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Review
The role of reinforcement architecture on impact damage mechanisms and post-impact compression behaviour

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This review considers the link between the damage tolerance of composite laminates and the nature and organization of the fibre reinforcement. This embraces composites made from unidirectional prepregs through composites based on a variety of textile forms such as woven fabrics, multiaxial fabrics, braids and knits. The objective has been firstly to detail how the differing varieties of composite exhibit different properties under impact conditions and under subsequent loading after impact. This includes both fracture mechanisms and data such as energy absorption, and peak failure loads. The second objective is to describe the links that have been found between these properties and the specific fibre architectures and damage development processes in the various composite forms. The post impact compression properties are highlighted as this is the area of greatest interest by end-users. The review describes the different forms of textiles that are used for composite reinforcement, considers different impact conditions (e.g. low velocity and ballistics), general materials variables such as fibre and resin type, and ultimately looks at specific textile systems. Some consideration is also given to the value and role of numerical modelling in the field of damage formation and damage tolerance. Clear differences have been found in the literature between composites based on different textile forms in terms of damage states after impact and the consequences of this damage on subsequent properties. While the literature is clearly incomplete at this time there is sufficient information available to indicate that control of fibre organization by the use of textiles may be an effective method of optimizing composite properties for specific end use properties.

1. Introduction
Unidirectional carbon fibre and glass fibre reinforced plastics are attractive materials to the aerospace and automotive industries for their high specific strength and stiffness, good formability, corrosion and fatigue resistance [1, 2]. Performance in structural applications may be optimized by tailoring the orientation of the pre-impregnated warp sheet prior to fabrication. However, in response to the need for a higher level of damage tolerance in structural composites, several concepts have been proposed. These include: increased fibre fracture strain, tougher resins, hybridization, modification of fibre/matrix interface and the use of textile processes [3]. The textile process offers the designer preforms that can be manufactured to suit, e.g. woven, knitted, stitched or braided fabrics.

The use of textile fabric in composite structures, rather than the prepreg tape unidirectional material, offers advantages such as ease of handling, excellent drape, thicker fibre forms and the ability to be used in the resin transfer moulding (RTM) process or prepregged [4]. Additionally, the toughness of these composite materials are improved over their unidirectional counterparts [5, 6]. However, fibre displacement/damage or re-orientation from the in-plane occurs as a consequence of the textile process, which translates into a reduction of the in-plane properties of the final composite [7–9].

The use of reinforcement architecture through the textile process enables complex geometries to be produced as an integral structure [10, 11]. The three-dimensional class of textile preform may provide good translation of in-plane mechanical properties whilst retaining high interlaminar toughness. Other groups (stitched fabrics) which still rely on a textile process, but are not an integral fibre structure distinguish themselves by through-the-thickness fibre reinforcement which reduces the problem of delamination [12, 13].

The following discussion classifies the fibre architecture of the composite and the subsequent mechanical performance, impact resistance and residual strength capability, i.e. damage tolerance in compression.

2. Fibre architecture and fabrication influence on properties

The architecture of the textile preform, or the fibre orientation and the level of structural integrity, determine the fibre packing density (fibre volume fraction) and the translation of the fibre properties to the composite structure. For the purpose of this investigation, the structural preforms may be classified into three categories: unidirectional prepreg tapes, planar (2-D) and 3-D (fully integrated and through-thickness-reinforced) systems. It should be noted that within each fabric class numerous structural geometries can result, e.g. woven fabrics: plain or satin weaves etc.

2.1. Unidirectional prepreg tape

The continuous unidirectional filament system, due to the degree of fibre continuity and linearity, has a high level of property translation efficiency. These materials are commercially available in a semi-processed form, i.e. pre-impregnated with resin, which holds the fibres together while on a "backing sheet". Manufacture of these composites requires controlled application of temperature and pressure for a specified time to optimize performance. The laminated structure results in low interlaminar strength and poor toughness when subject to out-of-plane loading. This low toughness has been partially overcome by the use of different matrix systems, sometimes at an increase in cost, e.g. thermoplastics, but in compensation there is often a reduction in the in-plane mechanical properties particularly at elevated temperatures and hot/wet environments. Much of the performance of laminated composites is a trade off between strength and modulus for improved toughness [14], although modifiers have been shown to enhance damage resistance [15]. A wealth of literature is available on the prediction of properties and analysis of strength using classical lamination theory [16–18].

2.2. Planar 2-D fabric

The planar structure of 2-D fabrics still provide continuous fibres, but, with reduced mechanical efficiency due to fibre crimping [7]. Single filaments have been observed in resin-rich regions due to the textile processes applied to the tows [19, 20]. The woven fabrics are formed on a loom by interlacing two or more sets of yarns. Knits and braids also are available as planar structural preforms, however while the technique of manufacture will not be discussed for these materials a brief description is given by Chou [21]. Furthermore, only woven fabrics are discussed as these materials are more often used in high performance applications. In the instance of plain weaves, the weft alternately goes over and under successive warps while special weave patterns may be produced by altering the number of warp tows to cross a weft tow, Fig. 1. These fabrics may also be conveniently prepregged and manufactured in a similar fashion to unidirectional tapes. The use of resin transfer moulding (RTM) provides greater flexibility given the ability of the textile process to produce a near net preform.

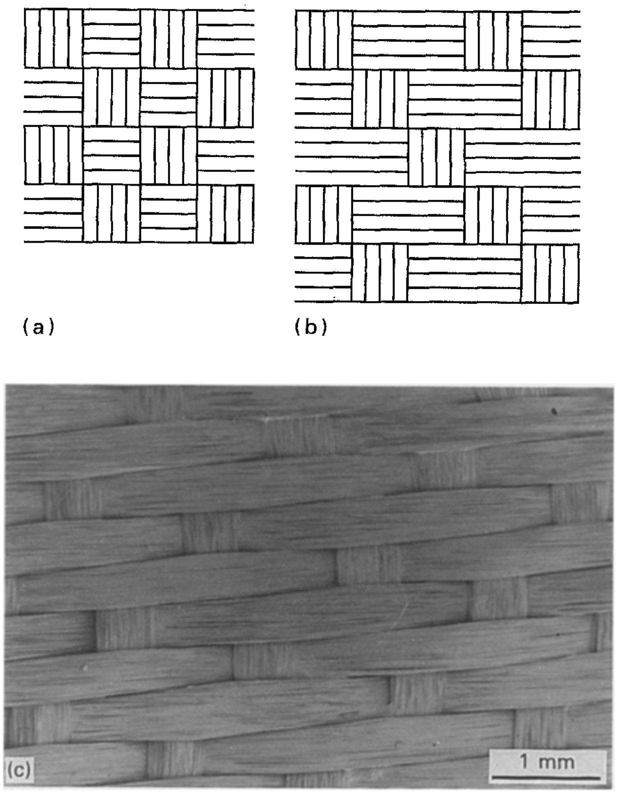


Figure 1 Woven fabric styles (a) schematic of plain weave, (b) schematic of twill weave and (c) SEM micrograph of an eight-harness satin weave.

An improvement in composite toughness is achieved, in comparison with the unidirectional equivalent, as a by product of the weave, i.e. fibre bridging occurs [22]. The analytical prediction of woven fabric composite moduli is based on the "mosaic", "crimp" and "bridging" models [21, 23–30], Fig. 2. In the mosaic model, a woven fabric composite is idealized as an assemblage of cross-ply laminates [23, 30] which assumes linearity of the fibres and does not account for the resin rich region at the tow cross over points, Fig. 2a. The crimp model takes into account the undulation of fibres in the plain weave composite [25, 26, 29]. The undulation shape may be defined mathematically and it is assumed that classical lamination theory is applicable to each infinitesimal slice of length, Fig. 2b. The bridging model is used for more complicated weave patterns and is essentially an extension of the crimp or undulation model [24, 28, 30]. In regions where the tows cross each other the crimp model is used while in the surrounding region the fibres are assumed linear, Fig. 2c, and as a result, this technique is only valid for satin weaves of four or more harness. More detailed information on each of the analysis techniques may be found in the references.

2.3. 3-D fabric

3-D fabrics can be considered to fall into two categories. Fully integrated systems with a continuous fibre architecture, can be manufactured by 3-D interlock braiding. This results in a significant element of structural reinforcement in all three dimensions, and makes the composite highly delamination resistant. Alternatively, a more recent class of fabric are the multiaxial

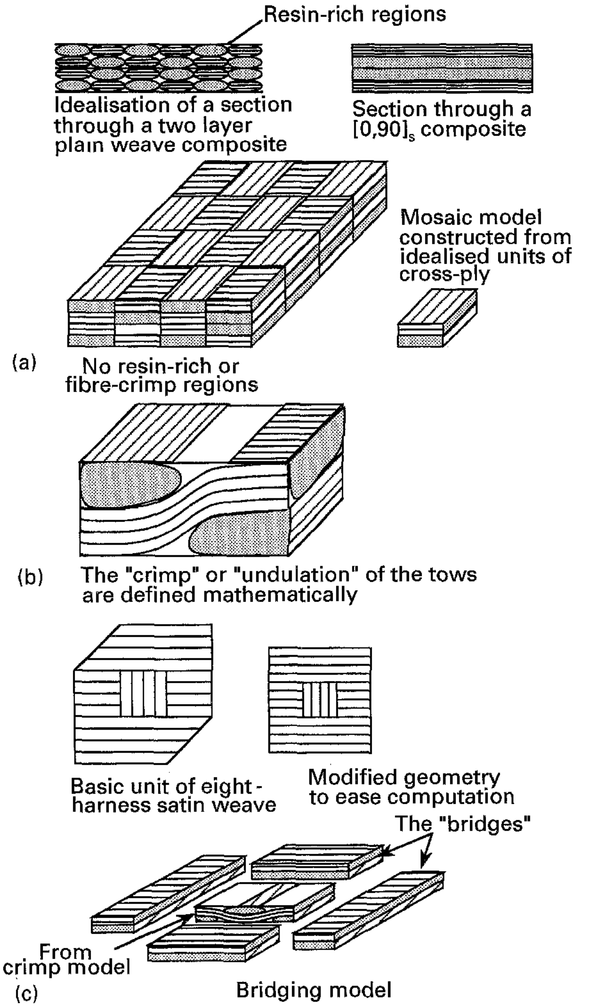


Figure 2 Schematic representations of the (a) "mosaic", (b) "crimp" and (c) "bridging" (after Chou and Ishikawa [30]) models for woven fabric composites.

(non-crimp) fabrics, which are "warp knitted' fabrics. These consist of layers of aligned plies at various orientations with the fabric held together not by the intersection of the tows but by a secondary fibre yarn knitted around the structural fibres [31], Fig. 3. This has the effect of producing a fabric with a near zero crimp state in the reinforcing fibres with a regular 2-D direction in fibre content which may be a high or low performance fibre yarn. Stitched fabrics can also be constructed from "2-D woven fabric " plies to facilitate handling during stacking and stitching. Some kinking of the reinforcement tows occurs during weaving [32] and some fibre damage occurs during stitching [33]. The mechanisms of failure in the "non-crimped" (stitched) class of materials differs from the unidirectional laminated form. Through a combination of matrix microcracking, fibre–matrix debonding, fibre breakage, stress redistribution and delamination, the failure process in the stitched composites usually proceeds gradually, as opposed to a sudden catastrophic mode, exhibited by unidirectional composites [34]. This review is focused primarily on damage tolerance, thus for descriptions of manufacturing techniques refer to Chou [21], Ko [35], Byun and Chou [36] and Maiden and Ebersole [37]. Braided composites are not layered structures and generally exhibit resin rich

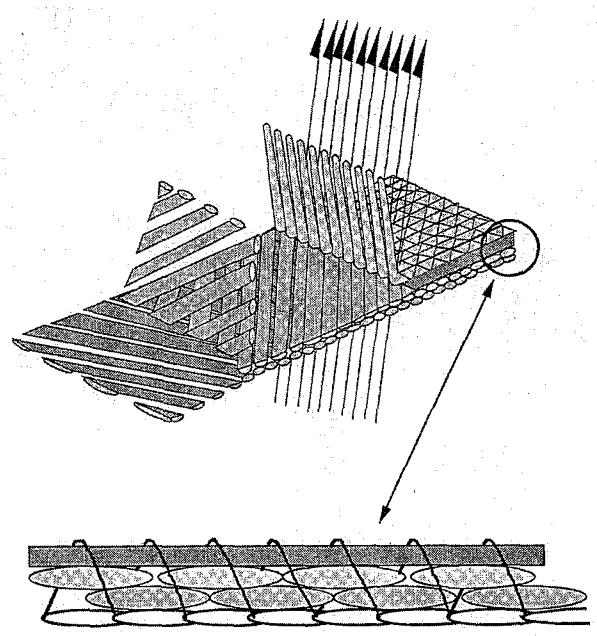


Figure 3 Schematic idealization of the construction of multiaxial (non-crimp) fabrics.

pockets and fibre tow interfaces which may account for their relatively low in-plane strengths [38]. The superior reinforcing ability of straight fibres over the curved tows was also observed by Ramakrishna and Hull [39] in 2-D knitted fabric composites. Damage to the in-plane yarns by the stitching process for the non-crimped materials may reduce in-plane properties of the composite, however, the interlaminar strengths for this class of material is expected to be high, a consequence of the through-the-thickness stitching in the laminate structure.

3. General considerations of impact behaviour and damage tolerance

3.1. Damage tolerance

Damage tolerance is described by Baker *et al.* [1] to be "the ability of a structure to contain representative weakening defects under representative loading and environment without suffering excessive reduction in residual strength, for some stipulated period of service".

Damage tolerance becomes an issue when the service performance of these materials is considered. Structures are designed to meet a set of specific service criteria, but frequently these structures are subjected to an unspecified range of miscellaneous events which fall outside these design parameters and can lead to damage, e.g. accidental blows, occasional overload, misuse and abuse.

Metals, the traditional structural material of the aerospace industry can cope with such instances to a degree by virtue of their inherent ability to yield. Damage is frequently benign and visible. Composites do not typically possess this ability to contain or react to damage in this way. The result of accidental blows such as low energy impacts is dissipated by elastic deformation of the structure, followed by processes such as heat and sound, but if the structural response exceeds the components' ability to deform elastically, then other processes such as fracture occur.

The control of damage tolerance in composites is therefore the control of the fracture processes. This control can be exercised by incorporating limited ability to yield into the composite, and by controlling fibre architecture. This latter approach does not necessarily restrict the extent of cracking, but will allow the nature and distribution of cracks to be controlled to minimize their effect on the mechanical performance of the structure. In notched and unnotched 2-D braided coupons, strength has been observed to be architecture dependent, suggesting reinforcement architecture influences damage mechanisms [40]. The most logical route to control fibre architecture is of course the rational exploitation of textile forms.

3.2. Impact testing

The following discussion focuses on the damage resulting from low velocity, low energy non penetrating blows, such that any material trends discussed for these conditions (e.g. the role of the matrix toughness) may not apply to the more severe impacts resulting in penetration.

A wide range of approaches to the problem of collision by a moving body on an elastic plate have been investigated. These include two isotropic bodies (Hertzian), one a half space, the other generally spherical [41,42] to the more complex situation of an orthotropic plate [43] and experimental laws [44]. While these techniques model the contact loads and deflections encountered, they do not cater for damage processes, consequently the predicted maximum force will exceed the actual maximum force [45].

Equating the kinetic energy of the contacting mass with that stored by the plate at maximum displacement allows prediction of the maximum force

$$\frac{1}{2}Mv^2 = \int_0^{\delta_{max}} F\,d\delta + \int_0^{\alpha_{max}} F\,d\alpha$$

where the plate is treated as a spring, $F = k\delta$ and the contact indentation, α, is related to the contact force by Hertz's law, $F = n\alpha^{3/2}$ with $n = \frac{4}{3} r^{1/2} E_{22}$ (r – indenter radius).

Substitution and integration yields

$$\frac{1}{2}Mv^2 = \frac{1}{2k}F_{max}^2 + \frac{2}{5n^{2/3}}F_{max}^{5/3}$$

Alternatively, the contact force–time history may be recorded using an instrumented striker, Fig. 4, with the plate motion determined using the equation of motion, $F = Ma$. Integration gives velocity, displacement and energy. A more detailed discussion on the modelling of a panel's response to impact may be found in Abrate [46].

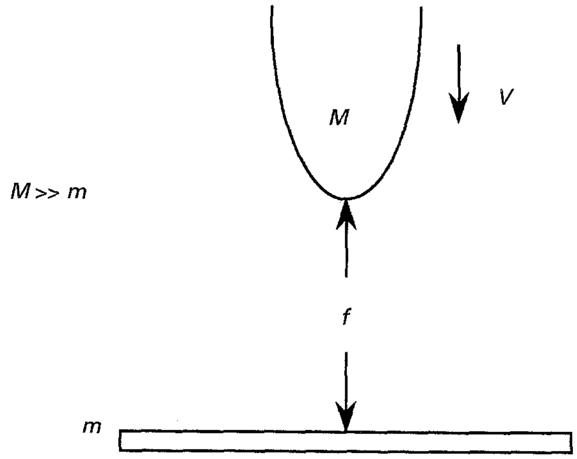


Figure 4 Free body diagram for a typical impact event.

$$F = Mg - f = M\frac{dv}{dt}$$

$$\int_v^{v_0} dv = g\int_0^t dt - \frac{1}{M}\int_0^t f\,dt$$

$$v = v_0 + gt - \frac{1}{M}\int_0^t f\,dt$$

$$\int_0^x dx = v_0\int_0^t dt + g\int_0^t t\,dt - \frac{1}{M}\int_0^t\int_0^t f\,dt$$

$$x = vt + \frac{1}{2}gt^2 - \frac{1}{M_0}\int_0^t\int_0^t f\,dt$$

$$\int_0^E dE = \int_0^t f\,dx$$

$$E = v_0\int_0^t f\,dt + g\int_0^t ft\,dt - \frac{1}{2M}\left(\int_0^t f\,dt\right)^2$$

For high mass low velocity impact some care should be taken with the interpretation of the force data as it may be composed of many small superimposed oscillations due to the plate vibrating against the impactor during contact [45,47,48].

The damage mechanisms observed under experimental conditions have played a particularly important role in the aerospace industry as it has shown that potential weight savings associated with the use of composite materials may not be fully realized due to the inability to exploit the allowable strains of these materials [49].

Methods such Charpy and Izod have been used to determine the toughness of notched homogeneous and isotropic materials, however, these techniques are of limited value (suited to ranking, see Cantwell and

Morton [50] and Li and Harding [51]) where composite materials are concerned due to their complex failure processes. Charpy and Izod tests destroy the specimens and give a measure of the energy absorbed in the process, but these tests on bar-like coupon specimens do not fully reflect the behaviour of plate-like structures. Furthermore, because composite materials are sensitive to out-of-plane loads, the residual mechanical properties after an impact that is not destructive or produces visible damage is essential to the evaluation of a composite system.

Consequently, methods have been developed to simulate in-service conditions. Techniques such as dropweight tests (simulating dropped tools and hail stones) and ballistic tests (simulating runway stones) are designed to facilitate the determination of the residual mechanical properties of the material [52]. Alternative impact methods are available, for example Li and Harding [51] used the Hopkinson pressure bar to investigate the low velocity transverse impact damage on plain-weave reinforced epoxy laminates.

In an evaluation of 10 test methods (pendulum to dropweight) undertaken by Kakarala and Roche [53] using several materials ranging from unfilled thermoplastics to reinforced thermosets, they found results from different impact methods did not correlate. However, similar damage and stress states were observed among tests using the same principal.

Other methods that may be of interest have been investigated by Farley [54], Williams and Rhodes [55] and 't Hart and Frijins [56]. Farley [54] considered the effect of structural geometry and material properties on the crash impact of composites for helicopter applications, while Williams and Rhodes [55] and 't Hart and Frijins [56] investigated the effect of pre-load on the impacted specimen. They found that impacted specimens under load are less capable of deforming elastically and thus fail at a lower impact energy. However, while the latter's work may be of more relevance it is reasonable to assume the stress states at periods of likely non catastrophic impact are low (e.g. dropped tooling on an aircraft wing during maintenance).

The focus for the understanding and characterization of the states of damage arising from impact have been almost exclusively experimental. This is generally conducted at low levels on the design evolutionary scale due to the cost involved in testing full scale structures. With the current capabilities of computer hardware and software there have been attempts to simulate the impact process, e.g. Joshi and Sun [57, 58] and Lakshminarayana *et al.* [59, 60]. The solution time step in the model should be small enough to define the contact force history. It is therefore essential to arrive at a fine enough mesh and small time step that will assure the solution converges. Lakshminarayana *et al.* [60] demonstrated that modal analysis is generally unsuitable in view of the high orders necessary to represent the structure's deformation with time. This technique, like the analytical methods discussed earlier does not account for failure processes and their effect on material properties (e.g. local stiffness, strength etc.) that will affect the ability of the structure to deform and consequently influence stress redistribution. Joshi and Sun [57, 58] directly integrated the equations of motion using a linear analysis for the solution of the internal stress distribution as a function of time, however, they acknowledge the need for large deformation (non-linear) theory to adequately represent the displacement field and hence distribution of stress. This was also confirmed by the investigations of Lakshminarayana *et al.* [59, 60].

Lakshminarayana *et al.* [59] attempted to predict the nature and extent of impact induced damage by first determining the localized 3-D time dependent stress state around the point of impact and incorporating this in an appropriate failure criteria. The analysis did not consider delamination onset and growth but only matrix cracking and fibre fracture. They found the non-linear analysis reliably predicted the impact response and damage geometry. However, the chronological appearance of cracks after the onset of impact can not be predicted by a single simple representation of the dynamic state of stress. Such a procedure implies that cracks which appear first in a layer are independent of cracks in other layers, but the stress field is modified by the appearance of a crack in a ply. To model the physical process requires modelling of fractures (matrix, delamination, fibre breaks, etc.) after each time step [57–60]. Providing the "reduced" properties after the impact analysis may be retained this ability to "switch off" various material properties of the finite element will provide a more realistic model for the subsequent simulation of mechanical testing.

3.3. Post impact mechanical performance

In order to quantify the effect damage (delamination, fibre breaks, etc.) has on the laminate, mechanical testing is required. Depending on the material and structural variables, the damage state from an impact event may consist of various amounts of fibre failure, matrix cracks and delaminations.

Damage dominated by fibre failure results in load redistribution that can affect both tensile and compressive residual strength. Damage causing matrix failure results in load redistribution that primarily affects compressive residual strength [61].

Dorey [62] and Curtis [63] compared the residual tensile and compressive strengths after dropweight impact on a carbon fibre epoxy laminate produced from prepreg tape. Low energy impacts caused delaminations which significantly reduced compression strength which was attributed to local buckling instabilities of the delaminated plies. The tensile strength was little affected until fibre fractures were caused, at higher incident energies, Fig. 5. Bishop and Dorey [64] also found that impact energies causing a 60% decrease in compression strength did not affect tensile strength. Broken 0° fibres, in addition to delamination, produced at higher energies resulted in a 25% reduction in tensile strength, but with no further loss in residual compressive strength. The asymptotic effect of residual compression strength may be an artefact of the test and attributable to the impact induced saturation of damage in the sample. This

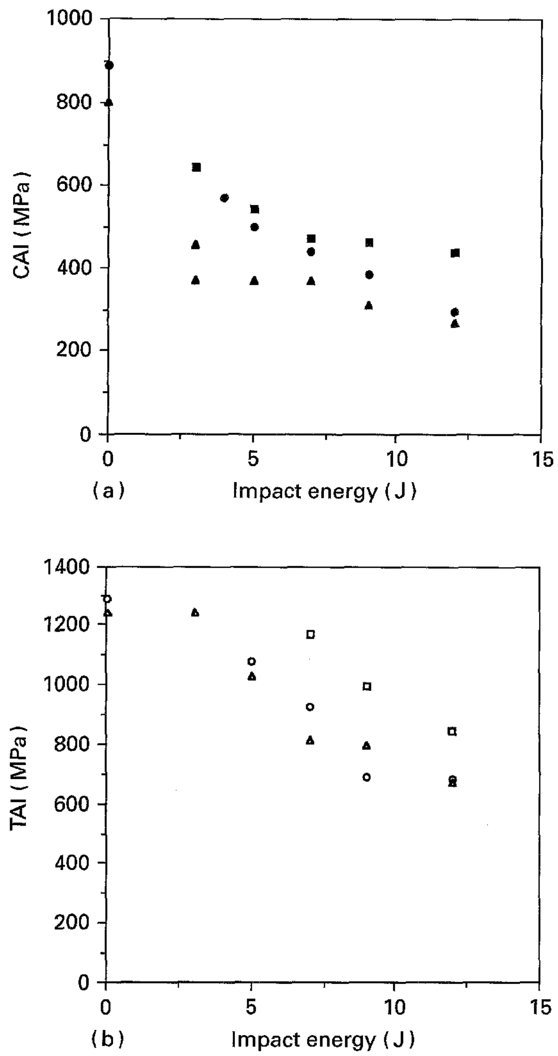


Figure 5 Residual tensile and compressive strengths of carbon fibre thermoset and thermoplastic matrix composites as a function of impact energy (after Curtis [63]). (a) ● Tough carbon/epoxy; ▲ carbon/epoxy; and ■ carbon/PEEK. (b) ○ Tough carbon/epoxy; △ carbon/epoxy; and □ carbon/PEEK.

effect may be governed by the support conditions of the impact facility or the boundary conditions in the compression after impact fixture.

Caprino [65] proposed a model based on linear elastic fracture mechanics to predict the residual strength, σ_r (tension and compression) of impacted unidirectional prepreg tape laminates. The damage zone is equivalent to a hole.

$$\sigma_r/\sigma_0 = (X_0/X)^m$$

In the above, σ_0 represents the strength of the undamaged material, X_0 the dimension of a characteristic defect and X, the dimension of damage. The parameters m and X_0 must be determined experimentally and depend uniquely on the material. The above equation is valid for

$$X \geqslant X_0$$

When $X = X_0$ the damaged material's strength is equal to its undamaged strength, i.e. a threshold value of defect insensitivity. The characteristic damage variable, X, was further related to the incident energy, U, of the striker.

$$X = kU^n$$

similarly for X_0 resulting in an equation directly relating impact energy with residual strength.

$$\sigma_r/\sigma_0 = (U_0/U)^\alpha$$

The parameters U_0 and α must be experimentally determined. For impact energies, U, less than the critical energy, U_0, the residual strength is equal to the undamaged strength (i.e. insensitive to defects below a critical value).

Despite the complex fracture mechanisms resulting from a low velocity, low energy impact and the subsequent complex fracture processes under loading (discussed later), Caprino [65] found the curve fitting procedure for a linear elastic fracture mechanics model representing a hole to describe (i.e. exponential decrease in residual strength after impact) the experimental data in residual tensile and compressive strength.

Compression strength is one of the most difficult properties to determine. Hahn and Williams [66] give a comprehensive overview of the theoretical failure mechanisms in unidirectional prepreg tape composites. A slight eccentric load will cause premature buckling failure rather than the intrinsic compressive failure [67]. For an overview of buckling failure, the reader is referred to Leissa [68, 69]. For thin flat specimens geometric instabilities may be avoided by providing multiple side supports. The supports reduce premature failure by localized "brooming" at the ends of the specimen [17].

Various test fixtures are available to obtain the compression strength of a composite laminate. Ryder and Black [70] discuss the testing of large gauge coupons. The Celanese and IITRI (Illinois Institute of Technology Research Institute) use short unsupported test sections to inhibit geometric buckling. These fixtures, however, are not designed for post-impact compression tests. The damage area of an impacted laminate generally extends beyond the gauge length required by these methods to prevent instabilities. Anti-buckling guides are used as an alternative to fully support relatively long test sections. Examples of such methods include: Boeing [71], CRAG [72], NASA [73], and SACMA [74]. More detail about the relevant test fixtures and procedures may be found in Gedney *et al.* [75] (Celanese), Whitney *et al.* [76] and Adams and Odom [77] (IITRI). Data cited in this review from internal sources was generated using a miniaturized jig developed at Queen Mary and Westfield College (QMW) [78].

Woolstencroft *et al.* [79], Gedney *et al.* [75] and Prandy *et al.* [80] found the measured compression strengths to vary with test fixture. Compression failure mechanisms in damaged composites (laminated prepreg tape) have been the subject of a large amount of numerical and experimental modelling. In 2-D and 3-D composites, failure mechanisms may be further complicated by the more intricate fibre structure. Furthermore, the contribution of differing damage states under compression loading may subsequently influence the residual strength. Consequently these new fabrics may be more critical or less damage tolerant under a different loading spectrum compared with the laminated composites.

3.4. Damage mechanisms

Impact damage is influenced by test conditions (striker shape, density or hardness and structural support) and laminate properties [41, 81–83]. The constituent properties; fibre/matrix, their interface, modulus (including ply stacking sequence and mass/velocity impact combinations) and the strain energy release rates G_{IC} and G_{IIC} may also be important, particularly for the one and two dimensional fibre architecture. Cantwell and Morton [84] suggest it is "impossible" to compare results from other sources, while Lindsay and Wilkins [85] "doubt" the ability to translate data from a laboratory specimen to information the designer can use. The complexity of the problem is shown in Fig. 6. In a study that investigated the effect of gripping conditions (loose and fixed) and opening diameters (40 and 80 mm) on the impact damage and residual strength Verpoest *et al.* [86] identified the diameter of the gripping device as the most influential test parameter. Soulezelle [87] also found the gripping conditions to affect the energy at which first damage occurs. This phenomenon is principally because the support/gripping conditions affect the impact response (deformation) and hence the relative amount of energy that may be absorbed elastically by the specimen prior to any failure criteria being met.

The impact induced damage state of fibre reinforced composite systems are primarily a function of: (a) velocity at impact, (b) material system, and (c) thickness or stiffness. A further parameter is stacking sequence or fibre architecture, which will be discussed separately.

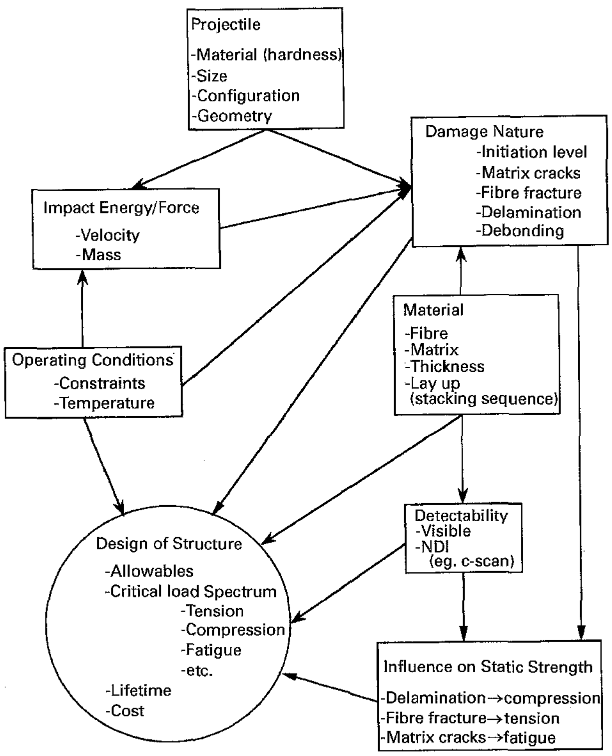


Figure 6 Factors involved in the design of composite materials.

3.4.1. Impact velocity (stress wave effects)

There is a considerable difference in target behaviour between impacts at "low" and "ballistic" velocities which can result in damage states that may be dominated by stress wave effects [45].

Godwin and Davies [88] and Robinson and Davies [89] demonstrated by considering the stress wave, which propagates at the speed of sound, c (approximately 2000 m s^{-1} for epoxy composites), and the compressive failure strain, ε, of the material (typically 0.5–1.0%) that at impact velocities, V_0, of 10–20 m s^{-1} there is a transition to a stress wave dominated impact.

$$V_0 = c\varepsilon$$

An additional condition for "low velocity" collisions is that impactor and target act like a single degree of freedom system with the target behaving as a spring of negligible mass [90].

Furthermore, the damage extent and mechanisms have been modelled satisfactorily using quasi-static tests [91–95], indicating that the limit (< 10 m s^{-1}) and the conditions that define "low velocity" impact in the typically rate-insensitive materials used are accurate. This is because these conditions imply that stress waves have time to propagate and reflect many times which results in a deformation mode approaching the static solution [45].

Considerably different structural responses and subsequent damage states arising from specimens impacted at high and low velocity have been observed [45, 96]. Under low velocity loading the contact time is relatively long and the target response is global, thus the structure's geometry determines the energy absorbing/failure mechanisms, Fig. 7a. The former is in agreement with the above condition (Swanson [90]) and the latter, geometry, will be discussed further in Section 3.4.3. Conversely, under high velocity impact loading the geometric parameters of the specimen, i.e. width, length and thickness, have little effect on the impact behaviour due to the localized response of the sample, Fig. 7b. The resulting damage is generally

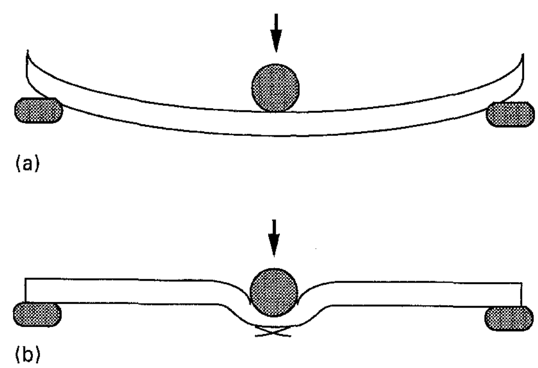


Figure 7 Schematic representation of the response to impact under (a) low velocity and (b) high velocity conditions.

more severe than that encountered under low velocity conditions [84, 97]. The behaviour of composites with higher-order fibre structures may be similar, but this requires experimental verification.

3.4.2. Material system

Srinivasan *et al.* [98] via a series of impact tests on epoxy and thermoplastic reinforced composites identified the influence of resin material on damage resistance. This effect has been observed by numerous workers [92, 99–103] in that thermoset composites generally exhibit more extensive damage for a given impact blow compared with the thermoplastic matrix composites. The reduced damage in the thermoplastics may be explained by the materials' higher G_{IC} and G_{IIC} values [104–108]. In addition, differences have been observed within the thermoplastic family [103, 109, 110]. Similar behaviour was observed by Ma *et al.* [111] for a PEEK and PPS woven carbon fabric. Srinivasan *et al.* [98, 110], found the difference between PEEK/IM-7 and PEEK/AS-4 to be negligible, implying that the damage resistance is a strong function of the matrix. Bibo *et al.* [103] also identified the dependence on matrix under low velocity non-penetrating impacts, but found through-penetration impacts to be independent of resin type for carbon fibre composites which confirmed a similar trend reported by Babic *et al.* [112] for random glass-fibre composites. Similar behaviour has also been observed in the 2-D and 3-D woven class of composite. Brandt *et al.* [113] subjected unidirectional prepreg tape, 2-D and 3-D woven thermoset and thermoplastic composites to impact. They found the components to behave very differently depending on the material used. The surface indication of damage was more apparent in all thermoplastics compared to the thermosets. However, the formation of delamination area was more extensive in the thermosets. No delaminations occurred in the 3-D woven fabrics.

It should be clarified that the matrix dependent behaviour is observed for a given reinforcement type. The ability of a composite to absorb energy is also dependent on the strain energy (area under the stress–strain curve) to failure of the fibres [52, 107]. For a given matrix, Cantwell *et al.* [114], found the damage area after impact to be considerably less in a high strain (1.53%) fibre composite than the standard high strength fibre composite. For the same energy in an impact blow, glass reinforced epoxy systems exhibit less delamination damage than a carbon fibre reinforced epoxy system, Fig. 8a. This may be attributable to the greater flexural stiffness of the carbon fibre reinforced laminates resulting in higher impact forces (as defined in the earlier equation), Fig. 8b. However, the differences in damage extent between the reinforcement types (i.e. glass or carbon) is reduced if compared on the basis of impact force, Fig. 8c.

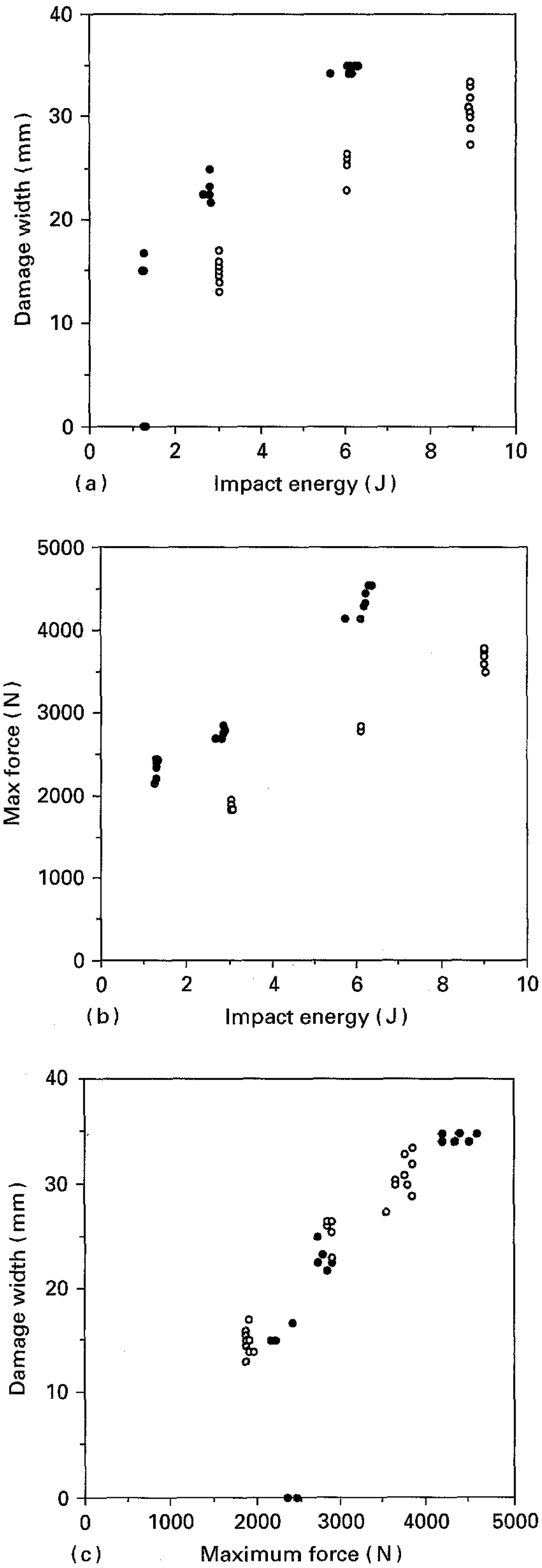


Figure 8 (a) Damage width and (b) maximum force as a function of impact energy. (c) Damage width versus maximum force. ● UDPT carbon/epoxy and ○ UDPT glass/epoxy.

The interface between fibre and matrix directly affects the stiffness and energy absorbing mechanisms of a composite. Dorey [62] and Stuart and Altstadt [115] showed lower levels of fibre treatment resulted in larger areas of impact induced delamination, but this is compensated for by reduced notched sensitivity in tension. Although drawn from compression testing of braided materials with differing percentages of laid in 0° fibres (Liao *et al.* [116]) using glass, carbon and Kevlar, it was found at all percentages of braiding the stress–strain curves for glass and carbon were linear while Kevlar exhibited a pseudo-plastic behaviour. This was attributed to the poor interfacial bonding between the fibre surface and epoxy matrix and as a consequence may also be influential under impact loading.

It is probable that the role of fibre, matrix and interface is similar regardless of the reinforcing architecture, within a given class of reinforcement geometry. However this cannot be stated conclusively at this time due to lack of sufficient experimental evidence in the literature to date.

3.4.3. Thickness/stiffness

The ability of a composite to deform elastically influences the macroscopic initiation of damage. Generally the contact induced stress field beneath the indenter results in matrix micro-cracks (discussed later), however the first externally visual failure mode is referred to here.

Cantwell and Morton [117] identified two damage modes in carbon/epoxy, a top surface contact failure in (stiff) short thick targets and a lower surface flexural failure in (flexible) long thin targets. Impact testing of thick graphite/epoxy samples resulted in broken fibres in the contact region [118].

An increase in specimen length provides an increase in energy absorbing capability [89, 108] and the failure mode changes from the top surface contact in the stiffer beams to a splitting mode between the lower surface fibres in the more flexible specimens [87, 107, 119]. Changing the stiffness in a quasi-isotropic lay-up by varying the percentage of 0° plies also showed that more energy was required to damage the less stiff specimens [87]. Finn *et al.* [108] noted that delamination decreased with increasing specimen length, but found it not to affect damage initiation. These failure modes are governed by the conditions of "low velocity" impacts and the ability of the target to be "globally aware" of the event.

Limited data is available in the literature on the influence of stiffness on impact of 2-D and 3-D fibre composites. Due to the highly integrated nature of some 3-D composites the extent or change from contact to flexural failure can not be deduced from observations of laminated composites.

4. Effect of reinforcement architecture on impact damage

Fibre orientation plays a significant factor in the initiation and propagation of damage under impact conditions. In view of the differing damage mechanisms arising in each fibre construction type, they will be treated independently.

4.1. Unidirectional prepreg tape

Damage initiation is often identified as the first load drop on a force–time trace, usually corresponding to the point at which delamination occurs [94, 95, 120]. This definition is useful when considering damage tolerance as delaminations appear to be the critical form of damage responsible for noticeable reductions in residual properties. However, other forms of damage occur prior to delamination namely transverse matrix and shear cracks [54, 121–123]. Matrix cracking is not detected by examining the force–time curves as they occur at too high frequencies and do not result in sufficiently large load drops [91]; however, using accelerometry and ultrasonic C-scan Kaczmarek and Maison [95] and Lammerant and Verpoest [124], using quasi-static impact tests and finite element methods were able to show that matrix cracking proceeds delamination. Moreover, numerical analyses have shown that delaminations occur at lower loads if shear/transverse matrix cracks are present which result in high interlaminar tensile and shear stresses at the crack tip [123, 125]. The presence of the shear cracks has also been shown to be a factor in prescribing the characteristic "peanut" shape. Fig. 9, of the subsequent delaminations [123, 126]. Such subcritical damage has been shown not to affect residual compression or tensile strength [98, 99, 122, 127, 128] consequently damage initiation, here, is regarded as first damage that significantly affects residual strength. It must be recognized however that this subcritical damage may be critical in reducing fatigue life through growth to a critical flaw size.

Both the onset and extent of delamination is determined by ply orientation, i.e. difference in angle between plies or sublaminates of similar orientation [120, 129, 130]. In accordance with bending stiffness mismatch theory, delamination area increases as the ply or sublaminate mismatch angle increases

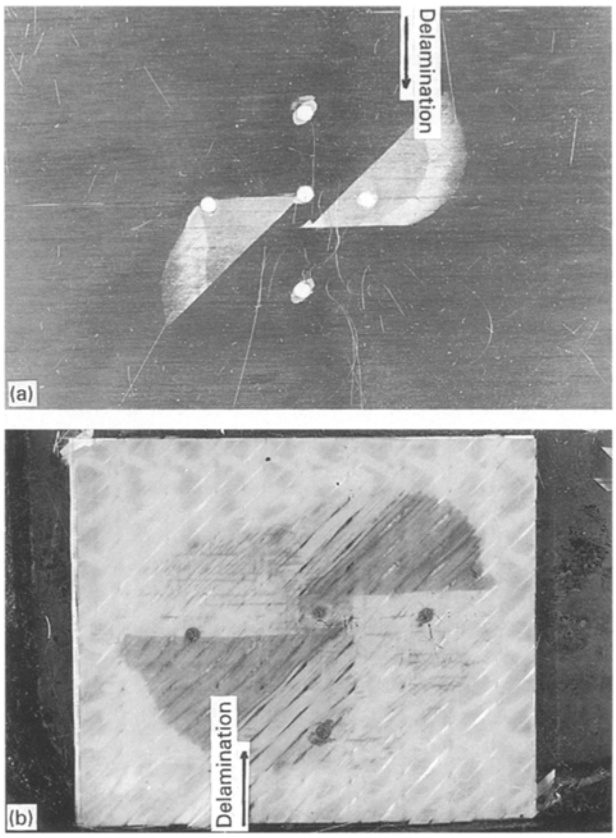


Figure 9 The shape of delaminations between individual plies using deply technique. (a) Quasi-isotropic carbon fibre epoxy laminate manufactured from unidirectional prepreg tape and (b) glass fibre prepregged epoxy multiaxial (non-crimp) fabric.

[130–132]. In addition, delaminations are not observed at ply interfaces of the same orientation [130] and is largest for mismatch angle [133] of 90°. Finn *et al.* [108, 134] concur up to mismatch angles of 30°, but found that beyond that there was no effect on damage initiation load or delamination size.

Thicker sublaminates subjected to the same impact energy have more delamination area [93, 108, 133]. This is confirmed by Clark's [135] model, Fig. 10, where local separation or closure forces between adjacent plies is considered. Thicker sublaminates are stiffer and result in greater forces between laminae. While the macroscopic analysis identifies the general (quadrant) distribution of interlaminar fractures based on Clark's [135] model after impact, no evidence as to the relative amounts, shape or geometry of the internal damage is obtained.

The major forms of damage are matrix cracks, delamination and at higher incident energies fibre fracture. Fibre debonding has also been observed in glass fibre/epoxy composites because of low interfacial strength [107]. In addition, for thermoplastic matrix composites a contact-induced plastic indentation occurs with associated matrix fracture that is aligned with the direction of the outer ply fibres [91, 103, 136].

Lesser and Filippov [137] noted that matrix cracking appears in a high concentration in the vicinity of the impact location and decay rapidly with distance from the impact site. This pattern has been described as a "pine tree" or "cone" of damage whose apex originates at the point of impact [94, 110, 138], Fig. 11a and b. Transverse cracks, Fig. 11c, are evident emerging from the tensile surface into the specimen and at the impact face and occur symmetrically about the contact region, while the shear cracks, Fig. 11d, are observed to occur internally and connect planes of delamination. This lends credence to the supposition that shear cracks occur before the fractures between planes of lamination and initiate delamination. Indeed, a 2-D finite element analysis of a composite containing a shear crack showed that before delamination occurred, there was a significant transverse normal stress (mode I) concentration at the tips of the intraply crack. However, with the growth of the delamination, a rapid transition occurs in which the mode I delamination driving force decreases dramatically relative to the shear component. The damage

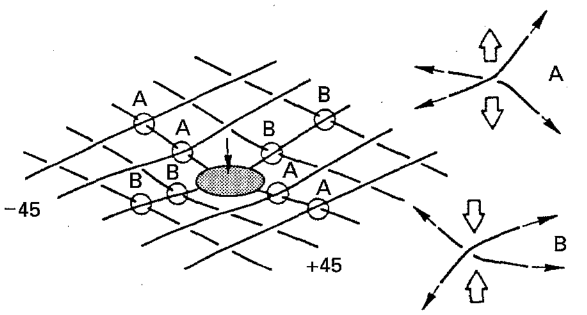


Figure 10 Model for the delamination mechanism. Peel forces acting at "A" promote fracture, while the interlaminar compression at "B" inhibits delamination (after Clark [135]).

mechanism may therefore be a mode II dominated fracture process initiated at the intraply (shear/transverse) crack by a mode I component [139].

Studies of the fracture surfaces of delaminated material have found regions with hackle markings characteristic of mode-II, shear cracking in thermosetting matrix laminates [49, 104, 140]. Directly, similar markings have not been found in thermoplastic composites [110] but this may result from relative changes in interfacial strength rather than any fundamental change in failure mechanism.

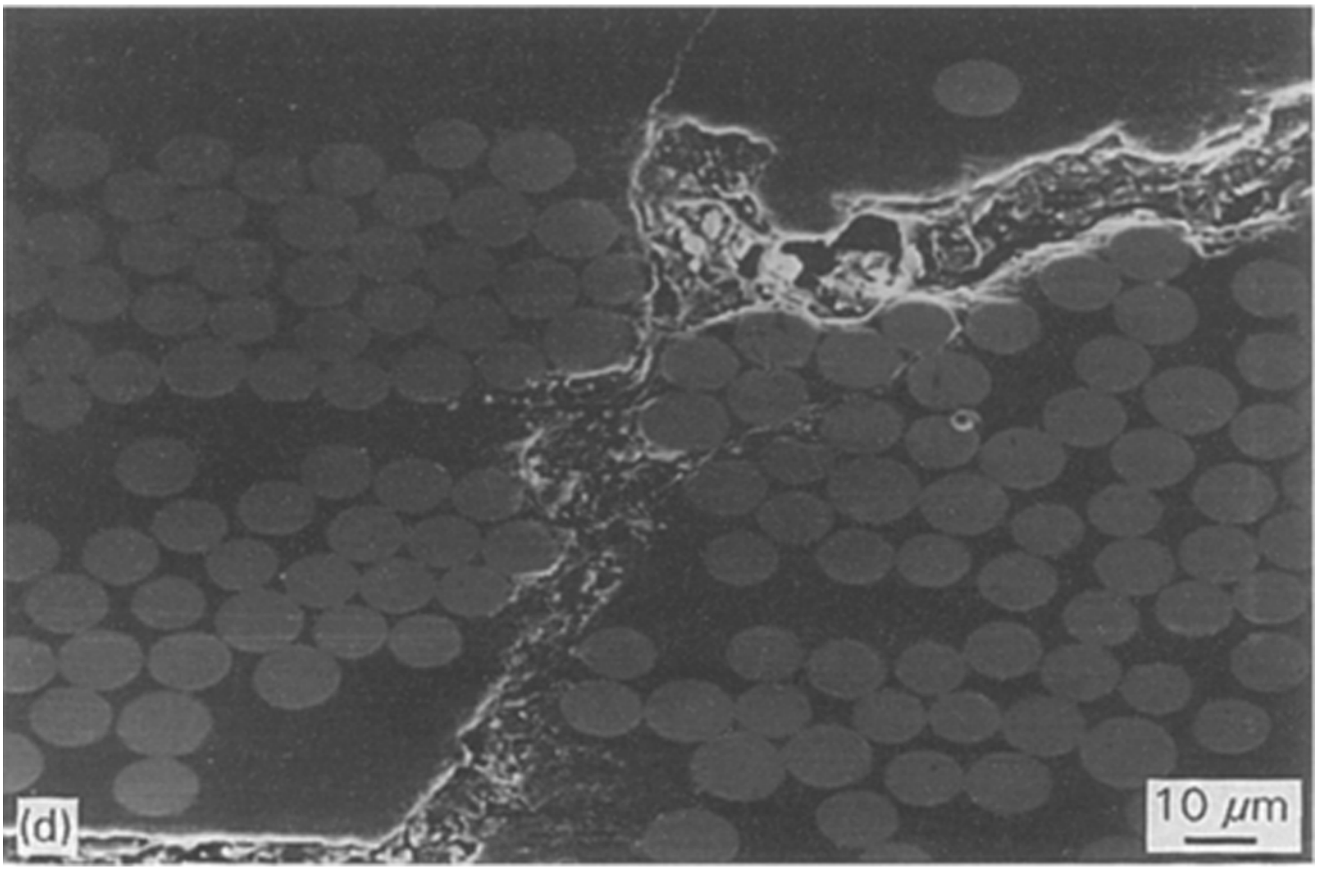
4.2. Planar 2-D fabric

In general for these composites authors have carried out a quantitative comparative study between laminated prepreg tape or fully integrated 3-D composites, thus failure mechanisms and processes unique to these materials are not fully investigated.

Teti *et al.* [141] associate first failure in a plain woven fabric with a change in slope on the force–displacement curve, however, Ko and Hartman [3] and Chou *et al.* [142] identified in the initial part of the curve an incipient damage point, Fig. 12a, which is more consistent with the previous definition. A sudden load drop occurs after reaching the maximum load [3,141], with a more gradual load reduction after maximum load occurring in plain weaves compared with satin weaves [3].

Curtis and Bishop [143] and Strait *et al.* [132] noted that the threshold energy for the initiation of damage in woven fabrics was similar to that for the prepreg tape composites, however the damage was more extensive in the non-woven material. Bibo *et al.* [144], however, observed a clear difference in the level of energy absorbed at damage initiation between quasi-isotropic laminates manufactured from an eight-harness satin weave and unidirectional prepreg tape, with the woven fabric requiring more energy. This difference in observations between Curtis and Bishop [143] and Bibo *et al.* [144] may be attributed to the methodology in evaluating when delamination damage occurs. Bibo *et al.* [144] associate damage onset with the first observable load drop on a force–displacement trace, which is consistent with previous definitions. Briscoe *et al.* [22] and Briscoe and Williams [145] demonstrated that as a consequence of the weaving process fibre bridging occurs and enhances the toughness of the composite. The interlaminar fracture toughness of woven fabrics exceed that of unidirectionally reinforced composites [146, 147].

The nature of damage, like that of the force-displacement trace, is similar to that of the non-crimp fabrics and laminated prepreg tape composites, Fig. 12b and c. After impact there is delamination damage between layers, Fig. 13a, [64, 111, 143, 148] and at the point of impact there is tensile cracking towards the back surface, Fig. 13b, and contact induced crushing, Fig. 13c, and compression buckling close to the front surface [143]. Furthermore, fibre fracture results in a region beneath the point of collision in the composite. This localized region of fibre fracture may be



(b)

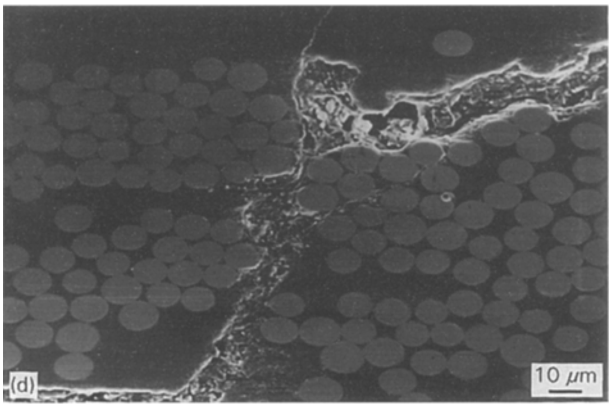


Figure 11 (a) SEM micrograph and (b) schematic at the point of contact showing the "pine tree" nature of impact induced damage. Higher magnification SEM micrograph (c) of the contact induced transverse cracks and (d) a shear fracture linking planes of delamination.

an inherent part of the structure of a weave, i.e. there are areas of stress concentration at locations of weave cross over points [107, 132, 149, 150].

Unfortunately the literature at present appears to be missing a detailed microscopic analysis of the distribution or orientation of the internal damage in the woven fabrics. Despite the apparent macroscopic similarities of the woven fabrics with the unidirectional prepreg systems, i.e. force–time trace, matrix cracks, delaminations and fibre fracture, it is unlikely that they occur in the same amounts and form a "cone" of damage. This is because delaminations and shear cracks are limited by the weave [64, 151], thus the zone of delamination is much narrower [144]. Bibo *et al.* [144] observed a pyramidal type protrusion on the tensile face, Fig. 13b, with very little evidence of delamination and shear/transverse cracks outside the immediate vicinity of this region. The damage extent may also be reduced in the fabrics by the intrinsic nature of the textile process, that is, the layers in woven fabrics are orthogonal and interlinked, thus effectively eliminating the largest mismatch angle, 90°, and these materials have higher fracture toughnesses than their prepreg tape laminate composite counter parts. Moreover, the damage distribution may vary within the fabric types due to changes in weave style.

4.3. 3-D fabric

Within this group there are numerous reinforcement constructions, and the two principally different forms, fully integrated and stitched, are discussed individually as their damage mechanisms are influenced by the degree of reinforcement architecture.

4.3.1. Fully integrated

The damage phenomena in these materials are as different from the previous systems as their force–displacement traces are, Fig. 14. The impact curve shows a gradual increase in load up to the maximum plateau, at which damage initiation is coincident [3, 142, 152] beyond which a more gradual decrease in load occurs. The force–time history for through-penetration of the braided composites investigated by Gause and Alper [153], however, were similar to the laminated composites with identical in-plane fibre orientation. This may be explained by the fact that while there was

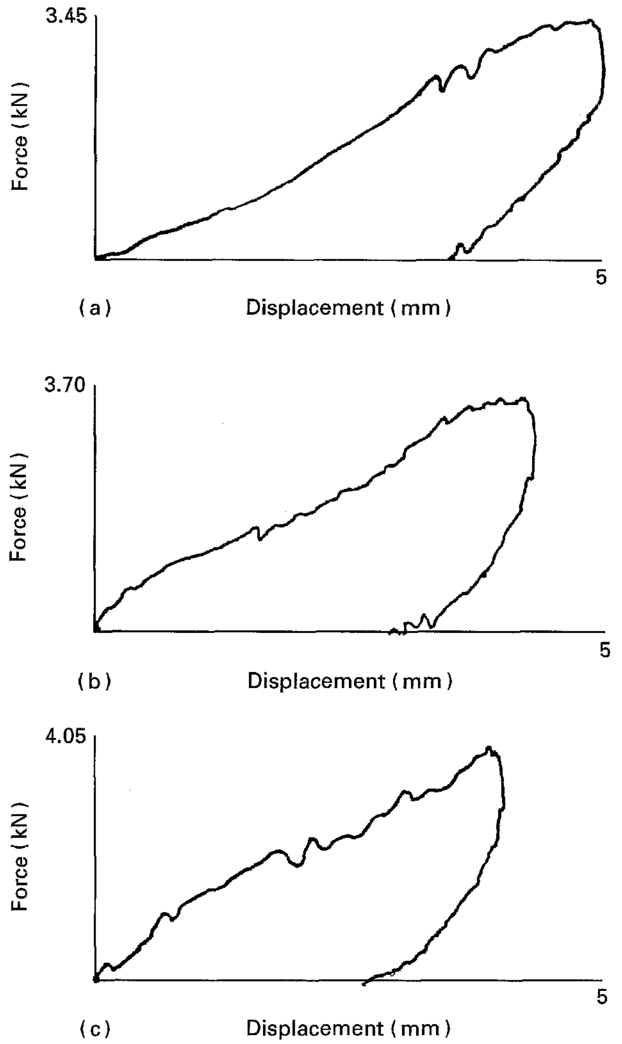


Figure 12 Force–displacement curves for an eight harness satin weave woven fabric, unidirectional prepreg tape and non-crimp fabric generated during a falling weight impact test. (a) Woven fabric (impact energy, 9.0 J and thickness, 2 mm). (b) Prepreg tape (impact energy, 8.9 J and thickness, 2 mm). (c) Non-crimp fabric (impact energy, 9.0 J and thickness, 3 mm).

interlacing in the braided structure, the thickness direction translations of those yarns were absent or very gradual. There are differences in behaviour within this group as Chou *et al.* [142] and Ko and Hartman [3] identified for a 3-D weave and braid subjected to impact and Li *et al.* [154] demonstrated that damage in the braids is affected by the unit cell size. These differences, due to fibre orientation effects are in tune with the variations exhibited by the unidirectional laminated composites subjected to alterations in stacking sequence.

In general these materials are incapable of delaminating as the composite does not have any planes for delamination, i.e. the fibre structure is fully integrated [155, 156]. Damage observed from a low velocity impact on a braided composite plate consists of a two tier damage zone such that in the first zone the nature of damage is primarily surface matrix cracking in resin pockets, these resin-rich regions are effective crack inhibitors [148, 157], fibre tow breakage and debonding of matrix and fibre filaments within broken fibre tows, while the second damage zone consists of

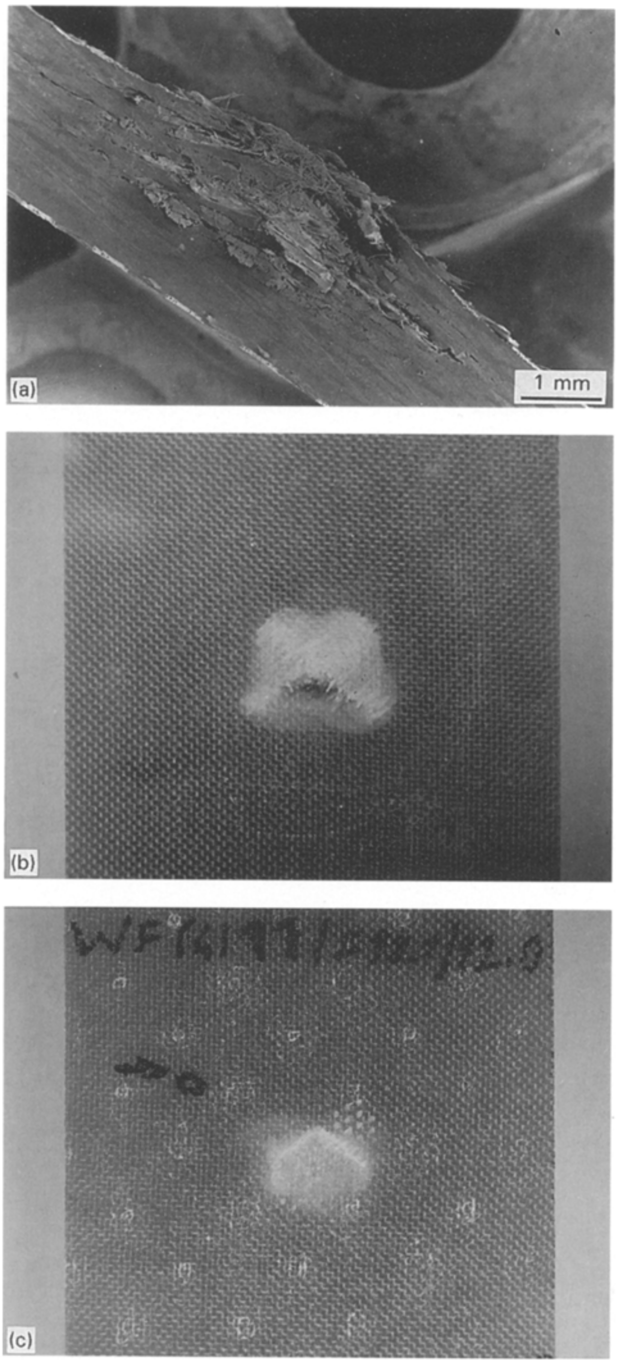


Figure 13 Impact damage in an eight-harness satin weave, glass fibre epoxy composite (a) SEM micrograph of cross-section through damaged region, (b) photograph of reverse (tensile) surface of impacted specimen and (c) photograph of contact surface of specimen.

separation of fibre tows. Broken fibre bundles were concentrated in areas of fibre crimp [38].

4.3.2. Stitched construction

The force–displacement response and damage, when viewed externally resemble their "cousins" (unidirectional prepreg tape laminates) composites [144]. Impact data of stitched and unstitched eight-harness satin weave commingled carbon/thermoplastic fabric showed damage initiation (characteristic load drop) to occur at lower energies in the stitched material, however, their extent of damage was less for the same impact energy [158].

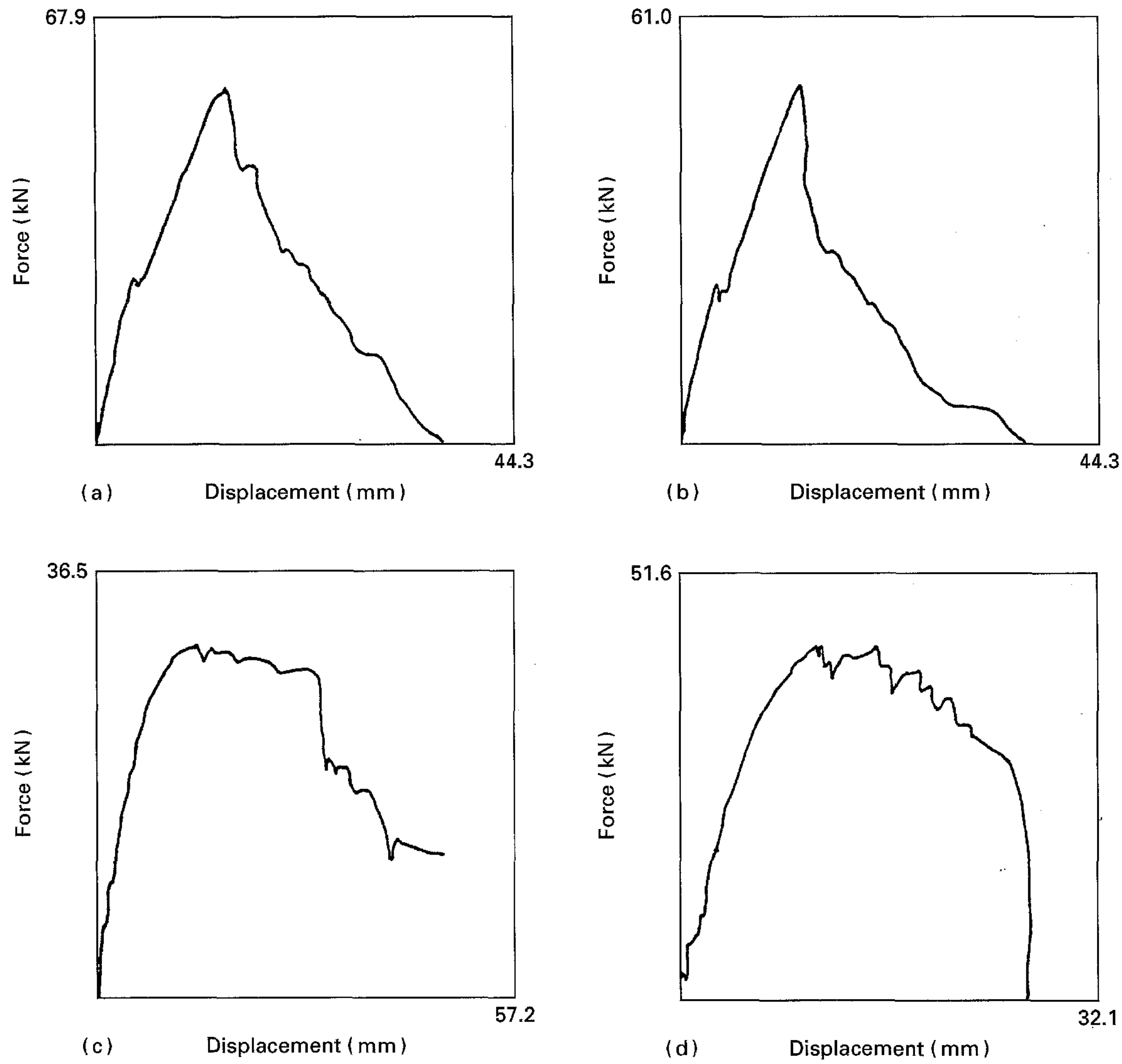


Figure 14 Through penetration impact, force–displacement curves for glass reinforced composites (nominal impact energy, 800 J) (a) plain weave (thickness; 12.7 mm), (b) satin weave (thickness; 12.7 mm), (c) 3-D braid (thickness; 12.7 mm) and (d) 3-D braid with laid in longitudinal yarns (thickness; 17.1 mm) (after Ko and Hartman [3]).

Impact induced damage regions are elliptical in shape and the extent is a function of the stitching yarn and the degree of stitch pitch [12, 13, 159]. Externally, impact induced damage comprised crushed material and intralaminar matrix cracks under the contact site [160]. Farley and Dickinson [161, 162] removed (machined) the stitching loop and consequently the crimp in the two surface plies. The damage was found to be comparable to the unmachined sample even though impacted at higher energies per unit thickness, thus it is suggested that the surface loop of through-the-thickness reinforcement does not participate in controlling damage. The through-the-thickness damage was found to consist of a cone of damage formed beneath the point of impact. The amount of crushed material decreased with increasing depth into the specimen whereas interlaminar delaminations increased with depth. The through-thickness reinforcement decreased the cone angle and reduced damage in comparison with its equivalent lamination construction [160]. Destructive inspection involving sectioning and polishing of a sample of a hand-lay non-crimp laminate through the impact site, indicated that intralaminar/interlaminar matrix cracks coalesce around fibre bundles, Fig. 15, and create a network through the composite resembling planes of delamination [163]. This phenomenon may be explained by the initial weaving process which binds the tows together for subsequent stitching or through the knitting yarn. Under a different loading spectrum the stitching has been observed to be the source of damage initiation, with damage propagating along both their length and across the width and affected through the thickness damage growth [164]. The influence of the through-thickness reinforcement is indicated by its ability to maintain the transverse damage nearly constant over the depth of the composite. The stitching has been shown to be effective in crack bridging [165–170]. Under mode I, albeit possibly not the critical fracture mechanism under impact, the stitching yarns carry most of the load at the crack tip, reducing the stress intensity in the surrounding matrix. The stitching provides a crack-closure force, thus higher loads are required to propagate the crack through the matrix [167, 168, 170]. Lebrun *et al.* [171], however, found that fibre bridging effects were negligible when the

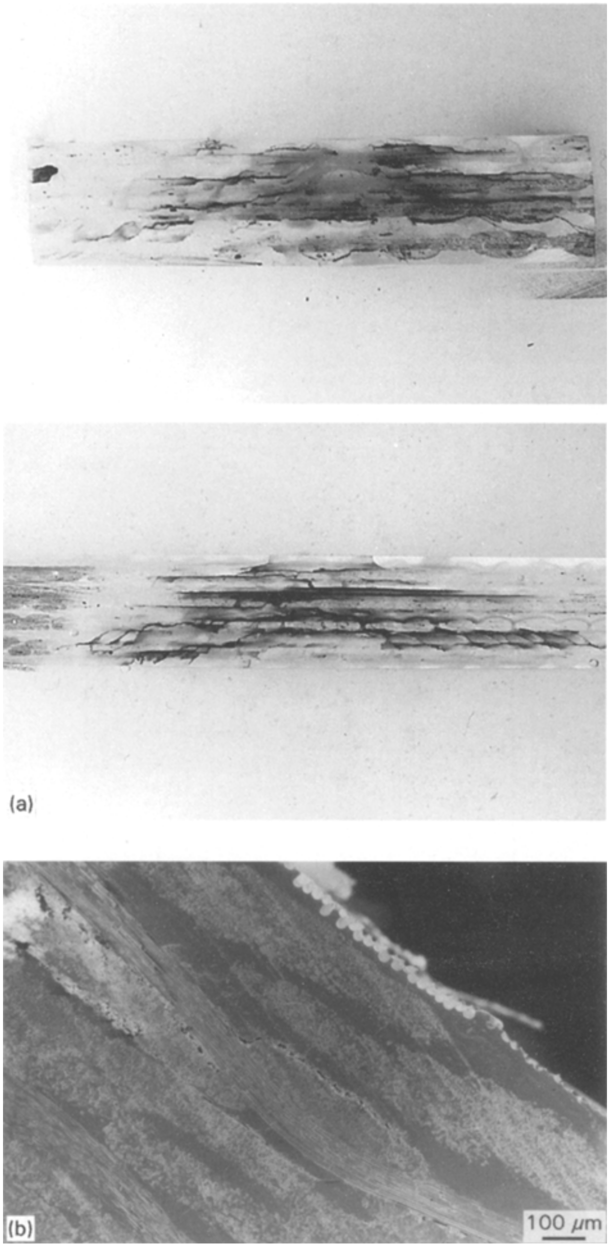


Figure 15 (a) Optical micrographs showing cracks around fibre bundles in a multiaxial (non-crimp) fabric hand lay up polyester at different impact energy levels and (b) SEM micrograph of fractures around a bundle in a glass reinforced epoxy eight-harness satin weave.

delamination was at 0°/0° and a maximum at 90°/90° interfaces. Regardless, the interlaminar fracture toughness of the stitched material is higher than that of the unstitched material [13, 167–170]. In practice, this translates into superior damage resistance. The damage area of stitched composites is less than that for unstitched systems [12, 13, 163, 172]. In addition, stitching has been shown to improve energy absorption [4,172–175].

As indicated above, the stitch and its yarn are important parameters in the subsequent behaviour of the composite to impact conditions. The knot strength indicates the fibres' ability to crimp without breaking while the denier reflects the volume per cent of stitching fibre and also the degree of in-plane fibre disruption [12, 168, 176]. The tensile strength and modulus are also relevant parameters although these yarns do not contribute to any structural reinforcement as achieved by 3-D orthogonal weaves.

The stitch pitch is also a contributing factor; too many stitches per unit length and the laminate's integrity is reduced, while too few and the stitch is ineffective in suppressing the extent of damage induced by impact [13], Fig. 16. In addition the shape of the delamination area changes when the stitch pattern is altered.

A good overview of the effects of stitching may be found in Dransfield *et al.* [177].

5. Effect of reinforcement architecture on compression and post-impact compression behaviour

The investigation of compression (tests and failure mechanisms) requires a significant review in its own right, as evidenced by the document published by the US Department of Transportation [178]. The complexity of the problem increases with plates as buckling failures are possible [68, 69, 179]. Since the review is primarily concerned with the residual compression strength after impact, an indication of the effect the textile process has on undamaged strengths relative to laminated prepreg tape composites is given.

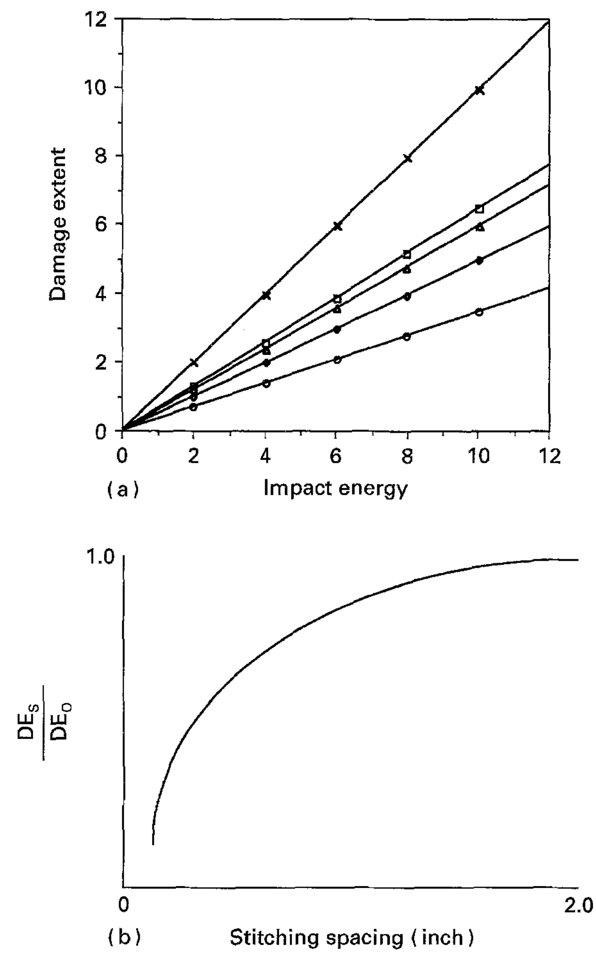


Figure 16 Influence of stitching parameters on extent of impact delamination damage (after Pelstring and Madan [13]). (a) Stitch spacing; × unstitched; □ 50.8 mm; △ 25.4 mm; ◇ 12.7 mm; and ○ 6.4 mm. (b) DE_s = damage extent with stitching and DE_0 = damage extent without stitching.

Comparisons of undamaged compression strength in absolute terms are not practical in view of the variations in the literature of fibre/resin combinations and lay-ups. In general it is acknowledged that woven fabrics are weaker than their laminated prepreg tape equivalent [0, 90] 's construction. Results indicate that woven fabric's strength are approximately 10% weaker than the non-woven material [7, 143]. Additional plate compression data of woven composites may be found in Ghasemi Nejhad and Chou [180] and Ghasemi Nejhad and Parvizi-Majidi [181]. Much of this strength reduction may be attributed to reduced fibre volume fraction and tow crimp or waviness [149, 150].

For the 3-D constructions (weaves and braids) the undamaged compression strength reduction relative to the prepreg tape composites (basic material) is generally more severe, which is associated with fewer in-plane fibres capable of carrying load [9, 113, 116, 152, 182]. This effect has been observed in braided composites such that preforms with small orientation angles have higher undamaged compression strengths than those manufactured with (large orientation angles) short pitch lengths [154]. It is postulated that macro-buckling of fibre bundles is the prevailing failure mode of braided composites. The kinked fibre bundles propagate into the surrounding matrix material, which then yields in a shear mode [36]. A brief examination of the failure mechanisms in 3-D weaves is given by Cox *et al.* [32].

The majority of data available for stitched composites has been produced using preforms constructed by a secondary stitching process that locks together individual layers of discrete unidirectional fabrics arranged in specific orientations to correspond to quasi-isotropic laminates. This manufacturing route inevitably results in some damage to and displacement of individual fibre bundles during the secondary stitching process. It is difficult to directly compare the initial compression data obtained from these materials with alternative systems such as those based on unidirectional prepreg laminates and woven fabric laminates, as the fibre volume fractions, fibre orientation and manufacturing processes are all different. The indications are that undamaged compression strengths are relatively low compared to those that may be achieved in comparable composites produced from other textiles.

In contrast, stitched composites where the stitching is an integral part of the textile construction, such as the multiaxial warp-knitted (non-crimp) fabrics would appear to offer only a limited reduction in properties compared to unidirectional prepregs (data adjusted for volume fraction). In some instances it even appears that properties are higher in multiaxial warp knitted (non-crimp) fabrics than is predicted from hypothetical laminates based on unidirectional layers of fibres in identical matrices.

To date post-impact mechanical tests have been conducted at coupon level, which is a consequence of the high material costs involved [183]. Subcomponent testing has been undertaken by Horton and Demuts [184], Card and Rhodes [185] and Madan and Shuart [186], however, this is to satisfy safety requirements and is specific to that design. Coupon data is even less useful as design data, but provides comparative results and facilitates analysis leading to a better understanding of the mechanisms involved under a representative loading spectrum.

Each class of material is treated separately in view of the different damage states (discussed earlier) arising from low velocity impact and the natural variations in failure mechanisms associated with fibre architecture.

5.1. Unidirectional prepreg tape

Residual strength is typically plotted against impact energy, and this highlights the ability of a tougher composite to minimize impact induced delamination damage and by virtue of higher fracture toughnesses appear superior, Fig. 17a. Tougher systems exhibit reduced damage for a given blow of impact energy, which translates to superior strength retention. In order to account for this, residual strength may also be plotted against the damage area or width as determined from a plan view from C-scan (this is less than

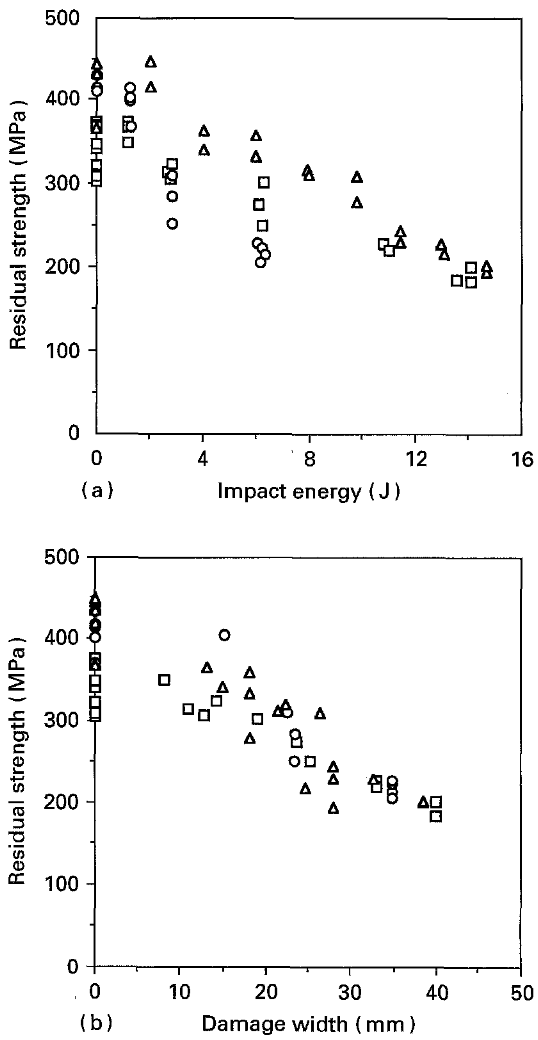


Figure 17 Compression after impact strength of toughened epoxy and two thermoplastics as a function of (a) impact energy and (b) damage width (after Bibo *et al.* [128]). Key: □ Radel; ○ epoxy; and △ APC-2.

the total delamination area due to shadow effects [36,187,188]), Fig. 17b. Consequently, relationships have been proposed to explain damage tolerance in terms of the materials toughness [189–191]. In line with this went efforts to toughen resins without penalizing other mechanical properties [14,15]. Current understanding proposes that G_{IIC} controls the extent of impact damage [139,191–194] while G_{IC} dominates during the post-impact compression phase.

Given that damage propagates perpendicular to the applied stress (Cairns *et al.* [195] and Tratt [196]), Prichard and Hogg [100] used a characteristic length parameter to quantify the extent of damage in the width or propagation direction. The results of their work, on two quasi-isotropic lay-up prepreg tape systems (toughened epoxy and thermoplastic matrices) of different toughness, demonstrated that current models based on the intrinsic properties of the material are over simplified. The data demonstrated that composites with significantly different fracture toughness exhibit equivalent compression after impact strength. This is supported by Bravenec *et al.* [189] who complied data for compression after impact strength and mode I fracture toughness from various sources. There was no correlation and the relationship appeared to be random in nature.

More recently attempts have been directed at trying to model (experimentally and numerically) and predict failure mechanisms. Experimental modelling has been directed at samples with an inclusion simulating delamination damage [197,198]. Simulation of ply cracks and fibre fracture have been shown not to affect the subsequent strength greatly under compression [199,200]. Furthermore, a single inclusion does not adequately resemble an impact damage state and consequently the residual strength is unaffected by the incorporated defect. Experiment has shown that by incorporating flaws representing delaminations in a pattern resembling impact induced damage (i.e. cone formation, etc. described earlier) that good correlation has been achieved with actual impacted samples [200]. Although the results are impressive for the flaws with a representative damage distribution, care should be taken with their interpretation as the inclusions are often square or circular and not the geometry (peanut) usually observed. The results however, vindicate the assumption that fibre fracture and shear/transverse matrix cracks resulting from impact are not instrumental in instigating the complete collapse of the specimen under load, and single delaminations have little effect on the compressive strength [201]. Despite this, the majority of theoretical analyses of the behaviour of a delaminated laminate have revolved around a "thin film" on a "parent" material [202–207] with a simple mode I fracture mechanics approach for the determination of crack growth [203–205] and/or a buckling analysis of the "thin film". Adan *et al.* [208] and Gottesman *et al.* [209] developed models for the more realistic case of multiply delaminated coupons; however, the former's analysis considered only the buckling behaviour while the latter's model was based on an iterative procedure using lamination theory to determine material properties and strengths or sublaminate buckling, but neither considered the propagation of the delaminations.

For the purposes of finite element modelling, compression after impact specimens may be assumed to incorporate delamination damage only [36,196,210]. This is fortuitous as shear/transverse matrix cracks and fibre fracture are impractical to model as the microstructure (i.e. fibre and resin) is assumed homogeneous, that is elements are defined with orthotropic properties. The failure analysis is based on sublaminate buckling and a simplified fracture criterion (usually mode I or a mixed mode) to determine delamination growth. Global or geometric instability of the damaged sample does not occur [211,212]. This is because the laminate, after impact, is asymmetric or anisotropic, so when in-plane loads are applied out-of-plane deformations occur, giving rise to peel and interlaminar shear stresses. Hence sublaminate buckling is not necessarily synonymous with failure of the laminates. Cairns *et al.* [195] and Greenhalgh [213] showed using the shadow moiré technique that out-of-plane displacements of the sublaminates increase with load prior to the propagation of delaminations. Cairns *et al.* [195] noted during experimentation that the larger delaminations could undergo a relatively stable growth phase during loading while the smaller delaminations exhibited a more catastrophic growth. Within the literature reviewed to date varying degrees of success are indicated using a sublaminate buckling or fracture mechanics approach to the growth of delaminations. Although the above process is intuitively more realistic, alternative methods have been proposed [214–217] that do not rely on the intrinsic material properties of G_{IC} and/or G_{IIC}. These models assume a reduced modulus in the damage region resulting in stress concentrations at the interface with the stiffer material.

5.2. Planar 2-D fabric

The residual compression strength of woven fabrics in percentage and absolute terms is superior to its unidirectional prepreg tape [0,90] equivalent when compared on the bases of incident energy due to the containment of shear cracking and delamination [64,209]. The cause of the lower strength retention for the non-woven material is undoubtedly the more extensive delamination, caused for a given impact energy, since this leads to reduced sublaminate stability [143].

When the woven fabrics were tested as angle (± 45) plies, the effect of impact was negligible, thus the damage caused by impact did not affect the ability of the layers to carry shear loads [143], Fig. 18.

Ghasemi Nejhad and Parvizi-Majidi [181] observed parallels in behaviour with laminates manufactured from unidirectional prepreg tape, whereby for a given woven substrate, the subsequent residual compression strength is dependent upon the toughness of the resin or damage incurred. A large strength loss was observed for the less tough material at low impact energy and then levelled off while for the tougher material, a more gradual decrease in strength was

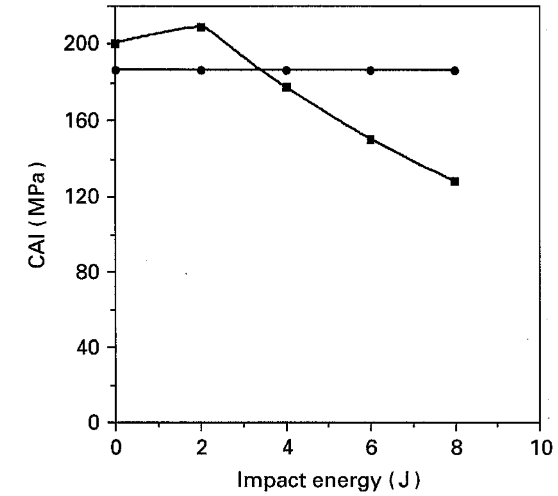


Figure 18 Effect of impact damage on the ability of woven and non-woven [± 45] laminates to carry compression loads after impact (after Curtis and Bishop [143]). Key: ● Woven fabric and ■ UDPT equivalent.

observed with increasing incident energy. Utilizing the data provided for each material in terms of the damage extent and residual strength for a given impact energy, residual strength versus damage diameter curves were generated. The damage diameter was evaluated by assuming the damage area to be circular. Plotting the compression after impact damage against damage diameter did not clearly support the trends identified for unidirectional laminates, Fig. 19. That is, while the residual strengths appear proportional to this characteristic damage dimension, the resistance to propagation of damage for the two materials was different.

The failure mechanisms of post impact compressed specimens was a shear mode on a plane oriented 45–60° with respect to the direction of the applied load. The shear mode was accompanied by "kinking" of the fibres at the line of fracture which passes laterally through the impact region, i.e. damage propagation is perpendicular to the direction of load application. The kinking is virtually a fibre micro-buckling mode; no brooming or delamination was observed at the failed section of the panels [181].

5.3. 3-D fabric

The 3-D weaves and braids possess higher compression after impact strengths than their respective 2-D constructions for a given impact energy, despite having undamaged strengths less than that of the 2-D materials [9, 113, 182, 218].

Investigations on the effect of unit cell size in braided composites showed that as pitch length increased, the residual strength decreased, even though the undamaged strength is higher in the longer pitched samples (explained earlier). This is in accordance with the amounts of damage induced for the pitch lengths concerned. A further effect realized was as tow size is reduced, damage tolerance is improved, however the effect of tow size for the same pitch length is not clear [154].

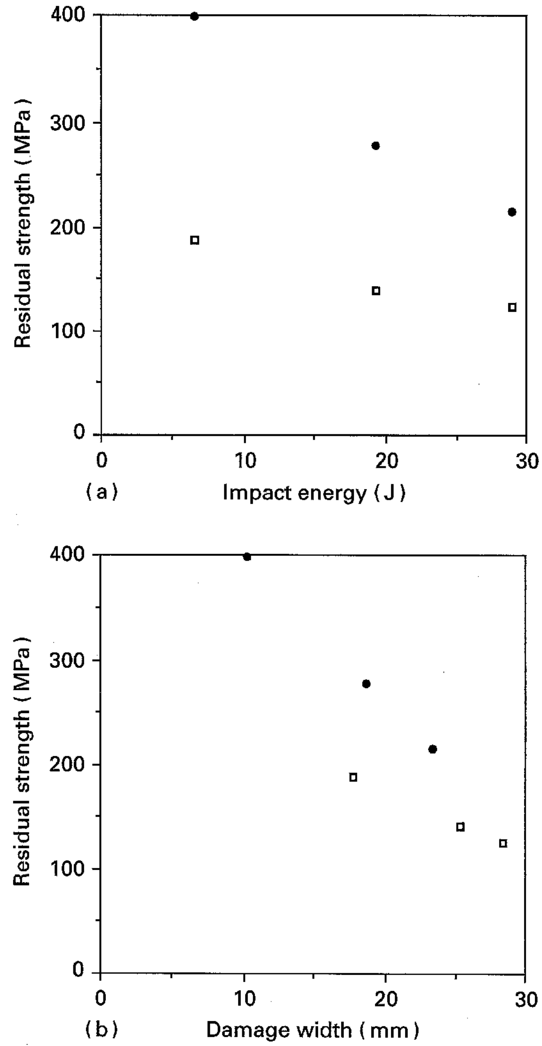


Figure 19 Compression after impact strength of woven thermoplastic composites as a function of impact energy and damage width (after Ghasemi Nejhad and Parvizi-Majidi [181]). Key: ● PEEK and □ PPS.

The use of a characteristic damage length as defined for laminated composites is supported by the findings of Li *et al.* [154], however, the physical meaning of such a term is unclear for this class of materials as they do not have any planes of delamination. Fortuitously, the two zones of damage identified from impact tests may provide a characteristic variable to which damage may be linked with residual strength in an analogous fashion to laminated prepreg tape composites. Support for this may lie in the results of Gong and Sankar [38], the relatively minor damage mode, separation of fibre tows which is prevalent in impact tests, compared with tow breakage was analogous to delamination damage in laminated composites. Furthermore, a braided composite with the same bending stiffness in the primary direction as a quasi-isotropic laminate has a lower compression strength, but superior damage tolerance as a function of incident energy.

More experimental characterization of these materials is necessary to identify the influence of tow size and weave or braid pattern on static strength, the extent of individual damage modes and their subsequent effect on the ability of the composite to carry load.

The stitched material has a higher compressive strength retention after impact than the unidirectional prepreg tape laminates for the same impact energy [120, 160, 169, 176]. As discussed for the unidirectional laminates, the apparent improvement in performance is minimized when the compression after impact strength is expressed as a function of the amount of damage induced. Although based on damaged area, Farley and Dickinson [161, 162] demonstrated the residual strength of samples with the crimped surface plies machined off to be slightly better when compared on the basis of the percentage of 0° plies with unmachined samples. The removal of the crimp did not alter the failure mechanism, transverse shear. However, in work conducted at QMW on an experimental prepreg multiaxial (non-crimp) material, with weighting of the results by undamaged strength for both systems the results showed, as a function of damage width, the multiaxial (non-crimp) material possessed a greater resistance to damage propagation than the prepreg tape equivalent [163].

Compression after impact failure modes are typically interlaminar crack growth or transverse shear failure. Interlaminar crack growth is caused by the bending/buckling of delaminated layers in the damaged region and is a function of the impact induced delamination extent. The greater the degree of delamination, the lower the load required to bend/buckle the delaminated layer(s) and hence promote crack growth. Transverse shear failed samples exhibit little interlaminar crack growth. Load is principally carried by the undamaged portion of the specimen. This failure mode is caused by the eccentric load path developed due to the conical shaped (through-the-thickness) impact damage. The eccentric load produces a local bending moment which causes the transverse shear [160]. This may be supported by observations of open hole tests [219] which show stitching to suppress crack initiation [220] and/or inhibits sublaminar instability [165, 166, 221]. This follows the trend observed by Kay and Hogg [163] for residual strength versus damage width, whereby the stitching delays or inhibits crack growth [13].

Further investigation of the stitched composites is currently underway at QMW to identify the influence of stitch yarn, density and pattern [12, 34, 168] on the mechanical properties, impact resistance (management of fracture processes) and subsequent post-impact compressive behaviour (control of fracture propagation).

6. Summary

Currently, the major issue of the damage tolerance of fibre composites is the ability of a structure to withstand compression loading after a non-critical impact blow. Composite structures based on unidirectional prepreg tape, which represent the bulk of high performance applications in the aerospace industry, suffer from significant reductions in compression strength after impact, due to the creation of delaminations in the impact event. Other forms of damage that arise, such as transverse and shear cracking are of second order importance. They both promote the formation of delaminations and influence the shape of delaminations, but they do not contribute much in themselves to the overall reduction in compression strength of the laminate.

Composites produced from a variety of textile forms, with 2-D and 3-D architecture, exhibit behaviour that increasingly diverges from that of the unidirectional prepreg tape based systems in terms of damage development during impact and subsequent damage tolerance, the more the textile structure changes from the simple stacking of unidirectional layers to an integrated 3-D arrangement.

Using fibres in textile forms can provide a method for changing both impact behaviour and subsequent compression performance. In both cases the change in behaviour relative to unidirectional prepreg tape composites arise from differences in the nature and dispersion of cracking generated in the impact event and also in the mode of propagation of that damage during subsequent in-plane compression. In extreme cases, such as interlocked 3-D braids, this arises from the elimination of weak planes for delaminations. In woven and multiaxial knitted (non-crimp) fabrics the changes in damage result from combinations of bridging between layers, direct stitching and preferential cracking around bundles.

The exploitation of textile forms presents an opportunity for continuing the development of composites with ever increasing performance and damage tolerance in particular. At this time increasing the damage tolerance of composites by matrix toughening appears to be a route that has now been exhausted after years of considerable success. It seems improbable that further increases in damage tolerance could be achieved without an unacceptable reduction in other composite attributes such as hot/wet compression strength and other temperature sensitive properties.

If this objective is to be realized, however, future effort must be directed into two areas. The first is to ensure that improvements in damage tolerance do not result in unexpected reductions in properties elsewhere. Temperature related properties are unlikely to be affected by textile architecture, but it may be that instead of post-impact compression, the focus of damage tolerance could revert to post-impact tension, or fatigue issues could begin to emerge if the nature of damage occurring in composites structures is changed.

The second area for future effort must come in the design of textile forms specifically to introduce damage tolerance, or more specifically to manage the damage that is created in a composite.

An effective use of textile forms must come from the recognition that damage will form in a composite but that effective damage management lies in directing this damage to non critical areas in forms that will not propagate or represent a major structural defect during subsequent loading.

At present most of the increasing volume of research directed towards textile forms for composites is relatively passive in influencing the nature of the textile form itself. The textile industry has numerous forms which are being taken "off the shelf" by the composites industry and evaluated for their properties when combined with carbon or glass fibres and a suitable matrix.

What is clearly desirable is an ability to design a textile form specifically to ensure a predictable and benign form of damage from service abuse. This requires continued efforts to understand the links between fibre architecture and cracking, the relative ease or difficulty in subsequently propagating different damage forms, and the effect of those damage forms on subsequent properties.

Hand in hand with a refocusing on textile design, rather than merely textile utilization should come a change in focus of modelling efforts, especially those based on finite element analysis.

Current work in the field is devoted to modelling the behaviour of laminates based on unidirectional prepreg tapes. This really translates into predicting the post-impact compression strength of impacted laminates and a considerable wealth of data exists against which the predictions can be assessed. Given the complexity of the damage states involved, in terms of the totality of cracking; delaminations, shear cracks, etc., the dispersion of delaminations, their size and shape, much of the work centres on attempts to optimize the ease of calculation with the accuracy of the prediction. The damage states are simplified in the models, and often do not readily compare to physical reality, but reasonable predictions can be made. This is a logical process which should lead ultimately to simplified models allowing the prediction of the residual properties of representative structures. The work however assumes a particular composite form and is primarily focused on the performance of the structure (i.e. the collapse of the test coupon).

What might be needed in the future is an attempt to lead textile design by producing models that predict the performance of conceptual textile forms, both in terms of what sort of damage might be created and how stable that damage might be. The modelling process should therefore become materials focused and provide the numerical justification for pursuing various textile approaches to the most efficient use of fibres in composites.

Inevitably this approach would require a far more realistic approach that attempts to move closer towards incorporating physical reality rather than over-idealized forms and an increasing emphasis will also be required on the impact event as well as the subsequent compression.

7. Conclusions

The damage sustained by a composite during impact, and other forms of service abuse, is a function of its fibre architecture. Similarly, the behaviour of a composite during post-impact compression loading is a function of that damage, its susceptibility to propagation and the severity of its effects on reducing compression strength. By providing a route to controlling fibre architecture, textile forms offer a route to controlling damage tolerance.

This facility coupled with the enormous manufacturing options presented by textile forms, e.g. net shape production of performs, drapeable fabrics, etc., will continue to increase the ability of composites to compete for new and more demanding applications.

The exploitation of textile forms to achieve greater damage tolerance without compromising other properties requires continuing study of the links between fibre architecture and damage development and propagation. At present textile forms with 3-D, through thickness reinforcement appear to offer the best potential for damage tolerant structures but the best route to optimizing damage tolerance and in-plane properties has not yet been defined.

The field is clearly not yet mature. Much work remains to be done in exploring the full potential of textile forms. This might include incorporating different types of fibres in a single perform, the selective modification of some tows but not others (e.g. for matrix compatibility, bond strength, etc.) and in the detailed organization of fibre architecture. Continuing research into damage mechanisms and modelling of damage processes should ultimately allow a full utilization of textile technology by the composites industry and a downgrading of the damage tolerance problem for composite structures.

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