Low Velocity Impact Testing on Laminated Composites

S. I. B. Syed Abdullah

Abstract The response of composite laminates from transverse impact loading is known to vary with the speed of impact. In Low Velocity Impact (LVI) conditions, boundary effects usually dominate since the impact duration is longer between the laminate and the impactor. The global damage modes in LVI is also distinctly unique, whereby large deflections often occur, which depend highly on the shear properties (both in-plane and interlaminar) of the material. Therefore, characterisation of impact resistance and damage on LVI conditions are crucial before material selection for structural design. In this chapter, the LVI behaviour of composite laminates under LVI loading is investigated. The type of damage under LVI is also highlighted and discussed to obtain a detailed understanding of the impactor mass and velocity effects. The extent of delamination is studied using ultrasonic C-scan and radiograph images. Finally, where possible, fractographic studies have been undertaken to understand the influence of the interlaminar toughness on the impact resistance.

Keywords Low velocity impact · Impact resistance · Fracture toughness · Delamination · Impact damage

1 Impact on Composite Structures

The problem of impact on composite structures has been a subject of review for more than three decades. To date, many review papers have been written by researchers (Davies and Olsson 2004; Richardson and Wisheart 1996; Argawal et al. 2014; Abrate 1991a, b, 1998; Vaidya 2011; Silberschmidt 2016; Cantwell and Morton 1991) reporting the advances observed in the field of impact mechanics on composite materials. These advances, which include those made in damage prediction using numerical methods such as FEM, have strengthened our understanding of more damage tolerant structures, designed for various applications. Mathematical models such as the spring-mass model, the energy-balance model and the Delamination Threshold Load (DTL) (Schoeppner and Abrate, 2000a, b; Donadon and Falzon 2006) have

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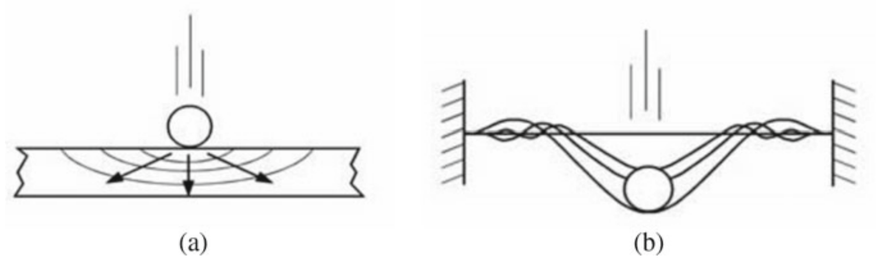

Fig. 1 Response types for impacted composite structures, **a** Stress wave dominated event, **b** Boundary dominated event (Davies and Olsson 2004)

greatly helped in determining the performance of the composite structure under transverse impact loading. Foreign Object Damage (FOD), characterised by the velocity of impact, is of importance due to its relevance in real-life applications. Research normally centres around two types of velocity regime—low and high velocity impact. Although argument exists with regards to the distinction of velocity range between low and high velocity impact, researchers have agreed that the threshold velocity defining low velocity impact is up to 10 m/s^{-1} (Davies and Olsson 2004; Richardson and Wisheart 1996; Cantwell and Morton 1991; Sjoblom and Hartness 1988), and more generally by Abrate (1998) for velocities below 100 m/s^{-1}. Additionally, Davies and Robinson (Robinson and Davies 1992) define Low Velocity Impact (LVI) as one in which through-thickness stress waves in the specimen play no significant part in the stress distribution at any time during the impact event. Hence, a global deformation can be observed in the laminate, shown in Fig. 1b, due to the long impact duration. In contrast, a High Velocity Impact (HVI) event is usually stress wave dominated, Fig. 1a, therefore the effects of boundary conditions can be neglected (Davies and Olsson 2004; Abrate 1991a, b, 1998).

Godwin and Davies (1988) proposed a simple technique to evaluate the transition velocity in which stress wave effects dominate. The relationship is based on the propagation of stress waves from the front face, which then progress towards the rear of the laminate. Therefore, the compressive strain, ε_c, can be calculated by using the relationship given by Robinson and Davies (1992):

$$\varepsilon_c = \frac{V_i}{C_z} = \frac{V_i}{\sqrt{\frac{E_z}{\rho}}} \tag{1}$$

where V_i is the impact velocity, C_z is the through-thickness speed of sound in the material, V_i is the out-of-plane Young's modulus, and ρ is the material's density. For instance, the transition velocity into a stress wave dominated event would occur at an impact velocity of approximately 20 ms^{-1} and above for a Uni-Directional (UD) CF/Epoxy composite with $E_y = E_z = 9.4$ GPa, $\varepsilon_c = 0.6\%$, and $\rho = 1.3$ g/cm^3.

Under High Velocity Impact (HVI) load, the longitudinal and transverse stress waves propagate from the point of impact upon initial contact between the projectile and the composite. The longitudinal wave propagates towards the edge of the laminate and is reflected back towards the centre, whilst the transverse wave propagates towards the composite back face, usually travelling at a velocity much lower than the longitudinal wave. The repeated reflections of the longitudinal stress waves typically decrease in intensity throughout the impact event, which was shown experimentally by Pandya et al. (2008).

When the impact velocity (or impact energy) is considerably lower than the V50 (or penetration energy for LVI loading), minimal to no damage will be observed in the laminate. V50 is defined as the velocity at which a projectile has a 50% probability of penetrating the target material (in this case composite laminates) (Cuniff 1999a, b). The contact stress between the projectile (or impactor) and the laminate would generally induce minimal damage in the form of matrix cracking (Cantwell and Morton 1989a, b). These cracks would saturate as a result of coalescence between multiple microcracks (Olsson 2001; Berthelot 2003; Puck and Schurman 1998; Williams et al. 2003). Depending on the thickness of the laminate, the damage pattern would differ due to the difference in the energy absorbing mechanism. For a typical brittle fibre-matrix system such as Carbon Fibre/Epoxy (CF/Epoxy) and Glass Fibre/Epoxy (GF/Epoxy), the cross-sectional damage normally resembles the so-called "pine tree" pattern. For thin laminates, a reversed pine tree pattern, Fig. 2a, can usually be observed from the cross-section of the impact area. This is due to the high bending load at the rear side of the laminate, hence initiating shear matrix failure. On the contrary, matrix cracks in thick laminates will often result in a pine tree pattern, Fig. 2b. This is because the normal in-plane stresses have exceeded the transverse tensile stress of the front plies, therefore initiating matrix failure. For both pine tree patterns, extensive experimental evidence exists for CF/Epoxy composites (Cantwell and Morton 1985; Jih 1993; de Freitas et al. 2000; Bouvet et al. 2009; Garcia-Rodriguez et al. 2000), and GF/Epoxy (Zhou 2003; Shyr and Pan 2003; Liaw and Delale 2007; Crupi et al. 2016).

The saturation of matrix cracks would typically lead to interlaminar failure, often known as delamination damage, which severely degrades the laminate's capability to carry further loads. An obvious drop in the load-time history (or load–displacement history) response (Schoeppner and Abrate, 2000a, b) can be observed indicating a delamination has occurred. For some materials (usually brittle materials such as CF/Epoxy and GF/Epoxy), violent oscillations may also be observed following

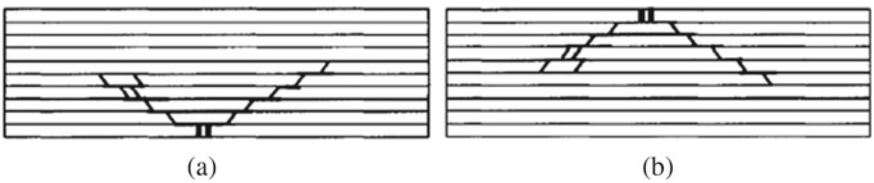

Fig. 2 **a** Reversed pine tree pattern, **b** Pine tree pattern (Abrate 1998)

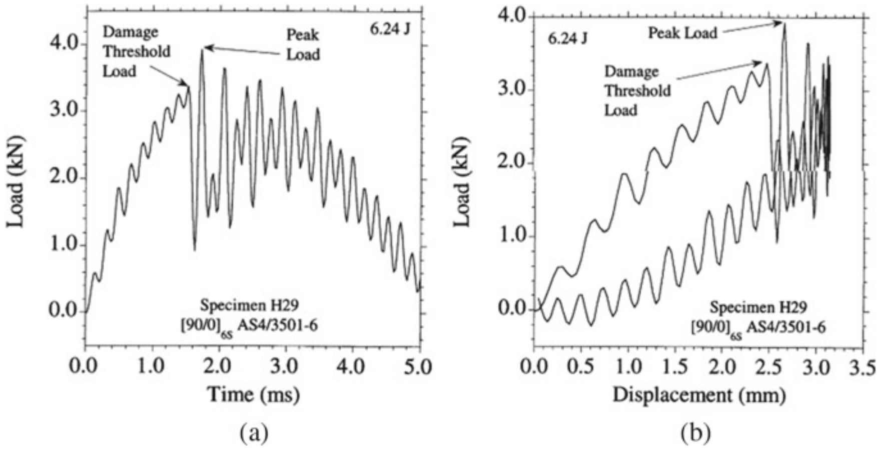

Fig. 3 Delamination Threshold Load (DTL) for a typical carbon/epoxy laminate, **a** Load-time history, **b** load–deflection history. Schoeppner and Abrate (2000a, b)

delamination damage, Fig. 3, due to the unstable propagation of delamination [24]. Also, a noticeable change in stiffness can be seen in the load–displacement history [20], illustrated clearly in Fig. 3b.

A closed form solution for the prediction of delamination damage has also been defined by Davies and Robinson (1992), utilising fracture mechanics given by:

$$P_c^2 = \frac{8\pi^2 E_f h^3}{9\left(1 - \upsilon^2\right)} G_{IIc} \tag{2}$$

where P_c is the critical load to initiate delamination, G_{IIc} is the Mode II critical strain energy, υ is the Poisson's ratio, E_f is the laminate flexural modulus, and h is the laminate thickness. The high dependence on the shear properties for LVI loading is apparent, resulting in the inclusion of Mode II critical strain energy in Eq. (2). This is not surprising since LVI is a flexural dominated event, hence interlaminar shear properties are central to the impact performance of laminated composites. A closed form approximation has also been derived by Olsson et al. (2006; Olsson 2010) to predict delamination initiation and propagation under HVI loading. It was found that the threshold load to initiate delamination is approximately 21% higher compared to its quasi-static value. This is due to the dynamic nature of the HVI event, inducing very high contact stresses localizing at the point of impact (Olsson et al. 2006).

It must be noted that delamination failure normally occurs only at interfaces with different fibre orientation, Fig. 4a (Richardson and Wisheart 1996; Abrate 1998; Liu 1988). This is associated with the adjacent plies which have different fibre orientations possessing different bending stiffnesses, therefore promoting delamination failure due to the property mismatch, Fig. 4b. From this, peanut-shaped delamination extending along the fibres can normally be observed, increasing in size with

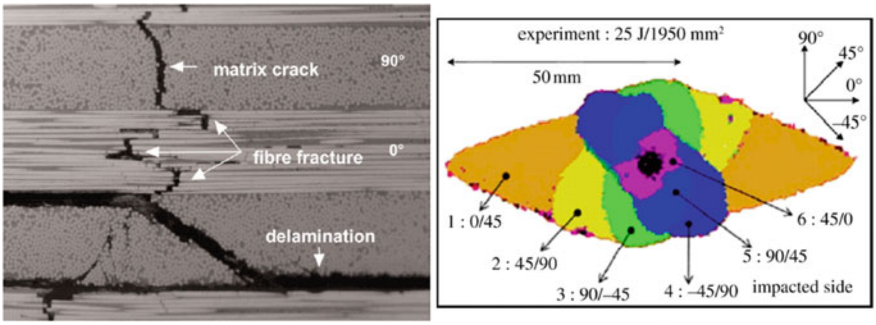

Fig. 4 **a** Variation of damage in low velocity impact loading (Davies and Olsson 2004), **b** Peanut-shaped delamination with respect to fibre orientation (Wisnom 2012)

an increase in the misalignment angle (i.e. a larger size should be present for a 0°/90° alignment compared to a 0°/45° fibre alignment) (Davies and Olsson 2004; Richardson and Wisheart 1996; Abrate 1998; Wisnom 2012).

If the velocity is increased (i.e. closer to the laminate ballistic limit or penetration energy), further damage would be seen due to higher contact stresses as well as a greater amount of energy transfer between the projectile and the laminate. For thin laminates, the energy transfer will normally result in tensile straining of the fibres, usually known as a cone formation at the rear of the laminate (Naik and Shrirao 2004; Shaktivesh et al. 2013; Naik et al. 2005). This is due to the high flexural load experienced by the laminate, with high tensile strain, particularly at the laminate back face. Penetration of the laminate is therefore governed by the amount of fibre failure; full penetration indicates that all fibres are broken, and no penetration indicates that zero or a minimal number of fibres have failed (Naik and Shrirao 2004). Fibre failure normally initiates from the laminate back-face, extending through-the-thickness towards the front face (Davies and Olsson 2004; Richardson and Wisheart 1996; Abrate 1998). For thick laminates, penetration is normally associated with shear failure, specifically in the transverse direction resulting in a shear-plug formation (Richardson and Wisheart 1996). The depth of penetration depends on the severity of damage and includes other factors such as the type and shape of the impactor (Richardson and Wisheart 1996; Abrate 1998; Cantwell and Morton 1991; Shyr and Pan 2003; Mitrevski et al. 2015, 2006). The energy related to laminate penetration is proposed by Dorey (1987) given as:

$$E_p = \pi \gamma t d \tag{3}$$

where E_p is the penetration energy, γ is a coefficient property related to the fibre fracture energy, d is the diameter of the impactor, and t is the laminate thickness. It is clear from Eq. (3), that the energy absorption due to penetration increases with increasing thickness. This has been confirmed experimentally by researchers (Richardson and

Wisheart 1996; Abrate 1998; Shyr and Pan 2003; Cantwell and Morton 1989a, b; Caprino and Lopresto 2001), and have concluded that the energy absorption (hence impact resistance) significantly increases with increasing thickness.

2 Polymer Fibre Composites Under Impact

The use of a polymer fibre as an alternative to conventional brittle fibres such as CF and GF offer a potential solution to the poor impact performance of composite materials. Its superior mechanical properties often result in an enhanced impact performance, both under LVI and HVI. A key step in understanding the impact performance of high-performance fibre composites was proposed by Cuniff (1999a, b) using a dimensionless analysis based on the fibre specific toughness and longitudinal strain wave velocity, given by:

$$U^* = \frac{\sigma \varepsilon}{2\rho}\sqrt{\frac{E}{\rho}} \tag{4}$$

where U^* is the so-called Cuniff velocity", σ is the fibre ultimate axial tensile strength, ε is the fibre ultimate tensile strain, ρ is the fibre density, and E is the fibre tensile modulus. Note the linear assumption in Eq. (4), which could be misleading for fibre (or composite) possessing a non-linear tensile response. Moreover, through-thickness stresses, as well as the in-plane shear properties were not considered in Eq. (4).

This may present a significant difference in the composite impact performance if the selection is based purely on U^*. However, Eq. (4) does allow for a quick analysis in selecting the appropriate fibre for composite design. Table 2.1 presents the calculated Cuniff velocity for selected commercially available fibres.

From Table 1, it can be easily observed that IM7 possesses the highest U^* when compared to the selected fibres. This should indicate that IM7 should possess excellent impact performance although in reality, this is not the case. It was found that IM7 composite was the worst performing composite when compared to Vectran and S2-Glass composites. This result is associated with several factors. First, the low tensile strain-to-failure of IM7 (≈1.9% (Hexcel 2018)) compared to Vectran (≈3.8% (Kuraray America 2006)) and S2-Glass (5.7% (AGY 2006)) limits the 'elongation' of the fibre to prevent impactor penetration. A recent investigation by Heimbs et al.

Table 1 Cuniff velocity for chosen fibres (Syed Abdullah 2019, 2021)

Fibre type	U^* (m/s)
IM7 (CF)	722
S2-Glass (GF)	634
Vectran	616

(2018) found that Dyneema HB26 showed superior HVI performance compared to CF and GF composites, whilst a comparable LVI performance with GF composite. Furthermore, in both impact events (LVI and HVI), CF composite exhibits the worst impact performance among all three composite materials. The superior performance can be partly attributed to the relatively high tensile strain-to-failure of Dyneema HB26 laminate, in addition to other factors such as the low $\pm 45°$ in-plane shear properties (both stiffness and strength). This is consistent with the works of Reddy et al. (2017) when investigating the impact performance (LVI and HVI) of GF/Epoxy and Dyneema HB26 laminates. The influence of shear strength on the ballistic performance has also been investigated by Karthikeyan et al. (2013), which found a significantly higher V_{50} for the uncured IM7/8552 (CF/Epoxy) if compared to its cured counterpart. More importantly, the low $\pm 45°$ in-plane shear properties have resulted in a superior impact performance of Dyneema HB26 and HB50 laminates; at least 50% higher than the uncured IM7/8552.

Recently, Hazzard et al. (2017) investigated the effect of fibre orientation of Dyneema HB26 under LVI loading. A large Back Face Deflection (BFD) due to the low $\pm 45°$ in-plane shear properties were observed for the 0°/90° orientation of Dyneema HB26 laminates. However, no penetration was observed even for large impact energies of up to 150 J impacted on a 2 mm thick laminate. The effects of fibre orientation of Dyneema HB26 under HVI loading have also been investigated by Karthikeyan et al. (2016), whereby a large BFD was observed for the 0°/90° orientation compared to other types of layup such as Unidirectional (UD) and helicoidal (hybridised 0°/90° orientation). In spite of this, the 0°/90° orientation yielded the highest V_{50} compared to the other orientation types. It must be noted that a large BFD may not necessarily be a positive characteristic concerning composite structural design. In certain structural design, particularly in defence applications, a large BFD may imply a severe weakness due to the 'soft' nature of the structure, enabling rapid (and excessive) membrane loading at the point of impact.

Secondly, the inherently large fracture energy of polymer fibre composites partly contributes to its superior impact performance. An earlier work by Park and Jang (2004) highlights the considerably large impact absorption energy of Aramid when compared to GF composites (6.3 J and 58.43 J for a 2 mm thick GF and Aramid laminates, respectively) under LVI loading. A similar observation was seen by Sikarwar and co-workers (2017) when investigating the behaviour GF/Epoxy and Kevlar/Epoxy laminates, where a considerably higher specific energy absorption capacity was found for Kevlar/Epoxy laminates ($\approx$ 36% higher) when compared to Glass/Epoxy laminates. This is consistent with the findings of Evci and Gulgec (2012), where a superior impact performance was observed for Aramid laminates if compared to GF composites under LVI.

The high fracture energy of polymer fibres (or polymer composite systems) enables a large energy absorption under impact loading, partly due to its unique fibre structure. For instance, the so-called 'skin–core' structure of Vectran fibre typically results in a tougher system, due to its inherent energy absorption mechanisms such as fibre pull-out and interfacial sliding. A recent investigation on the tensile fracture properties of Vectran/Epoxy laminates (Syed Abdullah 2018) revealed a much higher value of fracture toughness; up to 48.26% and 95.27% higher for initiation and propagation for some CF/Epoxy (Pinho and Robinson 2012) composite system, and 9.93 and 68.6% higher for initiation and propagation for some GF/Epoxy (Katafiasz et al. 2019) composite system. The high fracture energy of Vectran is partly attributed to the 'skin–core' nature of the fibre, resulting from its manufacturing process.

During fibre extrusion, the 'skin' cools much quicker than the 'core' resulting in a highly oriented crystalline region at the 'skin', whilst a less oriented, amorphous region at the 'core'. Hence, a noticeable difference in the mechanical properties may be found, whereby the 'skin' is much stiffer when compared to the 'core', resulting in a path of failure which initiated first at the 'skin' and then propagating towards the 'core'.

Thirdly, the sensitivity to strain-rate effects was evident, particularly in polymer fibre composites. A recent report by Singh (2018) attempts to collate existing strain-rate research on high performance fibre composites. It was found that GF (or GF composites) display considerable sensitivity towards strain-rate, with enhancements close to 200% observed at 103/s compared to the quasi-static case. A similar response was seen for Kevlar fibres, although its enhancement is not as significant as GF, with an increase of up to 50% in tensile strength. This is consistent with the works of Wang and Xia (1998), when investigating the effect of strain-rate (0.0001/s–1350/s) on Kevlar 49 fibre bundles. The authors found an approximately 22% increase on its tensile strength modulus and strength, including a 15% increase in its tensile strain-to-failure. Tan et al. (2010) investigated the tensile response of Twaron CT716 yarns at strain-rates of up to 480/s and found a 28 and 36% increase on the tensile modulus and strength, respectively. In a follow-up study, Koh et al. (2010) investigated the strain-rate effects of Spectra 900 yarns and found a 150% increase in the tensile modulus and 40% increase in the tensile strength.

Despite the excellent impact performance of polymer fibres, the significantly low compressive properties severely restrict their use as a structural component. This is due to the weak (van der Waals) bonds between molecules, which easily fails when the fibres are loaded in compression. Specifically for Vectran, the fibrillar nature of the fibre tends to de-fibrillate almost instantly under compression loads, resulting in a compressive property of approximately 10% of its tensile component (Donald et al. 2006). Hence, it is common that polymer fibre composites should be utilised as a reinforcement in a hybrid composite system.

Park and Jang (2001, 2004) investigated the impact performance of Aramid fibre/glass fibre hybrid composite and found a markedly higher impact performance when compared to the monolithic GF composite system. Sikarwar et al. (2017) observed a 20% increase in the HVI performance of GF/Kevlar hybrid composites when compared to the monolithic GF composite system. Bandaru et al. (2001) investigated the ballistic performance hybrid thermoplastic Kevlar and basalt composite system and found a 26.27% increase in the impact performance compared to Basalt monolithic composite system. A recent review by Iannucci (2018) highlights the advantages of composite hybridisation by incorporating polymer fibres with conventional materials such as CF and GF. Some of the benefits include potential weight savings and enhanced impact performance, whilst achieving the desired compressive properties to be used as a structural component.

3 Post Impact Residual Strength

The post-impact residual strength of composites, also known as the damage tolerance, is an important parameter when designing structures involving composite materials. This is important since the damaged composite is expected to maintain its original strength and stiffness. The damage tolerance of a composite is usually studied by determining the effect of different impact energies on their residual strength, the Compression After Impact (CAI) test being the experimental test of components damaged under LVI loads. First, the composite is subjected to an LVI test, using a drop weight impact tower which records the load and impactor vertical displacement. Next, the damaged composite is placed in a bespoke rig, usually following a specified dimension from the ASTM D7137 standard (American Standard for Testing Materials (ASTM) 2017). It must be noted that other tests exist to quantify the composite damage tolerance, such as the Tension After Impact (TAI) (El-Zein and Reifsnider 1990), although the test is rarely performed due to the difficulties associated with the test.

Sanchez-Saez et al. (2005) investigated the CAI behaviour of thin CF/Epoxy laminate (1.6–2.2 mm thick) with three different laminate layups (QI, CP, and Woven), and found that the woven laminate possesses the highest CAI strength compared to the other layups. Khondker et al. (2005) studied the CAI behaviour of weft-knit architecture and concluded that the weft-knit composite exhibits a considerably superior CAI performance, if compared to UD and braided architecture. This is mainly attributed to the enhanced structural integrity of the weft-knitted fabric, resulting in a composite with an improved impact resistance thus suffering minimal damage during initial impact. Hart et al. (2017) compared the damage tolerance of 2D and 3D woven Glass/Epoxy composite under CAI and Flexure After Impact (FAI) loading and found that the FAI testing yields 70% reduction in flexural strength compared to only 20% reduction in compressive strength. In addition, it was found that the 3D woven composite possesses a superior damage tolerance due to the presence of the through-thickness stitches (Z-binder) in the composite, significantly improving its delamination behaviour. Tan et al. (2012) studied the effect of through-thickness stitching in CF/Epoxy composite and concluded that superior CAI performance was observed for composites having through-thickness stitches, especially those with high density stitches. This is mainly because of the enhancement in the delamination behaviour, due to the presence of through-thickness stitches in the composite.

4 Factors Influencing the Impact Performance

5 The Fibre Constituent

Perhaps the most important factor in determining the performance of a composite is its fibre constituent. The choice of fibre as a component in a composite greatly determines its impact performance, since the fibre is the primary load bearer in a laminate. For instance, the use of fibres with a large tensile strain-to-failure would often result in a better impact performance. This is since the large tensile strain-to-failure enables a larger energy absorption by way of a large BFD, hence preventing impactor/projectile penetration (Hazzard et al. 2017). This is shown in Figs. 5 and 7, where it can be observed that Vectran/MTM57 exhibits a much larger BFD compared to S2-Glass/MTM57. In addition to a large tensile strain-to-failure, the polymeric nature of Vectran fibres results in a lower $\pm 45°$ in-plane shear properties (stiffness and strength), thus promoting a large BFD to occur. Under LVI loading, the interlaminar fracture behaviour is important, particularly the Mode II (G_{IIc}) fracture toughness. This is because under LVI loading, flexural deformation dominates hence laminates with a low G_{IIc} would outperform those with a higher G_{IIc}.

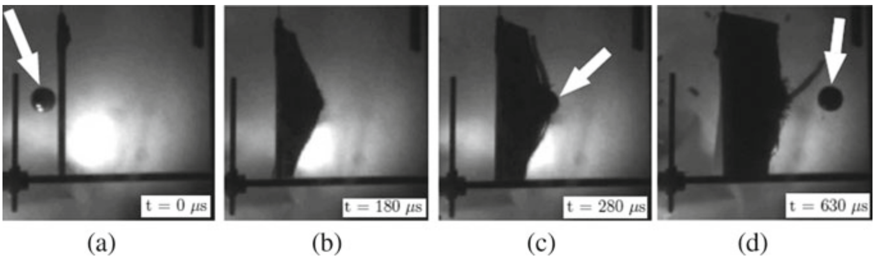

Fig. 5 Montage of HVI events on Vectran/MTM57 at 171.2 m/s, **a** before impact, **b** during impact, **c** partial penetration, **d** penetrated. White arrow indicates projectile (Syed Abdullah 2019, 2021)

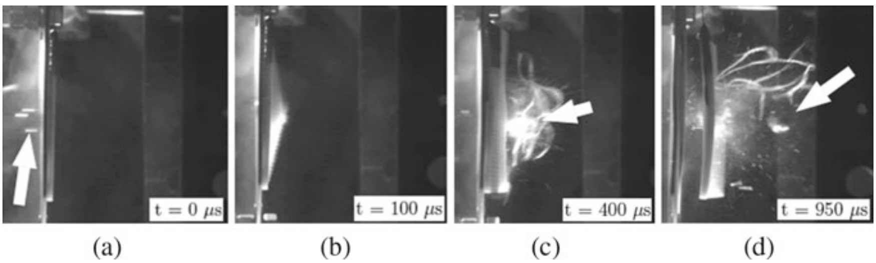

Fig. 6 Montage of HVI events on S2-Glass/MTM57 at 183.08 m/s, **a** before impact, **b** during impact, **c** partial penetration, **d** penetrated. White arrow indicates projectile (Syed Abdullah 2019, 2021)

6 The Matrix Constituent

It has been shown that the use of thermoplastic matrix results in an improved impact performance if compared to its thermosetting matrix counterpart. This is because there is generally no crosslinking in thermoplastic matrices when heated above its liquid to glass temperature (T_g). In fact, compared to thermosetting matrices, thermoplastic matrix will soften when heated above T_g and harden when cooled down. This mechanism is usually repeatable, whereby thermoplastic materials can be reheated again and formed into desired shape. In contrast, thermosetting materials will harden when heated above its critical temperature and will not soften again on reheating (Cowie 1991). Therefore, thermoplastic materials are often tougher compared to thermosetting materials.

Vieille et al. (2013) compared the LVI performance of CF/Epoxy (thermosetting matrix) and CF/PEEK (thermoplastic matrix) and found that the CF/PEEK composite performed considerably better compared to its CF/Epoxy counterpart. Arikan and Sayman (2015) investigated the impact behaviour of E-glass fibre reinforced Polypropylene and epoxy matrix composites and found that the composite with thermoplastic composites performed considerably better under LVI loading. In addition, the ‘ductile’ nature of the thermoplastic matrix greatly reduces delamination damage from occurring. Dorey et al. (1985) compared the LVI performance of epoxy and Polyetherketone (PEEK) matrix with carbon fibre reinforcement. While damage were less extensive for CF/PEEK composites, it was found that the damage in the carbon fibre/PEEK composite comparable to that of carbon fibre/epoxy. Recently, Sonnenfeld et al. (2017) investigated the use of thermosetting-thermoplastic combination with carbon fibre reinforcement and found a significant reduction in damage due to LVI loading.

7 Fabric Architecture and Layup Orientation

The fabric architecture of a composite could greatly contribute to its impact performance. For instance, Non-Crimp Fabrics (NCF) will have a superior impact behaviour if compared to Uni-Directional (UD) composites, although its layup orientation is comparable. This is because the stitches present in NCF based composite will significantly improve its delamination behaviour, resulting in a superior LVI performance (Greenhalgh 2009). In addition, the improvement in its delamination behaviour is particularly beneficial due to the dominance of flexural deformation in LVI conditions.

The effect of layup orientation is also influential in determining the impact performance of a composite. Depending on the layup, the orientation may contribute to the flexural behaviour of the composite. Under impact loading, a more ‘compliant’ composite is desirable to absorb the impact energy transferred from the impactor/projectile to the laminate. This is usually done by incorporating a higher number of ±45° plies in the laminate. As a result, the laminate will be less stiffer in flexure thus able to deform further to prevent penetration from occurring. Hazzard et al. (2017) investigated the effect of three different layup orientation (cross-ply, quasi-isotropic, and helicoidal) and found that the helicoidal orientation has the most superior LVI performance compared to the other two layups. Note that Cross-Ply (CP) laminates are obtained when each plies are arranged in a 0° and 90° only, whilst the Quasi-Isotropic (QI) layup is defined as a laminated in which the plies are arranged in such a way that the in-plane properties will behave as an isotropic composite, though the through thickness properties is not isotropic. Sharma et al. (2019) found that damage exerted by a Cross-Ply (CP) layup is considerably higher compared to a Quasi-Isotropic (QI) layup under LVI loading. In addition, the QI layup was also able to absorb a higher impact energy under LVI loading. Zhou et al. (2019) studied the effect of different stacking sequence and ply orientation on CF/Epoxy composite and found a significant dependence in the stacking sequence and ply orientation on the LVI performance. It was found that the QI layup is considerably superior compared to the CP layup under LVI loading, consistent with the findings of the previous researchers.

As mentioned earlier, the LVI performance of composite laminates is highly dependent on its fabric architecture. Recently, Miao et al. (2019) studied the effect of three different fabric architecture (UD, Woven, and 3D composite) and found that the 3D composite possesses better impact resistance compared to the UD and Woven composite. This is because the Z-binder which is present in a 3D composite act as initiation site due to the weak fibre-matrix bonding, resulting in an enhanced impact resistance for the composite. Similarly, Saleh et al. (2019) investigated the impact performance of three different fabric architecture (NCF, 2D and 3D woven composite) under multiple LVI and found that damage was the least in 3D woven composite. In addition, the residual strength under CAI loading was the highest for 3D woven composite if compared to the other two laminates used (NCF and 2D woven). A similar observation was also found by Syed Abdullah (2019, 2021), when investigating the LVI performance of three different monolithic composite (IM7/8552, S2-Glass/Epoxy, and Vectran/Epoxy) having similar weight (≈90 g) using two different fabric architecture (NCF and UD). It was found that both S2-Glass/Epoxy and Vectran/Epoxy possesses superior impact properties compared to the IM7/8552. This was due to several reasons. First, both S2-Glass/Epoxy and Vectran/Epoxy composite has a relatively larger tensile strain-to-failure compared to IM7/8552 (2.6% and 2.8% respectively for Vectran/Epoxy and S2-Glass/Epoxy, compared to 1.4% for IM7/8552), thus being able to absorb more energy from the impactor. Secondly, the NCF architecture significantly enhances the delamination behaviour, due to the presence of stitches in the NCF fabric which diverts the propagation of crack. In addition, the fibrillar nature of Vectran fibres promotes extensive fibre bridging in between each ply.

Figure 7 presents a fractographic observation of S2-Glass/Epoxy composite under Mode I Interlaminar Fracture Toughness (IFT) testing, where the propagation of crack was diverted parallel with the direction of the stitch, thus requiring a higher

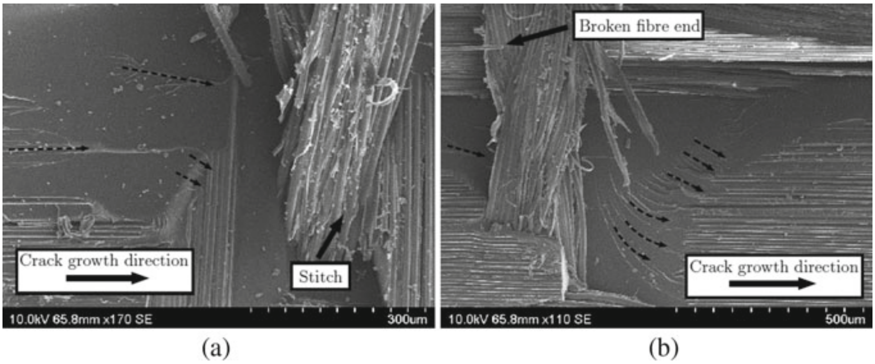

Fig. 7 Fractographic observation on the mode I behaviour of S2-glass/epoxy composite. **a** Crack growth morphology before a stitch, **b** crack growth morphology after the stitch. Notice the diversion of crack when propagating close to the stitch (Syed Abdullah 2019, 2021)

energy to further propagate the crack. Consequently, the delamination behaviour of the composite is significantly enhanced, resulting in a superior LVI performance.

8 Summary

While there may be many favourable attributes of composite material such is its high strength-to-weight ratio compared to conventional materials such as steel or aluminium, the susceptibility of composite towards impact loading is an inherent weakness which needs to be improved. The inherent weakness is mainly due to the small plastic component in the material, thus limiting its ability to a absorb a large amount of impact energy. This is especially true for classical brittle fibre-matrix system, such as CF/Epoxy, in addition to its relatively small tensile strain-to-failure. Under LVI, damage such as the BVID can increase the chances of catastrophic failure or sudden damage since no clear visual evidence of permanent indentation can be seen. It is quite possible for a composite structure to suffer fibre damage and massive delamination between plies, therefore more work is needed to detect the severity of damage (such as using a non-destructive testing approach). Therefore, it is important to improve the impact performance, by using an alternative approach such as hybridisation, where two (or more) types of reinforcements are included in a matrix material. Some of the reinforcements include polymer-based fibres such as Aramid and Vectran, which possess a large amount of plastic component in the material, thus being able to absorb more impact energy. In addition, hybridisation can also improve the compressive properties of polymer fibres, which is inherently poor, thus preventing it from being used as a structural component.

Perhaps more importantly is the need to understand the mechanics of failure due to a number of reasons. First, with a comprehensive understanding, engineers could exploit the full potential of the composite, without the need to over-design to ensure that the composite structure meets all safety requirements. This would result in a more cost-effective design, with minimal material consumption and fewer man-hours. Secondly, an accurate prediction of failure can be made, leading to a more reliable design since the mechanics of the composite are fully understood. Thirdly, physically based constitutive equations can be derived, which can be implemented as a material model for Finite Element Method (FEM) modelling. The material model can then be used to reduce the number of experimental tests required, leading to a reduction in the total design cost.

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