Mechanisms of penetration in polyethylene reinforced cross-ply laminates

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Abstract

The mechanisms of progressive penetration for two ultrahigh molecular weight polyethylene (UHMWPE) reinforced laminates have been investigated. One used an UHMWPE fiber reinforcement while the other utilized molecularly aligned tape. Both materials had similar out of plane compressive strengths, but the fiber system had a 40% higher in plane tensile strength than the tape. Laminated, 6 mm thick plates with a $[0^\circ/90^\circ]$ ply architecture were impacted by a 12.7 mm diameter sphere under conditions that either allowed out of plane plate deflection or eliminated this deflection by rear support of the target. The depth of penetration, and the ballistic limit in the rear-supported tests was identical for the two materials, and proceeded by progressive ply failure. However, tests in the edge clamped condition resulted in a substantially higher penetration resistance, especially for the higher tensile strength fiber-reinforced material. Edge clamped testing of a bilayer target, where the front third was comprised of the tape material and the remainder comprised fiber reinforced laminate, had the same ballistic limit as a target comprised of only the higher ply tensile strength fiber reinforced material. Penetration in both test support conditions was discovered to occur by tensile ply rupture under the projectile, consistent with a recently proposed mechanism for converting out of plane compression to in plane ply tension. Lateral displacement of plies was also observed near

26 the sides of impact craters in both materials, indicating the existence of a second mechanism
27 impeding penetration of the spherical shaped projectile.

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1. Introduction

It is well known that composite laminates comprising high tensile strength polymeric reinforcements and compliant polymer matrices possess very high ballistic penetration resistances when configured in a $[0^\circ/90^\circ]$ cross-ply architecture [1,2]. However, the mechanisms by which an impacting projectiles momentum and kinetic energy are dissipated during penetration are much less well understood, and are the focus of the study reported here. The response to transverse (out of plane) impact of a thin laminate has been analyzed by analogy with that of a single fiber [3–5]. The central impact of a single, end clamped fiber generates strain pulses that propagate away from the impact site. The fastest of these elastic disturbances travels along the fibers at the fibers longitudinal wave speed,

$$c_L = \sqrt{(\lambda + 2\mu)/\rho} \quad \text{Eq. 1}$$

where λ and μ are the Lamé constants and ρ the density of the fiber [6]. A shear wave with a lower velocity $c_H = \sqrt{\mu/\rho}$ travels behind the longitudinal disturbance, enabling the fiber to undergo transverse deflection in the direction of projectile motion. The analogous wave speeds in a composite laminate are governed by the stiffness constants of the laminate, and are therefore orientation dependent [7]. For a $[0^\circ/90^\circ]$ cross-ply lay-up, they are highest in the fiber directions, giving rise to a ‘pyramid shaped’ transverse deflection envelope during impact, Fig. 1(a), whose base width expands with time.

Using dimensional analysis, Cunniff [1] showed that the ballistic limit of a fiber scales with the product of its longitudinal wave speed and the strain energy per unit mass needed to fail a fiber in tension. This combination of material properties gives rise to a Cunniff velocity:

$$= \left[(E_f / \rho)^{1/2} \sigma_f \varepsilon_f / (2\rho) \right]^{1/3} \quad \text{Eq. 2}$$

where E_f , σ_f and ε_f are the Young's modulus, tensile strength and failure strain of a linear elastic fiber and ρ its density. A comprehensive compilation of the predicted Cunniff velocities of most high performance fibers has been given in reference [8] and indicates that ultra-high molecular weight polyethylene (UHMWPE) fibers should have a very high ballistic limit. However, this approach provides little insight into the mechanisms of penetration of $[0^\circ/90^\circ]$ laminated composites constructed from such fibers.

Phoenix and Porwall [4] analyzed the deflection of an impacted laminate in the thin membrane limit where the transverse deflection of the laminate was governed by (tensile) membrane stresses supported by the fibers within the laminate. In this limit, the fiber stress is independent of depth, and laminate perforation occurs when the membrane stress attains the ply tensile strength. All else being equal, laminates with higher ply tensile strengths will have a greater resistance to perforation. While the ballistic limit predictions of the Phoenix and Porwall membrane stretching model are consistent with the dimensional analysis of Cunniff [1], failure was treated as a binary process; the laminate was either undamaged or fully penetrated by an impacting projectile. No partial penetration is permitted by such an analysis. However, numerous experiments have shown that UHMWPE fiber reinforced laminates fail progressively, with a depth of penetration that increases with impact velocity [8–13].

A study by Heisserer [11] using HB26¹ grade Dyneema[®], showed the depth of penetration by a hard spherical projectile increased linearly with projectile kinetic energy. Recent studies by Karthikeyan and Russell [13] using a spherical projectile and by Nguyen et al. [14] using fragment simulating projectiles (FSP), confirmed the existence of progressive penetration before the laminates ballistic limit was attained. All these studies indicate the penetration of these materials occurs in two stages schematically illustrated in Fig. 1(b). The first “progressive” stage of penetration occurs early during the impact process. It is accompanied by minimal transverse deflection of the failed plies. The second corresponds to the out of plane deflection of the unpenetrated remainder of the laminate by a membrane stretching mechanism which results in fiber pull-in in the 0° and 90° fiber directions (converting a square sided panel into a pin cushion shape). Karthikeyan and Russell [13] estimated the second stage dissipated ~6.5 times more kinetic energy per perforated ply than that of the progressive penetration stage. Efforts to impede the progressive mode of penetration, forcing failure by the second membrane-stretching mode are therefore likely to result in substantial improvements to the ballistic resistance of these composite materials.

There are many mechanisms by which materials can be progressively penetrated by a projectile. For example, penetration can occur by shear plugging, and experimental evidence for this has been widely reported for carbon (CRFP) and glass (GRRP) fiber reinforced polymer composites constructed with impact resistant polymeric matrices [15–20]. This mechanism can be activated when the shear stress (normal to the fiber directions) created by the projectile reaches the dynamic shear strength of the laminate. A plug of material of roughly the diameter of contact surface is then formed beneath the projectile, Fig. 1(c). However, the literature currently

¹ A [0°/90°] cross-ply laminate consisting of 83 vol% UHMWPE SK76 fibers in a polyurethane matrix.

offers no observational evidence for the formation of such a shear plug in UHWMPE fiber reinforced composites. These materials have a very high fracture resistance under transverse shear loading because the fibers are very flexible (do not fracture upon bending) and their tensile strength is very high. Furthermore, He and Hutchinson [21] and more recently Noselli et al. [22] have shown that delamination at interply interfaces, rather than ply cracking, dominates the response of laminates made from highly anisotropic plies.

Recent experiments suggest an alternative mechanism governs the progressive phase of penetration. Indirect evidence from impact experiments indicated that reduction of the pressure imposed by a projectile on a laminate improves the resistance to progressive penetration [8,12,23,24]. Furthermore, quasi-static studies have shown that $[0^\circ/90^\circ]$ laminates uniformly compressed normal to the plane of the fibers fail by tensile fiber (and ply) failure [25–27]. Atwood et al. [25] showed how this counter-intuitive phenomenon can arise if the plastic Poisson expansion of a compressed ply transverse to the fiber direction is much greater than that parallel to the fibers, Fig. 1(e). A transfer of stress occurs from an expanding pair of 0° plies to the 90° ply located between them occurs. This shear lag loading mechanism then places the 90° ply in tension in the fiber direction (while the expansion of the 90° ply also loads the 0° ply in tension). As compression of the laminate progresses, the tensile strength of the ply is eventually reached and sudden fracture occurs. The compressive strength therefore directly depends upon the ply tensile strength and inter-laminar shear strength [25,26].

Other mechanisms could also be activated. For example, under some situations lateral displacement of material away from the nose of the projectile, analogous to ductile hole

enlargement [28,29], could occur. In the ductile materials in which hole enlargement occurs, the radial stress from the penetrating projectile plastically displaces material, leaving a cavity in the wake of the projectile. This ‘slip’ of material around a projectile has been frequently observed in dry woven composites made from UHWMPE and aramid fibers [30], but is retarded if the fibers are embedded in a resin with high shear strength resistance.

One way to test the significance of the indirect tension progressive penetration mechanism is to construct laminates from materials with similar out-of-plane compressive strengths, but with different tensile strengths. It is noted that UHMWPE tapes [31,32], manufactured under the trade names Tensylon[®] by DuPont and more recently grade BT10 Dyneema[®] by DSM, offer a means to compare laminates with similar compressive strengths but with different ply tensile strengths. Even though the tensile strength of the tape materials is substantially less than that of the fiber-based materials, their out of plane compressive strengths are similar to the fiber based grades of these materials. Additional insights can be gained by subjecting panels of the two materials to impacts under conditions that either permit or inhibit the membrane stretching response. A simple means for this is to place the laminate on a solid foundation so it cannot be displaced in the out-of-plane direction required to activate membrane stretching [9]. The study presented here shows that the depth of (progressive) penetration and ballistic limit of the fiber and tape-based laminates are similar when membrane stretching is prevented. The study then investigates the edge clamped impact response of the two laminates, and a bilayer target comprised of both materials, and reveals the presence of the same progressive penetration mode observed in the rear supported condition.

2. Materials and properties

The study was conducted using two laminated grades of UHWMPE Dyneema[®] (DSM, The Netherlands), Table 1. The grade HB50 laminates comprised a $[0^\circ/90^\circ]$ cross-ply lay-up of 60 μm thick plies of unidirectional grade SK76 Dyneema[®] fiber in a rubber-modified resin that accounted for 17 wt. % of the composite, Fig. 2(a). The properties of the HB50 laminate and its SK76 fiber have been extensively studied [12,23,26,33]. The second material, grade BT10 tape, consisted of 100 mm wide by 42 μm thick strips of UHWMPE solid-state tape. The strips were plain weave woven with a 200 mm periodicity, Fig. 3(a). Both materials were received as 4 ply thick ‘pre-pregs’ that could be consolidated into a laminate at a temperature of 127°C for HB50 and 130°C for BT10 using a pressure of 20.6 MPa. Prior to consolidation of the BT10 grade, the lightly adhered woven strips were first separated and then laminated to create a non-woven $[0^\circ/90^\circ]$ cross-ply lay-up, Fig. 3. This modified composite is subsequently referred to as BT10m, Fig. 2(b), while the term BT10 will refer to the as-received woven pre-preg.

Table 1. Laminate architecture and mechanical properties.

	HB50	BT10m
Reinforcement type	Fiber	Solid-state tape
Reinforcement grade	SK76	BT10
Resin [*]	SISTC	N/A
Fiber diameter/ tape thickness, t (μm)	17	42
Ply thickness, h (μm)	60	42
Longitudinal ply modulus, E_f (GPa)	70	100
Ply tensile strength, σ_f (GPa)	1.7	1.2
Inter-laminar shear strength, τ_0 (MPa)	0.35	3.5

^{*} Resin abbreviation: styrene-isoprene-styrene triblock copolymer (SISTC)

2.1. *Mechanical properties*

2.1.1. *Tension and shear*

Measurement of the tensile and shear strengths of the composites followed the procedures described by O'Masta et al. [26]. The unidirectional ply tensile strength was calculated from laminate tensile tests (taken to be twice the measured laminate's strength) since Russell et al. [33] revealed fiber misalignment and waviness and changes in fiber morphology reduce the ply strength compared to that expected from single fiber tests. The tensile tests used a $[0^\circ/90^\circ]_4$, dog-bone shaped coupon under a gripping pressure of 50 MPa and loaded at a nominal tensile strain rate of 10^{-3} s^{-1} . The shear tests utilized a $[0^\circ/90^\circ]_{40}$, 20 mm wide, double-notch specimen, with a 20 mm gauge length, loaded at a rate of 1 mm min^{-1} . Table 1 summarizes the measured tensile and shear strengths for the two materials. The measurements for HB50 compare well with values published elsewhere [12], and confirm that the HB50 laminate has a 40% higher tensile strength than BT10, but only a tenth of the BT10 laminates shear strength.

2.1.2. *Uniform compression strength*

When cross-ply $[0^\circ/90^\circ]$ laminates are placed under out of plane uniform compression, the anisotropic expansion between neighboring plies creates internal tensile stresses by a shear-lag mechanism [25]. The stress increases from zero at the edge of a sample to a constant stress over the shear lag length (inset of Fig. 4(b)). Therefore, the effective compressive strength increases with the in-plane length of the sample, L , approaching a plateau strength governed by the ply tensile strength as the shear lag length becomes small compared to that of the sample. However, the plateau strength of thin laminates can be reduced by the presence of missing fiber defects in fiber reinforced Dyneema[®] [24].

Six 1 mm thick samples (consisting of 16 HB50 plies and 24 BT10m plies) with $L = 6$ to 15 mm and two others with $L = 20$ and 25 mm that were 240 plies thick were prepared from both materials. Representative compressive stress-strain curves of HB50 and BT10m samples are plotted in Fig. 4(a) together with their tangent modulus. The stress for both grades monotonically increased with strain, with the tangent modulus of the BT10m sample being approximately twice that of HB50. The samples catastrophically failed at a peak strength, σ_c . The compressive strength, obtained by averaging test results from five specimens of each material and sample length combination, are plotted in Fig. 4(b). The strength of both materials monotonically increased with L as expected. However, the BT10m laminate had a higher compressive strength than HB50 despite its lower ply tensile strength. This is attributed in part to a smaller shear lag length for BT10m and to less efficient conversion of compressive stress into tension of the reinforcement, as a consequence of the low plastic shear resistance of HB50 [25]. This (fortuitous) difference in response to compression loading results in the two materials having a similar compressive strengths in the large L -limit (a 6% difference at $L = 25$ mm), but markedly different tensile strengths.

2.1.3. *Punch penetration*

The quasi-static punch resistance of each laminate was investigated with the sphere used subsequently for the impact study. A hardened, 52100 chrome-steel sphere (CCR Products LLC, West Hartford, CT), measuring 12.7 mm in diameter, centrally loaded a 100 mm x 100 mm x 6 mm laminated panel placed on a hardened, A2 steel plate, at a displacement rate of 20 mm min⁻¹, Fig. 5 (a). For HB50, the load, F , monotonically increased with depth of penetration, α , reaching a peak load of 25 kN, Fig. 5(a). Attainment of the peak load coincided with a highly unstable

failure event, similar to that observed during uniform compression, and was accompanied by a large drop in load. This load-to-failure sequence was repeated with continued penetration. Figure 5(a) shows that the BT10m panels exhibited a similar response to penetration and failed at the same peak load, consistent with the uniform compression study. Cross-sectioned HB50 and BT10m samples after 2 mm of penetration are shown in Fig. 5(b) and (c), respectively.

3. Impact study

The penetration behavior during impact of both HB50 and BT10m grade laminates and hybrid samples made up of both grades was investigated using the hardened steel sphere. Some tests were performed with the laminates back-supported on a hard foundation while others utilized edge clamped boundaries that allowed samples to undergo out of plane (transverse) deflection.

3.1. *Sample fabrication*

Preliminary studies indicated the impact of a rear-supported laminate engaged a small area of the target, whereas the entire laminate was engaged in deformation during edge-clamped testing. The rear-supported targets were therefore fabricated to an in-plane size of 100 mm x 100 mm, while the edge clamped targets were 200 mm x 200 mm, Fig. 6. For both target types, the plates were hot-consolidated to a nominal thickness of 6 mm (an areal density of 5.82 kg m^{-2}), which required a lay-up of $[0^\circ/90^\circ]_{50}$ for HB50 and $[0^\circ/90^\circ]_{72}$ for BT10m. A third, hybrid sample was fabricated by replacing the front (impacted side) third of an HB50 sample with BT10m resulting in a $[(0^\circ/90^\circ)_{24\text{B}}/(0^\circ/90^\circ)_{33\text{H}}]$ lay-up.

Fabrication of the edge clamped, BT10m target required the use of a special procedure. A single 100 mm wide tape was unwoven from the BT10 preg, Fig 3(a). To assemble a 200 mm wide ply, a strip of tape was split into two 50 mm wide strips, which were then placed adjacent to either side of a 100 mm wide strip, Fig. 3(b). The strips were held in place by dowel pins and a second set of 50/100/50 mm strips was laid on top, but rotated 90° to the first set, forming a second ply. This process was repeated to assemble the $[0^\circ/90^\circ]_{72}$ lay-up. Once consolidated, the excess length of material used for alignment was removed using a band saw, Fig. 3(c). This fabrication method ensured that the seams between adjacent strips were distant from the impact site.

Prior to testing, a 5 mm spaced square grid pattern was drawn on the front face of all the targets. The black grid was aligned with the reinforcement orientations, and its contrast against the white (Dyneema®) target provided a visual aid for interpreting high-speed video data collected during the impact process.

3.2. *Impact test protocol*

Each rear supported target was clamped to a 100 mm x 100 mm x 20 mm, A2 steel plate (hardened to 62 HRC) using two C-clamps at opposing corners, Fig. 6(a). Each edge clamped target was gripped between two 254 mm x 254 mm x 5 mm steel plates, Fig. 6(b). Each steel plate had a 150 mm x 150 mm central hole with sixteen clearance holes for 10 mm bolts around the periphery of the center hole. Bolts, passing through the two steel plates and the target, clamped the target in place.

The ballistic impact tests were performed at Chesapeake Testing (Belcamp, MD) using the set-up schematically drawn in Fig. 7. The hardened, steel sphere, measuring 12.7 mm in diameter and 8.4 g in weight, impacted the targets normally (0° obliquity) and centrally. The impact velocity, V_i , was calculated from the time of flight for the projectile to travel between two paper break screens. A model v1610 Phantom high-speed camera (Wyane, NJ) recorded an oblique view of the front of the target to observe the deformation of the outer ply. The camera was set to an exposure time of $0.45\ \mu\text{s}$ with an inter-frame delay of $1.52\ \mu\text{s}$, which required the image resolution to be reduced to 128×48 pix. A second model v7 Phantom video camera, recorded a side profile view of the target with an exposure time of $1.8\ \mu\text{s}$ and an inter-frame delay of $18.75\ \mu\text{s}$. This camera provided images for calculation of the incident and residual velocity, V_r , of any projectile that perforated the edge-clamped laminates. The incident velocities measured by the break screens and the high-speed camera agreed to within $2\ \text{m s}^{-1}$ over the velocity range tested.

The ballistic limit, V_{bl} , of each target type (material and gripping condition) was defined as the average of the lowest impact velocity to perforate the target and the highest impact velocity to penetrate, but not perforate, the target. A penetration onset velocity, V_o , was also defined as the average of the lowest impact velocity to penetrate the target and the highest impact velocity that did not penetrate the target. The low launch velocity limit of the gun prevented identification of V_o for the rear-supported targets, so this was obtained by extrapolation of the data.

3.3. *Post-test characterization*

X-ray computed tomography (XCT) was used for non-destructive examination of some partially penetrated targets. Edge clamped targets, which were larger and often contained embedded projectiles, were using a voxel size of 62 μm . Rear supported targets, having a smaller affected area and without an embedded projectile, were examined using a higher resolution, Xradia (Pleasanton, CA, USA) model μXCT -200 machine. The instrument recorded an 8.3 mm diameter sphere with a voxel size of 8.1 μm . Digital stitching of multiple scans provided images of a larger volume. Some rear-supported targets were also sectioned to expose damage along a transverse (Y - Z) plane passing through the impact crater. Each sample was first embedded in Ultrathin 2 Castable Epoxy (Pace Technologies Corp., Tucson, AZ), sectioned by a band-saw, and finally polished using progressively higher grit SiC paper (240, 320, 400, 600 and 800 grit). Samples were then imaged using a Nikon model D7000 camera equipped with a 105 mm Micro-Nikkor lens and an extension tube.

The depth of penetration was measured for each sample and recorded as the parameter, f , defined as the thickness of failed material divided by the initial target thickness. The thickness was calculated by counting the number of plies that had failed.

4. Impact results

4.1. *Rear supported samples*

Nine HB50 and eight BT10m rear supported targets were impacted by a 12.7 mm diameter spherical projectile at velocities of 67 to 299 m s^{-1} . The depth of penetration, represented by the perforated laminate fraction, f , is plotted as a function of impact velocity, V_i ,

in Fig. 8(b). Both materials behaved similarly. For HB50, the depth of penetration increased linearly from $f=0.22$ at the lowest impact velocity, ($V_i = 75 \text{ m s}^{-1}$) to its ballistic limit, $V_{bl} = 215 \pm 10 \text{ m s}^{-1}$. BT10m exhibited the same dependence of penetration depth upon impact velocity, Fig. 8(b), and had a ballistic limit that, within the experimental uncertainty, as the HB50 material, Table 2. The rear supported hybrid target had a similar penetration performance to HB50 and BT10m. It is noted that extrapolation of the penetration data to $f=0$ gives a penetration onset velocity of $\sim 45 \text{ m s}^{-1}$ for the three material systems.

Table 2. Target velocity limits, V_0 and V_{bl} , for edge clamped (EC) and rear supported (RS) tests

Target	Penetration onset $V_0 \text{ (m s}^{-1}\text{)}$		Ballistic limit $V_{bl} \text{ (m s}^{-1}\text{)}$	
	RS	EC	RS	EC
HB50	< 75	296 ± 37	215 ± 10	531 ± 16
BT10m	< 83	176 ± 4	194 ± 12	360 ± 33
Hybrid	< 74	181 ± 32	-	536 ± 13

An optical image of an HB50 target impacted at 75 m s^{-1} and sectioned through the center of the impact site is shown in Fig. 9(a). The projectile created a circular-shaped crater in the target, with failed plies defining the perimeter of the impact crater. The 90° plies (the darker grey shade ply whose fibers are in the Y -direction) that had fractured under the projectile had recoiled and buckled at the side of the impact crater during rebound, leading to a bulging of the region. The 0° plies appear to have thickened at the side of the crater and contributed to the bulging of material around the edge of the projectile impacted region. This bulge zone was about 2 mm in width in the X - and Y -directions at the edge of the crater. The optical image shows no evidence

of any ply damage beneath the impact crater, and was similar to that observed during quasi-static penetration with a punch, Fig. 5(b).

Increasing the impact velocity to 149 m s^{-1} , Fig. 9(b) resulted in a deeper and wider impact crater, and a $\sim 7 \text{ mm}$ wide bulge zone again containing microbuckled 90° plies and thickened 0° plies. Because the sectioned plane in Fig. 9(b) is slightly offset from the mid-plane of the impact site, the maximum depth of the impact crater was slightly wider and deeper than shown. The seven plies immediately below the bottom of the impact crater have thickened (in particular the 0° plies whose fibers were in the X -direction) and had rebounded upwards towards the impact crater. An out-flow of material from the impacted region into surrounding material and towards the impact surface is apparent. Sections through BT10m targets after impact at similar velocities revealed the same failure and deformation modes, Fig. 9(c) and (d). No evidence of shear plug formation could be seen in any of the samples.

Fig. 10(a) shows a μXCT reconstructed Y - Z cross-section along the impact crater mid-plane of a partially penetrated HB50 target. Failed ends from 90° plies are visible near the bottom of the crater². These failed plies had recoiled away from the impact crater into the surrounding material forming small microbuckled folds. This lateral displacement of failed plies was accompanied by delamination between the plies. Fig. 11 shows a similarly imaged BT10m target after impact at a similar velocity. The failure process appears to be identical to that of the HB50 material.

² Single fibers are difficult to identify given their similar density to the matrix and the voxel size of $8.1 \mu\text{m}$.

A magnified view of a transverse section near the bottom of the crater of the HB50 target is shown in Fig. 10(b). The 90° plies, whose fibers run into the page, show considerable thickening from in-plane extrusion of fibers transverse to the fiber direction. Fig. 10(c) shows an in-plane (X - Y) cross-section through the impact crater of a failed 90° ply. A ~1.5 mm wide ply strip had failed, and the fractured ends had been laterally displaced ~2.6 mm from their original position. Fibers to either side of the failed portion were also laterally displaced around the impact crater, with displacements accompanied by ply splitting.

The width of the failed portion of a ply, and that of the crater as a function of depth in an HB50 target impacted at 143 m s^{-1} were also investigated. Fig. 12(a) shows a photograph of the rear side of the fourth ply below the impact surface after removal from the target. The residual opening was square, with a crater width defined in Fig. 12(b). This width was greater than the width of the ply's failed fibers because of significant lateral (X -direction) displacement of the intact Y -direction fibers as facilitated by splitting. Fig. 12(c) shows both the width of the crater and of that of failed fibers as a function of depth below the impact surface. The width of the crater was well approximated by the cord length along the mid-plane of the projectile, while that of the failed fibers was 2 to 3 mm smaller. Similar splitting of the plies was observed in the BT10m targets, and indicated substantial lateral displacement of unfailed material accompanied penetration in both materials.

Fig. 13(a) shows a high-speed image sequence of the oblique front view of a rear supported HB50 target impact at $V_i = 225 \text{ m s}^{-1}$. A vertically displaced fiduciary marker has been identified at $t = 6.1 \text{ } \mu\text{s}$ after impact. In the following frame, the marker has been pushed away

from the projectile, and ply failure is visible by $t = 12.2 \mu\text{s}$. Hence, it can be concluded that the laminate failed at $9 \pm 3 \mu\text{s}$ after initiation of impact.

4.2. *Edge clamped targets*

Nine HB50, twelve BT10m and nine hybrid edge clamped targets were impacted at velocities of 80 to 700 m s^{-1} . The depth of penetration (for impacts below the panel ballistic limits) and the residual velocity for fully perforated samples are shown on Fig. 8. This laminate support condition resulted in substantially higher ballistic limits for both materials and the hybrid laminate, Table 2. For the HB50 laminates, the start of progressive penetration was delayed, from around 40 m s^{-1} for rear supported samples to $296 \pm 37 \text{ m s}^{-1}$ for the edge clamped test. The fraction of failed plies increased with impact velocity, and the laminate was completely perforated at $V_{bl} = 531 \pm 16 \text{ m s}^{-1}$. For the BT10m laminates, progressive penetration began at an impact velocity of $176 \pm 4 \text{ m s}^{-1}$ and the ballistic limit was $360 \pm 33 \text{ m s}^{-1}$; substantially lower than HB50, Table 2. The residual velocities of the perforated targets are plotted in Fig. 8(a).

For the bilayer target, where the front third of an HB50 target was replaced with BT10m, penetration was initially similar to that of the BT10m material located on its impact face, Fig. 8(b). As the impact velocity was increased above V_0 , the depth of penetration ceased increasing at $f \approx 1/3$, and a large delamination plane formed at the BT10m/HB50 interface, Fig. 14(d). The hybrid target impact performance then became governed by the HB50 performance for depths of penetration $f > 1/3$, Fig. 8. The ballistic limits of the hybrid and HB50 targets were the same, Table 2.

A XCT cross-section through the impact site of a BT10m target just above its V_0 limit is shown in Fig. 14(a). The entire laminate permanently deflected in the Z-direction by outward propagation of transverse hinges as depicted in Fig. 1(a). The plies below the projectile remained intact. A microbuckling zone, similar to that seen with the rear-supported targets (Fig. 9), formed within the failed plies adjacent to the perimeter of the projectile. In contrast to rear-supported targets, delamination between plies extended several centimeters outwards from the projectile. Increasing the impact velocity led to more perforated plies, which delaminated during recoil and buckled as they rebounded against the side of the projectile, Fig. 14(b) and (c). The formation of a shear plug was not observed at any impact velocity.

The HB50 laminates suffered a similar deformation and failure sequence as the impact velocity was increased, Fig. 14(f). Separation of the intact plies from the failed plies within the laminate was more pronounced. The ply kinks within the intact portion of the laminate are thought to have occurred after peak deflection as the laminate rang-down to rest. Measurements of the failed ply widths for both materials again showed the failure widths were smaller than the diameter of the projectile, consistent with elastic recoil of the fiber following a tensile fracture. The transverse (X - or Y -direction) width of the failed plies was 5 to 9 mm.

A high-speed image sequence of the oblique front view for an edge clamped, HB50 target after impact at $V_i = 259 \text{ m s}^{-1}$ is shown in Fig. 13(b). Pull-in was present behind the extensional wave front, as observed with the rear-supported target impacted at a similar velocity (Fig. 13(a)), but no plies had failed. Instead, the laminate transversely deflected, as observed by the base of the deflection pyramid outlined in the last frame. An edge clamped HB50 target impacted above

the penetration onset velocity is shown in Fig. 13(c). Evidence of ply failure is visible by $t = 4.6$ μs , and perhaps as early as $t = 3$ μs after the start of impact.

5. Discussion

Ideally, the interpretation of the results would begin with a numerically study of the impact process for the two materials. However, numerical assessments with even approximate constitutive models are greatly complicated by the very high anisotropy present, and the many length and time scales contributing to the response. Instead, we use shock propagation scaling to identify the key physical aspects of the problem, and then investigate the mechanisms by which penetration was facilitated in the two material systems.

5.1. *Onset of penetration*

When supported on a rigid foundation, penetration of both the HB50 and BT10m targets was initiated at an impact velocity of ~ 45 m s^{-1} . When the impact velocity was increased to $V_i = 75$ m s^{-1} , $\sim 20\%$ of the 6 mm thick laminates had been penetrated. Post-mortem analysis of partially perforated samples, Fig. 9(a) and (c), revealed similar damage to the samples quasi-statically loaded by the same diameter hemispherical punch, Fig. 5(b) and (c). The punch penetration tests also showed both materials to have similar indentation responses, with the load, F , monotonically increasing with indentation depth, α , until it reached a similar peak load at failure in both materials, Fig. 5(a).

If all of the kinetic energy from a projectile impact were converted to the work, W , needed to overcome the material's resistance to penetration, the penetration onset velocity, V_0 , can be approximated from a work-energy balance,

$$\int_0^{\alpha_0} F d\alpha = m_p V_0^2 / 2 \quad \text{Eq. 3}$$

where α_0 is the indentation depth at the initiation of ply failure from the punch study and m_p is the projectile mass from the impact study. This analysis is simplified by the preclusion of any bending or stretching mechanisms for the rear support condition, and assumes material rate effects can be neglected. After solving for V_0 in Eq. 3 and integrating the load-displacement curve, Fig. 5(a), the predicted V_0 was $\sim 45 \text{ m s}^{-1}$ for both materials. This is almost identical to the V_0 extrapolated from experimental measurements, Fig. 8(b). The impact velocity required to initiate failure in the rear-supported case is therefore governed by the quasi-static work required to indent the laminate, and given the similar resistance to quasi-static indentation of the two materials, both had similar impact performances, Fig. 8(b).

In the edge clamped condition, the HB50 and BT10m targets were again progressively penetrated, Fig. 8(b), with local failure below the projectile, Fig. 14. However, the penetration onset velocity, V_0 , for both materials was significantly higher than the rear supported case. Furthermore, the V_0 of the HB50 laminate ($296 \pm 37 \text{ m s}^{-1}$) was substantially higher than that of BT10m ($176 \pm 4 \text{ m s}^{-1}$). These differences in V_0 can be reconciled if the onset of penetration is controlled by the dynamic impact pressure.

Consider the impact of a stationary, edge clamped laminate by a rigid projectile that launches a transient compressive shock pulse into the target, Fig. 15. The peak pressure, P_H , scales as $\rho c_T V_i$, where ρ is the laminate's initial density and c_T is the laminate's through thickness shock velocity (which tends to its longitudinal elastic wave speed in the weak shock limit) [34]. As HB50 and BT10m have the same density and resistance to indentation failure, Fig. 5(a), the impact velocity for onset of ply failure, V_0 , would scale with c_T^{-1} if the dynamic pressure controlled the onset of failure. The shock wave speed, c_T , for a highly shocked material depends upon the stress state of a material, but will scale as $\sqrt{E_T}$, where E_T is the out of plane tangent modulus. For a qualitative comparison, the out of plane uniform compressive stress-strain responses shown in Fig. 4(a), can be used to estimate the modulus, E_T . They give a $\sqrt{E_T}$ ratio of HB50 to BT10m at a strain 0.3 of ~ 0.75 . Table 2 shows that under edge clamped loading, V_0 for BT10m was lower than HB50, and the V_0 ratio of the materials was ~ 0.6 ; consistent with the dynamic pressure governing V_0 for HB50 and BT10m when edge clamped.

Recall, this scaling of V_0 in the two materials was not observed under rear supported testing, where the onset of penetration occurred at the same velocity for the two materials. Furthermore, the low value of V_0 in the rear-supported case would result in insufficient shock pressure to cause failure if the same mechanism were responsible for the onset of penetration in the two loading scenarios (Table 2). To resolve this dichotomy, recall that in Fig. 13 the time from initiation of the impact until the first ply failed was substantially greater in the rear-supported case (9 ± 3 μ s versus 5 μ s for the edge clamped boundary condition). Fig. 15(a) schematically illustrates propagation of a shock pulse in a laminate for the two boundary conditions studied. For a rear-supported test case, the shock impedance of the steel support plate

will exceed that of the target, and the compressive shock will be back reflected as a compressive pulse into the laminate towards the projectile, Fig. 15(c). The pressure just beneath the projectile then rises, eventually approaching the quasi-static value for contact between a rigid sphere and the laminate body, Fig. 15(b). In contrast, when the shock front first reaches the rear of an edge clamped target, the low inertia laminate is able to suffer an out of plane back face deflection. A tensile wave is then reflected back towards the projectile, and partially relieves the shock pressure, Fig. 15(d). The onset of penetration for the edge clamped target must therefore occur before this reflection arrives at projectile (at $\sim 6 \mu\text{s}$ after impact for $c_T \approx 2 \text{ mm } \mu\text{s}$), which is consistent with the high-speed observations in Fig. 13.

To summarize, the onset of penetration of rear-supported targets began after sufficient time had elapsed for the stress to approach the quasi-static indentation limit. Consistent with penetration being governed by the materials compressive strength (which is the same for HB50 and BT10m), the penetration onset velocity of the two materials was then the same in rear-supported case. Edge clamped targets required a higher impact velocity than back supported samples because the failure stress of the laminate had to be surpassed before arrival of the back face reflected (tensile sign) shock pulse. Furthermore, the higher tangent modulus of the BT10m material increased the maximum shock pressure at constant impact velocity resulting in a lower V_0 for this material when edge clamped. The generality of these findings are likely to be affected by the target thickness, given its effect upon the transit time of shock pulses that traverse the sample, and by the projectile shape which governs the spatial form of the stress field.

5.2. *Failure mechanisms*

The experimental observations of rear supported tested samples, Fig. 9-12, indicate that the samples failed by tensile rupture of the reinforcement directly under the center of the projectile, and that this occurred under the action of a through thickness compressive stress field, in a loading scenario where the possibility of a membrane-stretching mode of deformation was denied.

A schematic illustration of a failure process that is consistent with these previous observations is given in Fig. 16. It invokes the indirect tension mechanism observed in uniform compression studies in combination with a lateral displacement mechanism [30]. Fig 16(a) shows the early contact situation on the Z - Y plane prior to reinforcement rupture. The projectile impact creates a contact stress with strongly compressive character directly beneath the projectile. The anisotropic Poisson expansion of 0° plies in the Y -direction loads the 90° plies in tension, and vice versa, with the highest tensile stress occurring in the 90° ply closest to the projectile. When this 90° ply fails in indirect tension, Fig. 16(b), the ply recoils from the fracture site, and the projectile is able to advance into the sample, but with a velocity reduced by the work done in advancing a ply thickness against the compressive resistance of the material. The eventual rebound of the fractured ply results in a series of buckles forming to accommodate the length of ply impeded by the projectile's presence (half cord length at the original ply depth).

An alternate mechanism that could have activated tensile rupture would have been if the extensional strain required of the plies to conform to the contact length of the projectile, Fig. 16(a), reached the failure strain of the ply ($\sim 2\%$). Since failure initiation under both quasi-static loading and impact loadings were the same when rear supported, we examine the quasi-static

penetration case, which failed after 1.5mm of indentation, Fig. 5(a). This corresponds to a projected loading area of 8.2 mm and an increase in the length of the reinforcement under the projectile of 0.7mm. The reinforcement, however, is likely to be pulled inwards, towards the projectile contact surface from outside the loading area since the interlaminar shear strength is so low, Table 1. For HB50, the load sustained by shear per projectile unit width is 35 N/mm, while 110 N/mm would be needed to reach the ply failure strength (1.7 GPa). Therefore, reinforcement will pull in to relieve the strain regardless of indentation depth. Furthermore, the lower ply strength and higher shear strength of BT10m would result in a lower penetration resistance than for HB50. The identical penetration versus impact velocity response of back-supported HB50 and BT10m materials hence suggests penetration is controlled by the laminate compressive strength and not the ply tensile strength.

Since the plies below the sides of the projectile contact surface do not reach as high of a pressure, they do not undergo indirect tension failure. Instead, as schematically illustrated in Fig. 16(b) and (c) and observed experimentally in Fig. 10(c), they are laterally displaced in a manner analogous to that observed during ductile hole enlargement in some metals; though here material displacement required ply splitting and delamination, as observed with the μ -XCT technique. It has been argued that reinforcement slip lowers a laminate's resistance to penetration [30]. This may also be the case here since penetration by indirect tension requires application of a stress comparable to the reinforcement strength (of order 1 GPa) while lateral displacement is matrix strength controlled and can be activated with a stress of a few MPa (a 10^3 -fold difference).

Examination of the edge clamped targets, Fig. 14, indicated two mechanisms of response. Initially, projectile penetration occurred by tensile rupture of plies directly beneath the nose of the projectile in combination with lateral displacement near the edges of the projectile. The experimental observations of this process, Fig. 14(a) to (c) and (f), were consistent with the mechanism observed in a back supported test scenario. However, unlike the rear-supported tests, the unfailed portion of partially perforated laminates suffered large out of laminate plane deflections and had delaminated from the perforated part of the laminate. A schematic illustration of the proposed failure mechanisms are shown in Fig. 17. Upon initial impact, Fig. 17(a), the dynamic pressure promptly induces indirect tension fracture of plies directly beneath the center of the contact surface). Once the compressive shock pulse had propagated across the sample, Fig. 16(d), the laminate could begin to move, and the pressure exerted by the projectile then drops because of the work it had performed to penetrate the first plies, and because the impact velocity in the local laminate frame of reference began to decrease due laminate acceleration. As the intact (1- f) fraction of the laminate continued to deflect, a membrane stress developed within it. The work done in extending the laminate against the membrane stretching resistance could then further retard the projectile motion. It is noted that during the projectile arrest phase, a local contact and membrane stress exists, and the combined stress may have enabled additional penetration during laminate deflection, Fig. 17(b).

The edge clamped hybrid target, with the front third of the target thickness consisting of BT10m and the remainder HB50, had the same ply perforation onset velocity, V_0 , as BT10m, Fig. 8(b). This is consistent with V_0 being governed by the dynamic impact pressure. After penetrating the BT10m portion of the hybrid, the depth of penetration followed the same

dependence with impact velocity as the higher performance HB50 target. These results suggest that a benefit may exist to a multi-material composite lamination, and several groups have investigated multi-material composite concepts [35–39]. The present study suggests the placement of plies with the lowest impedance and the greatest resistance to indirect tension (compressive strength) nearest the impact side, where deflection is minimal, would be beneficial. The higher tensile strength plies would be placed nearest the rear side of a laminate to sustain higher membrane stresses. It is also interesting to note that for some projectile nose geometries, use of a material with too weak of an inter-laminar shear strength may result in low ballistic resistance with failure dominated by lateral ply displacement.

6. Concluding remarks

An investigation of the deformation and fracture mechanisms that accommodate progressive projectile penetration of $[0^\circ/90^\circ]$ UHMWPE reinforced plastic composites has been performed. The study exploited differences in the response of HB50 fiber- and BT10 tape-reinforced grades of Dyneema[®]. The HB50 and BT10m laminates had similar transverse (out-of-plane) compressive strengths, while HB50 had a 40% higher ply tensile strength. Ballistic targets, measuring 6 mm in thickness, were made from each material as well as from a bilayer hybrid, where the front third of the thickness was BT10m. The impact of each target by a 12.7 mm diameter, hardened (non-deforming) sphere was examined in test scenarios where the target was either supported on a stiff foundation to prevent transverse deflection or clamped along its edges to permit out of plane deflection.

When rear supported, the velocity needed to initiate penetration and the dependence of the depth of penetration upon impact velocity were the same for the two materials consistent with material compressive response governing penetration. Indirect tension of the reinforcement and lateral displacement of material from the tip of the projectile accommodated penetration. No shear plug mechanism was observed with either material.

When end supported, both materials had a higher penetration onset velocity and a higher impact performance compared to the back supported test case. The velocity of penetration initiation scaled inversely with shock pulse speed (HB50 required a higher velocity to initiate ply failure than BT10m); while the ballistic limit increased with the tensile strength of the material. The samples exhibited progressive failure as the impact velocity increased, and failure was initiated rapidly before shock pulse reflections from the back face reached the region of contact. The mechanism of progressive penetration was consistent with indirect tension failure in combination with lateral displacement at the sides of the projectile and membrane stretching of the unfailed portion of the laminate. These observations suggest future investigation of multi-material (hybrid) laminates that combine materials with highest compressive strength and lowest impedance near the impact site and those with highest tensile strength at the rear may result in significant impact performance benefits.

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Figure 1. Schematic illustrations of the response of (a) thin and (b) thick $[0^\circ/90^\circ]$ laminates following impact by a spherical non-deforming projectile. Progressive failure of a thick laminate occurs by (c) shear plugging or (d) indirect tensile failure of the fibers due to anisotropic Poisson expansion of adjacent plies in a compressed $[0^\circ/90^\circ]$ laminate (e) or (f) lateral displacement of the reinforcement.

Figure 2. Optical micrographs of the X - Z plane cross-sections of (a) an HB50 and (b) a BT10m consolidated laminate.

Figure 3. Schematic illustration of the fabrication of the BT10m edge clamped sample: (a) The as-received BT10 grade plain woven tape was peeled apart into its constituent single ply tapes; (b) the tapes were layered to form a $[0^\circ/90^\circ]_{72}$ lay-up; (c) after hot consolidation, tabs used to maintain alignment were removed.

Figure 4. (a) Uniform compressive stress versus strain responses for $L = 12$ mm BT10m tape and HB50 fibrous laminates with $[0^\circ/90^\circ]$ lay-up orientations. (b) The variation of the uniform compressive strength, σ_c , with in-plane sample length, L , for the two laminates.

Figure 5. (a) The indentation load, F , as a function of indentation depth, α , when a 12.7 mm diameter sphere was pushed into a rear supported laminate. Optical micrographs of cross-sectioned HB50 (b) and BT10m (c) samples loaded to $\alpha \approx 2$ mm.

Figure 6. Schematic illustrations of $[0^\circ/90^\circ]$ laminated target plates situated for an impact while either (a) rear supported or (b) edge clamped.

Figure 7. Schematic illustration of the steel sphere impact experiment using an edge clamped sample geometry.

Figure 8. (a) The residual velocity, V_r , and (b) the perforated laminate fraction, f , are plotted as functions of the impact velocity, V_i , for both rear supported and edge clamped targets for the two materials and the hybrid system. Solid lines are drawn to help reveal trends.

Figure 9. Optical micrographs of cross-sectioned, rear supported targets. Micrographs (a) and (b) correspond to HB50 targets impacted at 75 m s^{-1} and 149 m s^{-1} , respectively, while (c) and (d) correspond to BT10m targets impacted at 84 m s^{-1} and 152 m s^{-1} , respectively. All micrographs are at the same magnification.

Figure 10. (a-c) Orthogonal micro-XCT reconstructed cross-sections of a rear supported, HB50 target after impact at $V_i = 117 \text{ m s}^{-1}$. Section planes are identified in (d).

Figure 11. Micro-XCT reconstructed cross-section of a rear supported, BT10m target after impact at $V_i = 115 \text{ m s}^{-1}$.

Figure 12. (a) Photograph of in-plane (X - Y) section of a ply ($u_z = -0.24 \text{ mm}$) peeled from a rear supported, HB50 target after impact at $V_i = 143 \text{ m s}^{-1}$. (+) Z is into page. (b) Schematic illustration used to define the crater and failure widths. (c) The measured widths are plotted as functions of the original Z -position, u_z , relative to the target's impact surface.

Figure 13. High speed image sequences showing the oblique front view of the following three HB50 targets: (a) rear supported and impacted at 225 m s^{-1} , and edge clamped and impacted at either (b) 260 m s^{-1} or (c) 505 m s^{-1} . Grid lines are spaced 5 mm apart and are aligned with the $[0^\circ/90^\circ]$ ply orientations.

Figure 14. XCT reconstructed cross-sections of edge clamped (a-c) BT10m, (d-e) hybrid and (f) HB50 targets.

Fig 15. Schematic illustrations of laminate impact by a non-deforming spherical projectile. (a) A transient pressure wave with amplitude P_H is launched into the laminate. (b) A contact stress, P_S , is formed near the laminate surface. (c) Reflection of the compressive shock at a laminate – rigid

foundation interface doubles the pressure of the reflected compression shock. (d) Reflection at the rear of the edge clamped laminate results in a sign change (compression to tension) to the pressure pulse.

Figure 16. (a) Ply failure and lateral displacement beneath the contact region are necessary to facilitate projectile penetration. (b) The compressive stress imposed by the projectile indirectly loads the plies in tension causing ply tensile failure, lateral recoil and rebound against the projectile. Penetration is also accompanied by local delamination, ply thickening and laminate bulging after laminate penetration. (c) A schematic illustration showing the region of *X*-direction fiber fracture, micro-buckling of the rebounded ply and *Y*-direction ply displacement at the sides of the contact.

Fig 17. Penetration mechanisms of an edge clamped target. (a) The case shortly after initial impact before shock reflection at the rear face. (b) Shows the situation after the shock pulse has traversed the laminate and the intact portion suffers out of plane deflection that is resisted by membrane stretching.

(a) Thin laminate

Figure 1

(b) Thick laminate **(c) Shear plugging** **(d) Indirect tension** **(e) Poisson expansion** **(f) Lateral displacement**

(a) HB26

Figure 2

(b) BT10m

(a) Figure 5

(a) Rear supported

Front view

Figure 6
Profile view

(b) Edge clamped

Rear view

Profile view

All dimensions are in mm

Figure 7

Figure 8

Figure 11

(a) Figure 12

(b)

(c)

(a) Rear supported,
 $V_i = 225 \text{ m s}^{-1}$

(b) Edge clamped,
 $V_i = 259 \text{ m s}^{-1}$

(c) Edge clamped,
 $V_i = 506 \text{ m s}^{-1}$

Figure 14

BT10m samples

10 mm

Hybrid samples

HB50 sample

(a) Figure 16

(b)

(c)