Failure in composite materials

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The economic and efficient exploitation of composite materials in critical load bearing applications relies on the ability to predict safe operational lives without excessive conservatism. Developing life prediction and monitoring techniques in these complex, inhomogeneous materials requires an under- standing of the various failure mechanisms which can take place. This article describes a range of damage mechanisms which are observed in polymer, metal and ceramic matrix composites.

The term composites now covers a wide range of existing and emerging engineering materials. Different types of composite exhibit a wide variety of failure mechanisms. However, a common feature of these diverse materials is their inhomogeneous and frequently markedly anisotropic nature, resulting in fracture behaviour unlike that of conventional metallic alloys. As a result, current fracture mechanics based analyses and test procedures are often found to be unsuitable for describing the behaviour of composites. Some of the important features of the fracture processes which occur in composites are described in this article. Understanding damage accumulation pro- cesses is an exciting challenge to materials scientists and engineers.

POLYMER MATRIX COMPOSITES

Polymer matrix composites, particularly in the form of laminates, have become established as engineering materials in many failure critical applications, for example in aerospace, in transport generally and in chemical plant. New areas of application, eg offshore[2] , are being actively sought, and recent material developments include high temperature and thermoplastic matrices[3]. Despite all these advances, the definition of failure in a fibre reinforced composite, and the monitoring and prediction of component life, remain major problems.

The difficulties arise because, in most situations, fibre reinforced materials do not fail by the initiation and propagation of a single dominant crack. During service, damage accumulates throughout the material until it reaches some 'critical' level, which might be determined by an unacceptable drop in modulus, or by complete separation in certain load controlled situations. Even when complete separation into two or more parts occurs, the failure process is complex.

The reasons for the non-localised accumulation of damage throughout the material are the statistical dependence of the strength of the brittle reinforcing fibres (eg glass and carbon) and the different properties of the matrix, reinforcement and interfacial regions, Tables 1 and 2.

Material	E (GNm^{-2})	Poisson's ratio ν
Carbon fibres		
Along fibres	200–800	
Perpendicular to firbres	10–20	
Glass	76	0.25
SiC	400–600	0.19
Al_2O_3	400	0.26
TiB_2	540	0.21
Al	70	0.35
Ti	120	0.36
Epoxy resins	3–6	0.38–0.4
Polyester resins	2–4.5	0.37–0.39
Carbon fibre/epoxy unidirectional laminae (V_f=0.60)		
Parallel to fibres	220	0.28
Perpendicular to fibres	7	0.016
Glass fibre/polyester unidirectional laminae (V_f=0.50)		
Parallel to fibres	38	0.26
Perpendicular to fibres	10	0.06
SiC Monofilament/6061 Al (V_f=0.50)		
Parallel to filaments	204	0.27
Perpendicular to filaments	118	0.12

For ν values, parallel and perpendicular are directions of applied tensile loading, and for laminae, the two strain values involved are those in the plane of the lamina

Table 1: Elastic modulus and Poisson's ratio for various reinforcing fibres, polymer matrix materials and polymer composites.

Material	Thermal expansion coefficient $\alpha \times 10^{-6}$ (°C^{-1})
Carbon fibres	
Along fibre axis	−0.1−−1.2
Radial	7−12
Kevlar fibres	
Along fibre axis	−2
Radial	59
Glass	4.9
SiC	4.5
Al_2O_3	7.6
TiB_2	5−9
Al	23.5
Ti	8.9
Epoxy resins	70
Polyester resins	100−200
Carbon fibre/epoxy unidirectional laminae (V_f=0.60)†	
Parallel to fibres	−0.2
Perpendicular to fibres	30
Glass fibre/polyester unidirectional laminae (V_f=0.50)†	
Parallel to fibres	11

†V_f=volume fraction of fibres

Table 2: Thermal expansion coefficients of various reinforcing fibres, polymer matrix materials and polymer composites (approximate figures only).

Unidirectional composites

In a unidirectionally reinforced material, eg a pultruded rod, the types of damage which occur involve fibre fracture, inter- facial debonding and matrix cracking. The stiff, brittle fibres are 'Griffith' materials. Fibre fracture initiates from surface defects, and the fibre strength distributions can be modelled using Weibull statistics[4]. The longer the fibre, the lower the strength because the more likely it is to contain a defect of a particular size. The distribution of defects means that within a group of fibres the weakest points in neighbouring fibres are unlikely to be adjacent to each other.

When fibre fracture occurs within a composite, the damage can spread in several ways. Where a very strong fibre/ matrix bond is combined with a brittle matrix, the effect of stress concentration at the crack tip can cause the crack to propagate across the whole section. This is an unusual situation in polymer composites, and one which must be avoided in ceramic matrix systems where the main function of the reinforcement is to provide toughening. Alternatively, the matrix around the crack can yield (thus blunting the crack and reducing the stress concentration) and/or shear failure can occur in the interfacial region, allowing the unloaded fibre to start to shrink back into the matrix, fig 1. Which of the three mechanisms illustrated in figure 1a, b or c occurs will depend on the relative values of the stresses σ_1, σ_2 and τ developed at the crack tip, fig 2, and on the fibre breaking strength, matrix shear strength and interfacial shear and tensile strengths. In polymer matrix systems interfacial shear failure usually occurs, to an extent determined by the interfacial strength and the energy released when the fibre fails.

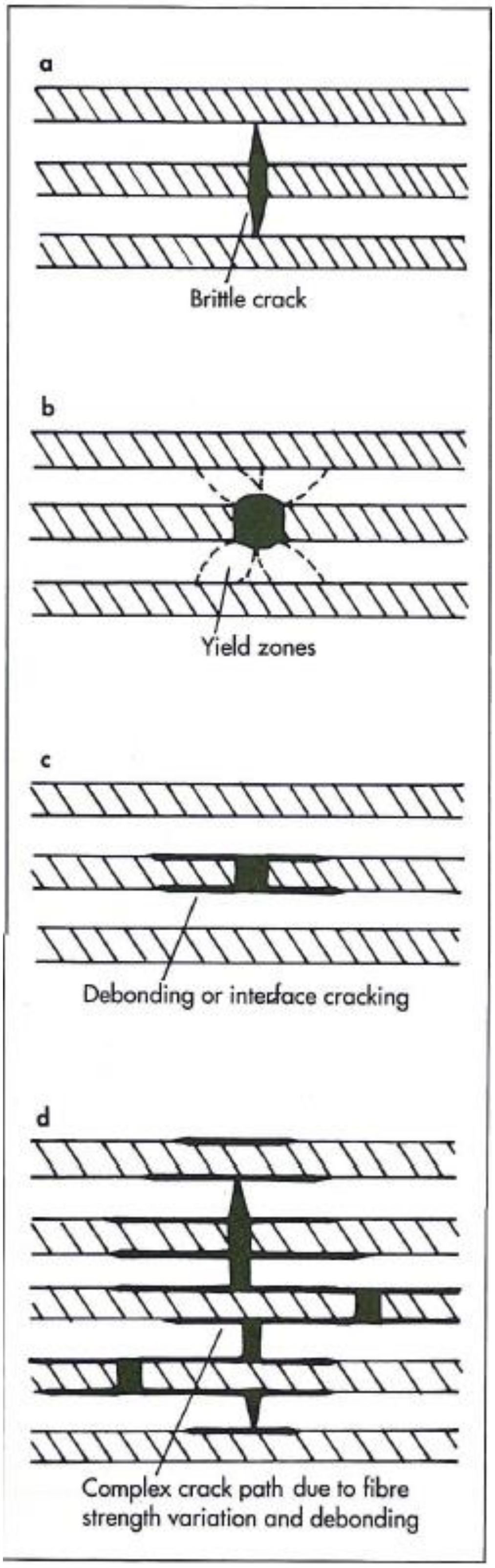

Fig.1 Fibre failure propagation[4,5]: a brittle cracking of matrix (high interfacial strength), b shear yielding of matrix, c interfacial failure, d propagation of fracture involving a and c.

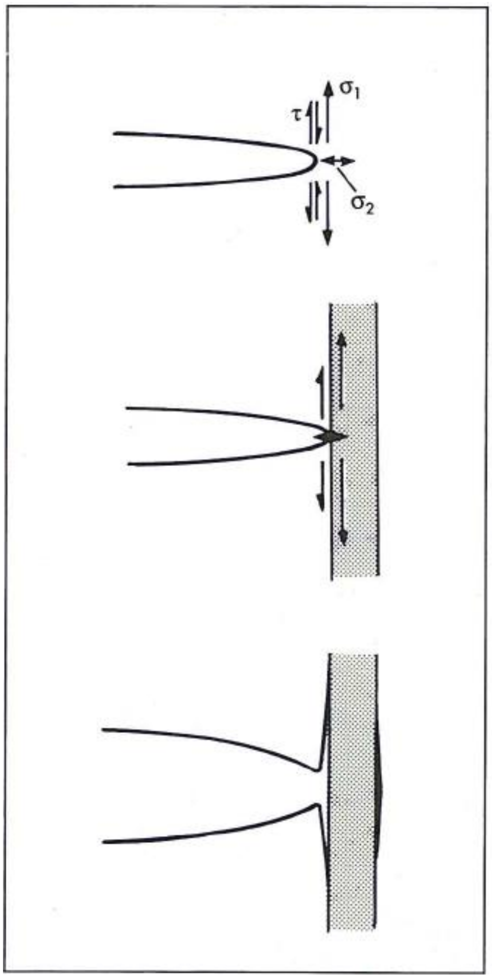

Fig 2 Stresses developed at the tip of a notch or crack in a composite

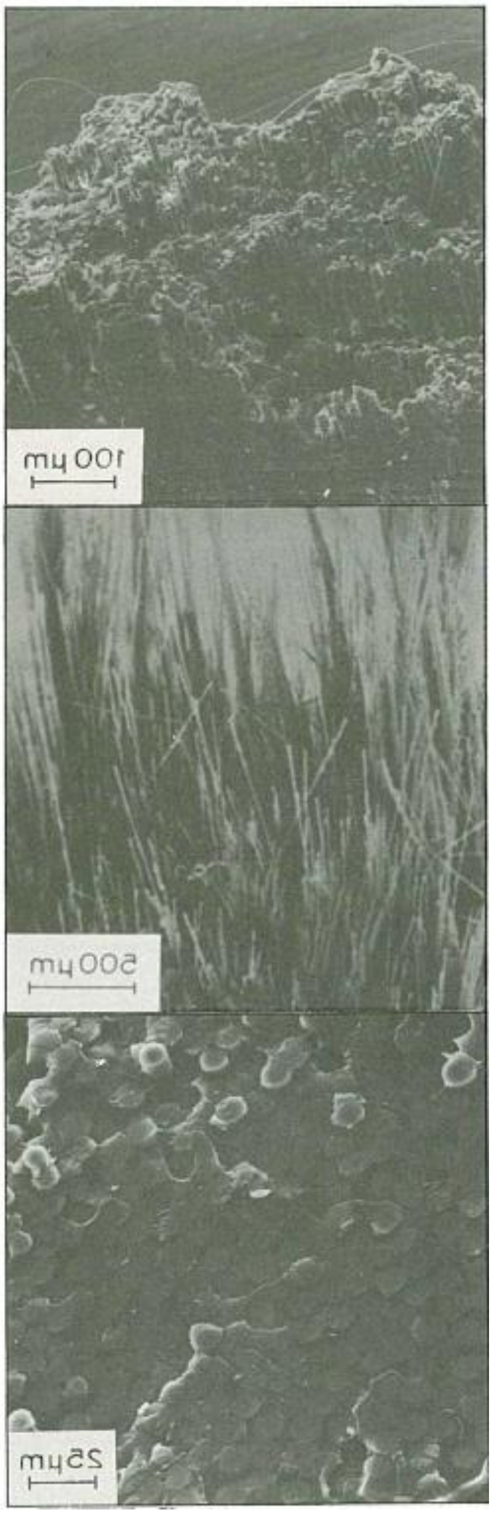

Fig 3 Tensile failure in unidirectional composites: a (top) strong interfacial bonding in glass (lbrel polyester (courtesy J N Price); b (centre) brush-like failure with extensive (lbre pull-out in glass (lbrel polyester (courtesy D Hull); c (bottom) flat stress corrosion fracture in glass fibre/polyester tested in dilute HCI (courtesy J N Price)

Fracture surfaces characteristic of strongly and weakly bonded interfaces are shown in figure 3. A strong interface gives a relatively smooth surface, fig 3a, whereas extensive debonding gives a brush-like appearance with extensive pull-out, fig 3b. This emphasises the importance of interfacial strength in controlling fracture behaviour and toughness. A considerable amount of energy is absorbed in pulling fibres out of the matrix.

Failure of the type shown in figure 1a is illustrated in figure 3c. A glass fibre rein- forced polyester rod has been loaded in an aqueous environment containing HCL The acid in the crack attacks the adjacent fibre surfaces, introducing numerous defects and thus drastically reducing the value of $\sigma 1$ at which the crack can propagate through the fibres. This produces a flat stress corrosion failure at low applied loads.

Cross-ply laminates

In cross-ply laminates containing plies with fibres parallel to the loading direction (0° plies) and perpendicular to the loading direction (90° plies), further mechanisms of damage accumulation arise for loading in tension. Delamination and matrix cracking in the form of longitudinal split- ting and transverse ply cracking occur, in addition to fibre fracture and interface de-bonding, fig 4.

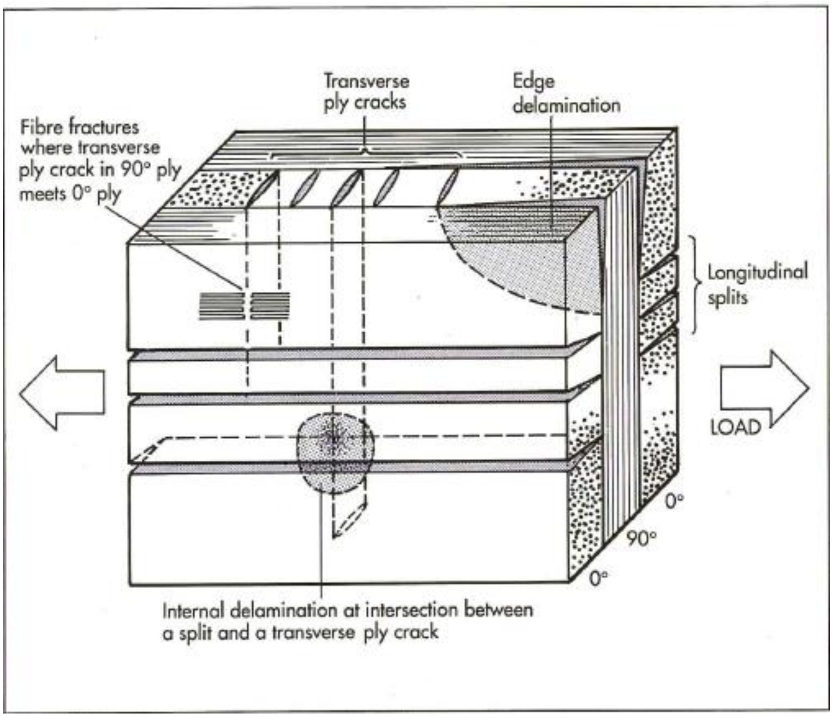

Fig 4 Delamination and matrix cracking are two of the damage mechanisms in a cross-ply laminate

Longitudinal splitting in the 0° plies develops because of the Poisson's ratio mismatch between the 0° and the 90° plies. As the 0° plies extend, they try to contract inwards but are restrained by the 90° plies which carry less load and have high stiff- ness along the fibre direction (Table 2). This results in tensile stresses in the 0° plies, perpendicular to the loading axis, causing longitudinal splits through fibre/ matrix debonding and matrix fracture.

Transverse ply cracking is another type of matrix cracking which occurs in the 90° plies. This usually starts at low stresses and again involves fibre/matrix debonding and matrix cracking, fig 5. As the load increases, the cracks propagate rapidly across the ply and increase in number until they reach a saturation state of regularly spaced cracks. The spacing at this point is characteristic of the particular laminate lay-up, and has been termed the 'characteristic damage state[8 9] . For both longitudinal splitting and transverse ply cracking, residual thermal stresses are important. On cooling after curing, there is greater contraction perpendicular to the fibre direction than parallel to the fibres because of the much lower thermal expansion coefficients of fibres than matrices (Table 1). This produces residual tensile stresses perpendicular to the fibres which assist both types of cracking. Indeed, in carbon fibre reinforced composites with brittle matrices and high cure temperatures, cracks can develop on cooling after production.

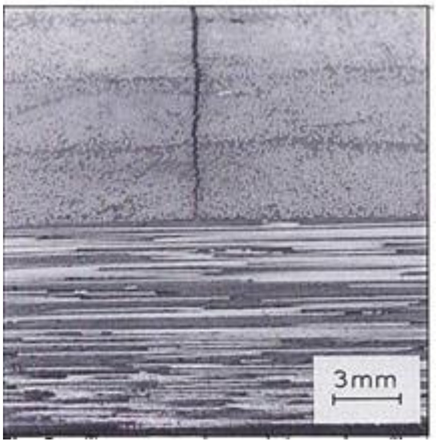

Fig 5 Transverse ply crack in carbon fibre/ bismaleimide resin (picture courtesy of T M Jennings and D Hull

Delaminations initiate at free surfaces; these can be test-piece edges, internal defects or holes, fig 6. They are again a result of the modulus and Poisson's ratio mismatch between adjacent plies. The transverse and shear stresses in each ply must drop to zero at the free surface, giving rise to tensile interlaminar stresses 10. Interior delaminations can propagate from unhanded areas between plies or at the intersections of other types of damage such as longitudinal splits and transverse ply cracks, fig 4.

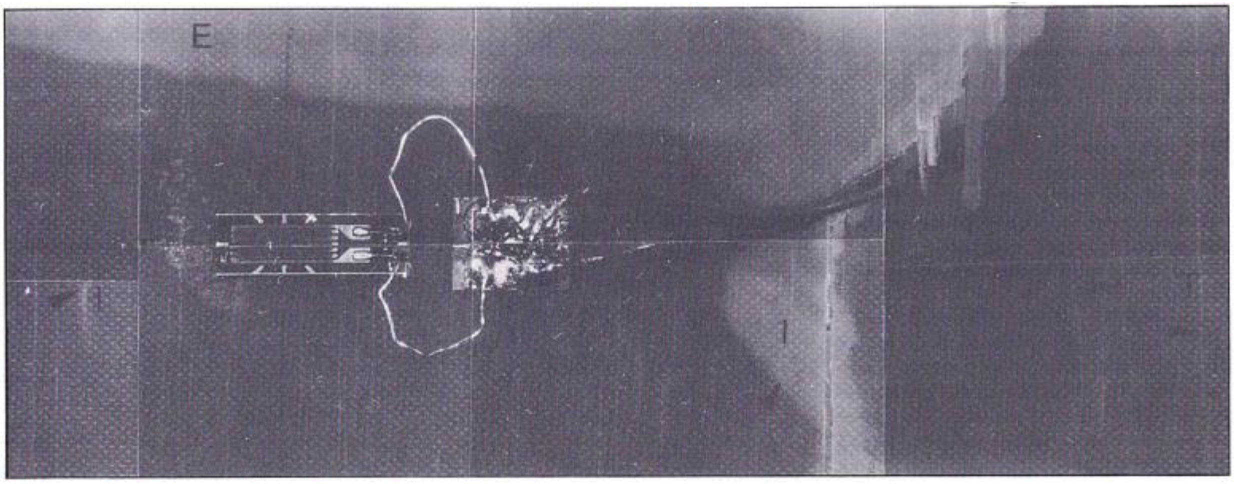

Fig 6 Edge, E, and internal, I, delamination in a glass fibre/epoxy laminate (courtesy K-H Leong): transverse ply cracks, T, in outer 90° plies

The sequence in which damage builds up depends on the types of fibre and matrix, the surface treatment and interfacial bonding, the particular lay-up and ply thicknesses.

Other failure modes

The elements of damage described above for tensile loading occur in other loading modes, although the detail of how they accumulate varies. In tensile fatigue loading, damage accumulation is similar to that for monotonic tension, although high frequency cycling can give rise to resin heating (particularly in glass and Kevlar reinforced materials with low thermal conductivity) and modification of matrix properties.

Impact loading can be particularly problematic. Very little evidence of the damage may be visible on the surface (BVID barely visible impact damage), but the delamination and cracking subsurface can cause significant strength reduction.

In compression, fibre buckling and the formation of kink bands has been observed, fig 7, with characteristics bending failures visible in the brittle fibres, showing regions of tensile and compressive fracture. In carbon fibre/epoxy systems, shear failure can also occur in compression, with the transition from buckling to shear depending on matrix shear modulus and fibre shear strength[11].

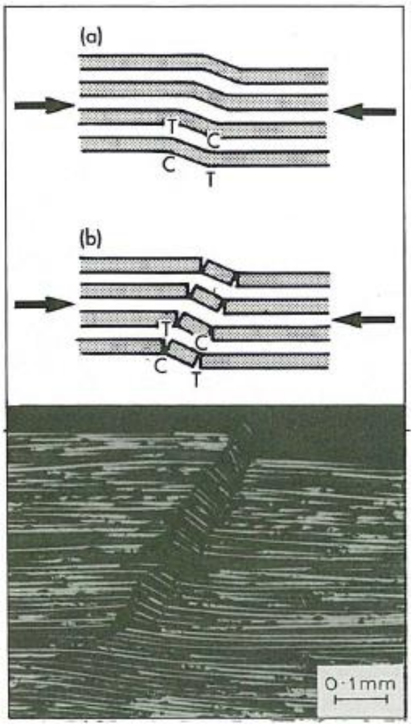

Fig 7 Fibre buckling and kink band formation 4: (top) buckling, (centre) fibre fracture, (bottom) kink band (courtesy of T V Parry and A S Wronski)

Damage accumulation in continuous fibre reinforced polymer composites is a complex process. The individual damage elements can be identified and, in some cases, their propagation has been success- fully modelled using fracture mechanics at the microscopic level, for example in the propagation of transverse ply cracks in 90° plies in fatigue [12]. Difficulties arise with the prediction of the growth of damage resulting from the interactions between damage in neighbouring plies. Two such situations are the initiation of internal delamination where transverse ply cracks cross longitudinal splits, and fibre failure in a 0° ply as a result of the stress concentration at the edge of a transverse ply crack in a neighbouring 90° ply, as shown in figure 4. A row of fractured fibres in a 0° ply, caused by transverse ply cracking in 90° ply and revealed by a de-ply technique, is shown in figure 8.

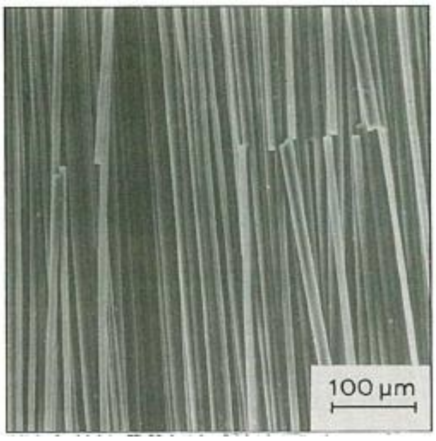

Fig 8 Fibre fractures in a 0° ply as a result of transverse ply cracking in the adjacent 90° ply. Carbon fibre/epoxy, revealed using a de-ply technique (courtesy of T M Jennings and D Hull)

Monitoring damage development by stiffness reduction[5], and the relation of stiffness reduction to residual strength, may provide a route to assessment of residual component life. However, it needs to be applied with caution because some damage mechanisms have more effect on stiffness than others. Further understanding of damage accumulation in polymer composites is still required, together with the development of · the new branch of mechanics known as 'damage mechanics' [13] to predict composite behaviour.

METAL MATRIX COMPOSITES

There are two main areas of metal matrix composite (MMC) development where applications are beginning to emerge: discontinuously reinforced aluminium alloys, and SiC monofilament reinforced aluminium and titanium matrices. Initial interest in whisker reinforcement has declined because of the health hazard posed by handling the whiskers.

The role of the reinforcement in metal matrix composites is similar to that in polymer systems, providing increases in stiffness and strength in low density matrices. In addition, the presence of stiff particles or fibres improves elevated temperature properties and thus extends the range of temperatures over which the light alloys can be used. However, one of the major problems with metal matrix composites is the reduction in ductility and toughness obtained in the composite compared with the matrix alloy, Table 3.

Property	Matrix alloy (8090)	MMC (8090+ 20 wt%SiC)
σ_y (MNm^{-2})	360–450	500–525
UTS (MNm^{-2})	460–520	550–570
Elongation (%)	~4	~3
K_{Ic} ($MNm^{-3/2}$)	20–40	10–25
E (GNm^{-2})	80	104

Table 3: Comparison of properties of aluminium-lithium (8090) based metal matrix composite with those of unreinforced matrix material

Discontinuous reinforcement

A limited number of applications have been reported for discontinuous systems. For example, Saffil (alumina) fibre reinforced aluminium castings are being used for piston crowns, and the first flight-qualified MMC component, electrical racking, is made from an extruded aluminium/SiC particulate system[14].

Discontinuously reinforced systems are generally isotropic and their fracture behaviour is therefore much closer to that of conventional alloys. However, the presence of the reinforcement has a number of important effects on matrix microstructure and properties, and on crack path. Reinforcements have lower thermal expansion coefficients than the metal matrices (Table 1). On cooling down from processing and heat treatment this induces compressive stresses within the particles and tension in the matrix. These residual stresses are partly relieved by the emission of dislocations, and high dislocation densities develop within the matrix. For spherical particles, the elastic residual stresses should be uniform and hydrostatic, but irregular shaped particles and short fibres lead to local stress concentrations and deviatoric tensile stress components. The presence of the reinforcement inhibits grain growth, and the combination of high dislocation density, fine grain size and the constraining effect of the rigid particles results in high initial work hardening rates [15]. The characteristic tendency of underaged aluminium alloys to develop intense planar slip bands is suppressed in the composite materials.

The dominant mechanism of tensile failure in aluminium/silicon carbide systems is by void formation. Void nucleation can occur

1. at cracked reinforcement particles,
2. adjacent to the reinforcement at stress concentrating features such as angular particles and fibre ends,
3. around the reinforcement particles themselves if the interfacial bonding is weak, or
4. around other particles within the matrix introduced by processing or heat-treatment.

Fracture of the reinforcement is encouraged by low strength reinforcement (eg coarse particle sizes), high strength matrices and high reinforcement aspect ratios, where load transfer to the reinforcement can occur by fibre loading (eg in Saffil).

In SiC reinforced aluminium, interfacial bonding is strong and case 3 is not common. The tensile fracture surface of a 15wt% 3 µm SiC reinforced 8090 alloy, fig 9, shows little evidence of voiding around SiC particles. Failure occurs by linking of voids formed around finer particles (egoxides) within the matrix, ie involving mechanisms 2 and 4 above. Once void nucleation occurs, void growth is rapid in the hydrostatic stress field within the matrix. In a recent study [17] heavily oxidised SiC particles were used to make a compo- site. The resulting material suffered void formation at SiC/matrix interfaces at very low strains.

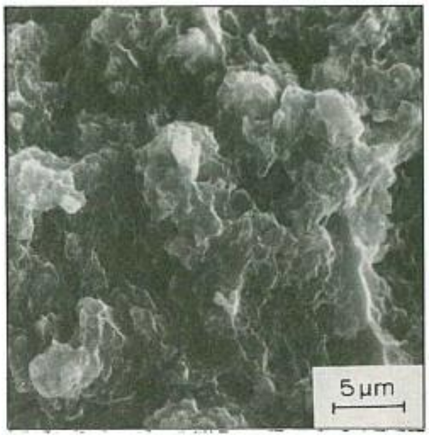

Fig 12 failure in Al/SiC particulate MMC made by a powder processing route. There is little evidence of voiding around the SiC particles (courtesy D M Knowles)

During fatigue crack growth in particulate composites, fig 10, the crack will avoid the stiff reinforcement unless particle cracking or interfacial failure attract the crack-tip. For example, the fracture of coarse SiC particles ahead of the crack-tip controls the crack path in the aluminium 6061/SiC composite, fig 11.

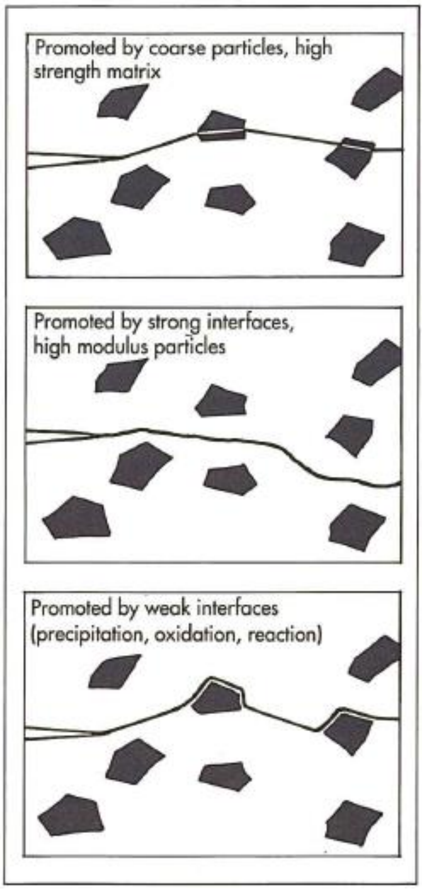

Fig 10 Crack-tip/reinforcing particle interactions during fatigue crack growth in particulate composites

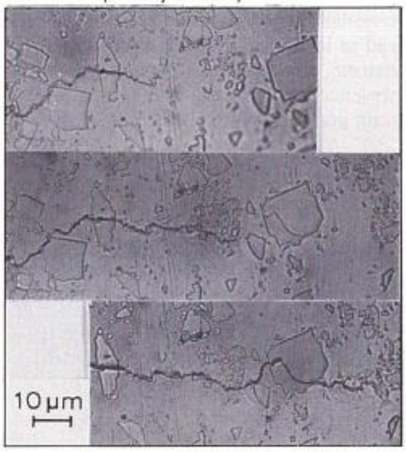

Fig 11 Particle fracture ahead of a fatigue crack in 6061 At/SiC (courtesy S Kumai

Unlike continuously reinforced materials, discontinuously reinforced compo- sites can be tested and analysed using many of the techniques developed for conventional alloys. However, the presence of the reinforcement does affect the mechanism of failure, even when that failure occurs predominantly in the matrix phase. This is demonstrated by a comparison of fatigue crack propagation at low growth rates in AI 8090/SiC and the matrix 8090 alloy, fig 12. In both materials, crack propagation occurs through the matrix, but the homogeneous slip character in the composite changes the mode of crack growth and gives growth rates which are almost a factor of two higher.

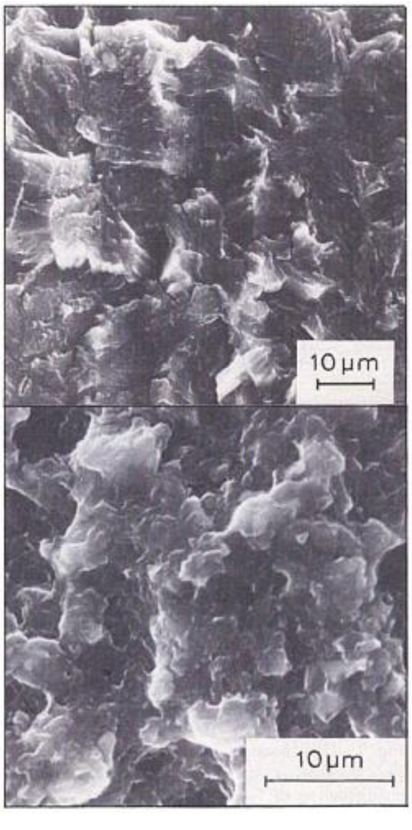

Fig 12 Fracture surfaces at low growth rates: (top) 8090 matrix alloy, showing effects of planar slip, and (bottom) 8090/SiC showing matrix failure via homogeneous deformation (courtesy D M Knowles)

It is becoming clear that particle size, shape and distribution, interfacial strength and matrix cleanliness are important in determining toughness and crack growth in MMCs; it is now necessary to learn how to control them.

Continuous fibre reinforcement

Recent developments in continuously reinforced systems have involved the use of monofilaments of SiC (~100 µm in diameter) in aluminium and titanium matrices. In terms of applications, the principal areas of interest are in aircraft, missiles and engines.
The strength and stiffness of unidirectionally reinforced material parallel to the fibres is very appealing: 6061 Al with 50% SiC monofilament gives a tensile strength of 1460 MNm^{-2} and a Young's modulus of 204 GNm^{-2} at a density of 2.9 Mgm^{-3} [17]. However, such material shows some similarities to the behaviour encountered in polymer composite laminates. Variations in moduli and Poisson's ratio between plies cause delamination and matrix cracking, and this is also promoted by the high thermal residual stresses resulting from the mismatch between the thermal expansion coefficients of the matrix and the reinforcement (Tables 1 and 2).

An additional problem arises in metal matrix systems due to reaction between the fibres and the matrix during processing and operation at elevated temperatures. Extensive reaction, particularly with SiC in titanium matrices, leads to the formation of brittle reaction products at the fibre/matrix interface. Cracking in the interfacial zone can either lead to interfacial failure and low transverse ply strength, or act as an initiation site for fibre fracture at low stresses, thus having a very deleterious effect on composite strength, fig 13. To overcome this problem, considerable re- search effort is now being concentrated on developing fibre coatings both to act as diffusion barriers and to provide interfacial regions with controlled properties [18].

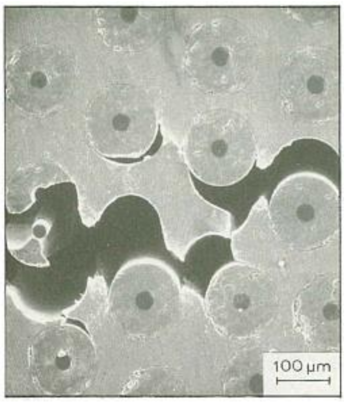

Fig 13 Transverse cracking in a brittle interfacial zone around SiC monofilaments in a titanium matrix (courtesy J G Robinson and British Petroleum)

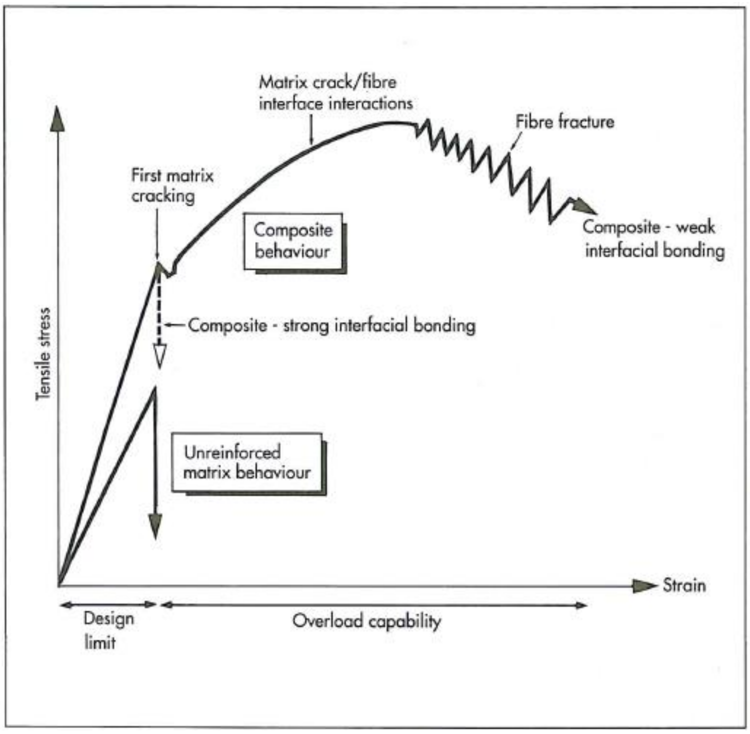

Fig 14 Continuous fibre reinforcement promotes 'graceful' or gradual Failure in ceramic matrix composites[19]. Three important elements of the failure process are labelled: matrix cracking, interfacial failure and fibre fracture

CERAMIC MATRIX COMPOSITES

Ceramic matrix composites, (CMCs) are the least developed (excluding the carbon-carbon materials), yet most eagerly anticipated composite systems. A host of applications awaits the arrival of high temperature materials which are environmentally stable, stiff, wear resistant and also tough.

In contrast to polymer and metal matrix composites, the primary role of the reinforcement in CMCs is to improve toughness. In continuous fibre reinforced systems this involves a change from a sudden, catatrophic failure mode to one of progressive damage accumulation, offering the possibility of an overload capability 18 or the detection of the early stages of damage and removal from service before final fracture. The problems arising here are similar to those in continuous fibre reinforced polymer composites, ie relating the nature and level of damage that has built up to the change in response of the material to the monitoring system being used.

Particulate reinforcement

In particulate reinforced systems, the role of the reinforcement is often to modify the crack path through crack deflection. Forcing the crack tip away from the plane of maximum tensile stress reduces the local crack-tip stress intensity and increases the length of the crack path.

In the commercially available titanium boride/silicon carbide system, the difference in thermal expansion coefficient results in tensile stresses in the TiB_2 particles and compression in the SiC matrix. This promotes toughness through crack clamping in the SiC, whilst interfacial failure and the difference in modulus between the two materials encourage crack deflection. Typical values of fracture toughness (K_{le}) in monolithic SiC are between 3 and 4 $MNm^{-3/2}$, whereas this can be increased to in excess of 8 $MN^{-3/2}$ in SiC/TiB_2 composites[18]. Where whisker reinforcement is used, crack bridging effects and whisker pull-out can also be significant aspects which promote toughness [19].

Continuous fibre reinforcement

The role of continuous fibre reinforcement in promoting gradual or 'graceful' failure is indicated in figure 14. Matrix cracking interfacial failure and fibre fracture are important elements of the failure process. Much of the toughening (and hence strength in brittle materials) is derived from fibre pull-out, and so control of interfacial strength is crucial in ceramic matrix materials. This is illustrated dramatically by the micrographs in figure 15, which should be compared with those for polymer composites in figure 3. The material is a SiC fibre (Nicalon) reinforced glass ceramic. In figure 15a and b, the surface treatment has led to too strong an interfacial bond and so a crack has been able to run straight through fibres and matrix, whereas in figure 15c and d, a weaker interface has promoted interfacial failure and fibre pull-out. The strength of the sample in figure 15c was three times that of the one illustrated in figure 15a.

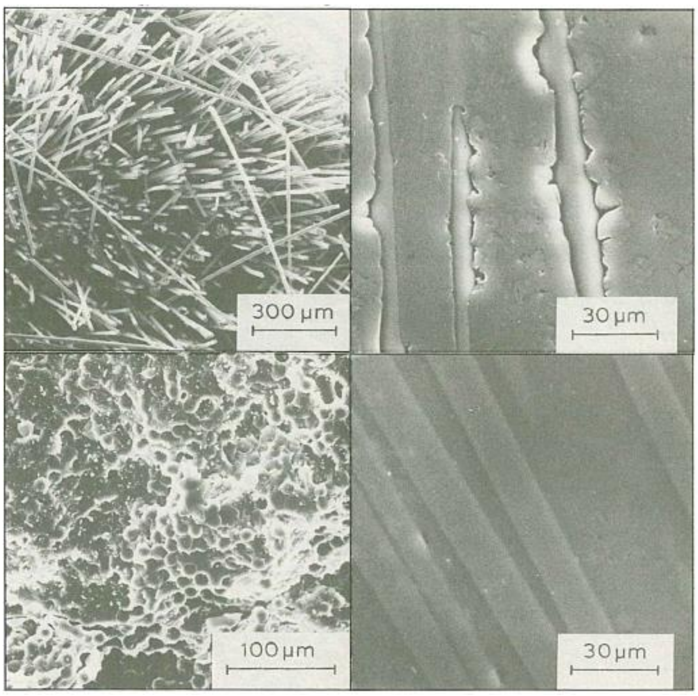

Fig 15 Effect of fibre surface treatment on fracture surface appearance in a Nicalon reinforced glass ceramic (top left and right) debonding and pull-out with a weak interface, (bottom left and right) strong bonding resulting in a flat, brittle failure (courtesy D W Shin and K M Knowles)

In common with MMCs, the high processing temperatures used for CMCs can lead to interfacial reaction and the formation of brittle interfacial layers, with deleterious effects on fibre strength. Again, fibre coatings are being examined as a means of controlling the interfacial failure process.

CONCLUSIONS

Continuous fibre reinforced materials show complex damage development and interaction processes due to their anisotropic nature. Further modelling of damage development is needed, together with an improved understanding of the factors controlling the different damage mechanisms, and the way in which these con- tribute to changes in properties such as reductions in stiffness or strength. Much of this work will be relevant to polymer, metal and ceramic matrix systems.

In discontinuously reinforced materials, much of the current fracture mechanics based modelling can be used. To improve defect tolerance, a better understanding is required of the way in which the reinforcements control fracture mechanisms, both directly, eg by crack path deflection, and indirectly, eg by modification of matrix deformation behaviour.

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