STRATEGIES FOR IMPROVING THE DAMAGE TOLERANCE OF FUTURE COMPOSITE MATERIALS

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Abstract

Improving the damage tolerance of a composite is identified as a key goal for future developments. A change in emphasis is suggested with a move away from the philosophy of improving performance by increasing the resistance of the laminate to damage formation, which should be accompanied by attempts to control the nature of damage such that energy can be absorbed through cracking in a way that is not deleterious to the final properties of the structure.

Different strategies are proposed to achieve this, consisting of specific attempts to maximise the resistance of the laminate to damage growth by improving interlaminar toughness, G_I (prop) or by introducing new architecture into the composite such that the nature of damage is modified. Examples are provided of the use of interlaminar veils, 3D textile forms and the use of specific fiber hybridisation to facilitate the improved resistance to crack propagation.

Introduction

There is a continuing need to develop composites that are damage-tolerant in service. Damage tolerance in this context means that a composite part can suffer some form of service abuse that results in damage without the service performance of that part falling below an acceptable level. The need for damage tolerant composites transcends all parts of our industry and is common to structures made from glass fiber and carbon fiber, for aerospace to marine and wind turbine applications. Perhaps the most high profile and historically most critical applications have been carbon fiber composites in primary aerospace structures. Here the problem of damage is compounded by the fact that damage is commonly sub-surface, undetectable by eye, but can have a considerable detrimental effect on key laminate properties such as compression after impact [Bibo, Hogg, Backhouse and Mills, 1998].

The importance of damage tolerance in the aerospace sector is increasing given the trend to manufacture more primary civil aircraft structures from CFRP, including the fuselage which is susceptible to significant airfield abuse (baggage handling, vehicles etc) and from large hail stones. However damage tolerance in general is also of prime concern to the wind turbine industry and automotive sector. It is simply not economical to replace damaged wind turbine blades in service, especially if the turbines are deployed off-shore [Det Norske Veritas, 2008]. The automotive sector is used to operating with metallic materials and there is no culture of (or infrastructure for) in-service non-destructive inspection of structures that would be required if the predicted widespread usage of composites for automotive structures becomes a reality but based on the use of composites with poor damage tolerance.

The industry has made considerable progress for a thirty year time period in improving the damage tolerance of all classes of composites. This is exemplified by the improvements in carbon fiber composites intended for service as primary structure in aerospace where the damage tolerance of the composite laminate is usually quantified by measuring the compression after impact strength according to some recognised standard (ASTM). Low energy impact is a significant in-service problem which results in difficult to detect damage, primarily consisting of delaminations, Fig.1, which can seriously reduce the compression strength of a composite laminate [Saito and Kimpara,2006, Prichard and Hogg,1990].

The compression after impact properties have increased when measured using the Boeing standard test method from a level of about 170 MPa to almost 400 MPa in this period. The improvements have been largely the result of increases in resin toughness, with associated increases in interlaminar strength [Bascom, Bitner, Moulton and Siebert, 1980]. The developments, Fig.2, have included the introduction of formulated epoxies, followed by rubber-toughened systems [Davy, Hashemi, and Kinloch 1989], thermoplastic toughened systems [Di Pasquale, Motto, Rocca, Carter, McGrail, and Acierno, 1997], through to thermoplastic, phase inverted thermoplastic-epoxies, monolithic thermoplastics and interleaf toughened epoxies [Wong, Lin, McGrail, Peijs, and Hogg, 2010].

It is noticeable however that no significant improvements in damage tolerance have been developed over the last ten years, possibly as no specific new targets have been set by the industry. Boeing had stipulated a challenging CAI value of 350 MPa for selection of prepreg systems for use on the Boeing 777 empennage which was achieved by Toray with their 3900 resin prepregged with T800 fibers (Boeing). Most work undertaken recently by industry has focused on developing composite systems (materials and preforms) that can be used in alternative out-of-autoclave processes to deliver equivalent damage tolerance to conventional autoclave processed materials. It might be concluded that the process of achieving increased toughness via improvements in resin toughening have reached the end of its life and that there is little room for further development. If this is indeed the case then alternative strategies for developing an enhanced damage tolerance must be considered. In aerospace the desirability of developing a carbon fiber fuselage for medium size single aisle commercial aircraft, when the fuselage thickness is very low, means that composites systems with a damage tolerance substantially above the level achieved to date are required. While the composite fuselage of a large twin aisle aircraft is tough enough to withstand damage from hail stones, service abuse etc, a thinner single aisle aircraft would not be capable of safe operation. Whereas this is a particular issue for the aerospace sector, any alternative approach to increasing damage tolerance will be welcomed by all sectors of the composite industry. Sectors such as the wind and marine industries that rely largely on infusion processing are restricted from using epoxies toughened with large quantities of thermoplastics as the resins are too viscous to be infused. Variants of infusion processing are now also being adopted by the aerospace sector, emphasising the need for alternative solutions to generating damage tolerant composites.

Damage Tolerance : Mechanisms

The compression after impact test used to measure the damage tolerance of composites consists of two separate events- a low energy out of plane impact test, and a subsequent in-plane compression test. A traditional method of presenting the data obtained from such testing is plotting the compression strength versus the kinetic energy of the striker, sometimes normalised to account for variations in the target laminate thickness. This presentation obscures the relative performance of the composite under each part of the test protocol. Observation by many research groups over the years has shown that variations in materials constituents can influence the extent and nature of damage formed in the initial impact test [e.g Hull and Shi, 1993, Shyr and Pan, 2003 and Naik and Logan 1999]. The damage is primarily delaminations that form between different layers of the composite. In a quasi-isotropic laminate the individual delaminations take a peanut shape in each layer and combine overall to provide a roughly circular damage zone when detected using NDE techniques such as ultrasonic C-scan, Fig.1. The magnitude of this damage has been characterised by different groups according to either the area of the damage zone or the width of the damage zone [Saito and Kimpara, 2006, Prichard and Hogg, 1990].

In the second part of the test the delaminations will propagate towards the extremity of the plate loaded in compression, at which point the laminate may buckle in the center with major failure taking place usually via an in-plane shear fracture. While it has not proved very easy to observe the propagation of the delamination during compression failure of carbon fiber composites due to their opacity, it has proved feasible to study the growth in glass fiber systems which show that the delaminations only propagate perpendicular to the loading direction. This has prompted the use of damage width as being the key damage parameter [Prichard and Hogg, 1990]. It is therefore possible to deconstruct the compressing after impact test and observe the effects of materials variables such as resin toughness on (a) the initial width of the delaminations after the initial impact as a function of kinetic energy of the striker and (b) the compression after impact strength of the laminate as a function of the width of the delaminations.

What is noticeable is that when the data has been examined in this way, the prime influence of increasing resin toughness is to reduce the extent of the initial delaminations in the laminate. When laminates are then compared on the basis of the damage width, irrespective of what level of impact blow was required to create that damage, then laminates with a different toughening system, result in almost identical compression strengths, Fig.3. This is an important observation in the light of the apparent roadblock on improvements in damage tolerance that result from increases in resin toughness.

The philosophy adopted by the industry over the years has in effect been to improve damage tolerance by restricting the tendency of the composite to suffer damage in the first instance. If this philosophy cannot provide any room for improvement, then an alternative strategy needs to be considered, namely to allow damage to form (or at least recognize the inevitable damage at a sufficient impact blow) but to focus on reducing the consequences of that damage under subsequent loading. This could be approached by changing the nature of the damage itself such that it does not reduce the compression strength, and by increasing the resistance to growth of that damage provided by the laminate.

Strategies for Improving the Resistance to Damage Propagation

There are two independent strategies that can be considered in order to increase the resistance of a laminate to damage growth. The first of these is to incorporate into the laminate features and modifications that provide additional toughness in the propagation phase, whilst accepting that damage will be formed in essentially the same manner as in traditional composites. This implies that the laminate must achieve a higher value of interlaminar toughness $G1_{(prop)}$, a property that is not linked closely to the resistance of the laminate to the initial creation of that damage under impact conditions.

The second strategy is to deliberately modify the architecture of the composites such that when damage forms it does so in a different geometric pattern and that the nature of this new form of damage is that it is either harder to propagate and/or is less deleterious on the compression strength of the laminate.

Improving Resistance to Crack Growth

Strategies for increasing the resistance to crack growth could concentrate on increasing interlaminar mode-I toughness. A simplistic interpretation of the various parameters influencing the compression after impact performance suggests that the Mode-II toughness is primarily responsible for the extent of the initial damage formed during the impact event, while mode-I interlaminar toughness controls the susceptibility to crack propagation in the final compression phase [Prichard and Hogg, 1990]. Techniques such as through-thickness stitching or tufting can be considered in some situations as routes to improving Mode-I toughness, although these approaches have little effect on Mode-II toughness, Fig.4. The use of stitching/tufting implies an additional operation in the manufacturing process and as such is expensive and these methods also have the drawback of disturbing the underlying fibers, creating possibly defect sites and reducing the overall volume fraction of the reinforcement [Potluri, Sharif, Jetavat, Hogg and Foreman, 2009]. The technique has been shown to have merit in certain situations, especially for the local reinforcement of key interlayer connections (for example at T-sections) but it may not be suitable for application over a large area such as a wing skin or fuselage, and may be too intrusive in a thin laminate.

In some cases stitching has such a deleterious effect on the underlying composite structure that the susceptibility of the laminate to the formation of damage in the initial impact phase is increased. This is illustrated by the following data on infused carbon fiber epoxy cross-ply laminates where a variety of stitch pitches (separation) were used, Fig.5 [Ahmadnia, Daniel and Hogg, 1999]. The basic compression after impact strengths were reduced as a result of the stitching in all cases due to an increase in the initial impact damage size, but the resistance of the laminate to subsequent crack propagation was improved.

An alternative approach is to selectively reinforce the interlaminar region with a fibrous veil, Fig.6. In some respects this mimics the use of interleaved tough resin layers that are very effective in improving damage tolerance by minimising the initial damage area due to impact. A tough resin interlayer can yield in shear during impact.

However during the crack propagation phase such an interlayer is thought to be less effective as Mode-I crack growth could occur at the interface between a tough zone and the underlying composite. If a thin fibrous layer such as a low areal weight veil is used, especially where the fibers are thermoplastic fibers, then the ability to resist initial crack growth during impact is maintained. However in addition, due to the fibrous nature of a veil it is likely to interact with the underlying structural layers facilitating a degree of crack bridging which will enhance Mode-I toughness [Kuwata, 2010]. Data on the performance of veils in the interlaminar region is sparse but encouraging

[Hogg, Kuwata, Jamshidi, Hjadaei and Teztzis, 2010]. Some thermoplastic veils have demonstrated an ability to improve both the resistance to damage formation (impact) and propagation (compression) resulting in a significant improvement in compression after impact strength [Hogg et al], Fig.6. The effectiveness of an interlaminar veil seems to depend on how and where a delamination propagates in the laminate.

Stiff veils based on fibers such as carbon, seems to force the crack to propagate at the boundary between the structural layers and the resin rich zone adjacent to the veil. This only results in a marginal increase in toughness and minimal fiber bridging. A soft veil based on thermoplastic fibers in contrast induces crack propagation to occur at the interface between the veil and resin rich layer and within the veil itself. This maximises pull out and fiber bridging, results in a significant increase in Mode I and Mode II toughness and is responsible for the improvement in compression after impact properties. The various possibilities are shown schematically in Fig.7. This form of toughening also has the advantage of being compatible with out of autoclave vacuum infusion processing.

Mitigating the Consequences of Damage : Mesoscopic Control

The second strategy proposed is based on the concept that cracking is inevitable eventually. Instead of seeking to always improve the resistance to initial cracking, it may be more productive to allow or tolerate cracking but to ensure that this takes place in positions that are not likely to cause a significant reduction in compression strength. To facilitate energy absorption by cracking, without serious consequences, the damage must be diffused into the laminate, with sufficient undamaged material remaining in critical positions. The basic concept is illustrated simply in Fig.8. For an equivalent damage area, a concentration of cracks as long delaminations leads the structure susceptible to buckling failure, whilst a distribution or diffusion of damage would result in shear failure at higher loads.

A number of different concepts have been proposed to achieve this with one of the first practical attempts to introduce a controlled damage state being that of Sun and Norman in 1990. Their approach involved introducing adhesive strips into a cross ply laminate that would form a grid that inhibited crack growth on impact. This method reduced damage area significantly and resulted in a modest improvement in residual post impact compression strength, albeit at the expense of a lower initial undamaged compression strength, Fig.9.

Some other early observations were based on the performance of non-crimp fabrics. Early examples of Non-Crimp Fabrics (NCF's) prepared for the marine industry and consisting of glass fibers assembled using a polyester knitted mesh, show that cracking occurred in impact by growing around isolated bundles. The apparent effect was very similar to laminates made from similar lay-up orientations using unidirectional prepregs. The overall area of damage was not greatly changed. However the compression strengths of the laminates, as a function of the original strength were higher. The cracks around bundles gave the appearance of a macroscopic delamination but the individual bundles cracks had not coalesced and as such did not reduce the in-plane stiffness to the same degree as a major delamination, Fig.10. Compression after impact strength was therefore increased [Hogg, Ahmadnia and Guild, 1993].

This effect was emphasised when comparing composites manufactured from nominally identical non crimp fabrics via different process routes including prepreg/autoclave, resin transfer moulding and vacuum infusion. The RTM and Vacuum infusion laminates retained more of the discrete bundle structure after processing than prepregged NCFs and as a consequence exhibited superior compression after impact properties as a function of impact energy. Part of this was due to a reduction in damage area.

However, as is shown in Fig.11, when the same results are plotted as a function of damage width, it is apparent that the relative performance changes but there is still an improvement with the RTM and Vacuum infusion laminates suggesting that at least in part they are better at resisting crack growth than their prepregged analogues. It is interesting to note that woven fabric laminates, also feature in Fig.11, exhibit similar overall damage tolerance to the NCF prepregged laminates, but when the data is compared on the basis of damage width, the woven materials appear inferior. This suggest that the damage in the woven fabric is smaller for a given impact but the consequences are more significant and more likely to propagate. The retention of fibers in a discrete bundle format in non-crimp fabrics is encouraged by many factors including the presence of the fine knitting yarn that assembles the fibers and keeps them in place within the fabric. The knitting yarn may have a high tension which will ensure that the bundle integrity is retained. The recent industrial attempts to improve NCF fabric properties have focused on providing a uniform distribution of fibers within a layer thereby eliminating the discreet bundle geometry of these early fabrics. Another mechanism for controlling crack morphology could be the use of more complex textile architectures including the use of 3D woven fabrics with angle-interlocking formats, orthogonal weaves or layer to layer 3D structures, Fig.12.

In all such cases the yarns in the various directions are kept in position by the presence of other yarns in differing orientations and there is little or no possibility of bundles merging and spreading. The consequence is that, in addition to 3D fabrics possessing no obvious weak interlaminar plane, the nature of damage that forms in their composites has a very different format to that created in simpler 2D materials. In all such cases the yarns in the various directions are kept in position by the presence of other yarns in differing orientations and there is little or no possibility of bundles merging and spreading. This can be illustrated by the x-ray tomographic images that reveal the cracking patterns and cracking densities in different 2D and 3D versions of otherwise similar materials (S-2 glass/epoxy) and subjected to similar impact blows, Figs.13 and 14. The 2D fabrics themselves in this case are unusual. The bundles are very discrete and tightly woven. The impact event results in a number of discrete debonds from woven fiber yarns where the debond extends to the area of one tow that is crossed by an adjacent tow. This in effect creates small pockets of debonded bundles but does not result in an extensive single delamination as would be expected in as laminate based on unidirectional prepregs, Fig.14c. When the laminates shown in Fig.13 were subjected to compression after impact [Potluri, Hogg, Arshad, Jetavat and Jamshidi, 2011] it was determined that the 3D fabrics provided much greater resistance to damage formation in the impact event than both UD cross ply materials and 2D woven fabrics as shown in Fig.15.

However when tested in compression, Fig.16, the actual residual properties were inferior to the cross ply laminates, which seems to be a result of the lower undamaged compression strength in these materials, which can itself be linked to the presence of crimped and wavy fibers in the 3D architectures. The 3D fabric laminates tested are a little unrepresentative of commercial materials as their fiber volume fraction is low (a function of the yarn texture used). It is apparent that judicious control of the fiber architecture within a composite can influence the nature and subsequent growth of damage.

At this stage the nature and location of damage has not been deliberately controlled and fracture events are allowed to occur naturally. It can be envisaged that a next stage in this process would be the deliberate introduction of selective interfacial control at bundle level to trigger fracture at sites and positions that are considered benign within the overall composite architecture. This could allow the dissipation of energy within the structure without creating defects that are liable to propagate and lead of a significant reduction in properties.

Mitigating the Consequences of Damage: Microscopic Control

The dissipation of energy within a structure via controlled cracking could also be pursued by control of deformation processes at the micro (fiber) level. An interesting hybrid composite system has been explored over a number of years [Hogg, 2005], conceived as a route to providing through penetration impact and ballistic resistance. This has involved the use of hybrid yarns systems whereby a structural fiber is combined with a non-structural low cost but ductile fiber in a commingled yarn that is used to produce an otherwise conventional fabric, Fig.17a.

These materials when impregnated with a thermosetting resin provide significant improvements in the energy necessary to rupture a composite plate, in some cases by a factor of 2 to 3. Typical combinations of materials include glass fibers with polypropylene fibers and carbon fibers with polyamide fibers. An unexpected benefit may also result from the use of such hybrid systems in compression after impact. The data shown in Fig.17b illustrates this point for a carbon fiber/nylon commingled laminate infused with a low temperature epoxy. The compression after impact strength as a percentage of the original compression strength is much higher than that of the equivalent composite produced from a plain carbon fiber woven fabric. The result should be considered in context however. It demonstrates the greater resistance to damage propagation in this system, which may be due in part to some thermoplastic fiber bridging or possibly crack blunting and damage delocalisation as a result of multiple debonds between thermoplastic fibers and the thermosetting matrix. The absolute compression after impact strength of the hybrid composite is poor as the introduction of the thermoplastic fibers reduces the volume fraction of structural carbon fibers that can be contained in the final laminate, thereby reducing compression stiffness and compression strength. Hybridisation of tows may in principle however provide an additional mechanism for damage dissipation by cracking between dissimilar materials without reducing the strength of the material itself under the critical loading conditions imposed.

A Future Strategy

The control of damage and an improvement in the resistance to crack propagation can evidently be expected if composites are designed with new fabric additions and architectures. At present a systematic assessment of architectural effects on damage diffusion into a laminate has not been adequately explored.

There are numerous concepts that have been proposed and which show some promise as a route to improving these critical properties. However most of the results have been generated in an ad-hoc fashion. What is needed now is a modelling-led drive to examine the consequences of deliberate architectural modifications, the role of weak interfaces at bundle and fiber level, and the combination of different features with a view to providing a step improvement in properties.

Examination of an SEM micrograph, Fig.18, of a region from an impacted woven carbon fiber laminate shows extensive cracking within a fiber bundle. If the accompanying fractures in the orthogonal woven fiber bundles had been avoided, perhaps this composite would have absorbed more energy without suffering from significant reduction in compression strength. Selective local control of bond strength, between fibers and/or bundles could facilitate easy/early cracking in some locations to absorb energy and avoid damage in others.

The following image, Fig.19, of a hybrid carbon-nylon bundle also illustrates some possible routes to achieving selective and controlled cracking. The hybrid bundle of carbon and nylon fibers is wrapped with a thin layer of nylon fibers which would act as a weak interface allowing primary bundle cracking to occur. Such bundles could be introduced randomly throughout a laminate or spaced at critical locations. The ability to deploy relatively straight forward textile processes to organise fibers and bundles with specific properties suggest that a new toughening technology is waiting to be developed. What is missing at present however is the ability to predict the effects of local architectural control to introduce selective areas of weakness. All results to date have been based on empirical observations which, while proving insights, are not sufficient to produce an optimised new generation of materials.

This discussion has not as yet mentioned the possible use of nano-materials as an added route to toughening a resin and improving damage tolerance. This is not to say that such a route does not have some value. However while the addition of most nano-particulates and even carbon nanotubes, does indeed increase resin toughness, it does not seem to translate into a significant improvement in composite damage tolerance [Iqbala, Khana, Munirb and Kim, 2009]. This is likely to be because the ultimate benefit in resin toughness has already been achieved. However the addition of nano-materials in low quantities does not greatly influence the viscosity of the resin and as such, a nano-composite route is probably of greatest benefit in restoring the toughness of laminates made via an infusion process where resins are produced without conventional toughening additives because of viscosity based restrictions on the processing route. Hence nano-materials could be an important complement to complex 3D preforms allowing a base line resin toughness to be retained whilst the composite uses an infusion processes to consolidate a complex architecture. Nano-additives may also have a specific role to play as coatings on fibers to yarns to assist in interfacial control.

What is most likely is that the step change in composite damage tolerance, as defined by a compression after impact measurement, will need an integrated approach harnessing many different individual toughening mechanisms to result in the improvements required by the industry. The problem is made more complex by the need to ensure that any solution is cost effective and compatible with manufacturing processes, and that additional service requirements such as lightning strike protection can be integrated within a commercial system. The indications however are that progress can be made as long as the industry is prepared to accept new material forms and to move away from conventional unidirectional prepreg-based construction methods.

The idea of accepting that damage will happen and seeking to nullify the consequences of damage, as opposed to continually seeking to stop damage forming has been applied in other fields of materials and by nature. A classic example lies in the use and role of rubber particles in high impact modified polystyrene (HIPS). Polystyrene is a brittle polymer at room temperature that readily forms crazes that grow into cracks.

Rubber particles in HIPS promote craze formation but ensure that crazes grow between the rubber particles and are stabilised and do not readily propagate as cracks, converting a brittle polymer into a tough system, Fig.20. A similar process occurs in ABS, Fig.20. In nature, wood and bone have micro and ultra-structures that allow cracks to form but do not provide a path for subsequent propagation.

The challenge for composite engineers is to create systems based on similar concepts that work in structural fiber composites, without incurring any weight or other performance penalties.

References

1. Ahmadnia, A., Daniel, L and Hogg, P. J., "The Effect of Secondary Through-thickness Stitching on Damage and Energy Absorption in Glass and Carbon Fiber Laminates", Fracture and Damage Mechanics, Ed. M. H. Aliabadi., Queen Mary and Westfield College, ISBN 0 904188 54X, 1990, pp.489-503.

2. Argon, A. S., Cohen, R.E., in Kausch, H.H. (Ed.), Crazing in Polymers, p.301, Springer.

3. "Test Method Measuring the Damage Resistance of a Fiber-reinforced Polymer Matrix Composite to a Drop-weight Impact Event", American Society for Testing and Materials, West Conshohocken, Pennsylvania, 1990, ASTM, D 7136, p.301

4. Bascom, W. D., Bitner, J. L., Moulton, R. J. and Siebert, A. R., "The Interlaminar Fracture of Organic-matrix, Woven Reinforcement Composites", Composites, 11(1), 1980, pp.9-18.

5. Bibo. G.A., Hogg, P.J., Backhouse, R. and Mills, A., "Carbon Fiber Non-crimp Fabric Laminates for Cost Effective Damage Tolerant Structures", Composites Science and Technology, 58, 1998, pp.129-143.

6. Boeing, Materials specification BMS 8-276.

7. Cauchi Savona S., "Energy Absorbing Composites for Crash Energy Management", Ph.D Thesis, Queen Mary, University of London, 2005.

8. Davy, G., Hashemi, S and Kinloch, A. J., "The Fracture of a Rubber-modified Epoxy Polymer Containing Through-thickness and Surface Cracks", International Journal of Adhesion and Adhesives, 9 (2), 1989, pp.69-76

9. Det Norske Veritas, A.S., "Future Perspectives for Design and Testing of Wind Turbine Blades", "A White Paper on a Rational Approach to Defects and Damage Tolerance", Det Norske Veritas AS, Report No. WTDK - 6022. (http://www.dnv.us / binaries / damage_tolerance-white_paper_2008_tcm153-29 5072.pdf), 2008.

10. Di Pasquale, G., Motto, O., Rocca A., Carter, J. T., McGrail, P. T and Acierno, D., "New High-performance Thermoplastic Toughened Epoxy Thermosets", Polymer, 38 (17), 1997, pp.4345-4348.

11. Hogg, P. J., Ahmadnia, A and Guild, F. J., "The Mechanical Properties of Non-crimped Fabric-based Composites", Composites, 24, 1993, pp.423-432.

12. Hogg, P. J., "Toughening of Thermosetting Composites with Thermoplastic Fibers", Materials Science and Engineering: A, Vol.412, No.1-2, pp.97-103.

13. Hogg, P. J., Kuwata, M., Jamshidi, M., Hjadaei, A and Teztzis, D., "The Potential for Improved Durability in Vacuum Infused Composites by the Use of Interlayer Veils", Compo 2010, Harbin, China, 2010.

14. Hull, D and Shi, Y. B., "Damage Mechanism Characterization in Composite Damage Tolerance Investigations", Composite Structures, 23, 1993, pp.99-120.

15. Iqbal, K., Khan, S. U., Munir, A and Kim, J. K., "Impact Damage Resistance of CFRP with Nanoclay-filled Epoxy Matrix", Composites Science and Technology, 69 (11-12), 2009, pp.1949-1957.

16. Kuwata, M., "Mechanisms of Interlaminar Fracture tOughness Using Non-woven Veils as Interleaf Materials", Ph.D Thesis, Queen Mary, University of London, 2010.

17. Naik, R. A and Logan, C. P., "Damage Resistant Materials for Aeroengine Applications", AIAA Paper 99-1370, 40th AIAA/ASME/ASCE/AHS/ASC Structures, Structural Dynamics and Materials (SDM) Conference, St. Louis, MO, 1999.

18. Michler, G. H., "Determination of the Morphology and Mechanical Microprocesses in Polymer Combi-

nations by Electron Microscopy", Polymer, 27, 1986, pp.323-328.

19. Prichard, J. C and Hogg, P. J., "The Role of Impact Damage in Post-impact Compression Testing", Composites, 21, 1990, pp.503-511.

20. Potluri, P., Sharif, T., Jetavat, D., Hogg, P. J and Foreman, A., "Benchmarking of 3D Preforming Strategies", International Conference on Composite Materials ICCM17, Edinburgh, Jul 2009, IOM3.

21. Potluri, P., Hogg, P. J., Arshad, M., Jetavat, D and Jamshidi, P., "Influence of Fiber Architecture on Damage Tolerance in 3D Woven Composites", Deformation and Fracture of Composites (DFC-11), Cambridge, April, 2011.

22. Saito, H and Kimpara, I., "Evaluation of Impact Damage Mechanism of Multi-axial Stitched CFRP Laminate", Compos Part A, 37, 2006, pp.2226-2235.

23. Shyr, T. W and Pan, Y. H., "Impact Resistance and Damage Characteristics of Composite Laminates", Composite Structures, 62, 2003, pp.193-203.

24. Sun, C. T and Norman, T, L., "Design of a Laminated Composite with Controlled-damage Concept", Composites Science and Technology, 39 (4), 1990, pp.327-340.

25. Wong, W. Y., Lin, L., McGrail, P. T., Peijs, T and Hogg, P. J., "Improved Fracture Toughness of Carbon Fiber/Epoxy Composite Laminates Using, Dissolvable tHermoplastic Fibers", Composites Part A, 41, 2010, pp.759-767.

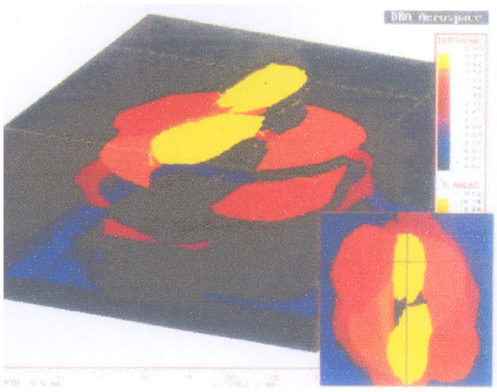


Fig.1 Time of Flight Ultrasonic C-scan Image of Impact Damage in a Quasi-Isotropic Carbon Fiber-Epoxy Laminate

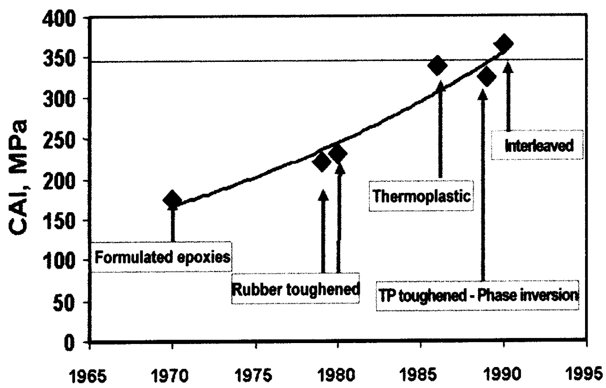


Fig.2a Typical Compression After Impact Test Results for Various Categories of Toughened Composites (Data Selected from Multiple Literature Sources)

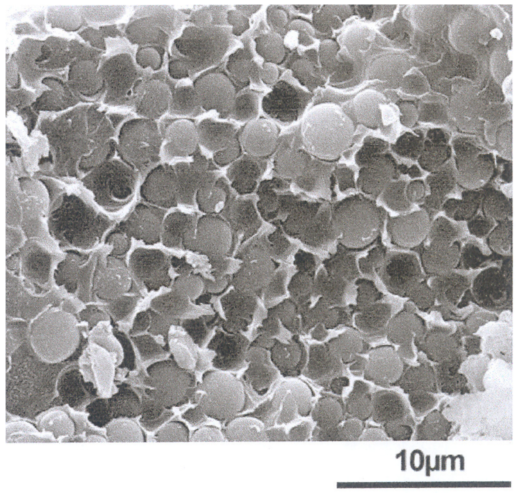


Fig.2b Fracture Surface of an Epoxy Resin Toughened with a Thermoplastic Additive, Showing Phase Inversion. Islands of Thermoset Polymer are Surrounded by a Tough Contiguous Thermoplastic Phase [Wong, Lin, McGrail, Peijs and Hogg, 2010]

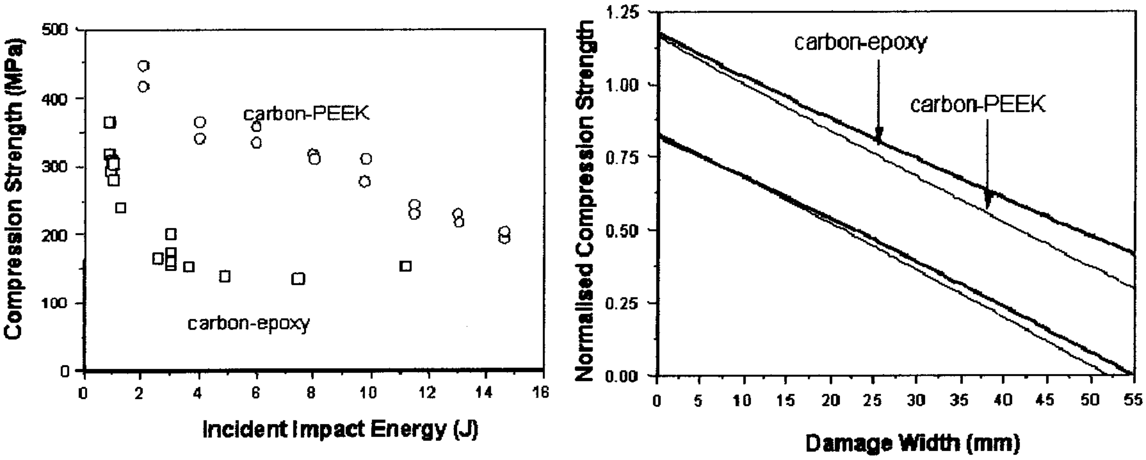


Fig.3 A Set of Data for the Compression After Impact Strength of a Conventional, Quasi-Isotropic Carbon Fiber Epoxy Laminate and an Equivalent Carbon Fiber PEEK Laminate. The first plot shows the data plotted as a function of impact energy and the PEEK laminates are clearly superior. The second plot shows 95% prediction internals for the same data normalised and plotted as a function of damage width showing that both data sets are now almost identical [from Prichard and Hogg, 1990]

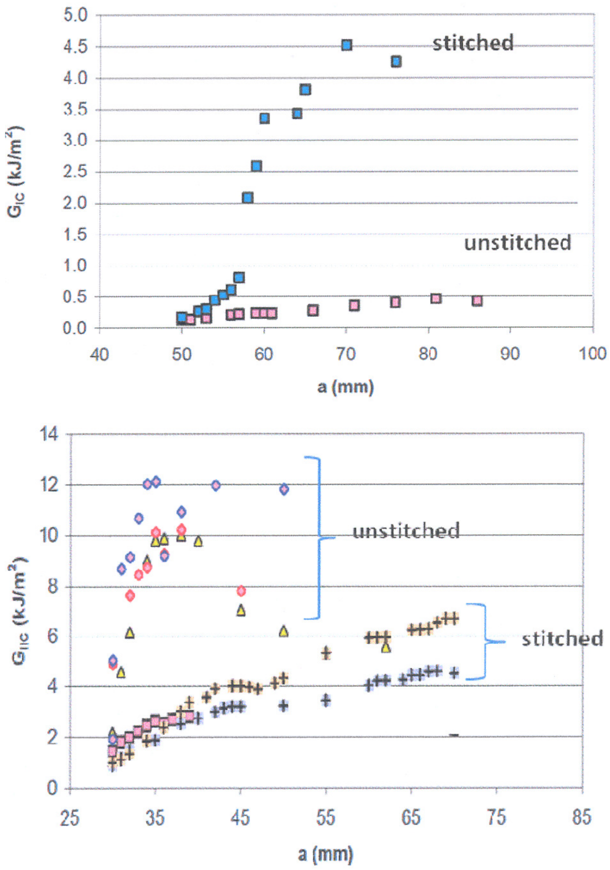


Fig.4 Two sets of data from similar resin/glass fiber laminates showing the effect of stitching on R-curves for F_{IIC} (UD laminates) and G_{IIC} (multi-axial), from Cauchi Savona, 2005

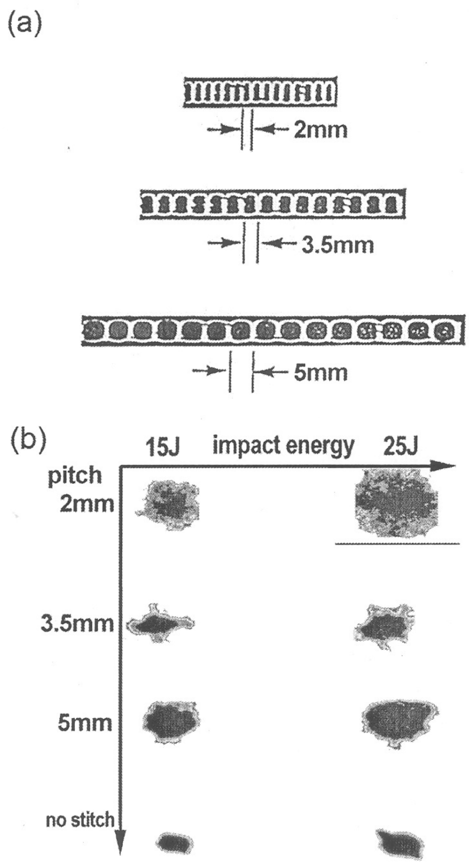


Fig.5 A Set of Data Illustrating the Effect of Stitching on CAI Properties. (a) The Pitch of the stitching is varied from 2 to 3.5 to 5mm. (b) The damage at a given impact blow is increased in the stitched laminates.

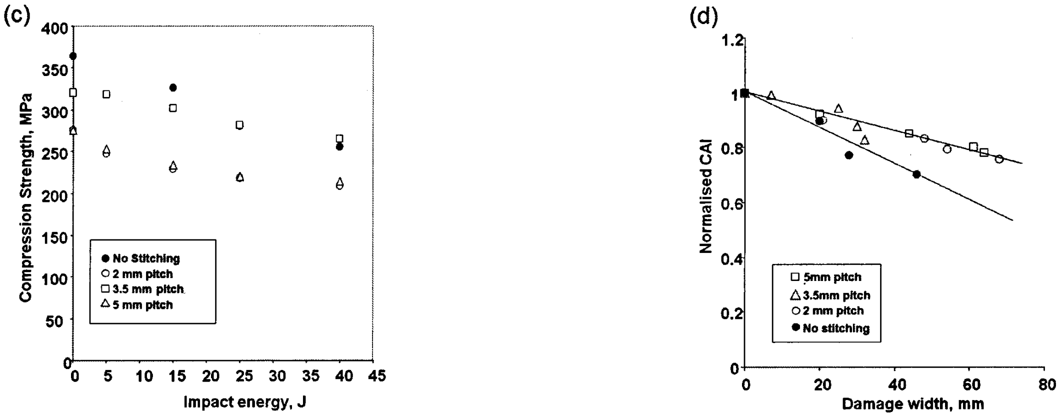


Fig.5 A Set of Data Illustrating the Effect of Stitching on CAI Properties. (c) The absolute compression after impact strength is reduced in the stitched laminates (open symbols) but (d) the resistance to crack propagation (delamination growth) is increased (Ahmadnia, et.al., 1999]

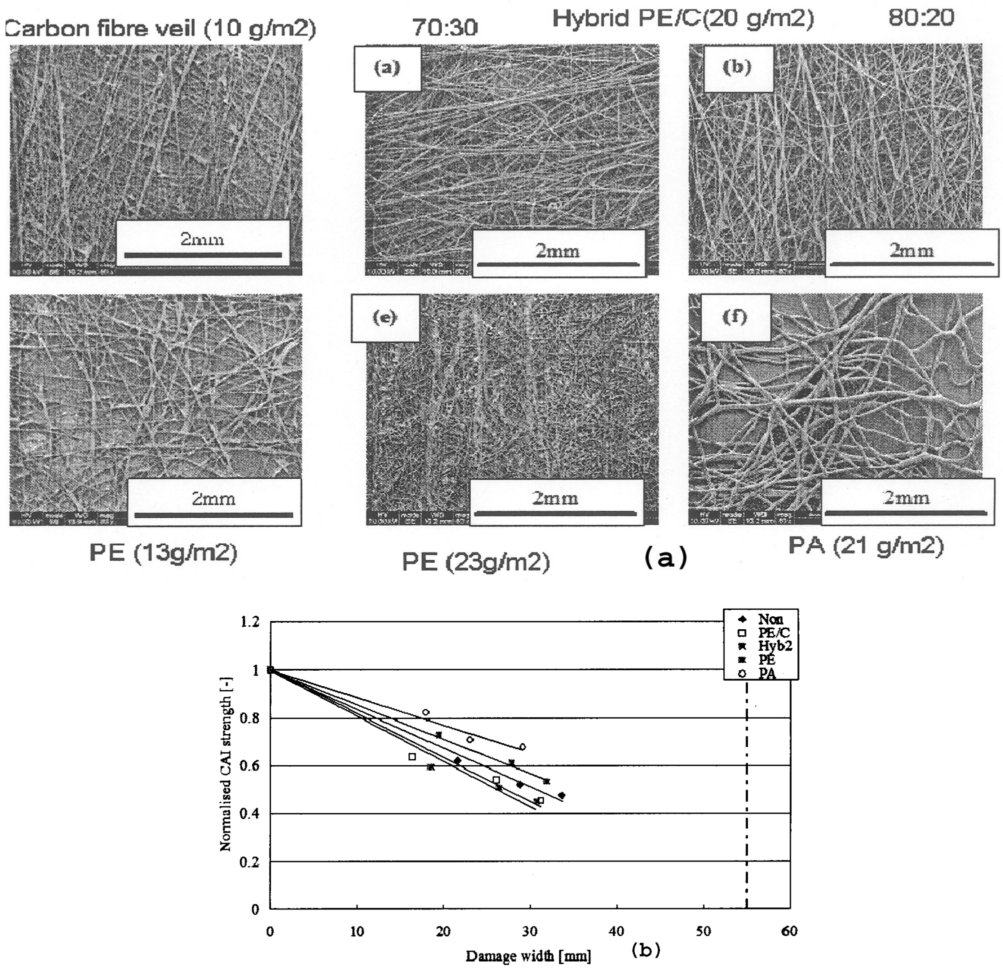
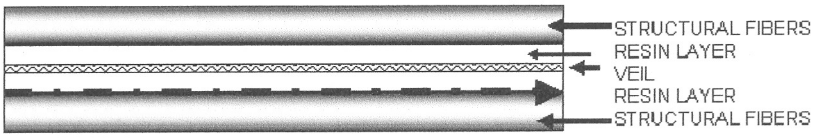
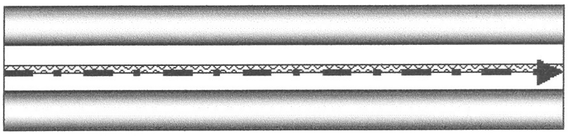


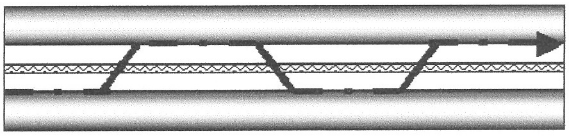
Fig.6 (a) Light weight veils based on thermoplastic fibers (polyester and polyamide) and carbon fibers, used to provide improved resistance to delamination growth in infused, woven carbon fiber epoxy laminates and (b) Normalised compression after impact data, plotted versus damage width [from Hogg, et.al., 2010]



a) Crack initiates and grows at interface between resin rich region bordering structural carbon fibre layers and the interleaf veil. Little improvement in toughness is likely over that exhibited by laminates without an interleaf veil. Favoured by stiff interleaf (.eg. Carbon)



b) Crack initiates and grows within the interleaf veil or at interface between veil and resin rich region. This is likely to encourage additional bridging and increase toughness. Favoured by softer interleaf veils (e.g. PE, PA)



c) Crack oscillates between interfaces, Favoured by hybrids and woven substrates. Intermediate case.

Fig.7 Possible Crack Paths in Laminates with Interlaminar Veils

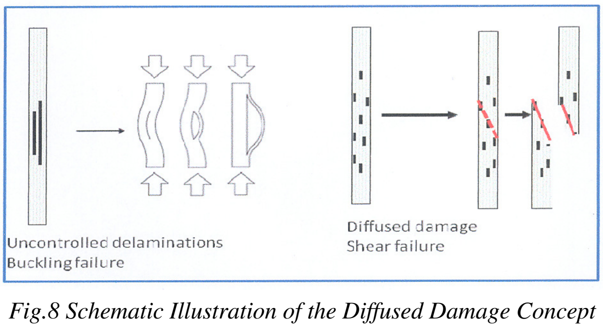


Fig.8 Schematic Illustration of the Diffused Damage Concept

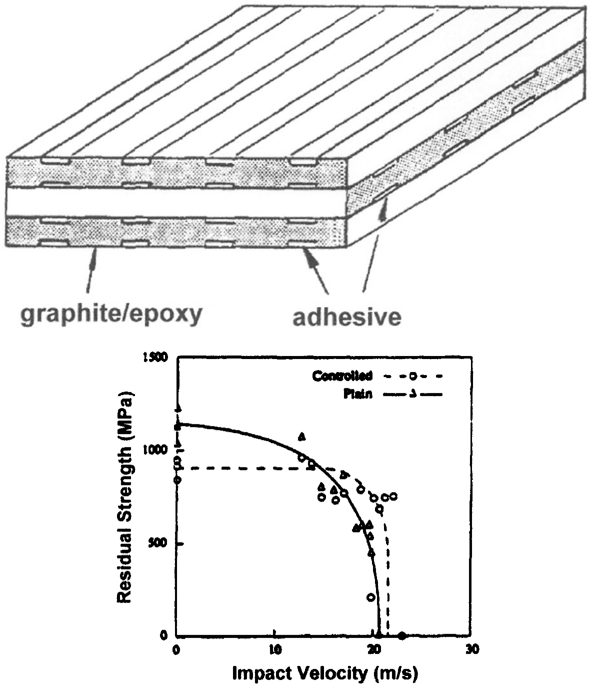


Fig.9 Controlled Damage Concept of Sun and Norman, 1990

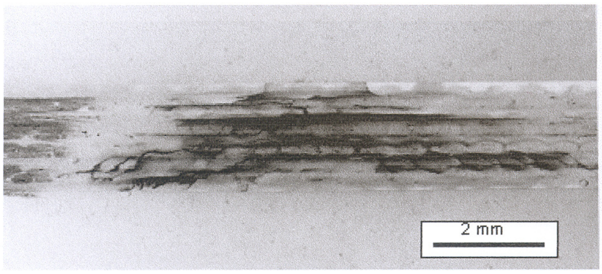


Fig.10 Cross section through an impacted glass fiber polyester resin composite (2mm thick) produced using an early non-crimp fabric. The apparent delaminations are comprised of assemblies of cracks that form around individual fiber bundles

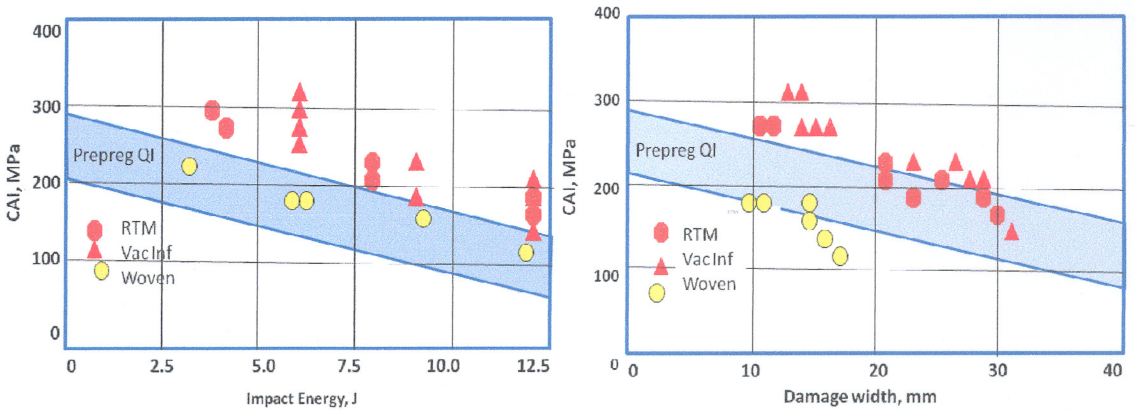


Fig.11 A comparison of CAI data for carbon fiber (quasi-isotropic) NCF laminates prepared via a prepregging route (scatter band only is shown), RTM and Vacuum infusion. Also shown for comparison is prepregged plain weave fabrics

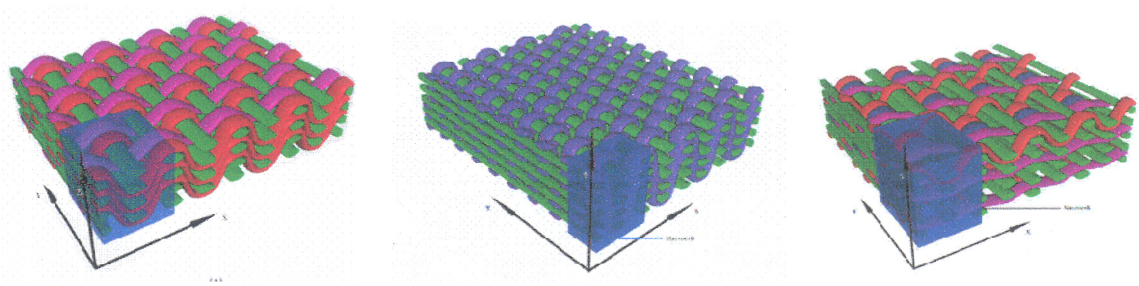
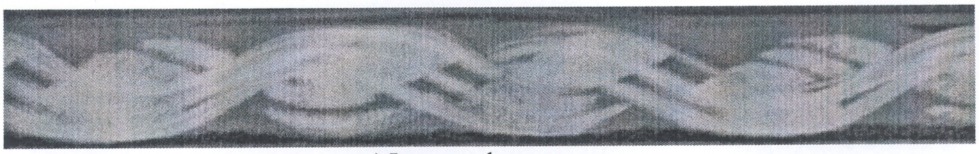
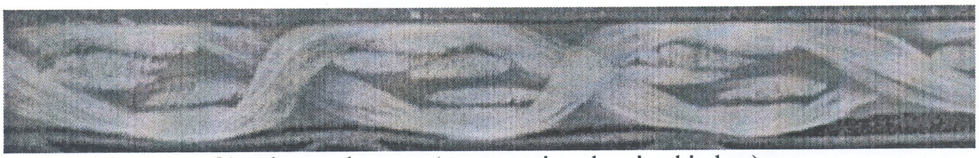


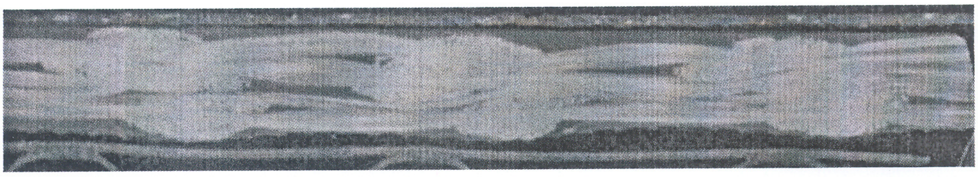
Fig.12 A Variety of Different 3D Woven Constructions with Differing Degrees of Interlayer Linkages



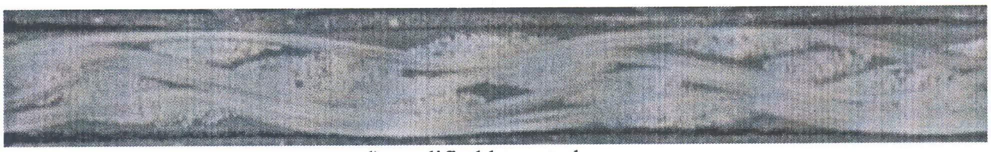
a) Layer-to-layer weave



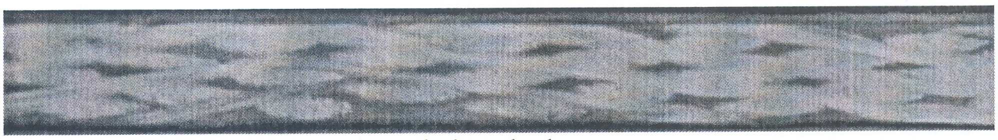
b) orthogonal weave (cross-section showing binders)



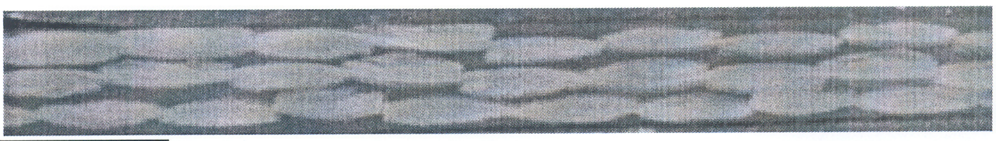
c) Orthogonal weave (cross-section showing stuffers)

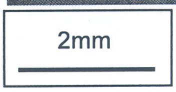


d) modified layer to layer



e) angle interlock weave





f) 3 layers of 2D plain weave

Fig.13 X-ray Tomograms of a Series of 2D and 3D Weaves

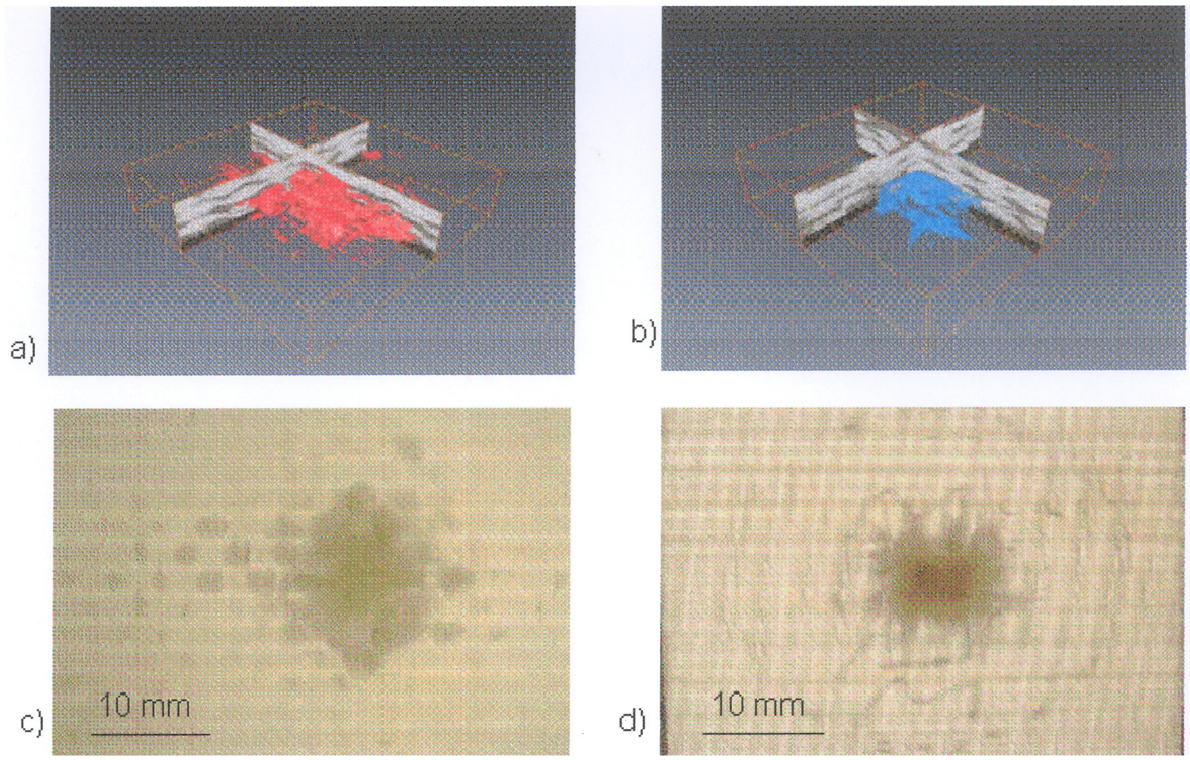


Fig.14 X-ray tomographic images showing cracking in (a) 2D and (b) a 3D woven S-2 glass fiber/epoxy composite subjected to equivalent impacts. The damage is more extensive and of a different format in the 2D laminate compared to the 3D sample. The optical transmission photographs, of the (c) 2D and (d) 3D damage areas are seen from above the impact site

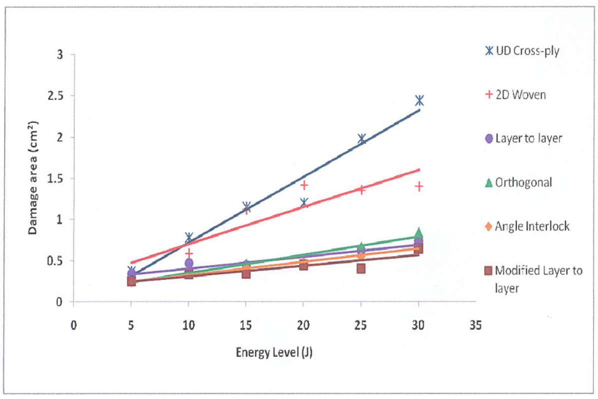


Fig.15 The Damage Area After Impact of the 2D and 3D Fabric Laminates Shown in Fig.13

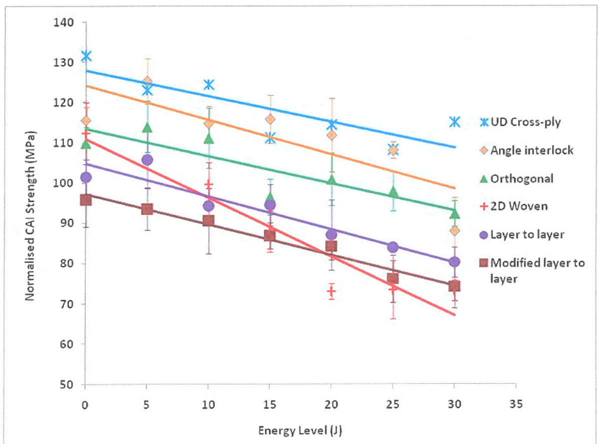


Fig.16 Residual Compression Strength After Impact of 3D Woven Laminates Compared to 2D Woven and Cross Ply Laminates (materials as per Fig.15)

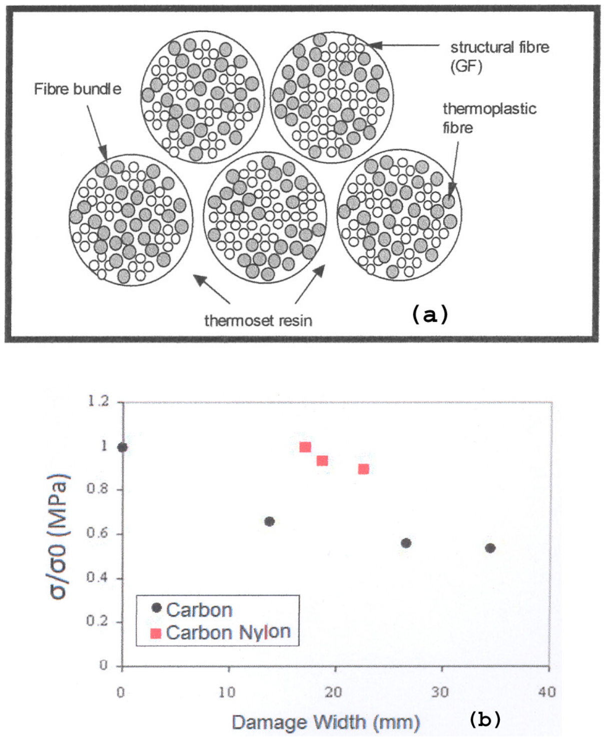


Fig 17 (a) A schematic showing the concept of a hybrid yarn toughened composite and (b) normalised compression after impact results for a hybrid yarn-epoxy laminate compared to a conventional carbon fiber laminate

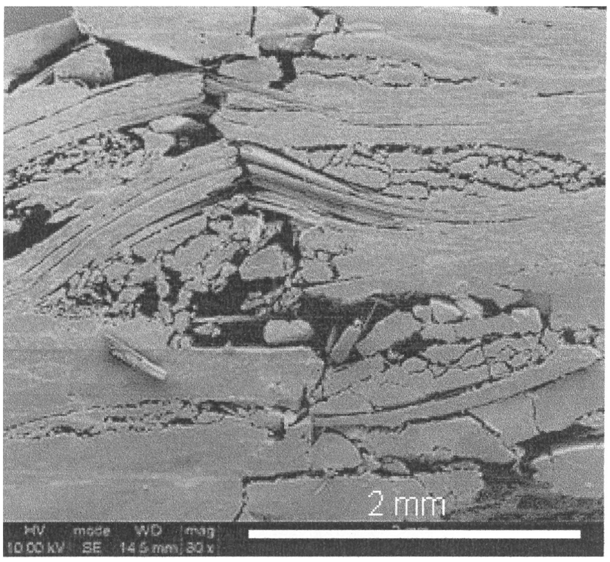


Fig.18 SEM image showing extensive cracking within a bundle of fibers in a woven fabric laminate, accompanied by cracking in the fibers in adjacent bundles

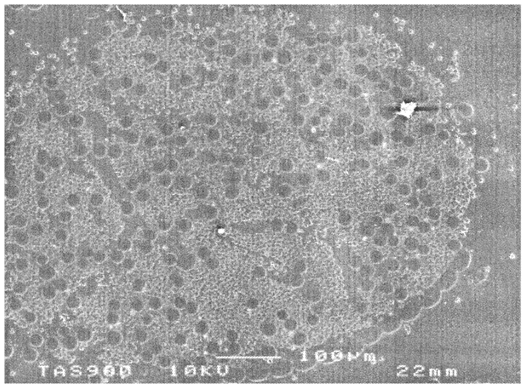


Fig.19 SEM image showing a single bundle in a hybrid carbon fiber/nylon fiber laminate where the bundle is surrounded by nylon fibers that act to retian bundle integrity, but could also act as a controlled weak interface

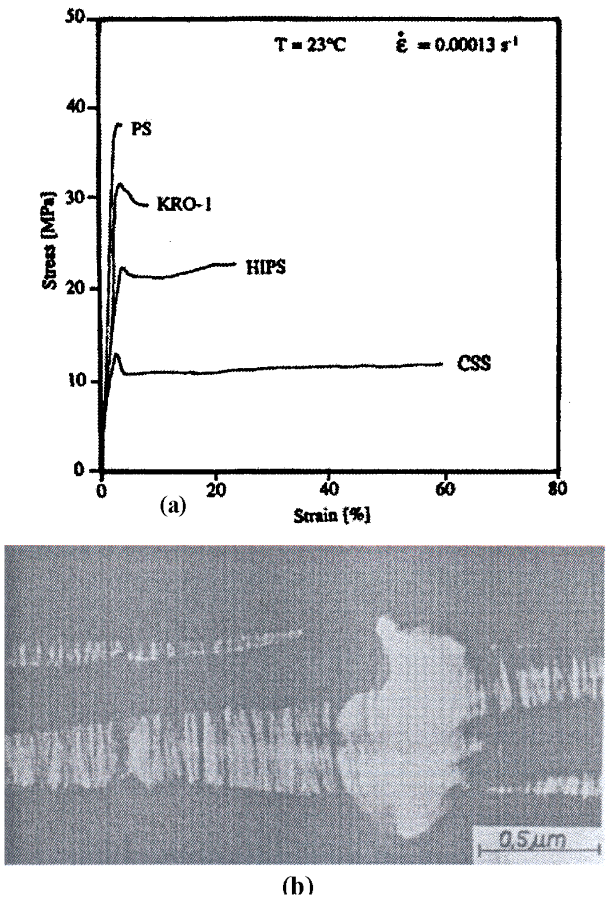


Fig.20 (a) Stress strain curves for polystyrene and derivatives toughened with rubber particles showing an increase in toughness [from Michler, 1986] and (b) an electron micrograph showing crazing induced by a rubber particle in ABS [from Argon and Cohen, 1990]