before accounting for laminate size and scale effects, it is important to know in advance the dominant failure mode (and the potential failure process) under particular conditions (load application and specimen geometry), which can be deduced using the flowchart shown in Figure 7-1. The identification of the dominant failure mode is required to accurately predict the effect of the laminate size and scale effect. Once the particular dominant failure mode has been identified, then the respective approach (such as Weibull Theory or Linear Elastic Fracture Mechanics[209]) which accounts for the size and scaling effects can be employed. Finally, it is suggested that when accounting for laminate size and scale effects, other factors should also be considered. In particular, there is no evidence in the literature whether manufacturing has an effect on the scaling of compressive strength, i.e. what is the fibre waviness, void percentage, thermal stresses in thicker laminates, factors which can greatly influence the compressive performance.
7.2 Hybrid Composites

7.2.1 Compression Testing

In this study the effect of hybridisation of multidirectional composite laminates on the compressive performance and failure process was investigated. The chosen approach was to compare the compressive performance of two monolithic carbon fibre/epoxy systems with that of their respective hybrids. These two material systems were CYTEC HTS/MTM44-1 and IMS/MTM44-1[157] in the form of prepreg tape with nominal thickness 0.250 mm (0.246 ± 0.011 mm and 0.249 ± 0.012 mm experimentally calculated – see Appendix D). Four (0/90/45/-45)\textsuperscript{2}S configurations were obtained for this study namely HTS, HTS\_IMS\_A, HTS\_IMS\_O and IMS, where HTS and IMS were the monolithic configurations and HTS\_IMS\_A and HTS\_IMS\_O were the hybrid configurations. The HTS was deemed the baseline monolithic configuration.

For this work three compression tests were employed: compact compression (CC), plain compression (PC) and sandwich panel compression. Initially, compact compression was utilised because of the reported stable crack growth and the absence of global buckling as well as to compare against the performance of the (0/90/45/-45)\textsuperscript{4}S multidirectional configuration made of IM7/8552 presented in Chapter 4.

The results from the compact compression test suggested that hybridisation led to an improvement of the compressive performance compared to that of the baseline monolithic configuration HTS (Figure 5-1). The incorporation of ±45\degree IMS plies in the HTS baseline configuration (HTS\_IMS\_A) moderately increased the stress capability of the 0\degree load-bearing plies by approximately 1\% yet led to the enhanced of the overall compressive performance approximately 8\% for strength and 20\% stiffness, due to the improvement of the in-plane shear support on the HTS 0\degree load-bearing plies and the delamination fracture toughness (Table 7-1). The incorporation of 0\degree and 90\degree IMS plies in the HTS configuration (HTS\_IMS\_O) significantly improved the stress capability of the 0\degree load-bearing plies by.
Discussion

approximately 40%, and also led to the enhancement of the overall compressive performance (approximately 18% for the strength and 17% for the stiffness). This should be attributed to the incorporation of the much stiffer IMS 0° load-bearing plies (Table 3-1) as well as the improvement of the delamination fracture toughness of the hybrid interfaces.

<table>
<thead>
<tr>
<th>Layup</th>
<th>Fibre Type - 0° Plies</th>
<th>Stress (MPa)</th>
<th>Difference (%)</th>
</tr>
</thead>
<tbody>
<tr>
<td>HTS</td>
<td>HTS</td>
<td>-580</td>
<td>N/A</td>
</tr>
<tr>
<td>HTS.IMS_A</td>
<td>HTS</td>
<td>-587</td>
<td>1%</td>
</tr>
<tr>
<td>HTS.IMS_O</td>
<td>IMS</td>
<td>-821</td>
<td>41%</td>
</tr>
<tr>
<td>IMS</td>
<td>IMS</td>
<td>-825</td>
<td>N/A</td>
</tr>
</tbody>
</table>

Table 7-1 Compressive stress at the 0° load-bearing plies of CC specimens just prior to failure based on CLT.

Moreover, while the average compressive strength of the HTS.IMS_A configurations lied within that of the two monolithic configurations, as the rule of mixtures would suggest, the HTS.IMS_O was well beyond it. On the contrary, the compliance of the HTS.IMS_O configuration was in agreement with the rule of mixtures, whereas the stiffness of the HTS.IMS_A configuration was higher than the theoretical predictions (Table 5-2). Such discrepancies should be expected since the theoretical predictions are purely based on linear elastic analysis where no interlaminar effects are considered. In addition, contrary to the representative load-displacement curve of the (0/90/45/-45)_4S multidirectional configuration made of IM7/8552 (Figure 4-1), in these four multidirectional configurations no abrupt drop in the stiffness was observed after failure (Figure 5-1), i.e. these configurations were able to withstand higher loads for larger displacements implying that they exhibited higher delamination fracture toughness.

In light of the experimental results, DIC and theoretical analysis, the fractographic analysis of the four multidirectional configurations revealed that delamination and in-plane shear fracture were the dominant failure modes and that extensive ply splitting occurred at the notch. In fact, the location of the dominant delaminations and the failure process differed,
implying that the layup, and thus hybridisation, influenced the stress distribution during compressive failure. Furthermore, less overall damage was observed in the fracture morphologies of these four representative configurations both at the notch and away from the notch compared to the representative fracture morphology of the $(0/90/45/-45)_4S$ multidirectional configuration made of IM7/8552 (Chapter 4). This was due to the fact that delamination was the dominant failure mode and thus limited sliding of the fractured surfaces occurred. This shift in the dominant failure mode, from in-plane shear fracture to delamination, which explains the deterioration of compression strength compared to the IM7/8552 configuration (Table 4-1 and Table 5-1), could be attributed to the difference in the ply thickness (100% increase), a factor which has been highlighted by Wisnom[214] to promote delamination prior to any other failure mode due to blunting of the stress concentration at the notch.

Finally, it was also observed that the residual stresses (as obtained by LAP) were much higher (approximately 60%) especially in the hybrid configurations, than the residual stresses reported in the $(0/90/45/-45)_4S$ multidirectional configuration. This indicated that these additional stresses played a significant role in the compressive performance of these four multidirectional configurations and may have also contributed to degradation of the compressive strength.

X-Ray radiography noted extensive longitudinal and off-axis ply splitting tangentially to the notch (Figure 5-5) in all four configurations, confirming the observations in the $(0/90)_8S$ and $(0/90/45/-45)_4S$ configurations made of IM7/8552. In fact, it was suggested that ply splitting was the first failure mode to occur irrespective of the layup due to the lack of the support on the surface and the high stress concentration at the notch. However, ply splitting does not directly cause unstable out-of-plane compression microbuckling (like delamination does) but can induce or interact with other failure modes such as translaminar fibre fracture and delamination[1]. Moreover, X-Ray radiography highlighted the formation of the step-like fractures at both $0^\circ$ and $\pm 45^\circ$ plies which acted as initiation sites for the in-plane shear
fractures of the ±45° and 0° plies respectively[1], similar to those observed by Potter[68] and Pinnell[72]. Eventually, X-Ray radiography suggested that the major delaminations formed at 90°/45° and 45°/-45° ply interfaces (Figure 5-5).

Subsequently, optical and scanning electron microscopy suggested that delamination was the dominant failure mode (on the contrary to the study presented in Chapter 4) and that hybridisation influenced the location of these key delaminations. Regarding the monolithic configurations, it was observed that the dominant delaminations in the HTS baseline configuration occurred at 45°/-45° and 90°/45° ply interfaces (Figure 5-5, Figure 5-6, Figure 5-14 and Figure 5-16), whilst in the IMS configuration the dominant delaminations occurred at 90°/45° and -45°/0° ply interfaces (Figure 5-5, Figure 5-12, Figure 5-15 and Figure 5-17). The location of the dominant delaminations were confirmed by the numerical analysis (Figure 5-28 and Figure 5-29), in accordance to the observations made by Pinnell[72], Purslow[73] and Prabhakar and Waas[85]. In light of the fractographic analysis of the two monolithic systems (namely HTS and IMS), it was observed that in the HTS_IMS_O hybrid configuration the dominant delaminations had occurred at non-hybrid interfaces (0°/90° and 45°/-45° ply interfaces – Figure 5-6) whereas in the HTS_IMS_A configuration delaminations had occurred at hybrid interfaces (90°/45° ply interface Figure 5-8). Moreover, in the HTS_IMS_A configuration, as soon as the hybrid interface had failed, multiple delaminations occurred (mainly in 45°/-45° non-hybrid ply interfaces) while in the HTS_IMS_O configuration in-plane shear fracture was induced instead. This suggested that the fracture toughness of the hybrid 90°/45° and -45°/0° ply interfaces, where 0°, 90° plies were of IMS and ±45 plies were of HTS, exhibited higher delamination fracture toughness than the same hybrid interfaces where 0°, 90° plies were of HTS and ±45 plies were of IMS.

Albeit the effect of residual stresses on hybridisation was also taken in account, the difference in the magnitude between the two hybrid configurations was negligible and thus it is thought that their contribution to the compressive performance was similar (Figure 5-22).
In addition, whilst the numerical analysis (Figure 5-28 and Figure 5-29) accurately predicted the location of the dominant delaminations for the monolithic configurations, the prediction of the dominant delaminations did not correspond to the fractographic observations for the hybrid configurations, indicating that the delamination fracture toughness values used for the cohesive zone may have not corresponded to the actual values. This could be attributed to the fact that these delamination fracture toughness values (Mode I and Mode II) were obtained for unidirectional and not multidirectional ply interfaces (Chapter 6). Such discrepancy may have been larger in the case of hybrid interfaces, despite the fact that the delamination fracture toughness values of the monolithic configurations were obtained in the same manner. Note that essentially in multidirectional ply interfaces fibre nesting cannot form and thus the interply thickness is higher which leads to higher delamination fracture toughness[88,215].

Even though the compact compression test yielded very interesting findings, the complex specimen geometry might have also had an effect on the apparent compressive performance of such multidirectional laminates. It is thought that the loading via the pins had not been directly applied to the notch but instead a moment was applied. This was evident by the extensive off-axis ply splitting which had formed in the sharp notch due to the large shear stresses in the ±45° plies. Furthermore, due to the complex geometry of the CC specimen and the load application via the pins (Figure 3-1), large tensile stresses were evident at the free edge. The effect of the CC geometry and the applied shear was also demonstrated by the theoretical analysis. Whilst a 115% degradation of the compressive strength was noted compared to the unnotched compressive strength (pure compression), the compressive strength was decreased by an additional 80% when shear was also applied (Table 5-3). Considering these reasons, plain compression (PC) was employed to alleviate these effects and further investigate how the load application and specimen geometry influenced the compressive performance.
In plain compression, the four $(0/90/45/-45)_2s$ multidirectional configurations ($HTS$, $HTS_{IMS\_A}$, $HTS_{IMS\_O}$ and $IMS$) exhibited different compressive performance compared to that observed in the compact compression test, that is the ranking differed. All four configurations exhibited similar force-displacement curves (although different compliances - Table 5-5) which were all characterised by an abrupt drop in the stiffness after the failure load had been reached (Figure 5-30). Hybridisation in this instance led to the degradation of the compressive performance. In particular, the monolithic HTS baseline configuration exhibited the highest strength but also the largest scatter from all four configurations. Similar to the compact compression, the incorporation of $\pm 45^\circ$ plies of IMS in the HTS baseline configuration slightly increased the stress capability of the $0^\circ$ load-bearing plies (Table 7-2), nevertheless it led to a 3% lower overall compressive strength, which should be attributed to the deterioration of the in-plane shear support on the HTS $0^\circ$ load-bearing plies. Ditto, the incorporation of $0^\circ$ and $90^\circ$ plies of IMS in the HTS configuration significantly enhanced the stress capability of the $0^\circ$ load-bearing plies by approximately 40% yet deteriorated the overall compressive strength by approximately 9%. Such a discrepancy could be caused by the deterioration of the in-plane shear support on the $0^\circ$ load-bearing plies or the delamination fracture toughness due to hybridisation.

<table>
<thead>
<tr>
<th>Layup</th>
<th>Fibre Type - $0^\circ$ Ply</th>
<th>Stress (MPa)</th>
<th>Difference (%)</th>
</tr>
</thead>
<tbody>
<tr>
<td>$HTS$</td>
<td>HTS</td>
<td>-978</td>
<td>N/A</td>
</tr>
<tr>
<td>$HTS_{IMS_A}$</td>
<td>HTS</td>
<td>-980</td>
<td>&lt;1%</td>
</tr>
<tr>
<td>$HTS_{IMS_O}$</td>
<td>IMS</td>
<td>-1396</td>
<td>42%</td>
</tr>
<tr>
<td>$IMS$</td>
<td>IMS</td>
<td>-1392</td>
<td>N/A</td>
</tr>
</tbody>
</table>

Table 7-2 Compressive stress of the $0^\circ$ load-bearing plies of PC specimens just prior to failure.

These results were not in accordance with the Classical Laminate Theory based analysis where it was suggested that the hybrid configurations exhibited superior compressive performance (Table 5-3). Essentially, according to the theoretical analysis the
HTS configuration should have exhibited the lowest compressive performance due to the less strong and stiff fibres in comparison to the IMS as well as their hybrids as the rule of mixtures would suggest (Table 5-6). The discrepancy between the stress capability of the load-bearing plies and the overall compressive strength indicates that other factors such as delamination fracture toughness and in-plane shear, played a significant role. In fact, in the Classical Laminate Theory based analysis the contribution of the interlaminar stresses was not considered which is thought to have greatly influenced the compressive performance, especially of the hybrid configurations, let alone the notch and free edge effects.

With regards to the fractographic analysis, initially X-Ray radiography noted that in the vicinity of the notch longitudinal and off-axis ply splitting had formed tangentially to the notch due to the low transverse and shear strength of the 0\(^\circ\) load bearing plies (longitudinal splits) as well as the large shear stresses in the ±45\(^\circ\) plies (off-axis splits). In this case the length of the longitudinal and off-axis ply splits in the plain compression specimens was shorter than that observed in the compact compression specimen, which can be attributed to the less sharp notch and the absence of shear load (introduced by the pins). Furthermore, the characteristic saw-tooth (or step-like) morphology of the compressive failure was observed both in the 0\(^\circ\) load-bearing plies and the ±45\(^\circ\) off-axis plies, however with irregular step size (Figure 5-34), implying that there was an interaction between the off-axis and longitudinal ply splitting with the translaminar failure of the 0\(^\circ\) load-bearing plies and the in-plane shear fracture of the ±45\(^\circ\) off-axis plies respectively.

Contrary to compact compression, the dynamic nature of the plain compression test and the higher compressive strengths led to fracture morphologies having greater amount of post-failure damage both in the vicinity of the notch and away from the notch. This made the interpretation of the fracture morphology arduous. Considering the fracture morphologies and the stress distribution prior to failure obtained by DIC, it can be suggested that the failure was unstable and short in duration where multiple delaminations where evident.
Delamination was also in this instance the dominant failure mechanisms which triggered the failure process in all four configurations. In both monolithic configurations, HTS and IMS, delamination in the 0°/90° ply interface was dominant, however these delaminations consequently induced different failure modes (Figure 5-35, Figure 5-36, Figure 5-41 and Figure 5-42). These observations were not in accordance to those made in the compact compression or those made by Pinnell[72], Purslow[73], indicating that the load application and the specimen geometry influenced the failure processes.

Regarding the hybrid configurations, the dominant delaminations in the HTS_IMS_A which exhibited superior compressive performance (compared to HTS_IMS_O) occurred at a non-hybrid interface (45°/-45° and 0°/90° – Figure 5-37 and Figure 5-38); whilst in the HTS_IMS_O the dominant delamination occurred at a hybrid interface (-45°/0° – Figure 5-39 and Figure 5-40). In fact, this delamination at a hybrid interface had a greater effect on the compressive performance than a delamination at a non-hybrid ply interface. In particular, in this hybrid interface, the surrounding material developed higher strain energy just prior to the formation of the delamination. As soon as the delamination had formed, the stress state changed dramatically and led to higher local instability, which implies that the hybrid interfaces exhibited higher delamination fracture toughness than the non-hybrid ply interfaces. This important observation led to the study on the delamination fracture toughness of hybrid interfaces, presented in Chapter 6. It should be noted here that no dominant delaminations were observed at 90°/45° ply interfaces, an interface where most dominant delaminations had occurred in the compact compression configurations (both monolithic and hybrid). This is indicative of the change in the stress state that hybridisation and load application had caused in the laminate.

To further investigate the influence of the load application and the specimen geometry on the compressive performance of these hybrid configurations, sandwich panel compression testing was carried out. In this test, four sandwich panels made of aluminium honeycomb core and skins made of HTS, HTS_IMS_A, HTS_IMS_O and IMS (0/90/45/-45)
configurations respectively, were manufactured (Chapter 3). The compressive performance of these four sandwich panels was in accordance with the results obtained by the plain compression test, but in this instance the-HTS\_IMS\_A configuration exhibited the highest strength and was followed by HTS, HTS\_IMS\_O and IMS configurations (Figure 5-49). As Table 7-3 shows, the incorporation of ±45° plies of IMS in the HTS baseline configuration slightly increased the stress capability of the HTS 0° load-bearing plies but improved the overall compressive strength by approximately 5%.

<table>
<thead>
<tr>
<th>Layup</th>
<th>Fibre Type - 0° Ply</th>
<th>Stress (MPa)</th>
<th>Difference (%)</th>
</tr>
</thead>
<tbody>
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<tr>
<td>HTS_IMS_A</td>
<td>HTS</td>
<td>-978</td>
<td>&lt;1%</td>
</tr>
<tr>
<td>HTS_IMS_O</td>
<td>IMS</td>
<td>-1394</td>
<td>43%</td>
</tr>
<tr>
<td>IMS</td>
<td>IMS</td>
<td>-1396</td>
<td>N/A</td>
</tr>
</tbody>
</table>

Table 7-3 Compressive stress of the 0° load-bearing plies of sandwich panel skins just prior to failure.

On the contrary, the incorporation of 0° and 90° plies of IMS substantially improved the stress capability of the IMS 0° load-bearing plies (approximately 40%) but deteriorated and the overall compressive strength by approximately 10%. Note that due to the high manufacturing cost of these sandwich panels, only one panel was tested per configuration and thus the results need to be considered with caution and may not be representative of the true behaviour.

With respect to the post-failure analysis of the sandwich panels, only X-Ray radiography was employed to investigate the fracture of the front skins and noted that in the vicinity of the notch, longitudinal and off-axis ply splitting was observed in accordance to the observations made in the compact and plain compression specimens. Even though the nature of the sandwich panel compression was dynamic, the delaminations which were observed on the front skins (as well as post-failure damage) were limited, contrary to compact and plain compression where delamination had been more extensive. The limited
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delamination should be attributed to the constraint that the aluminium honeycomb core offered to the skins, which effectively constrained the front skin from buckling out-of-plane.

In addition, the compression crack propagation was recorded using high-speed video (Figure 5-56) and it was noted that the higher the compressive strength the higher the crack speed and the higher the load the higher the strain energy. Moreover, considering the DIC data (Figure 5-52), just prior to failure it was also observed that the configurations which were at high strain state exhibited high crack propagation speed implying that the crack propagation speed was somehow related to the dramatic change in the stress state as soon as the failure had occurred. Although such a relation has been noted for tension, there is no evidence in the literature that such a relation exists in compression[5]. However, further studies are required to refine the observations on the relation between the crack propagation speed, and strain energy release during failure as well as the compressive strength.

At this point, a comparison on the overall effect of hybridisation on the 0° load-bearing plies stress capability and the laminate strength between the three compression tests can be made. A useful approach could be to compare the change in the 0° load-bearing plies stress capability and the laminate strength for both monolithic materials, HTS and IMS, contrary to the discussion previously, with the aim to highlight the overall effect of hybridisation on both the monolithic materials. Considering the results presented previously (Table 7-1, Table 7-2 and Table 7-3), an overall comparison is made in Figure 7-4 and Figure 7-5 which show the relative change in the stress capability of the 0° load-bearing plies and overall laminate strength of the two monolithic configurations, when the off-axis plies (noted HTS_±45 and IMS_±45) and cross plies (noted as HTS_0/90 and IMS_0/90) are replaced.

The plot presented in Figure 7-4 indicates that hybridisation has a significant positive effect on the stress capability of the 0° load-bearing plies, when stiffer and stronger 0° load-bearing plies are introduced in the layup. This of course should be expected since the overall
compression capability of the load-bearing plies is enhanced, yet this seems to be the case irrespective of the compression test and specimen/element geometry.

![Figure 7-4 Comparison on the effect of hybridisation on the \(0^\circ\) load-bearing plies stress capability between the three compression tests.](image)

On the contrary, the effect of hybridisation on the laminate strength of monolithic configurations seems to be more complex, that is there is no a clear trend among the various hybrid configurations across the various compression tests (Figure 7-5). The results shown in Figure 7-5 are indicative of the hybrid effect that was clearly demonstrated in this work with the means of experimental, theoretical and fractographic analysis. In particular, the following plot suggests that the replacement of particular plies \((0^\circ, 90^\circ\) or \(\pm 45^\circ\)), not only can have a positive or negative effect on the overall laminate strength but that it is also highly sensitive to the compression test and specimen geometry. Furthermore, considering Figure 7-4 and Figure 7-5, it is also apparent that factors other than the \(0^\circ\) load-bearing capability greatly influence the compression behaviour, such as in-plane shear, load and specimen geometry as well as delamination fracture toughness of the hybrid and non-hybrid interfaces. The latter is discussed in the following section.
Figure 7-5 Comparison on the effect of hybridisation on the overall laminate strength between the three compression tests.

7.2.2 Delamination Fracture Toughness

The effect of hybridisation on the compressive performance of multidirectional laminates which was highlighted in the compression tests, could neither be explained by the simplistic theoretical formulations nor the numerical analysis (albeit having been based on experimentally determined delamination fracture toughness values). In fact, it was observed that it was not the compressive strength and stiffness of the fibres alone which influenced the compressive performance (Table 3-1). This was evident especially in the plain compression test where the HTS configuration exhibited the highest strength even though the compression strength and stiffness of the HTS plies was much lower than that of the IMS. The fractographic analysis revealed that delamination, especially at the hybrid interfaces, was the key factor which influenced the compressive performance. However, since no literature was available on the effect of hybridisation (the same epoxy resin-different fibre types) on the compressive performance of multidirectional laminates and no reliable delamination fracture toughness values for the two systems used in this work (and of course
Discussion

A study on the delamination fracture toughness of these systems and their hybrids was conducted.

Three tests were employed to investigate the delamination resistance of these systems, Mode I (DCB), Mode II (ELS) and Mixed-Mode (75% Mode I) on three unidirectional configurations (HTS)\textsubscript{16}, (IMS)\textsubscript{16} and (IMS/HTS\textsubscript{4}/IMS/HTS/IMS/HTS\textsubscript{2}/IMS/HTS). In all three tests, a large increase in the delamination fracture toughness was observed in the hybrid configurations compared to the monolithic configurations. In addition, the scatter in all configurations and tests was limited (<8%) indicating a consistency in the delamination fracture toughness of the particular configurations and successful execution of these tests. In Mode I fracture toughness testing (DCB) of the hybrid configuration where the improvement was approximately 30% (with respect to the monolithic configurations - Table 6-1), a higher degree of fibre bridging and formation of cusps were observed (Figure 6-1, Figure 6-2, Figure 6-3 and Figure 6-4).

In Mode II fracture toughness test (ELS), the hybridisation also improved the delamination fracture toughness by approximately 40% (Table 6-2) and the fracture surfaces were characterised mainly by a larger amount of cusps with higher frequency than those observed in the monolithic configurations (Figure 6-5, Figure 6-6, Figure 6-7 and Figure 6-8). Finally, in Mixed-Mode I/II (75% Mode I) testing (MMB), the enhancement of the delamination fracture toughness was very large; approximately double the values of the monolithic configurations (Table 6-3). The fracture morphologies of the hybrid interfaces were characterised by a large amount of fibre bridging and cusps and the frequency of these cusps was greater than that observed in the HTS and IMS configurations (Figure 6-9, Figure 6-10, Figure 6-11 and Figure 6-12).

The features which were observed in fractographic analysis of the hybrid interfaces and contributed to the large increase of the delamination fracture toughness in Mode I, Mode II and Mixed-Mode I/II, can be grouped into two main categories. The first category
comprises those features induced by manufacturing such as resin-rich pockets and fibre nesting. The former was caused during the curing process and mainly affected the formation of the cusps, especially in the hybrid interfaces where excessive interply matrix was observed. Essentially, in the locations where resin-rich pockets were present, a few large cusps were observed[1]. Although these areas were limited in the monolithic configurations, the frequency of such large cusps was slightly higher in the hybrid configurations. The latter was observed in all three tests (Mode I and Mixed-Mode I/II) while in the hybrid interfaces the amount of cusps was higher. Since the IMS fibres had double the tow size compared to the HTS fibres, increased intermingling had been expected in the hybrid interfaces. However, the optical microscopy revealed that fibre nesting at the hybrid ply interfaces was not as extensive as it had been expected (Figure 6-40). Instead a resin layer was observed between the plies in areas where fibre nesting was not evident (one to two HTS fibre diameter thicknesses shown in Figure 6-40). It is thought that the features described above may have contributed to the increase of the delamination fracture toughness (in Mode I, Mode II and Mixed-Mode I/II), however the improvement due to these factors should be attributed to an inherent material effect [1].

On the contrary, in the second category there are features which contributed to the enhancement of the delamination fracture toughness in the hybrid interfaces due to a mechanism and not a manufacturing artefact, namely fibre bridging and cusps. The former is probably the most important feature of Mode I delamination (and Mixed Mode I/II). Fibre bridging essentially modifies the process zone ahead of the delamination crack tip and thus increases the delamination resistance. Hence the higher amount of fibre bridging observed in the hybrid interfaces compared to the non-hybrid interfaces also contributed to the large increase in delamination fracture toughness. However, it has been reported that fibre bridging is promoted by fibre nesting[1] and thus the increased amount of fibre bridging, observed in the hybrid interfaces, could mainly be attributed to fibre nesting due to the difference in size of HTS and IMS tows. Finally, at this point it should be noted that fibre
Discussion

bridging can also be an indication of incorrect processing, i.e. incorrect consolidation conditions[1].

With regards to the second feature, cusps, starting with Mode II delamination where the cusps are the prevalent feature, larger cusps with higher frequency were observed in the hybrid interfaces compared to those observed in the monolithic configurations. The morphology and the higher frequency of the cusps implied that the matrix in the hybrid interface had undergone greater plastic deformation than the matrix in the monolithic configurations, which consequently led to the improvement of the delamination fracture toughness. Similarly in Mixed-Mode I/II, a large amount of cusps was observed which could also be attributed to the larger plastic deformation of the matrix in the hybrid interface (let alone the larger amount of fibre bridging).

The most interesting feature was the formation of cusps in Mode I specimens. A logical explanation would be that accidentally Mode II loading was applied. However, it was ensured throughout the testing procedure that pure Mode I loading was applied onto the specimens, i.e. the loading blocks were bonded properly on the specimen surface. It is important to note that cusps were also observed at non-hybrid interfaces of the monolithic configurations, however they were sparse. In the literature it has been reported that localised cusps can be induced by fibre bridging[1,56,216-219]. These cusps form in the fibre tracks during crack opening as the fibres are pulled from the surfaces and are smaller than those formed in pure Mode II. Locally as the fibres are pulled out, shear is induced in the fibre/matrix interface. Cusp formation has also been reported by Garg in Mode I testing at temperatures close to the curing temperature[56]. Moreover, optical microscopy revealed that the hybrid interfaces had an undulated shape (Figure 6-39), which was not observed in the monolithic configurations. Essentially, due to this undulated shape, most likely induced by the large residual stresses induced during processing, the crack plane was no longer aligned with the global laminate plane and thus shear was induced. Considering the observations from the fractographic analysis of the hybrid and monolithic ply interfaces, it is
suggested that the formation of the cusps at the hybrid ply interfaces is attributed to fibre bridging (the sparse ones), the thick resin layer and most importantly the undulated shape of the interface (due to residual stresses), while the formation of sparse cusps at the monolithic ply interfaces is attributed to fibre bridging. Unfortunately, there are neither studies in the open literature which have employed these materials (HTS/MTM44-1, IMS/MTM44-1 and of course their hybrids) that report similar observations or implications, nor the supplier (CYTEC) has suggested anything that could be related to these observations.
Chapter 8 – Conclusions

The overall aim of this project was to undertake an in-depth investigation of the compressive failure mechanisms which occur in multidirectional composite laminates and consequently suggest novel approaches for improving the compressive performance by arresting/redirection of compressive damage. Firstly, Compact Compression tests were conducted to investigate the compressive performance of multidirectional composite laminates and fractographic analysis was employed to identify the dominant failure mechanisms and deduce the sequence of events which led to global failure, which had not been reported elsewhere in the literature. The identification of the key aspects of compressive failure led to the investigation of material-based approaches which could offer compressive crack arrest/redirection capabilities, such as piezoelectric actuation, carbon nanotube reinforcement and hybridisation. While piezoelectric actuation and carbon nanotube reinforcement were not pursued due to the current immaturity of these technologies and time constraints, hybridisation was extensively studied by means of experimental and fractographic analysis. In the hybridisation study, Compact Compression, Plain Compression and Sandwich Panel Compression were conducted to further understand compressive failure of both monolithic and hybrid laminates under different loading conditions, whilst Mode I, Mode II and Mixed Mode (75% Mode I) tests were carried out to investigate the delamination fracture toughness of both monolithic and hybrid ply interfaces.

The conclusions from this study are the following:

- The compressive performance of the compact compression specimens was greatly influenced by the layup. The $(0/90/45/-45)_4$ and $(-45/45/0/90)_4$ multidirectional configurations exhibited approximately 20% higher compressive strengths and compliances than the $(0/90)_8$ and $(90/0)_8$ cross-ply configurations.

- The enhancement of the compressive performance was attributed to the improvement of the effective shear stiffness of the laminate due the incorporation of
the off-axis plies, since it was observed that in the compact compression specimens a moment was applied at the notch via the loading pins.

- Delamination and in-plane shear fracture were the dominant failure modes throughout the different configurations, whilst longitudinal and off-axis ply splitting in the vicinity of the notch were also evident.

- The formation of delamination or in-plane shear fracture depended on whether the critical strain energy release rate for delamination was exceeded prior or after the in-plane shear strength respectively. In particular, in the (90/0)$_{8S}$ and (-45/45/0/90)$_{4S}$ configurations, delamination triggered the failure, whereas in the (0/90)$_{8S}$ and (0/90/45/-45)$_{4S}$ configurations the failure was triggered by in-plane shear fracture.

- The multidirectional configurations were found to be more prone to delaminations and post-failure damage which was attributed to the Poisson mismatch and shear-extension coupling, while a consistency in the angle of the translaminar shear fracture was observed throughout the four configurations (53 ± 2°). The extent of this translaminar fracture through the thickness greatly depended on the delamination formation, since the two failure modes interacted during the failure propagation.

- The current theoretical formulations and failure criteria cannot accurately predict the compressive performance of multidirectional laminates due to their inability to adequately model the interaction of the key failure modes, which plays an essential role in the failure process, as was highlighted in this study. The comprehension of the key aspects of compressive failure of multidirectional composites was essential in order to improve the compressive performance by suggesting novel compression crack arrest/redirect concepts, such as hybridisation.

- Hybridisation of (0/90/45/-45)$_{2S}$ multidirectional laminates was found to have influenced the compressive performance. The introduction of a second material in the
axial and off-axis plies had both positive and negative effect in the compressive strength. The theoretical and fractographic analysis highlighted that hybridisation changed the stress capability of the $0^\circ$ load-bearing plies, the support of the off-axis plies on the load-bearing plies and the delamination fracture toughness. Whilst the enhancement of the $0^\circ$ load-bearing plies is irrespective of compression test and specimen geometry, the overall laminate strength is highly sensitive to the load and specimen/element geometry as well as in-plane shear and delamination fracture toughness of hybrid and non-hybrid interfaces.

- Specimen and notch geometry as well as load application also influenced the effect of hybridisation, since no consistency in compressive strength trends was observed across compact and plain compression as well as sandwich panel compression.

- The fibre volume fraction and lamina thickness was found to vary approximately 1% across the four configurations (monolithic and hybrid), which was considered insufficient to have significantly contributed to the hybridisation effect.

- Fractographic analysis showed that delamination was the dominant failure mode both in monolithic ($HTS, IMS$) and hybrid ($HTS_{IMS \_A}, HTS_{IMS \_O}$) multidirectional configurations. It was also noted that hybridisation also affected the location of the key delaminations and that delamination at a hybrid interface was more detrimental with respect to the compressive performance than that at a monolithic interface, suggesting that hybrid interfaces exhibited higher delamination fracture toughness.

- Mode I, Mode II and Mixed-Mode I/II delamination fracture toughness tests at unidirectional ply interfaces confirmed the different behaviour of the monolithic and hybrid interfaces, previously suggested by the observations from the fractographic analysis in compact and plain compression specimens. In particular, a significant improvement of the delamination resistance was highlighted in all three modes at the hybrid interfaces.
• Increased fibre bridging, high number of shear cusps and evidence of undulated crack plane induced by the large residual stresses (i.e. inherent material effects) were found to be responsible for this large improvement. Even though a part of this improvement can be regarded as artificial, such as due to the undulated crack plane, a clear improvement of the delamination resistance at the hybrid interfaces was evident.

• Considering the high importance of delamination in the compression failure process and the significant improvement in the delamination resistance offered by hybridisation, it can be said that the introduction of a second material in selective interfaces can have a significant effect in the overall compressive behaviour of multidirectional composite laminates and therefore should be regarded as a concept with a great potential.

• Even though the modification of the layup and the introduction of a second material clearly influenced the compressive performance and delamination resistance of the multidirectional composite laminates in the various compression tests, evidence of compressive fracture crack arrest was not observed in this work. In fact, considering the observations made in this work, due to the highly complex nature of compressive failure (translaminar, intralaminar and interlaminar fracture) and the large instability which accompanies unstable (out-of-plane) compressive failure, to effectively arrest/redirect compressive failure a more sophisticated approach is probably needed to comply with crack growth tolerance. Such an approach should target on both improving the in-plane and out-of-plane performance, and an efficient way to do so could be a combination of concepts in the material as well as structural level.
Chapter 9 – Implications and Recommendations for Future Work

In this section, the implications of this study are presented and recommendations for the future are provided.

9.1 Compressive Failure of Multidirectional Composites

9.1.1 Experimental

Three compression tests were employed in this work to investigate the compressive failure of multidirectional composite laminates and it was noted that the compressive performance was sensitive to layup as well as specimen and notch geometry. Even though a considerable amount of information was acquired from this study and given the dearth of studies on the interpretation of compressive failure of multidirectional laminates, it is suggested that further experiments should be conducted on other types of specimens or elements (with or without notch) and different stress raiser geometries.

In addition, the compressive performance of $(\pm 45/0/90)_m$ laminates made of different types of carbon fibre/epoxy systems or other proportions of the different ply directions (soft and hard layups) should also be investigated to further support the scenarios presented in Figure 7-1. Moreover, regarding hybrid composites, the effect of hybridisation on the compressive failure should also be assessed in other configurations apart from the $(0/90/45/-45)_m$ which was employed in this study and other ply interfaces such as $0^\circ/90^\circ$ and $45^\circ/-45^\circ$. Nevertheless, to better understand the effect of hybridisation on the compressive performance, a thorough study should be conducted in multidirectional hybrid interfaces (Mode I, Mode II and Mixed-Mode I/II) and alleviate the effects of features representative of unidirectional ply interfaces such as fibre bridging. Another interesting idea would be to extent this concept to the hybridisation of material types and architecture types, i.e. hybridising prepreg tapes and woven fabrics of different carbon fibre/epoxy system. Finally, it would also be very important to assess and quantify the effect of the residual stresses on
Implications and Recommendations for Future Work

multidirectional ply interfaces, especially hybrid where the residual stresses are expected to be significantly higher, as noted in Chapter 5.

Finally, as it was made clear in Chapter 7, the scaling of composites is essential for composites design. Given the dearth of studies on this topic, it is suggested that the effect of scale and size on the compressive performance of multidirectional fibre-reinforced composite specimens should further be investigated especially in more complex specimen geometries such as plain and compact compression, with the aid of fractographic analysis. The latter could provide essential information as to how does the failure process is influence by scale and size. Unfortunately, considering all the numerous issues related to the size and scale effects, the high cost and the time required for such a study, the scaling of composites in compression was not investigated in this study.

9.1.2 Fractographic Analysis

The fractographic analysis in this study provided essential information about the dominant failure modes and how their interaction influenced the failure process. Even though post-failure techniques such as X-Ray radiography, optical and scanning electron microscopy are very powerful without which the interpretation of compressive failure would be very difficult, all utilise fractured specimens. Furthermore, given that during compressive failure extensive post-failure damage is induced due to the sliding of the failed surfaces over each other, the correct failure interpretation depends on the experience of the operator and thus is potentially subject to errors.

Although interrupted tests would be very useful in order to obtain the sequence of events which lead to failure, this is not practical in the techniques employed in this study. Regarding X-Ray radiography, if the specimen has not failed the penetrant ingression is difficult, whereas for electron and optical microscopy the specimen has to be dissected and the fracture surfaces may need to be separated, which is very difficult for a specimen that has not failed catastrophically. In addition, the limitation of X-Ray radiography to provide
Implications and Recommendations for Future Work

Information about the depth or the location of the damage through the laminate thickness, could be overcome by using stereo pairs or X-Ray Tomography, with the latter being able to provide much more information. Another alternative would be to test specimens using the in-situ SEM fixture employed by Gutkin[43], however specimens such as the CC are large for such a facility. Finally, in this study due to time constraints, only one sandwich panel per configuration was tested and on those no scanning electron and optical microscopy were employed for the fractographic analysis of the sandwich panels. Therefore, representative failure modes and sequence of events for each configuration could not be obtained as obtained for CC and PC configurations. Even though similar failure modes to those observed in plain compression were noted (ply splits at the notch and delaminations), further fractographic analysis is required.

9.1.3 Numerical Analysis
In addition to the experimental studies and fractographic analysis, ABAQUS was also employed to compare against the experimental results and assess the ability of FEA to accurately predict the compressive performance of multidirectional configurations. Even though, ABAQUS (Hashin and 2dVUMAT) approximated the strength of the various laminates, the prediction of the compliance (especially of the multidirectional configurations) and the sequence of the failure events were not accurate. This discrepancy should be attributed to the fact that probably the mechanical properties fed into the models did not correspond to the actual values and that the delamination fracture toughness values were obtained in unidirectional and not multidirectional ply interfaces. Therefore, there is a need to obtain more realistic mechanical properties for the IM7/8552, HTS/MTM44-1 and IMS/MTM44-1 systems than those reported by the suppliers, and an apt way to do that is to conduct a round-robin test. Another reason for such a discrepancy could also be the inability of the current failure criteria to accurately model the compressive performance of multidirectional laminates. However, the novel observations made in this project could be
utilized to formulate a more realistic failure criterion for the compressive failure of multidirectional composite laminates.

In addition, the numerical models of the various plain compression and sandwich panel compression configurations, which were not built in this study, could also provide essential information about the ability of ABAQUS to model the compressive performance especially of less complex geometries than the compact compression (i.e. plain and sandwich panel compression). Moreover, the potential formation of longitudinal and off-axis ply splitting at the notch was not modelled because it would make the modelling even more laborious and the simulation more expensive as well as due to the time constraints. However, in ABAQUS the modelling of ply splitting can be achieved using the XFEM (eXtended Finite Element Method) capability which would be very useful in order to compare against the standard FEM method and assess the effect of ply splitting on the failure process from a numerical analysis point of view.

9.2 Compression Failure Crack Arrest
As it was noted in this study, compressive failure of multidirectional composites is a very complex process, highly sensitive to various factors such as material and layup selection, specimen and notch geometry as well as load application and manufacturing artefacts. The formation of the dominant failure mechanisms such as delamination, fibre microbuckling and ply splitting, depends on the laminate stress state. Prior to failure, the various stress components are in equilibrium, but as soon as one of these components exceeds its critical value, compressive failure occurs. Therefore, to suggest a compressive crack arrest approach a series of factors need to be taken in account. It is important to note that due to the high complexity of the compressive failure, a unique feature may probably not offer crack arrest capability. This is because in composite laminates, compressive failure occurs due to a combination of interlaminar, translaminar and intralaminar fracture. So far, including this study, crack arrest concepts have been suggested for the individual types of fracture. Given the complexity of compressive failure, the most efficient approach would probably be to
employ a combination of novel concepts which would offer crack arrest capabilities against interlaminar failure (which causes unstable out-of-plane failure) as well as translaminar and intralaminar failure (stable in-plane failure). In fact, this has been the idea of the CRASHCOMPS project. To date, the four Crack Arrest and Self-Healing Exploratory Themes and two Core Themes of the CRASHCOMPS project, have suggested novel concepts and have produced a considerable amount of information on the respective fields[11-13,101-103,220-226]. The attempt to combine compressive failure crack arrest with self-healing is in progress, and it is thought that in a later stage of the CRASHCOMPS project the efficient synergy between crack arrest and self-healing will be reality.
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Finally, the author would like to acknowledge the constant and generous support of his beloved family, his friends Anastassia, Dan, Jonny, Ollie and Sumana as well as his colleagues Abeed, Carla, Ed, Geremy, Jovan, Leon and Liz during the beautiful and difficult moments of this work.
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Appendix A – Estimation of In-situ Lamina Thickness of the IM7/8552 System

According to Hexcel, the supplier of the IM7/8552 system, the nominal thickness of the laminate is 0.125 mm. However, after curing the actual (in-situ) lamina thickness (i.e. the lamina thickness within the laminate) may not be identical to the nominal [150]. For that reason, Nikon NIS Elements BR was utilised to determine experimentally the actual lamina thickness. This study was conducted at Swerea SICOMP AB, Sweden.

To obtain the in-situ lamina thickness, polished sections of Compact Compression specimens from the four different layups were used as input in the software. Since this study was conducted in a different microscopy system than the one from which the micrographs were acquired (Olympus BHM – see Chapter 3), to achieve accurate estimations, the embedded information (e.g. magnification of the micrographs, working distance etc.) were used to calibrate this system. Once the accuracy of the calibration was verified, using an object of known dimensions (supplied and calibrated by Nikon), the assessment of the actual ply thickness of the four layups was conducted. In this study, as ply thickness was regarded the distance between the middle of two consecutive interply resin layers (Figure A-1).

![Figure A-1 Definition of ply thickness at a (0/90/45/-45)_8S CC specimen (x50).](image)

This was considered by the author to be more representative than the distance between fibre boundaries. These measurements were taken from twenty locations across the cross
Representative images of the output from the Nikon NIS Elements BR as given in (Figure A-2) and (Figure A-3), while the obtained thicknesses of the four layups are shown in (Table A-1).

Figure A-2 Representative in-situ ply thickness estimation at a (0/90)$_{as}$ CC specimen (x5).

Figure A-3 Representative in-situ ply thickness estimation at a (0/90/45/-45)$_{as}$ CC specimen (x10).
Note that the thickness values tabulated in Table A-1 represent the average values of the lamina ply thicknesses which were obtained from the cross sections of two CC specimens of each layup.

<table>
<thead>
<tr>
<th>Layup</th>
<th>In-situ Lamina Thickness (mm)</th>
</tr>
</thead>
<tbody>
<tr>
<td>(0/90)$_{ss}$</td>
<td>0.121 ± 0.09</td>
</tr>
<tr>
<td>(90/0)$_{ss}$</td>
<td>0.124 ± 0.07</td>
</tr>
<tr>
<td>(0/90/45/-45)$_{4s}$</td>
<td>0.123 ± 0.09</td>
</tr>
<tr>
<td>(-45/45/0/90)$_{4s}$</td>
<td>0.123 ± 0.05</td>
</tr>
</tbody>
</table>

Table A-1 Average in-situ ply thickness values of the various layups based on twenty measurements.
Appendix B – Calculation of Fibre Volume Fraction in IM7/8552 System

The supplier of the IM7/8552 pre-preg, Hexcel, suggest in their data sheet that the volume fraction is 60% [150]. According to Hexcel, this system is manufactured with low-viscosity “zero-bleed” epoxy resin, which means that the actual fibre volume fraction is expected to be 60%. Moreover, the low viscosity of such a resin system results in exceptionally low porosity within the laminate. This study was conducted at Swerea SICOMP AB, Sweden.

The most often used methods to obtain the actual fibre volume fraction in a carbon-fibre reinforced composite laminate are chemical digestion (ASTM D3171) or resin burn-off (mainly used in glass fibre reinforced composites – ASTM D2584). However, these methods provide estimation of overall volume fractions in the laminate (i.e. fibre and resin as well as void content) rather than local to the local site. For this work it was thought that a subtler method was needed, which would focus on the volume fraction of the constituents in the vicinity of failure. Such a method was considered by the author more applicable to this particular work, rather than using the methods suggested in the literature. It should be noted though that optical microscopy is generally not recommended for the estimation of void content.

To obtain the actual ply fibre volume fraction and compare it against the value suggested by Hexcel, Nikon NIS Elements BR was employed. This software, which uses images acquired by optical microscopy, is based on light intensity thresholding to detect and count objects of interest (namely object count). In other words, NIS Elements BR translates a micrograph into areas of different light intensity and counts the respective area fractions. In that way, for a carbon fibre/epoxy system, where the fibre and resin exhibit different light intensities in a polished section, their volume fractions can be obtained. Since the method is based on the micrographs and their light intensity, the smoothness of the polished surfaces is essential. Moreover, a downside of this method is that artefacts, such as fibre chipping,
can the results. Therefore in order to avoid any interference in the estimations from
contamination or local surface imperfections, the estimations were carried out at specific
areas of the overall captured where the polishing quality was high. In that way the fibre
volume fraction estimation was more accurate. Such areas where the software could
distinguish the different phases are marked with white dotted lines.

The approach taken in this study was to use micrographs depicting plies in the
vicinity of damage, to provide a more realistic estimation of the fibre volume fraction in
undamaged plies (over fifteen fibre diameters from the damaged area). This was conducted
for all four laminates, \((0/90)_{ss}\), \((90/0)_{ss}\), \((0/90/45/-45)_{4S}\) and \((-45/45/0/90)_{4S}\). Representative
micrographs, before and after the application of intensity thresholding (object count) are
shown in (Figure B-1) and (Figure B-2) while the results are tabulated in (Table B-1). Note
that the fibre volume fraction values reported below are based on measurements on twenty
locations.

![Figure B-1](image)

**Figure B-1** (a) Original optical micrograph; (b) Optical micrograph after light intensity
thresholding at a \((0/90)_{ss}\) CC specimen \((\times 10)\).

In Figure B-1 and Figure B-2 the green and red colours correspond to the colours
which the software used to mask the selected areas of different light intensity in each ply, in
this case green for resin and red for fibres. Note that the intensity thresholding and thus the
object selection were manually controlled by the author, in order to mask (select) more
precisely the fibres and the resin. Considering the results (Table B-1), the fibre content at
Appendix B

Ply level in the different laminates was close to the value reported by Hexcel [Ref] and a relative consistency was observed among the different layups.

Figure B-2 (a) Original optical micrograph; (b) Optical micrograph after light intensity thresholding at a (0/90/45/-45)_{4S} CC specimen (x10).

<table>
<thead>
<tr>
<th>Layup</th>
<th>Ply Fibre Volume Fraction (%)</th>
</tr>
</thead>
<tbody>
<tr>
<td>(0/90)_{8S}</td>
<td>59.9 ± 0.5</td>
</tr>
<tr>
<td>(90/0)_{8S}</td>
<td>59.6 ± 0.7</td>
</tr>
<tr>
<td>(0/90/45/-45)_{4S}</td>
<td>59.3 ± 0.9</td>
</tr>
<tr>
<td>(-45/45/0/90)_{4S}</td>
<td>60.0 ± 0.3</td>
</tr>
</tbody>
</table>

Table B-1 Average in-situ fibre volume fraction values of CC specimens.
Digital Image Correlation is a method which has been thoroughly used across different disciplines, such as fracture mechanics, non-destructive evaluation, biomechanics to name a few. The versatility and accuracy of two and three-dimensional image correlation has been under scrutiny for nearly two decades. In composites, several studied have utilised image correlation mainly for structural analysis in the literature and a growing number of studies is employing such a technique since scientists and engineers begin to understand its potential and capabilities[156,193,227-239].

The calibration of a Digital Image Correlation system (in this case GOM Aramis) is a measuring process in which the system is adjusted in such a way to ensure the dimensional consistency of the measuring system. This is achieved with the aid of calibration objects such as the ones shown in Figure C-1. For the ARAMIS measuring system, two different calibration objects are generally used, calibration panels for small measuring volumes and calibration crosses for large measuring volumes. In the tests described in this work (Chapter 3), only calibration panels were used, a small calibration panel for the CC and PC tests (Figure C-1a) and a large calibration panel for the sandwich panel compression tests (Figure C-1b). Note that these calibration objects contain scale bar information, where the scale bar is the specified distance between two specific points.

Figure C-1 Calibration objects used with GOM Aramis DIC system.
Once the appropriate calibration object was chosen for the respective measuring volume (to achieve accurate measurements), the specimen surfaces were prepared according to the GOM’s recommendations. Since Digital Image Correlation is an optical method which allows accurate two and three dimensional measurement of changes in digital images (of specimens during loading) and then translates them into displacement and deformation, the quality of the recorded images is essential. To achieve this, DIC allocates coordinates to the image pixels, where these images usually represent the surface of a structure that has a stochastic speckle pattern (Figure C-2).

This process is done by the use of facets (square or rectangular) where the choice of the appropriate size is a compromise between accuracy and computation time and is also important for the strain computation and visualization. In particular, the default facet size in ARAMIS is 15×15 pixels (with a facet step of 13 pixels and a 2 pixel overlapping area), which allows for accurate strain computation. Indeed for the compact and plain compression tests, a 15×15 pixels facet size was regarded suitable while for the sandwich panel compression a larger size was used, 35×35 pixels, due to the different speckle size. Note that in general a facet size larger than the default (15×15 pixels) results in a more time consuming strain computation.

![Stochastic speckle pattern used in DIC](image)

Figure C-2 Stochastic speckle pattern used in DIC[240].
According to GOM[155,240], this stochastic pattern must stay intact during testing to follow the deformation of the specimen, must also be smooth to ease the pixel (facet) allocation and should not be reflective so good brightness and contrast are achieved. In this work, such a stochastic pattern was achieved using sprays with full cone nozzles, which were able to provide aerosol paint particles of approximately 5-15 μm for the CC and PC specimens and 20-35 μm for the sandwich panels.

After the speckle pattern was painted on the specimen surfaces, the calibration process recommended by GOM was followed[155,240]. In particular, the calibration panel was placed 10 cm in front of the specimen surface (for the CC and PC and 1.5 m for the sandwich panels) and the cameras were positioned at a 25° angle with respect to each other. Then a series of steps were followed, involving rotating and adjusting of the panel with respect to the specimen surface, in order to achieve a correct calibration. In fact, the deviation achieved during the calibration process was between 0.01 and 0.04 pixels, an acceptable range according to GOM recommendations[155,240]. Further details regarding the calibration process can be found in the Aramis manual [155,240].
Appendix D – Estimation of In-situ Lamina Thickness of the HTS/MTM44-1 and IMS/MTM44-1 systems

Along the same lines to the procedure described in Appendix A, the actual laminate thickness of the two systems, HTS/MTM44-1 and IMS/MTM44-1 was determined, using the same software, Nikon NIS Elements BR. Apart from the measured laminate thickness in the two monolithic materials, the average values for the two hybrid systems, HTS_IMS_A and HTS(IMS)O are also presented.

Representative in-situ ply thickness estimation values for the four layups are shown in Figure D-1, Figure D-2, Figure D-3 and Figure D-4 whilst average values of the estimated thicknesses for the four layups are shown in Table D-1. Note that the average ply thickness values shown in Table D-1 have been acquired from both CC and PC specimens of the four layups, to achieve a more realistic estimation. This study was conducted at Swerea SICOMP AB, Sweden.

![Image](image-url)

Figure D-1 Representative in-situ ply thickness estimation at an HTS (0/90/45/-45)₁₀ CC specimen (×20).
Figure D-2 Representative in-situ ply thickness estimation at an HTS.IMS.A \((0/90/45/-45)_{2S}\) CC specimen (x20).

Figure D-3 Representative in-situ ply thickness estimation at an HTS.IMS.O \((0/90/45/-45)_{2S}\) CC specimen (x20).
Table D-1 Average in-situ ply thickness values of CC and PC specimens.

<table>
<thead>
<tr>
<th>Layup</th>
<th>In-situ Lamina Thickness (mm)</th>
</tr>
</thead>
<tbody>
<tr>
<td>HTS</td>
<td>0.246 ± 0.011</td>
</tr>
<tr>
<td>HTS_IMS_A</td>
<td>0.248 ± 0.005</td>
</tr>
<tr>
<td>HTS_IMS_O</td>
<td>0.248 ± 0.008</td>
</tr>
<tr>
<td>IMS</td>
<td>0.249 ± 0.012</td>
</tr>
</tbody>
</table>

Figure D-4 Representative in-situ ply thickness estimation at an IMS (0/90/45/-45)_{2S} CC specimen (×20).
Appendix E – Calculation of Fibre Volume Fraction in HTS/MTM44-1 and IMS/MTM44-1 Systems

The light intensity thresholding (object count) method was used in this instance as well to determine the fibre volume fraction of the tow monolithic materials, HTS and IMS, as well as their respective hybrids, HTS_IMS_A and HTS_IMS_O. Representative micrographs, before and after the application of intensity thresholding (object count) are shown in Figure E-1 to Figure E-4 while the results are tabulated in Figure E-1. Note that the areas marked with white dotted line, are those which the software could distinguish as different phases, leading to more accurate estimations. This study was conducted at Swerea SICOMP AB, Sweden.
Figure E-3 (a) Original optical micrograph; (b) Optical micrograph after light intensity thresholding at an HTS_IMS_O (0/90/45/-45)_{2S} CC specimen (x10).

Figure E-4 (a) Original optical micrograph; (b) Optical micrograph after light intensity thresholding at an IMS (0/90/45/-45)_{2S} CC specimen (x10).

<table>
<thead>
<tr>
<th>Layup</th>
<th>Ply Fibre Volume Fraction (%)</th>
</tr>
</thead>
<tbody>
<tr>
<td>HTS</td>
<td>60.2 ± 0.7</td>
</tr>
<tr>
<td>HTS_IMS_A</td>
<td>59.8 ± 0.8</td>
</tr>
<tr>
<td>HTS_IMS_O</td>
<td>59.9 ± 0.5</td>
</tr>
<tr>
<td>IMS</td>
<td>59.5 ± 0.9</td>
</tr>
</tbody>
</table>

Table E-1 Average in-situ fibre volume fraction values of CC and PC specimens.
Appendix F – Bending Stiffness Estimation

The main criterion to choose the layup of the two arms for the delamination resistance tests was that both arms should exhibit almost identical bending stiffnesses, i.e. $D_{11}$ and then that the overall coupling stiffness matrix was equal to zero, $B_{ij}$. Moreover, the interface between the two arms had to be a hybrid interface, meaning that the bottom ply of the upper arm and the top ply of the lower arm were of different material, i.e. HTS/MTM44-1 and IMS/MTM44-1.

<table>
<thead>
<tr>
<th>Layup</th>
<th>$D_{11}$ (Nmm)</th>
<th>Upper vs. Lower Arm Difference in $D_{11}$ (%)</th>
</tr>
</thead>
<tbody>
<tr>
<td><strong>Upper Arm</strong></td>
<td></td>
<td></td>
</tr>
<tr>
<td>IMS/HTS/IMS</td>
<td>1.046*10^4</td>
<td>N/A</td>
</tr>
<tr>
<td><strong>Lower Arm</strong></td>
<td></td>
<td></td>
</tr>
<tr>
<td>HTS/IMS/HTS/IMS/HTS</td>
<td>5.895*10^3</td>
<td>43.64</td>
</tr>
<tr>
<td>HTS/IMS$_2$/HTS$_2$</td>
<td>5.776*10^3</td>
<td>44.78</td>
</tr>
<tr>
<td>HTS$_3$/IMS/HTS</td>
<td>5.265*10^3</td>
<td>49.67</td>
</tr>
<tr>
<td>HTS/IMS$_2$/HTS/IMS/HTS</td>
<td>1.068*10^4</td>
<td>2.10</td>
</tr>
<tr>
<td>HTS/IMS/HTS$_2$/IMS$_2$/HTS</td>
<td>1.037*10^4</td>
<td>0.85</td>
</tr>
<tr>
<td>HTS$_2$/IMS$_2$/HTS$_2$</td>
<td>9.846*10^3</td>
<td>5.87</td>
</tr>
<tr>
<td>HTS/IMS$_2$/HTS/IMS$_2$/HTS$_2$</td>
<td>1.715*10^4</td>
<td>63.96</td>
</tr>
<tr>
<td>HTS/IMS/HTS$_2$/IMS$_2$/HTS</td>
<td>1.611*10^4</td>
<td>54.02</td>
</tr>
<tr>
<td>HTS$_2$/IMS/HTS$_2$/IMS$_2$/HTS$_2$</td>
<td>1.528*10^4</td>
<td>46.08</td>
</tr>
</tbody>
</table>

Table F-1 Representative values of $D_{11}$ for different lower arm layups.

To cover a wide range of layups, arms with 5, 6 and 7 plies were investigated. While most combinations met one of the criteria, only one met both criteria and was deemed as the most suitable, (IMS/HTS$_4$/IMS//HTS/IMS/HTS$_2$/IMS/HTS). Table F-1 presents an overview of
representative results from the calculations carried out in LAP [162], since the entire set comprised 126 calculations.
Appendix G

Appendix G – Hashin Failure Criterion

The expressions for the four failure modes in the Hashin criterion are the following:

Matrix tension \( \left( \hat{\sigma}_{22} \geq 0 \right) \)
\[
F_m^t = \left( \frac{\hat{\sigma}_{22}}{Y^T} \right)^2 + \left( \frac{\hat{\tau}_{12}}{S^L} \right)^2
\]
Equation G-1

Matrix compression \( \left( \hat{\sigma}_{22} \leq 0 \right) \)
\[
F_m^c = \left( \frac{\hat{\sigma}_{22}}{2S^T} \right)^2 + \left[ \frac{\hat{Y}^C}{2S^T} \right]^2 - 1 \left[ \frac{\hat{\sigma}_{22}}{\hat{Y}^C} + \left( \frac{\hat{\tau}_{12}}{S^L} \right)^2 \right]
\]
Equation G-2

Fibre tension \( \left( \hat{\sigma}_{11} \geq 0 \right) \)
\[
F_f^t = \left( \frac{\hat{\sigma}_{11}}{X^T} \right)^2 + \alpha \left( \frac{\hat{\tau}_{12}}{S^L} \right)^2
\]
Equation G-3

Fibre compression \( \left( \hat{\sigma}_{11} \leq 0 \right) \)
\[
F_f^c = \left( \frac{\hat{\sigma}_{11}}{X^C} \right)^2
\]
Equation G-4

where \( Y^T \) is the transverse compressive strength, \( S^L \) is the longitudinal shear strength, \( S^T \) is the transverse shear strength, \( \hat{Y}^C \) is the transverse compressive strength, \( X^T \) is the longitudinal tensile strength, \( \hat{X}^C \) is the longitudinal compressive strength, \( \alpha \) is the coefficient that determines the contribution of the shear stress to the fibre tensile criterion and \( \hat{\sigma}_{11}, \hat{\sigma}_{22}, \hat{\tau}_{12} \) are the components of the effective tensor \( \hat{\sigma} \) which is used to evaluate the initiation criteria [191].
Appendix H – ABAQUS VUMAT Subroutine

In this code[203], damage initiates under compressive stress when the following expression is satisfied:

\[
\left( \frac{\tau_{\text{eff}}}{S^T} \right)^2 + \left( \frac{\tau_{\text{eff}}}{S^L} \right)^2 \leq 1
\]

Equation HH-1

where \( \tau_{\text{eff}} = \left(-\sigma_{22} \cos a \sin a - \eta^T \cos a \right) \)

Equation H-2

\[ S^T = Y^C \cos a \left( \sin a + \frac{\cos a_0}{\sin 2a_0} \right) \]

Equation H-3

\[ \tau_{\text{eff}} = \left( \cos a |r_{12}| + \eta^T \sigma_{22} \cos a \right) \]

Equation H-4

\[ S^L = S^C \]

Equation H-5

\[ \eta^T = -\frac{1}{\tan 2a_0} \]

Equation H-6

\[ \eta^L = -\frac{S^L \cos a_0}{Y^C \cos^2 2a_0} \]

Equation H-7

In the 2dVUMAT subroutine the following parameters were fed into the model:

<table>
<thead>
<tr>
<th>Parameter</th>
<th>Description</th>
<th>Parameter</th>
<th>Description</th>
</tr>
</thead>
<tbody>
<tr>
<td>( E_{11} )</td>
<td>Young’s modulus in fibre direction</td>
<td>( \beta_3 )</td>
<td>In-plane shear</td>
</tr>
<tr>
<td>( E_{22} )</td>
<td>Young’s modulus in matrix direction</td>
<td>( \beta_4 )</td>
<td>Fibre compression</td>
</tr>
<tr>
<td>( v_{12} )</td>
<td>Major Poisson’s ratio</td>
<td>( \beta_5 )</td>
<td>Matrix compression</td>
</tr>
<tr>
<td>( G_{12\text{init}} )</td>
<td>Initial in-plane shear modulus</td>
<td>( G_{1CT} )</td>
<td>Fibre tension</td>
</tr>
<tr>
<td>( G_{12\text{ult}} )</td>
<td>Ultimate in-plane shear modulus</td>
<td>( G_{2CT} )</td>
<td>Matrix tension</td>
</tr>
<tr>
<td>( \beta_1 )</td>
<td>Fibre tension</td>
<td>( f_{12\text{act}} )</td>
<td>In-plane shear</td>
</tr>
<tr>
<td></td>
<td></td>
<td></td>
<td>((\gamma_{\text{ult}} = f_{12\text{act}}Y_C)) can be set to zero</td>
</tr>
<tr>
<td>( \beta_2 )</td>
<td>Matrix tension</td>
<td>( G_{1CC} )</td>
<td>Fibre compression</td>
</tr>
<tr>
<td>Parameter</td>
<td>Description</td>
<td>Symbol</td>
<td>Description</td>
</tr>
<tr>
<td>-----------</td>
<td>---------------------------------------------------------</td>
<td>--------</td>
<td>---------------------------------------------------------</td>
</tr>
<tr>
<td>$G_{2CC}$</td>
<td>Matrix compression</td>
<td></td>
<td>Matrix compression post failure strength</td>
</tr>
<tr>
<td>$X_T$</td>
<td>Fibre tension</td>
<td></td>
<td>In-plane shear non-linearity parameter</td>
</tr>
<tr>
<td>$Y_T$</td>
<td>Matrix tension</td>
<td></td>
<td>In-plane shear rate dependency factor</td>
</tr>
<tr>
<td>$S_C$</td>
<td>In-plane shear</td>
<td></td>
<td>In-plane shear stress where non-linear behaviour starts</td>
</tr>
<tr>
<td>$X_C$</td>
<td>Fibre compression</td>
<td></td>
<td>Failure flag</td>
</tr>
<tr>
<td>$Y_C$</td>
<td>Matrix compression</td>
<td></td>
<td>Maximum damage (fibre tension always one)</td>
</tr>
<tr>
<td>$F_{interact}$</td>
<td>$1 = S_{12} - S_{12}^*$ interaction, $0 = no$</td>
<td></td>
<td>Shear or matrix strain for material point deletion</td>
</tr>
<tr>
<td>$X_{CF}$</td>
<td>Fibre compression post failure strength</td>
<td></td>
<td></td>
</tr>
</tbody>
</table>

**Table H-1 Input parameters for 2dVUMAT code.**
Appendix I – Cohesive Zone

In ABAQUS the available traction-separation model assumes initially linear elastic behaviour which is followed by damage initiation and evolution. The elastic behaviour assumed by this model which relates the stresses and strains is given by Equation I-1[191]:

\[
t = \begin{bmatrix} t_n \\ t_s \\ t_t \end{bmatrix} = \begin{bmatrix} K_{nn} & K_{ns} & K_{nt} \\ K_{ns} & K_{ss} & K_{st} \\ K_{nt} & K_{st} & K_{tt} \end{bmatrix} \begin{bmatrix} \varepsilon_n \\ \varepsilon_s \\ \varepsilon_t \end{bmatrix} = K\varepsilon \tag{Equation I-1}
\]

Where \( t_n, t_s, t_t \) are the three components of the nominal traction stress vector which represent the normal (3-direction) and the shear tractions (1,2-directions) and \( \varepsilon_n, \varepsilon_s, \varepsilon_t \) are the nominal strains that relate to the corresponding separations, \( \delta_n, \delta_s, \delta_t \), from the following expressions[191]:

\[
\varepsilon_n = \frac{\delta_n}{T_o} \tag{Equation I-2}
\]
\[
\varepsilon_s = \frac{\delta_s}{T_o} \tag{Equation I-3}
\]
\[
\varepsilon_t = \frac{\delta_t}{T_o} \tag{Equation I-4}
\]

where \( T_o \) is the original thickness of the cohesive element. Obviously if coupling between normal and shear components is not desired, the diagonal terms are equal to zero.

Considering Figure 3-16, as traction increases it reaches a point where the deformation is either purely normal to the interface or purely in the first or second shear direction. At that point, which corresponds to the nominal traction stresses \( t_n^0, t_s^0, t_t^0 \) and separations \( \delta_n^0, \delta_s^0, \delta_t^0 \), the initiation failure criterion is met and damage is introduced. In particular, damage initiates (Quads Damage) when:
where \( \langle t_n \rangle \) signifies that a compressive traction stress state cannot initiate damage.

As soon as the above criterion is met and damage initiates, the material stiffness starts to degrade and thus damage evolves. To represent the damage evolution in the cohesive zone ABAQUS uses a scalar variable, \( D \), which relates to the traction stress components as follows:

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

Equation I-5

where \( \langle t_n \rangle \) signifies that a compressive traction stress state cannot initiate damage.

To quantify the mode mix between normal and shear deformation, two approaches are available in ABAQUS, one based on energies and one based on traction. For the studies presented here the first approach was taken because the energy values for the input parameters were available in the literature[152]. To describe the energies associated with the traction components, three terms are introduced, \( G_n \), \( G_s \), \( G_t \), that represent the work done by the three traction components and are related to each other with the following expressions:
The definition of damage evolution in ABAQUS is done using two components. The first component is the energy that is dissipated during failure, $G^C$ (Figure I-1) and the second is the nature of damage evolution until failure, which in this case was specified by a linear power softening law.

\[
G_T = G_n + G_s + G_t \quad \text{Equation I-10}
\]
\[
G_s = G_s + G_t \quad \text{Equation I-11}
\]

According to this power law, failure in any mode (normal or shear) is caused due to an interaction of energies, which are related with $a$, the following power law:

\[
\left\{ \frac{G_n}{G_n^C} \right\}^a + \left\{ \frac{G_s}{G_s^C} \right\}^a + \left\{ \frac{G_t}{G_t^C} \right\}^a = 1 \quad \text{Equation I-12}
\]

where $G_n^C, G_s^C, G_t^C$ are the critical fracture energies required to cause failure in the normal, first and second shear directions. The response of the cohesive elements under a mixed-mode, as it was described above, can be summarized in Figure I-2.
In Figure I-2 the response of the cohesive elements in mixed-mode is represented in a traction versus normal and shear separation magnitudes three-dimensional plot. The triangles between the two vertical planes represent the damage evolution under mixed-mode whereas the triangles on the vertical planes represent the damage evolution under pure normal and shear mode. In terms of user control over the damage initiation and evolution of the cohesive zone ABAQUS requires the input of values for initiation and evolution. The values used in this model are given in Table 3-3 and Figure 3-5.