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FATIGUE FRACTURE BEHAVIOUR OF CARBON-FIBER-REINFORCED MODIFIED BISMALEIMIDE COMPOSITES

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Abstract

The fracture toughness and fatigue fracture behaviour of carbon-fiber-reinforced modified bismaleimide (BMI) composites have been studied. These composites were found to have higher fracture toughnes, better damage tolerance and longer fatigue life than carbon-fiber composites with epoxy matrices. Delamination is the major mode of failure in fatigue and it is controlled by the properties of the matrix and interface. The improved performance is due to the presence of thermoplastic particles in the modified BMI matrix which gives rise to enhanced fiber/matrix adhesion and more extensive plastic deformation. The fatigue behaviour also depends on the stacking sequence, with the multidirectional [45/90/−45/0] fiber-reinforced modified BMI composite having a lower crack propagation rate and longer fatigue life than the unidirectional laminate. This arises because of the constraint on the damage processes due to the different fiber orientation in the plies.

Keywords: fatigue, fracture toughness, modified bismaleimide, composites, microstructure

INTRODUCTION

The fatigue behaviour of composites is vastly different from that of homogeneous materials. In a homogeneous material under fatigue loading a crack, once initiated, propagates very quickly, thus leading to sudden failure of the material. On the other hand, the damage processes in composite materials are complex, consisting of matrix cracking, interface debonding, delamination and local fiber breakage. As a consequence of this long and complicated process, composite materials possess good damage tolerance and fatigue durability.

Since composite materials are frequently used in structures subjected to dynamic conditions it is important that their responses to cyclic loading be well understood. Until now, most of the studies have concentrated on axial tensile fatigue[1–3] and there has been very little work on bending fatigue behaviour. It is widely recognized that the fatigue properties of composite materials depend on the nature of the fiber, matrix, interfacial adhesion, stacking sequence, and loading mode, so it is necessary to combine bending fatigue experiments with analysis of the changes in the microstructure.

Epoxy resin is traditionally used as the matrix material in carbon-fiber-reinforced composites. It is brittle and has poor heat resistance, and thus can only be used up to 150°C. In recent years, bismaleimide (BMI) resins with excellent thermal properties have been developed[4] but, unfortunately, these materials are brittle and also difficult to process. More recently, a BMI resin modified by the incorporation of thermoplastic particles has been produced.[5,6] This material has excellent properties, including good toughness, high temperature resistance and ease of processing.

In the present work, the static and fatigue flexural fracture behaviour of carbon-fiber-reinforced composites with modified BMI matrices has been studied, and compared with the behaviour of composites with epoxy matrices. The effects of different stacking sequences have also been investigated. Crack propagation and microstructural failure were observed in a scanning electron microscope. From these results we attempt to provide an explanation for the toughening effects in composites with modified BMI matrices.

SPECIMEN PREPARATION AND EXPERIMENTAL METHODS

The matrix material used for preparing the unidirectional (U-B) and multidirectional (M-B) carbon-fibre-reinforced composites was a commercially available bismaleimide modified by compounding with PES. The modified bismaleimide (MBMI) was dissolved in a low-boiling-point solvent. Toray T300 carbon fiber unidirectional prepregs were made by the solvent method. The composites were cured at 177°C in an autoclave under a pressure of 5 bar for 2 h, then moved into an oven for post-curing at 120°C for 5 h. The elongation to break of MBMI (2·2%) matches that of T300 carbon fiber. The tensile strength and modulus of MBMI are 66 MPa and 3·5 GPa, respectively. All these values are higher than those of epoxy resin.[5,6] For comparison, unidirectional (U-E) and multidirectional (M-E) carbon-fiber-reinforced epoxy specimens were also studied. The nominal fiber volume fracton for all the specimens is 0·6 and the lay-up for the M-B and M-E specimens is $[45/90/-45/0]_{6s}$.

Figure 1 shows the specimen geometry and mode of loading. A notch of length $a = 4$ mm and width 0·12 mm was cut by the electric spark technique. The width w of the specimen is 20 mm, thus giving $a/w = 0{\cdot}2$. The bending fracture test was carried out according to ASTM E399.

The experiment was conducted on a computer-controlled MTS 818 universal testing machine. The frequency of the sinusoidal fatigue load was 20 Hz. The stress ratio R was set at 0·1 or 0·3. The initial damage and crack propagation during the fatigue process were observed by a travelling microscope. For a given increase in the number of fatigue cycles ΔN the increase in crack length Δl was measured, and this was repeated until the specimen failed. Fracture experiments were also carried out on small specimens in a Hitachi S570 scanning electron microscope to observe the development of the fracture process with time and the microstructural features after failure.

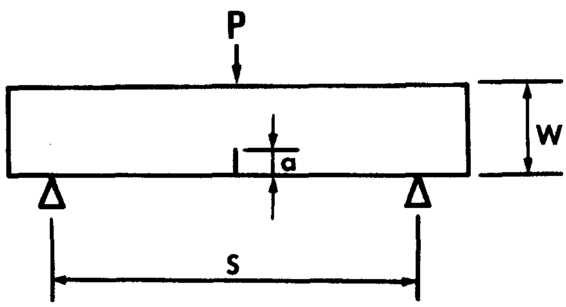


Fig. 1. Schematic diagram showing specimen geometry and loading mode.

RESULTS AND DISCUSSION

Static fracture toughness

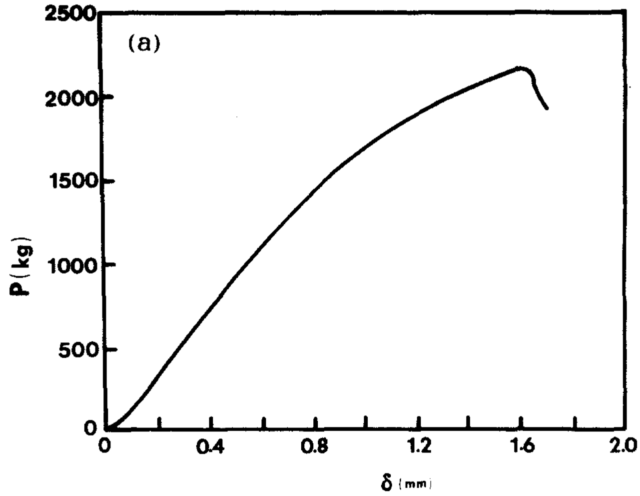
The load (P) versus displacement (δ) curves for the U-B and M-B specimens are shown in Fig. 2. From the observed fracture load P the fracture toughness K_c was calculated according to the following equation:[7]

$$K_c = (3PS/2tw^2)\sqrt{\pi a}\, Y(a/w) \qquad (1)$$

where t is the thickness and w the width of the specimen, S is the span, a is the length of the notch and $y(a/w)$ is given by:

$$Y(a/w) = 1{\cdot}09 - 1{\cdot}73(a/w) + 8{\cdot}2(a/w)^2 - 14{\cdot}2(a/w)^3 + 14{\cdot}6(a/w)^4 \qquad (2)$$

The average values for the fracture toughness K_c of the composites are given in Table 1. It is clear that U-B, the unidirectional composite with modified BMI matrix, has the highest K_c value. Its fracture toughness is about 30% higher than that of the unidirectional composite (U-E) with epoxy matrix. Similarly, for the



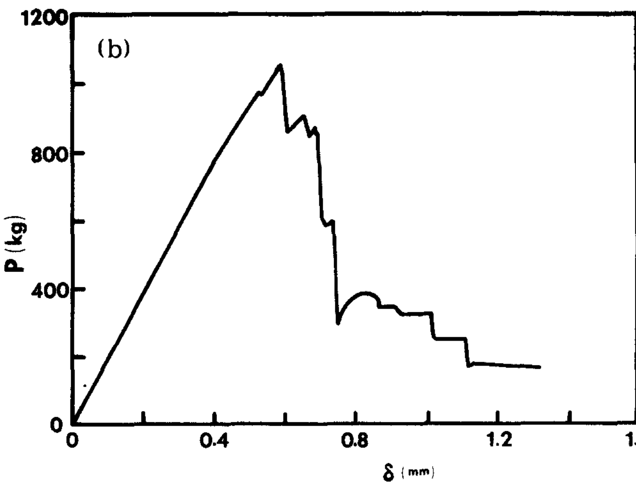


Fig. 2. Load, P, versus displacement, δ, curves of a T300/MBMI composite: (a) [0] specimen; (b) [45/90/−45/0] specimen.

Table 1. Fracture toughness of carbon-fiber-reinforced composites

Specimen	Material	Lay-up	a/w	K_c (MPa$\sqrt{m}$)	δ_{max} (mm)
U-B	T300/MBMI	[0]	0·2	46·3	1·64
U-E	T300/epoxy	[0]	0·2	32·0	1·71
M-B	T300/MBMI	[45/0/−45/90]	0·2	38·6	0·542
M-E	T300/epoxy	[45/0/−45/90]	0·2	25	0·610

multidirectional composites, the substitution of epoxy by MBMI also leads to a 35% increase in fracture toughness. These results clearly show that the use of a modified BMI matrix gives rise to a significant toughening effect.

Fatigue fracture performance
In order to evaluate the effect of different matrices on the fatigue fracture performance, fatigue measurements were performed at a maximum load equal to 60 or 65% of the monotonic fracture load, and with the stress ratio kept at 0·1 or 0·3. From the results shown in Table 2 it is seen that the fatigue fracture performance of the multidirectional composite with modified BMI matrix (M-B) is much better than the composite with epoxy matrix (M-E). After they have undergone the same number of cycles ($N = 1{\cdot}77 \times 10^5$) at 60% of the static fracture load level and a stress ratio of 0·1, the M-E specimen fails but M-B has a residual fracture load that is only 2% lower than its initial un-notched value. At 65% of the fracture load level and a stress ratio of 0·3, the M-E specimen ruptures after $1{\cdot}19 \times 10^4$ cycles while the M-B specimen ruptures after $2{\cdot}22 \times 10^5$ cycles, i.e. there is a 20-fold increase in the fatigue life. This reflects the effect of the toughened BMI matrix on the fatigue fracture behaviour.

Damage growth and failure modes
Generally, when the load reaches about 60% of the fracture load, microscopic damage occurs at the notch tip, then cracks appear at the interfaces and propagate with further increase of monotonic load or fatigue cycle. For unidirectional composites, the crack propagates along the direction of the 0° interface (Fig. 3) in a non-self-similar manner (also termed split growth Δl). The complex local stresses at the notch tip induce multiple damage modes including matrix cracking, interface debonding and fiber breakage due to bending (Fig. 4). Delamination plays a major role in fatigue damage and this leads to failure by splitting. The fatigue crack propagation rate is higher in unidirectional composites than multidirectional composites.

For the $[45/90/-45/0]_{6s}$ specimens the crack at the notch tip first propagates along 45°. In a scanning electron microscope under high magnification, 'wiggles' along 90° are observed in addition to the main crack along 45° (Figs 5 and 6). Under fatigue loading, damage accumulates gradually. Primary cracks alternate between +45° and −45°, and the length of the crack segment along each direction decreases as the crack propagates until finally the specimen fractures along the notch direction (Fig. 7). This is because, under bending fatigue fracture, there are more transverse cracks propagating through the ply thickness in the 90° layer than the 0° or 45° layer. Transverse ply cracking is one of the main factors in fatigue fracture. At the same time there is more fiber breakage in the 45° layer than the 0° layer. After the appearance of cracks in the matrix, debonding occurs along the fiber direction as a result of the normal

Table 2. Fatigue fracture performance of carbon-fiber-reinforced composites

Specimen	Material	Lay-up	Load level (%)	Stress ratio, R	Fatigue cycles (N)	Residual fracture load/un-natched fracture load (%)
U-B	T300/MBMI	[0]	60	0·1	$6{\cdot}73 \times 10^4$	Failure
U-B	T300/MBMI	[0]	65	0·1	$5{\cdot}66 \times 10^4$	Failure
U-B	T300/MBMI	[0]	65	0·3	$1{\cdot}08 \times 10^5$	Failure
M-B	T300/MBMI	[45/90/−45/0]	60	0·1	$1{\cdot}77 \times 10^5$	98
M-B	T300/MBMI	[45/90/−45/0]	65	0·1	$1{\cdot}24 \times 10^5$	Failure
M-B	T300/MBMI	[45/90/−45/0]	65	0·3	$2{\cdot}22 \times 10^5$	Failure
M-E	T300/epoxy	[45/90/−45/0]	60	0·1	$1{\cdot}77 \times 10^5$	Failure
M-E	T300/epoxy	[45/90/−45/0]	65	0·3	$1{\cdot}19 \times 10^4$	Failure

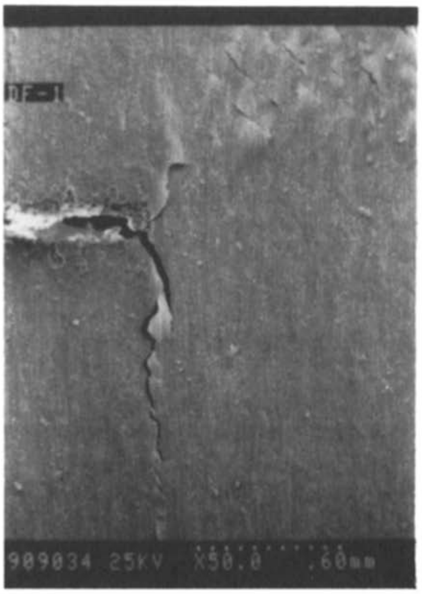


Fig. 3. Crack propagation along the 0° direction in a 300/MBMI [0] specimen.

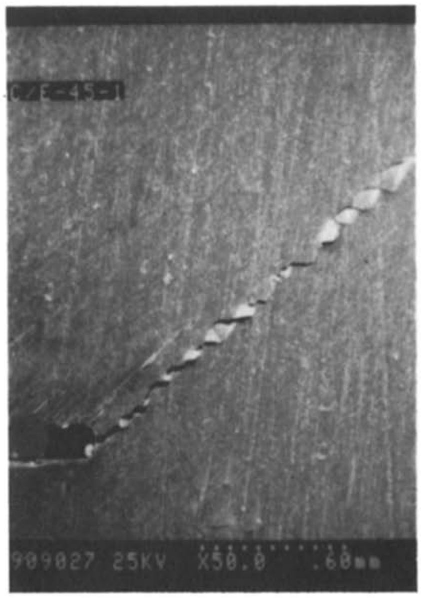


Fig. 5. Crack propagation in a T300/epoxy [45/90/−45/0] specimen.

Fig. 4. complex stress state near the notch tip of a T300/MBMI specimen.

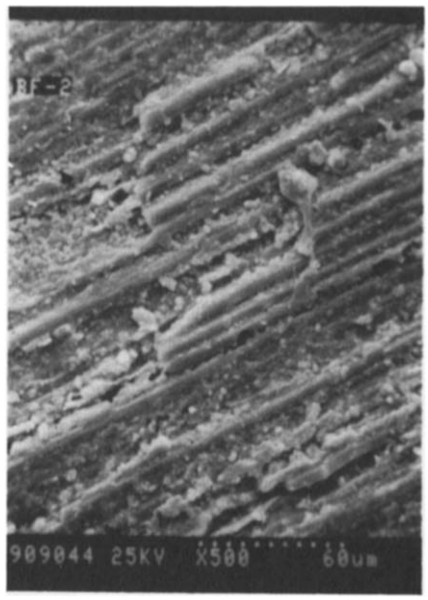


Fig. 6. Crack propagation in a T300/MBMI [45/90/−45/0] specimen.

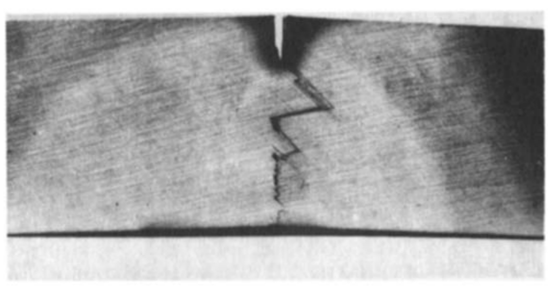


Fig. 7. Fracture failure mode of a T300/MBMI [45/90/−45/0] specimen under fatigue loading.

stress and interlaminar shear stress. Delamination is more extensive in the off-axis plies under fatigue loading than during static loading. Figure 8 shows that the fracture surface of the M-B specimen. The right-hand side is the prenotched surface. From the fatigue fracture surface adjacent to the notch tip shown on the left-hand side, it is clearly seen that there is fiber breakage in the 0° plies, fiber pull-out due to shearing in the ±45° plies and transverse separation surface in 90° plies.

During the stable growth stage of the fatigue tests (at 60% of the fracture level and a stress ratio of 0·1), the U-B specimens have an average crack propagation rate dl/dN of 5 mm/10^4 N, while the M-B and M-E

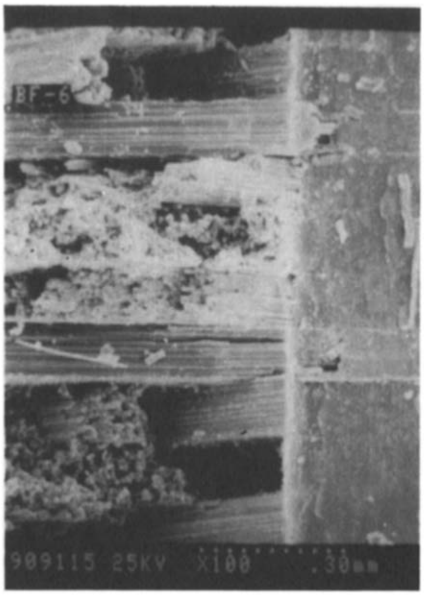


Fig. 8. Fatigue fracture surface of a T300/MBMI [45/90/−45/0] specimen.

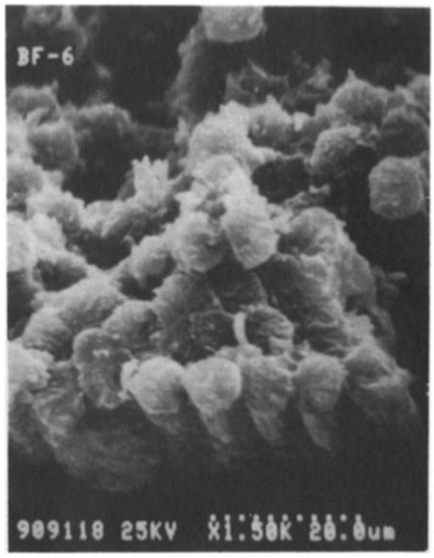


Fig. 9. Transverse fracture surface of a T300/MBMI specimen showing good interface bonding.

Fig. 10. Transverse fracture surface of a T300/epoxy specimen showing considerable fiber pull-out.

specimens have dl/dN values of 0·082 mm/104^{4}N and 0·11 mm/10^4 N, respectively. The dl/dN value is lower for the multidirectional composites because there are constraints on the damage processes when the fiber orientation varies from ply to ply. For the same reason the fatigue life is higher for the multidirectional composites than the unidirectional composites. The crack propagation rate dl/dN depends not only on the stacking sequence but also on interfacial adhesion, loading level and stress ratio.

Toughening mechanisms in fiber-reinforced composites with modified BMI matrix

From the experimental results it is clear that the fracture toughness and fatigue fracture performance of carbon-fiber-reinforced composites with modified BMI matrix are better than those with epoxy matrix.

In the following we attempt to understand the toughening effects of the modified BMI matrix by considering the microstructural features:

(1) After modification, the MBMI resin adheres well to the carbon fibers, thereby protecting the fiber surface. It is seen from the fracture surfaces shown in Figs 9 and 10 that there is less fiber pull-out in the specimen with modified BMI matrix. The thermoplastic particles in the modified BMI matrix have good elastic properties. Furthermore, the occurrence of microcracks in the matrix near the particle surface gives rise to significant energy absorption, thus preventing brittle fracture (Fig. 11). This leads to a substantial increase in the fracture toughness.

(2) Under bending fatigue loading, shear bands appear in the modified BMI matrix, thereby retarding matrix cracking and interface debonding. Figure 12 shows the shear bands and plastic deformation in the M-B specimen while Fig. 13 shows brittle cracking and interface debonding in the M-E composite.

CONCLUSION

The use of modified BMI matrix in fiber-reinforced composites improves the toughness and fibre/matrix adhesion, thereby enhancing the static and fatigue fracture performance. Carbon-fiber/modified BMI composites have higher fatigue life and damage tolerance than carbon-fiber/epoxy composites and can thus be used in a wider range of applications.

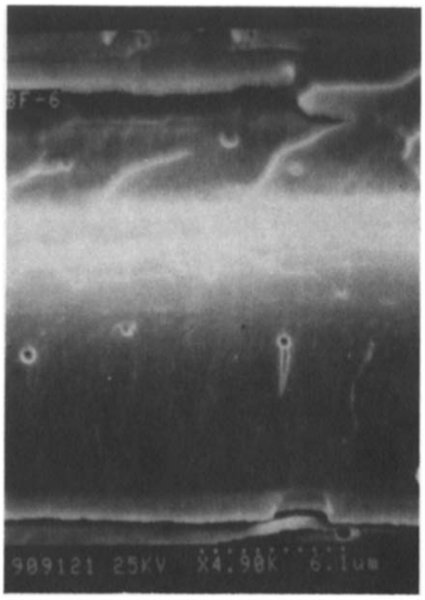


Fig. 11. Fracture surface of a T300/MBMI specimen showing good fiber/matrix adhesion and microcracks at the thermoplastic particles.

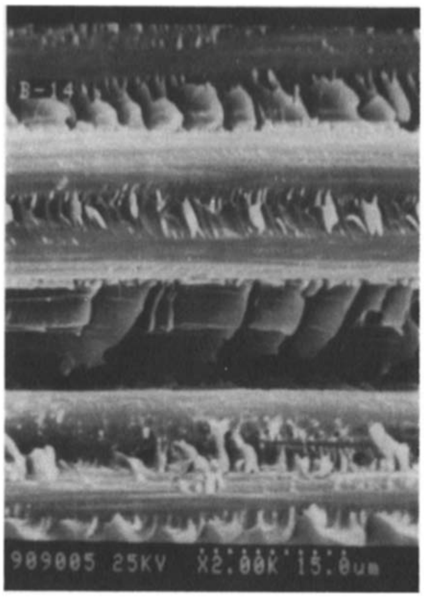


Fig. 12. Longitudinal fracture surface of a T300/MBMI specimen showing shear bands and plastic deformation in the matrix.

Delamination is the major mode of failure in fatigue and it is controlled by the properties of the matrix. The good adhesive property and plastic deformation of the modified BMI matrix give rise to a reduction in interface debonding, surface damage of the fibers, and matrix cracking, thereby delaying the occurrence of delamination and increasing the fatigue life.

Even after cracks have appeared near the notch tip, carbon-fiber/modified-BMI composite can still sustain a large number of fatigue cycles. The cracks propagate in a non-self-similar manner and the propagation rate

Fig. 13. Longitudinal fracture surface of a T300/epoxy specimen showing brittle cracking and interface debonding.

varies with the interfacial adhesive strength, lay-up sequence, fatigue load level and stress ratio. Low interfacial strength, unidirectional fiber arrangement, high load level and stress ratio give rise to high fatigue crack propagation rate.

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