JOURNAL OF MATERIALS SCIENCE **34** (1999) 5641 – 5648

Impact characterization of RTM composites-II: Damage mechanisms and damage evolution in plain weaves

VISTASP M. KARBHARI
Department of Structural Engineering, MC-0085, University of California, San Diego, La Jolla, CA 92093, USA
E-mail: vkarbhari@ucsd.edu

RICHARD W. RYDIN
The Why Not Corporation, Springfield, OH 45502, USA

Resin Transfer Molded Composites exhibit impact induced damage mechanisms and sequences different from those shown by laminated composites due to differences in layering and compaction of reinforcement. In addition to the classical modes of damage such as matrix cracking, delamination and fiber breakage, mechanisms such as inter-bundle, intra-bundle and void pocket cracking are also seen. The presence of damage types relative to the impact regime is discussed with reference to regions of the Inelastic Energy Curve for Impact. Fiber tow level based damage evolution is also investigated, and damage mechanisms and sequences are elucidated for E-glass/vinylester plain weave based composites. © *1999 Kluwer Academic Publishers*

1. Introduction

A critical issue in the design of composite structures is that of impact response to loadings typified by relatively high contact forces concentrated over a small area and of short duration. The fracture processes are diverse, depending on factors such as geometry and load configuration, materials, and fiber orientation/architecture, resulting in a complex set of event-response combinations, each of which can result in a different level of energy absorbing capability. With the increased use of composites, issues related to the determination and prevention of impact induced damage become more important, and there is an increasing need to develop an understanding of damage phenomena at the materials level. Experimental observations of damage in composites, induced by dynamic loadings such as low velocity impact began in the early 1970s [1, 2] and have become a focal point of many investigations during the last two and a half decades [3–9], during which laminated composites, representative of the thin-skin aerospace world were intensely studied. However, it was noted that tailorability in these composites was always limited by the predominant effect of delamination between plies, even after substantial modification through the use of interply toughening layers and hybrids. The use of tailored textile structural composites, even with 2D architectures, through the use of the Resin Transfer Molding (RTM) process, however, allows the composites designer greater latitude through the development of damage mechanisms and sequences tailored through fabric architecture in addition to the materials and orientation levels afforded with laminated composites.

At this stage it is worthwhile reviewing the major differences between laminated composites and composites fabricated by the RTM process. Laminates are simply stacks of preimpregnated material consisting of planer (or sets of) reinforcement encased in a layer of resin. During a controlled temperature and pressure cycle in an autoclave, individual laminae are diffusion bonded into an evenly distributed continuous composite structure as illustrated in Fig. 1. Reinforcement in adjacent layers does not intrinsically make intimate contact, thereby developing a layered (or laminated) system with distinct layering of resin impregnated tows and neat resin. If processed appropriately, the microstructure is ordered, and individual laminae are distinguishable, both physically and as related to damage initiation and evolution. In RTM, however, dry fabric layers are compacted in the tool cavity before the resin is injected, bringing fabric layers into varying degrees of intimate contact. The compaction step allows for the nesting of neighboring layers and for mechanical interlock at various regions between layers. Consequently there is a lack of distinct resin rich interlaminar zones, resulting in a structure that shows no distinct macro-scale regions of differentiation between layers, with structure varying locally depending on aspects of fabric architecture locally in contact (Fig. 1). The establishment of a representative unit cell is still possible, however a statistical averaging procedure is generally recommended due to local architectural variation. However, it may differ in subtle fashion based on geometrical positioning of the architectural aspects of adjacent layers. Resin zones will also vary in thickness, not just from

0022–2461

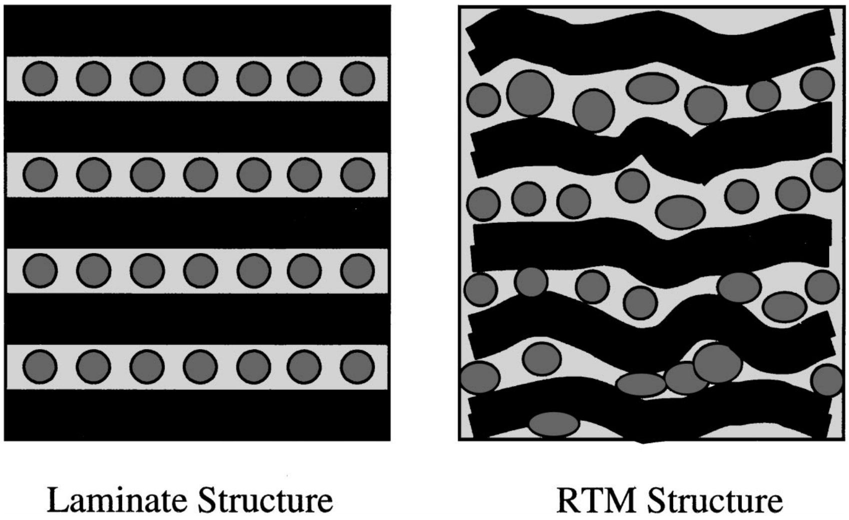


Figure 1 Schematic comparison of laminate and RTM based microstructures.

one layer to the next, but also within layers creating a structure more representative of a fibrous structural system than a uniformly distributed laminated material. Also, since resin infusion and impregnation occur after compaction, there is a greater possibility of local zones having incomplete wetout, resulting in a unique defect morphology.

This paper elucidates damage mechanisms and their growth in Resin Transfer Molded Composites with special emphasis on plain weave fabric architectures. In that vein, the focus is on the identification and description of damage mechanisms, with the comparison of impact response being for purposes of clarity alone with the reader being referred to [10–13] for further details of methods and processes.

2. Classical modes of damage

A major feature of the low velocity impact regime, such as characterized by a drop-weight impact test, is that both structural and materials response must be considered in analysis since both global plate reaction and local contact indentation reaction contribute to the overall response. Impact energy can overall be dissociated into three pools of energy, i.e. stored energy, absorbed energy and dissipated energy, of which the first two result in materials level response through damage, whereas the last is associated with mechanisms such as frictional sliding and damping losses. Most laminates respond to a low velocity impact event by bending or local compression and shear, with the dominant damage mechanisms being matrix cracking, fiber fracture, and delamination, of which the latter is the ultimate mode of failure seen in most cases. Two different delamination initiation mechanisms can be identified, both originating from matrix cracks which propagate to an interface (or resin rich zone) between reinforcement layers, as shown in Fig. 2. Transverse shear resultants, transverse normal forces and excessive bending deformation lead to such a mode with the bending stiffness mismatch and reinforcement orientation differences between adjacent plies controlling its growth [1, 4].

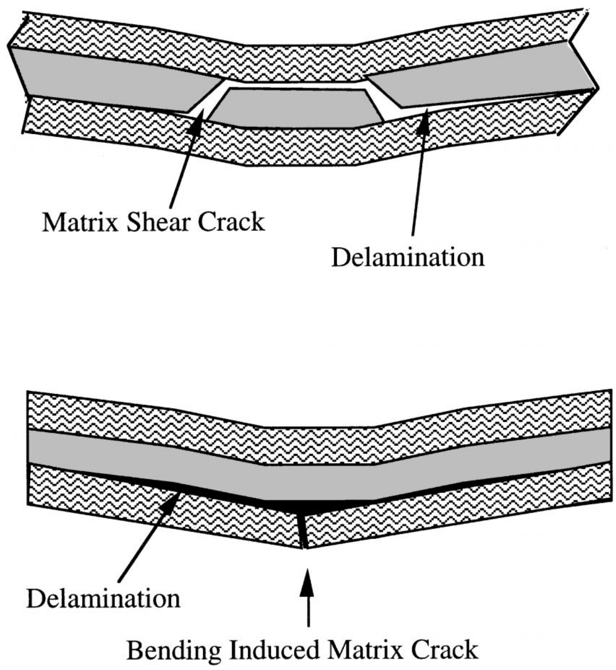


Figure 2 Delamination mechanisms in laminated composites.

Previous work on woven fabric systems [14] suggests that delaminations may also be initiated from microcracks originating at localized regions where fiber debonding has taken place. The driving force for debonding is excessive shear stress developed during load transfer resulting from an impact event. Matrix cracking is a complex fracture process that depends on local geometry and the externally applied stresses. Matrix cracks parallel to the reinforcing fibers are a manifestation of a low energy fracture path which causes crack blunting, whereas those at 45° are typically representative of compressive shear bands originating under the point of impact. Transverse cracks can be attributed to tensile stresses developed by large membrane reaction displacements. In light of the microstructural differences between laminated composites and RTM composites, especially as related to the previously mentioned phenomena of interply nesting and fibril entanglement, it is important to note that although layer based mechanisms of delamination still exist in high volume fraction RTM composites, they are infact replaced in prominence by other and combined mechanisms at the tow/bundle level.

3. Damage evolution in RTM composites

As discussed previously, damage mechanisms and modes in RTM Composites are a consequence of the fabric packing and architecture specific to such materials. In this section, we describe the evolution of damage following the three distinct regions described by Inelastic Energy Curves [10], using plain weave E-glass fabrics infused with a Vinylester resin system (Dow Derakane 411-C50) as the sample materials system. All samples were fabricated as flat plates cured in the tool itself at 98 °C for 30 min, followed by a postcure for 3 h at 120 °C. Impacts were administered using a hemispherical indentor on an instrumented drop weight impact tower. The IEC curves [10–12] serve as a graphical means of providing interrogational sensitivity to the characterization of materials impact response through the plotting of returned energy. Three major zones of response can be identified.

- *Region I*—purely elastic in nature with a one-to-one correspondence between incident and returned energy, and characterized only by superficial damage such as dimpling under the point of impact and minor matrix crazing.
- *Region II*—linear relationship between incident and returned energy with localized visible damage continuing up to the Linear Inelastic Limit (LIL), at which point damage is more extensive and the linear relationship no longer holds.
- *Region III*—embodies the puncturing of multiple fabric layers with layer separation as a result of prior, and consequent, damage development. Impact response is not highly predictable and can be greatly affected by local fabric architecture, damage state, and fabric-impactor geometry interactions.

3.1. Damage mechanisms in regions I and II

Region I of the IEC is characterized by a one-to-one correspondence between incident and returned energy, with little indication of permanent global deformation or damage accruing at the back surface. In Region II, further damage is seen as a form of energy absorption, but the overall integrity of the plate is not compromised. Three major types of damage are observed in this region of initial energies (i) crazing, (ii) limited fiber bundle fracture, and (iii) different combinations of debonding and matrix cracking comprising Intra-Bundle, Inter-Bundle and Void Pocket cracking.

Crazing: Crazes form in polymers under tensile loading when micro-voids are nucleated around microscopic and submicroscopic inhomogeneities which become sites of high stress concentration. These microvoids are frequently unable to coalesce into a true crack since they are stabilized by plastically deformed fibrils that bridge the craze. Cracks propagate slowly through the interpenetrating system of voids that form a craze since load may be transferred across the craze faces. The resultant crack jumps back and forth between faces depending on the highest stress at the fibril roots, leaving a characteristic hackle zone. Crazes may initiate and break down into cracks at stresses well below the bulk shear yield limit so they are not uncommon as an initial form of impact damage which leaves a characteristic haze due to a lower refractive index.

At low impact energies, crazing is observed to occur primarily within fiber bundles in plain weave systems, possibly due to residual stresses and the presence of microvoids resulting from incomplete wet out especially at regions of tow cross-over. Crazing within fiber bundles can lead to brittle fracture as cracks nucleate at intersections with free surfaces including the bundle surface. Several types of Intra and Inter-Bundle cracks may nucleate from bundle crazes. A physical artifact of crazing is the formation of a haze which combines with other forms of damage to provide a damage area which appears much larger than it really is, and thereby giving an inaccurate means of quantifying damage in these composites (using conventional Projected Damage Area plots).

A unique form of cracking/crazing is observed in the top ply of plain weave fabric reinforced systems. Thin, parallel, evenly spaced, but non-continuous cracks propagate along the surface of warp (or weft) tows at impact energies below the threshold value of K_1. The use of dye penetrant shows that the cracks have very little interaction with adjacent tows and are mainly formed along, and at the surface of, individual bundles. Although these cracks appear to be very similar to surface shrinkage cracks appearing in thick section nonwoven RTM composites, they are seen only after the impact event. These cracks could very well be due to the release of residual stresses developed within individual fiber bundles by a combination of cure shrinkage and thermal expansion mismatch between the glass fibers and the vinylester resin. It should however be mentioned that in a number of cases, crazing that appears after low energy impacts, disappears after the application of post-static compression or a second impact at a level sufficient to cause through thickness penetration.

Intra-bundle and inter-bundle cracking: The three dominant modes of damage seen in Region II represent combinations of fiber debonding and matrix cracking. A number of these are shown in the cross-sectional micrograph of a plain weave E-glass vinylester composite shown in Fig. 3. Region (1) depicts a matrix void through which a crack has passed, regions (2) and (3) depict intrabundle cracking, and region (4) depicts interply separation. The distinguishing feature of intrabundle cracks is that they are contained within a fiber bundle and propagate distances approaching several centimeters along the bundle as it undulates within the layer of fabric. In some cases multiple cracks parallel to each other are seen with almost uniform crack spacing and very thin crack widths, which in combination with the signal attenuation through glass preforms,

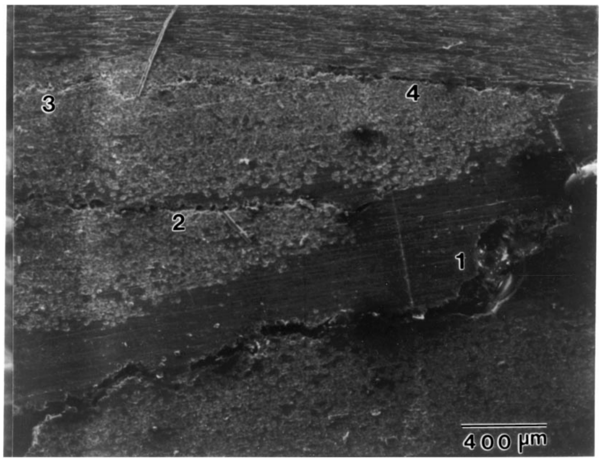


Figure 3 SEM micrograph showing the location of four distinct types of impact damage in a cross-section of an E-glass plain weave-Vinylester composite (40 X).

makes them transparent to normal incidence acousto-ultrasonic waves. These small cracks can serve as initiation sites for more pronounced interbundle cracking and interply separation. Intrabundle cracks can be classified into two types as shown schematically in Fig. 4, of which type I cracks (Fig. 5a) are inclined to the bundle axis, whereas type II cracks run parallel to the bundle axis (Fig. 5b), often being formed at the interface between two tows in intimate contact. The use of dry (unlubricated) compaction in RTM leads to the nesting of bundles from adjacent layers resulting in the interpenetration of individual fibers and the formation of a combined macro-bundle. Type II cracks are restricted to propagation distances of no further than the length of a single unit cell in woven fabrics due to the undulation of warp and weft tows. Based on limited investigation of their initiation and growth, it appears that Type II cracks are formed due to the relative motion of adjacent reinforcing layers in response to an impact event. Type II cracks are also seldom observed in isolation, existing rather in combination with Type I cracks. In the case of a macro-bundle formed by the compaction and interpenetration of fibers from two adjacent bundles, the impact event provides impetus for cracking along the weak plane or along a compaction gradient.

The evolution of intrabundle cracks can be related to the deformation of the bundle itself which can be considered as a composite with very high fiber volume fraction (Bundle fractions can approach 80%, whereas the fiber volume fraction in the composite is typically

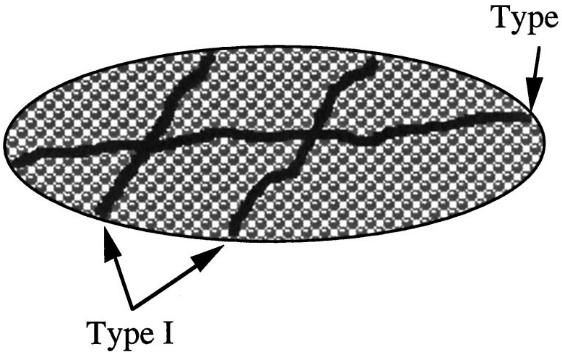
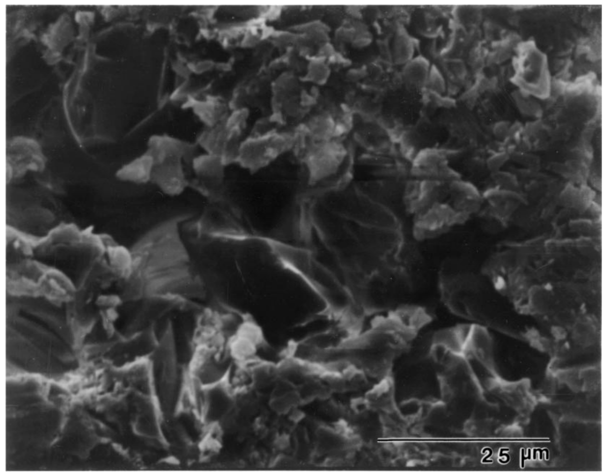
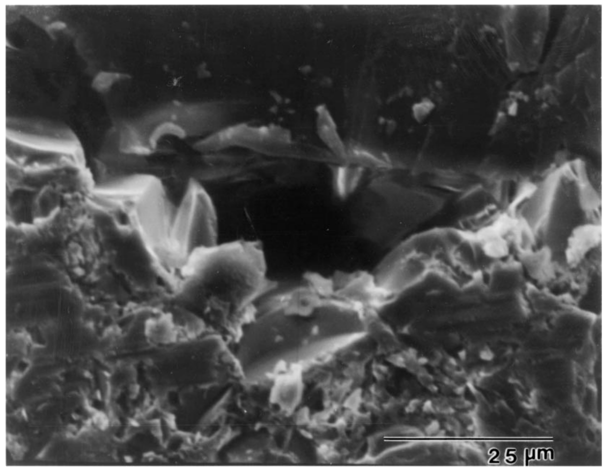


Figure 4 Schematic showing types of intra-bundle cracks.



(a)



(b)

Figure 5 (a) SEM micrograph of Type I Intrabundle cracking in a plain weave reinforced system (1250 X), (b) SEM micrograph of Type II Intrabundle cracking showing separation close to a void in a bundle (1250 X).

between 45–65%, depending on the type of fabric and compaction pressure used). The commensurably small volume fraction of matrix and sizing in a bundle facilitates the development of critical strains during deformation. Under the case of load transfer through the matrix at bundle crossovers in a plain weave, transverse tensile stresses result in pulling the fibrils in the bundle apart, hence causing the formation of multiple parallel cracks, through redistribution of fibril spacing. Further, under stress, bundles deform to accommodate load. As the load is applied, the bundle aspect ratio can change under the assumption that in highly packed fiber bundles, individual fibrils are elastically strained whereas the small volume of the connective matrix phase must undergo significant plastic deformation. The 45° incline of Type I cracks suggests a shear type failure initiated through such a mechanism.

Interply separation or interbundle cracking is shown in the schematic in Fig. 6a, and appears as a result of crack propagation between bundles, in a local region, between separate layers of reinforcing fabric. Interbundle cracks (Fig. 6b) are similar in form to Type II intrabundle cracks except that they closely resemble delaminations within matrix rich zones. An interbundle fracture surface is shown in Fig. 6c, comprising regions

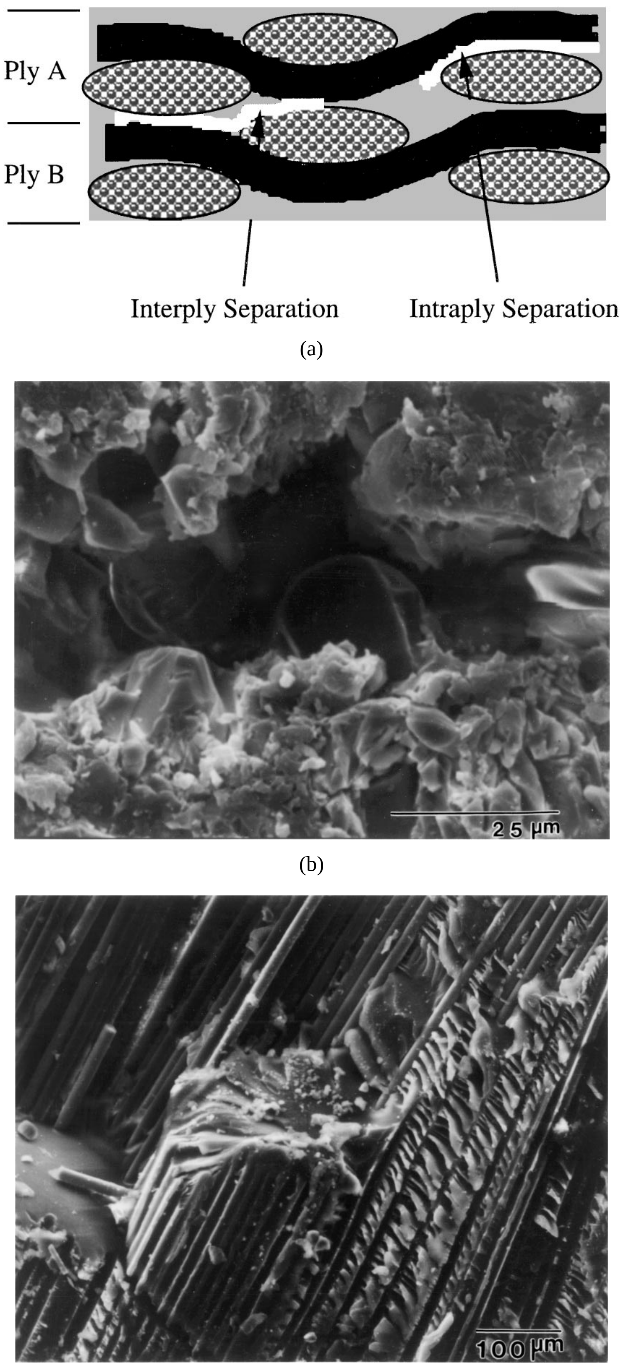


Figure 6 (a) Schematic showing Interbundle cracking, (b) SEM Micrograph of interbundle cracking (1250 X) and (c) SEM Micrograph of interbundle fracture surface showing combined debonding, matrix hackles and void pockets (160 X).

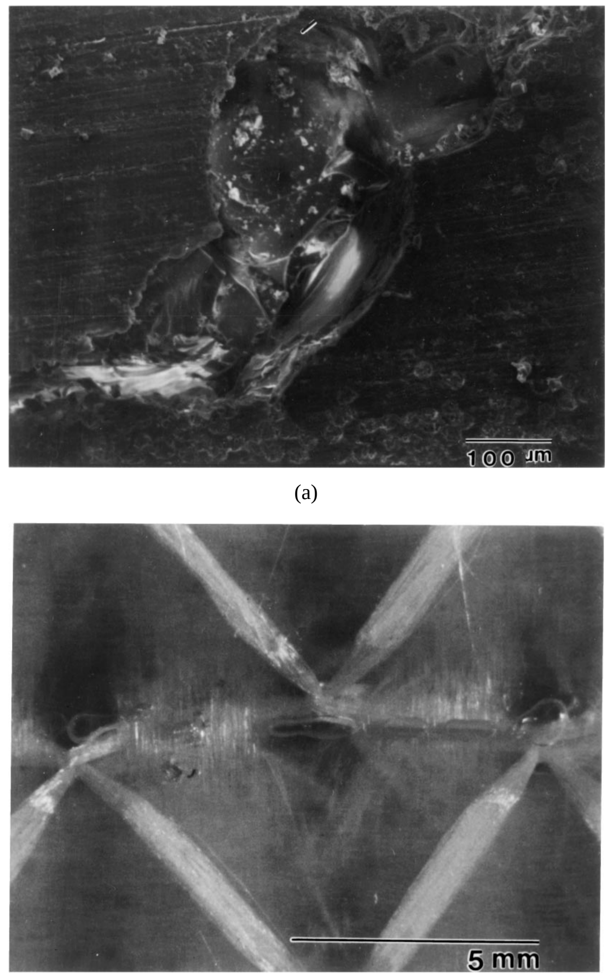


Figure 7 (a) SEM micrograph of cracks emanating from a void pocket in a plain weave fabric reinforced Vinylester composite (160 X), (b) Macrograph of Void pockets adjacent to knitting threads in a non-woven "knit" systen (8 X).

of fibril debonding in Mode I fashion from thin local resin interlayers interspersed with matrix rich regions which are either hackled due to craze nucleated cracking or are apparently strained but morphologically unaffected. Since fiber bundles in woven fabric are mechanically interlocked within layers and adjacent layers are nested due to the RTM process, there is no clear driving force for complete separation akin to delamination, but isolated sites of partial debonding do occur. Local matrix areas are hackled in appearance, indicating shearing after extensive plastic deformation.

Matrix void pockets: While microvoids are responsible for craze formation, macrovoid pockets serve as crack nucleation sites. Ellipsoidal voids of various dimensions and shapes are routinely observed scattered throughout a typical cross-section. A relatively large ellipsoidal defect shown in Fig. 7a has a major axis length of approximately 0.2 mm. Void pockets are found in resin rich regions formed adjacent to the intersection of bundles and are debilitating in that under low impact forces cracks may emanate from more than one location as shown at region 1 in Fig. 3. The degree of stress concentration is linked to the size and shape of the defect, which depends on the void location and fabric type. It should be noted that because of fabric architecture, the morphology and damage mechanisms accruing from these voids are different from voids commonly observed in the vicinity of knit threads that hold biaxial knits together in non-woven fabric systems as shown in Fig. 7b, wherein they may actually have greater potential as sites for crack initiation, than in woven fabric based composites.

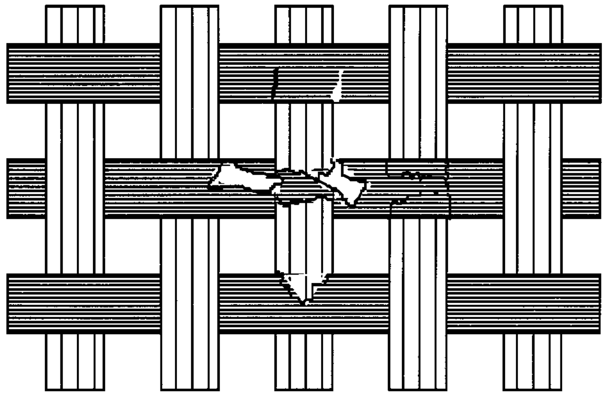
Void formation during thermoset processing depends primarily on resin purity, thermal gradients, production of volatiles, cure shrinkage and degree of wet out. A perfectly consolidated part should be void free while less that 1% by volume is considered ideal under laboratory conditions even for vacuum bagged composites.

Voids are an integral part of commercial composites so it behooves an understanding of their influence on material behavior, especially impact resistance. Cracks propagate depending on the local residual stress state and stress distribution during plate deflection. Since neither of these are simply derived it is difficult to completely assess the influence of voids on energy absorption within Regions I and II of the IEC.

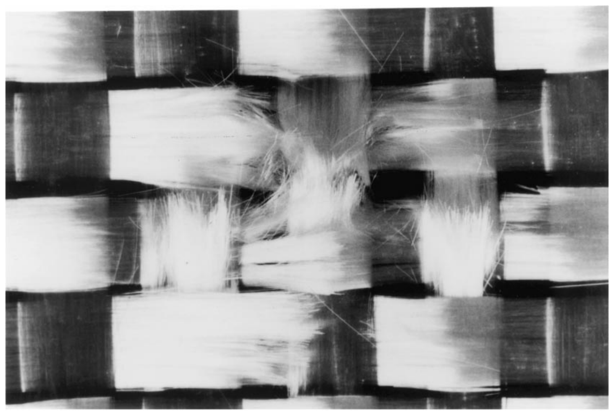
3.2. Characteristic damage within region III

As described in [10, 12] returned energy, projected damage area (PDA) and peak contact force, all deviate from the model of linear increase with impact energy levels, upon entering Region III of the IEC. Damage accumulation can be both local (under the point of impact) or global (due to extensive fiber breakage and ply separation), but is heavily influenced by the local fabric geometry, reinforcement characteristics, and impactor-fabric interactions.

Whereas limited fiber fracture is visible directly below the point of contact between the tup and the composite at lower energy levels (Regions I and II), fiber fracture of the form shown in Fig. 8 is now extensively seen. Fiber bundles may fracture either across the cross-section or with extensive pullout of fibers from either side resulting in greater energy absorption and a longer crack path with further fracture moving across the warp and weft directions. Woven fabric penetration is a sequential process of bundle fracture with the load path along the principle reinforcing bundles below the point of impact determining which neighboring bundles will fracture. As shown in Fig. 8b, fiber bundles can commence fracture from the middle as a result of excessive curvature strain from the advancing tup, or through progressive tearing from the edges. A typical distribution of impact damage in a plain weave system is illustrated in Fig. 9 which superposes projected damage area and the number of bundle fractures as a function of impact energy level. It is seen that very few bundles break before the attainment of the Linear Inelastic Limit (LIL) which corresponds to the end of Region II. However, as discussed in the previous section, considerable fiber-matrix debonding and local shearing of the matrix occurs. As impact energies increase beyond LIL there are continuous jumps in the number of broken bundles resulting from sequential layer penetration and resulting bundle fracture.



(a)



(b)

Figure 8 (a) Schematic depiction of bundle fracture in plain weave fabric, (b) Bundle fracture in a layer of plain weave fabric as seen after ashing of the resin.

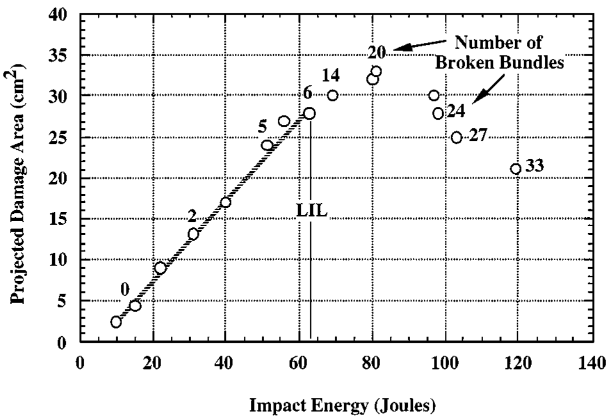


Figure 9 PDA and bundle fracture as a function of impact energy.

During the impact event, at higher energy levels, the impact tup physically penetrates the reinforcing layers through a combination of fiber breakage, fiber and matrix crushing, and slippage between gaps in the overall reinforcement architecture. A loosely held architecture may allow movement of bundles enabling puncture without significant fiber breakage, whereas a tighter and more closed and interwoven architecture would resist tup penetration to a greater extent. Weakly linked bundles in a nonwoven fabric may be pushed aside by the indentor/tup much more easily than the interlocked tows of a tightly woven fabric. This lateral movement of fibers in a bundle is shown in Fig. 10 which shows movement and snapping of individual fibers within an 18 oz/sq yd non-woven fabric bundle after the indentor/tup pushed the bundle out of its path during penetration. It should, however, be noted that although the latter results in an enhanced level of energy absorption through fiber and bundle fracture, the movement of fiber bundles in the former case results in more global ply separation and delamination, which in themselves are major energy sinks. As described in [12], the use of nonwoven fabrics with chopped strand mat backing can result in good impact resistance as well, but due to combinations of different mechanisms of damage.

The size of the impactor relative to the fabric architecture and bundle size determines resistance to ply penetration. A small number of fiber tow crossovers per representative length, as is shown by the use of heavier

Figure 10 SEM micrograph of fibril snapping during ply separation in a nonwoven fabric reinforced Vinylester composite (160 X).

tows, often results in larger intertow gaps which are not conducive to resisting penetration of a projectile between these gaps. As the size of the projectile increases in comparison to the fabric unit cell, more bundles are involved in the initial response, and they are also more tightly interlocked, resulting in greater resistance to penetration through slippage and movement. A special case of this is in the "Promat" fabrics [12] wherein the layer of chopped strand mat backing on each layer increases the nesting and interpenetration of fibers in adjacent layers of nonwoven fabric, causing two or more layers to behave as one in resisting penetration.

Sequential ply failure also relates to the combined effect that stacked reinforcing layers have on ply separation. If an upper layer shields lower plies from localized indentation stresses through separation, damage progresses in a layer by layer manner, with damage intensity and amount decreasing from top to bottom through the thickness. However, if adjacent plies are highly coupled, or do not shield the lower plies in sequence due to an open architecture, fiber bundle damage due to indentation and penetration accumulate continuously through the thickness. In some cases this could result in excessive damage in localized areas at energy levels well below LIL. In many composite systems, reinforcement layer penetration is thought to be resisted in part by the accumulation of debris from upper plies which is forced ahead of the impactor as it penetrates deeper into the plate. This results in an increase of damage area with depth analogous to the plugging action seen in monolithic materials as described by Backman and Goldsmith [15] wherein the mass of the sheared target is physically pushed ahead of the impactor/projectile. However, unlike in prepreg lamina based composites, this does not hold true in all cases for the RTM systems discussed herein due to the greater interplay between impactor size and fabric architecture. Layer by layer investigation of damage for a 24 oz plain weave system suggests that final puncture of the composite occurs when a consistent level of five-six bundle breaks is seen in each layer in a narrow zone below the point of impact. The apparent steady state is thought to result from short segments of fractured bundles bending downwards but not breaking free from their respective layers, thereby enabling penetration of the projectile but not resulting in an increase in the relative amount of debris moving in front of the impactor.

4. Microstructure effects

Based on the impact response of autoclave cured laminate composites, it is clear that factors such as tow size, fiber-matrix bond (or lack thereof), fiber volume fraction and the use of specially designed complaint layers between laminae, have a significant effect on impact response and damage tolerance. It is thus of considerable interest to investigate the effect of tow size, interphase configuration, and compaction, on impact response of plain weave RTM composites, with a view towards the microstructural tailoring of response. Results on the effects of fabric level interphasial layer tailoring are given in [13].

Fig. 11 plots projected damage area (PDA) as a function of impact energy for four plain weave systems, each molded to a thickness of 6.4 mm, corresponding to the details given in Table I. Since the plate thicknesses are the same, fiber volume fraction resulting

TABLE I Characteristics of plain weave reinforced composites

Plate designation	Fabric area weight oz/sq yd (nominal weight in gm/m^2)	Layup sequence	Fiber weight fraction (%)
A	18 (610.3)	4s[a]	47.1
B	18 (610.3)	5s	55.4
C	24 (813.8)	4s	58.9
D	36 (1220.7)	4s	72.5

[a]s = Symmetric layup.

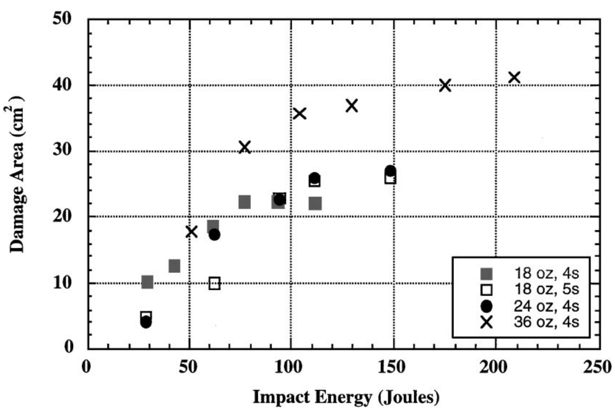


Figure 11 PDA as a function of impact energy for plain weave composites using fabrics of 18, 24 and 36 oz/sq yd weights.

from the use of the same number of layers of increasing areal weight also increases. Obviously, this is a result of larger tow size and greater compaction between layers. It can be seen that the 36 oz/sq yd (nominally 1220.7 gm/m^2) fabric results in the attainment of the highest overall impact energy level before penetration, but also shows the greatest PDA. This can be related to greater intrabundle and interbundle cracking made possible by the larger sized tows. The threshold impact energy, K_1, as defined in [2] also increases with fiber tow size, with thicker tows resisting catastrophic curvature strains imposed by the impactor to a greater extent than the smaller tows used in the lower weight fabrics. In general, the impact energy levels required for penetration increase with weight fraction for the three systems, A, C and D, having the same number of layers. The increased compaction with the higher weight fabrics results in greater fiber-to-fiber contact and interpenetration between layers resulting in a decreased tendency for delamination. A comparison of the two 18 oz/sq yd (nominally 610.3 gm/m^2) sets reveals that the 4s layup has greater damage at lower energy levels due to the presence of larger resin rich zones between layers than in the 5s layup, which follows the same reasoning.

5. Summary and conclusions

The progression of impact induced damage in resin transfer molded plain weave reinforced composites is seen to share some mechanisms with those resulting from impact events on unidirectional prepreg based laminated autoclave cured composites. In addition, however, mechanisms such as intrabundle and interbundle cracking have great significance. The compaction of preform layers in the dry state, prior to resin injection, results in significant interpenetration of layers by individual fibrils and nesting of adjacent layers, resulting in damage progression that emphasizes bundle level dynamics rather than laminae level mechanisms such as delamination. Mechanisms such as delamination still occur, but should in fact be considered in terms of layer separation, since they are accompanied by fiber fracture, fabric tearing and pullout, resulting from the effects of compaction, coupled with significant matrix shearing. Damage progression is often by a mix of local and global phenomena, with bundles providing modes for transmission of incipient energy, while simultaneously acting as crack arrestors at regions where undulations and cross-over occur due to construction specialties of plain weave fabrics.

The flexibility in microstructure design offered by such systems shows immense potential for tailored fabric architectures, optimized for impact response through aspects such as variation in fabric areal weight and bundle size, selective application of sizings on warp and weft bundles to provide greater shear coupling in one direction than the other, and the use of duplex effects through coatings on fabrics to enable energy absorption and dissipation through fibril and bundle sliding.

References

1. N. CRISTESCU, L. E. MALVERN and R. I. SIERAKOWSKI, *ASTM STP* **568** (1975) 159–172.
2. R. L. SIERAKOSKI, L. E. MALVERN and C. A. ROSS, in "Failure Modes in Composites III," edited by T. T. Chiao (AIME, New York, 1976) pp. 73–88.
3. W. J. CANTWELL and J. MORTON, *Composite Structures* **3** (1985) 241–257.
4. S. ABRATE, *Applied Mechanics Reviews* **44**(4) (1991) 155–189.
5. P. A. LAGACE and E. WOLF, *AIAA Journal* **33**(6) (1995) 1106–1113.
6. J. D. WINKEL and D. F. ADAMS, *Composites* **16**(4) (1985) 268–278.
7. B. Z. JANG, L. C. CHEN, C. Z. WANG, H. T. LIN and R. H. ZEE, *Composites Science and Technology* **34** (1989) 305–335.
8. P. W. R. BEAUMONT, P. G. RIEWALD and C. ZWEBEN, *ASTM STP* **568** (1975) 134–158.
9. J. G. WILLIAMS and M. D. RHODES, *ASTM STP* **787** (1982) 450–480.
10. R. W. RYDIN and V. M. KARBHARI, *Journal of Reinforced Plastics and Composites* **14**(11) (1995) 1175–1198.
11. R. W. RYDIN, M. B. BUSHMAN and V. M. KARBHARI, *Journal of Reinforced Plastics and Composites* **14**(2) (1995) 113–127.
12. V. M. KARBHARI, *Journal of Materials Science* **32** (1997) 4159.
13. R. W. RYDIN, P. C. VARELIDIS, C. D. PAPASPYRIDES and V. M. KARBHARI, *Journal of Composite Materials* **31**[2] (1997) 182.
14. G. J. DVORAK and N. LAWS, *Journal of Composite Materials* **21**(4) (1987) 309–329.
15. M. E. BACKMAN and W. GOLDSMITH, *International Journal of Engineering Sciences* **16**(1-A) (1978) 1–99.

Received 21 April 1997
and accepted 27 April 1999