THE MECHANISMS OF FATIGUE CRACK AND DELAMINATION ZONE GROWTH IN FIBRE REINFORCED ALUMINIUM-LITHIUM LAMINATES

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ABSTRACT

Research has been carried out investigating the fatigue response of fibre reinforced aluminium laminates which employ the aluminium-lithium alloy 8090. Microstructural analysis of fatigued laminates was performed, in order to develop an understanding of the interactions between the fatigue crack in the alloy layers and the observed delamination zone in the composite ply. These have been described in terms of the progress of the crack through the constituent materials and the influence of their mechanical properties.

INTRODUCTION

Efforts to improve the fatigue characteristics of ARALL-type laminates have led to the employment of the low density aluminium-lithium alloy 8090 instead of the more conventional 2xxx & 7xxx aluminium alloys used in ARALL laminates. The response of the new laminate to various load conditions requires characterisation, but a precondition is that the mechanisms of fatigue crack growth and damage accumulation in the composite layers is understood. The load response can then be interpreted within these terms of reference. The purpose of this paper is to present a qualitative model of the processes occurring in the laminate under fatigue conditions, based on the analysis of various fatigue tests.

EXPERIMENTAL WORK

The laminate panels were constructed in a 2-1 configuration as shown in Figure 1. The panels were laid up by hand at DRA Farnborough and cured in an autoclave at 150°C for 2 hours. The final test panels were 152mm by 350mm, with a saw-cut in the centre either 10 or 20mm long lying in the transverse direction.

The model presented here is based upon the analysis of fatigue tests carried out on laminates under a variety of different conditions, analysing the effect of load ratio ($R = \sigma_{max}/\sigma_{min}$), initial ΔK (stress intensity range based on the length of the saw-cut in the metal) and ΔP levels, frequency and reversed loading. The baseline test was $R = 0.1$, $f = 20Hz$, and $\Delta K_i \approx 9MPa\sqrt{m}$.

Fatigue tests were terminated when the crack length was sufficient to allow for the analysis of the condition of the fibres in the crack wake. The specimens were analyzed with a variety of techniques:

i) C-scanning in an immersion scanner,

ii) cross-sectioning, vacuum impregnation and polishing for examination by optical microscopy using the Nomarski differential interference contrast technique,

iii) removal of the alloy layers by immersion in concentrated Sodium Hydroxide; the exposed delamination zone on the surface of the composite layer was then examined in the Scanning Electron Microscope.

RESULTS AND DISCUSSION

The mechanics of fatigue crack growth within each metal layer is, metallurgically, the same as that for a monolithic sheet of the same configuration. The reduction in crack growth rate results from the physical processes that comprise the interactions between the metal and composite layers. These processes are strongly interrelated, but in order to simplify the discussion that follows they have been broken down into several stages.

Fatigue cracks grow in the metal plies; the cracks grow through the thickness of the plies and reach their internal surfaces adjoining the composite layer. Wherever the cracks reach this interface they grow through the resin-rich sub-layer and then branch along the interface of the resin-rich and fibre-rich sub-layers - Figure 2. There was some uncertainty about whether the failure of the fibre-resin interface was caused by crack branching or by Poisson contraction of the material at the crack tip inducing the failure of the interface. Examination of cross-sections close to the tip of a fatigue crack in the metal (Figures 2 & 3) showed a crack in the resin that deviated completely in one direction at the fibre-resin interface; if the failure was initiated by Poisson contraction then some evidence of crack deviation in both directions would be expected since the stresses in advance of the crack in the polymer will be symmetrical. This indicates that crack branching on the interface is the cause of the initial failure of the resin-fibre bond.

The nature of the branching depends upon the thickness of the resin-rich layer. In regions of low resin thickness the crack branches on the interface, Figure 4a, in areas of high resin thickness the crack may split before it reaches the interface, the two branches diverging slightly before they meet and follow the interface - Figure 4b. It should be remembered that these are 3-dimensional processes rather than 2-dimensional. Thus the term 'crack branches' refers to the movement of two crack fronts because of adhesive failure under shear as described by Marissen [2]. The progress of the crack fronts leave, in their wake, a region where the fibre-resin interface has debonded, the delamination zone. If the surface of the delamination zone is examined under SEM the 'trace' of the crack path can be seen, which highlights these two mechanisms. In Figure 5a the smooth area running across the middle of the figure is caused by the sudden failure of the interface and corresponds to Figure 4a; in

Figure 5b resin remains attached to the surface, corresponding to Figure 4b.

The occurrence of the interfacial failure prevents crack growth in the fibre-rich sub-layer and allows the fibres to retain their load bearing capabilities. The stresses in the fibres cause the crack branches to grow incrementally parallel to the load direction, as the fatigue crack in the metal plies grows perpendicular to the load direction, [1], Figure 6. The shape of the delamination zone seems to be dependent on the stresses induced by the crack face opening displacement of the fatigue cracks in the metal plies; the 'elliptical' shape of the delamination zones corresponding with the elliptical crack face opening distribution along the crack length. This is reinforced by work on a different system [3], where fibres directly reinforce a cracking matrix, resulted in a debonded zone of the same elliptical shape as found in this type of material.

Tests at different R-ratios, but the same initial ΔK have resulted in delamination zones of the same size indicating that delamination growth is a fatigue process. This is further reinforced by examination of the surface of the delamination zone - Figure (5a). A change in failure 'mode' is apparent with a smooth region representing the initial instantaneous failure of the interface changing to a rougher surface characteristic of the incremental growth of the delamination zone.

Table 1 shows the values of Young's moduli for the constituent materials and their differences. The value of ΔE for the carbon reinforced panel is larger than that for the glass reinforced case, this correspond with the fact that the delamination zones for carbon reinforced laminates are consistently larger than those with glass reinforcement.

Thus it seems that although the shape of the delamination zone is a function of the crack face opening distribution the actual size of the delamination zone is controlled by the stiffness mismatch between the constituent materials and the applied initial ΔK.

In earlier work on ARALL researchers [4 & 5] found that the fibres failed in the wake of the crack, at approximately 5mm behind the crack tip, although recent work has stated that fibre failure occurs only under low or negative R-ratios [6]. As yet no such failure mechanism has been found for the carbon reinforced laminates. Evidence of gross fibre failure was observed in the glass fibre reinforced panels, but this appeared at much larger distances behind the crack tip ($\approx$ 48mm). As yet no evidence of failure has been found in the carbon fibre composite layer. This is attributable in part to the increased stiffness of the 8090 alloy reducing the stresses in the composite ply.

It is thought that the critical factor controlling fibre failure on these materials would be the values of strain, especially in the low strain-to-failure carbon fibres. The high stiffness of the carbon fibres will result in lower stresses within the metal plies, a smaller crack face opening displacement, and consequently lower strains experienced by the fibres before the action of the delamination zone. Furthermore the larger delamination zones in the carbon reinforced laminates give a greater reduction in the strain experienced by the fibres in the crack wake than in the glass case. Approximate calculations have shown that the average strain in the carbon bridging fibres is half of that for the glass fibres. These two factors combine to prevent failure of the carbon fibres.

The presence of the intact fibres reduces the crack growth rate in the metal layers by stress redistribution away from the crack tip and by restraint of the crack face opening displacement [1 & 7]. It is interesting to see how the various components of the failure process can be related to the fatigue performance (Figure 7). The glass fibres have a lower stiffness than 8090; as a consequence the growth curve for the laminate initially follows the shape of the 8090 curve, but the intact fibres begin to bridge the crack effectively and the resulting change in the gradient of the curve can be observed at point A. The failure of the glass fibres in the crack wake would lead to a change in the magnitude of the bridging as the crack continues to grow, producing the second change in the gradient at point B. The carbon fibres, having greater stiffness than 8090, contribute immediately to fibre bridging and consistently lower growth rates are observed. However a transition can be seen, from a decreasing fatigue crack growth rate to one that is increasing, at point C. This is caused by a change in the balance between the two 'forces' that control the crack growth rate. One is the reduction in ΔK experienced by the crack tip owing to the presence of the bridging fibres, the other is the increase in ΔK as crack length increases. After point C the increase in ΔK with crack length predominates.

In the present work analysis of the crack growth data is restricted by the limitation of using a nominal ΔK, calculated as if for an unreinforced metal. This cannot account for fibre bridging and any second order contribution from delamination growth, both of which in turn are dependent on crack length and load conditions. Attempts to define a more appropriate presentation of fatigue data in laminates are described elsewhere [8].

CONCLUSIONS

The analysis undertaken so far indicates that replacing the conventional alloys with 8090 does not significantly alter the crack growth / delamination growth processes within the laminate. The fatigue crack grows from the metal layers into the resin-rich layer of the composite ply; at the fibre-resin interface the crack branches, preventing the fatigue crack growing into the fibre-rich layer. As the fatigue crack grows through the metal in the transverse direction, the crack branches form two crack fronts which grow along the interface in the longitudinal direction; this debonded area is the delamination zone. The presence of delamination zone increases the 'free length' of the wake fibres thus preventing failure of these fibres by reducing the induced strain. The intact fibres bridge the crack and reduce the fatigue crack growth rate in the metal layers. The change in alloy material alters only the magnitude of the stresses in each layer rather than the mehcanisms of failure.

ACKNOWLEDGEMENTS

The support of AD Science (Air), Ministry of Defense, together with the technical guidance of DRA Farnborough is gratefully acknowledged.

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Table 1

	8090	Carbon/epoxy	Glass/epoxy
E (GPa)	77	135	43
ΔE (GPa)	....	68	(-)34

ΔE is the difference between the stiffness of the alloy and the reinforcing fibres.

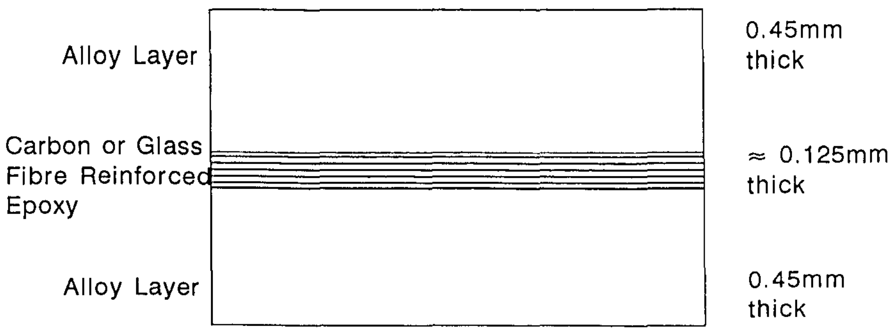

Figure 1: The 2-1 configuration of the laminates

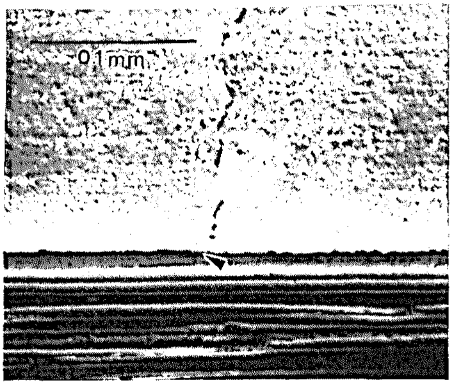

Figure 2

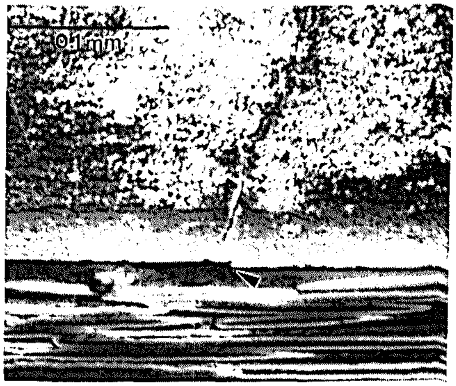

Figure3

Figures 2 & 3 both show irregular deviation of the crack at the fibre-resin interface.

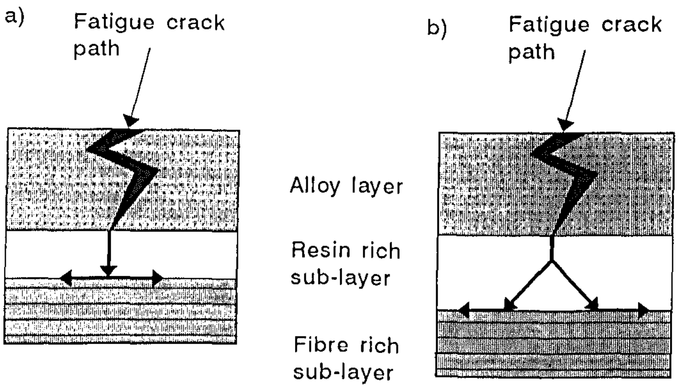

Figure 4: Schematic of crack path through a) a thin resin-rich layer and b) a thick resin-rich layer.

a) b)

Figure 5: Traces of fatigue crack path on the surface of the delamination zone a) for a thin resin-rich layer and b) for a thick resin-rich layer with resin remaining attached to the surface (X).

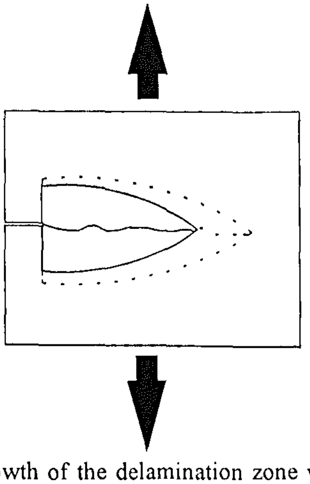

Figure (6): Growth of the delamination zone with crack length.

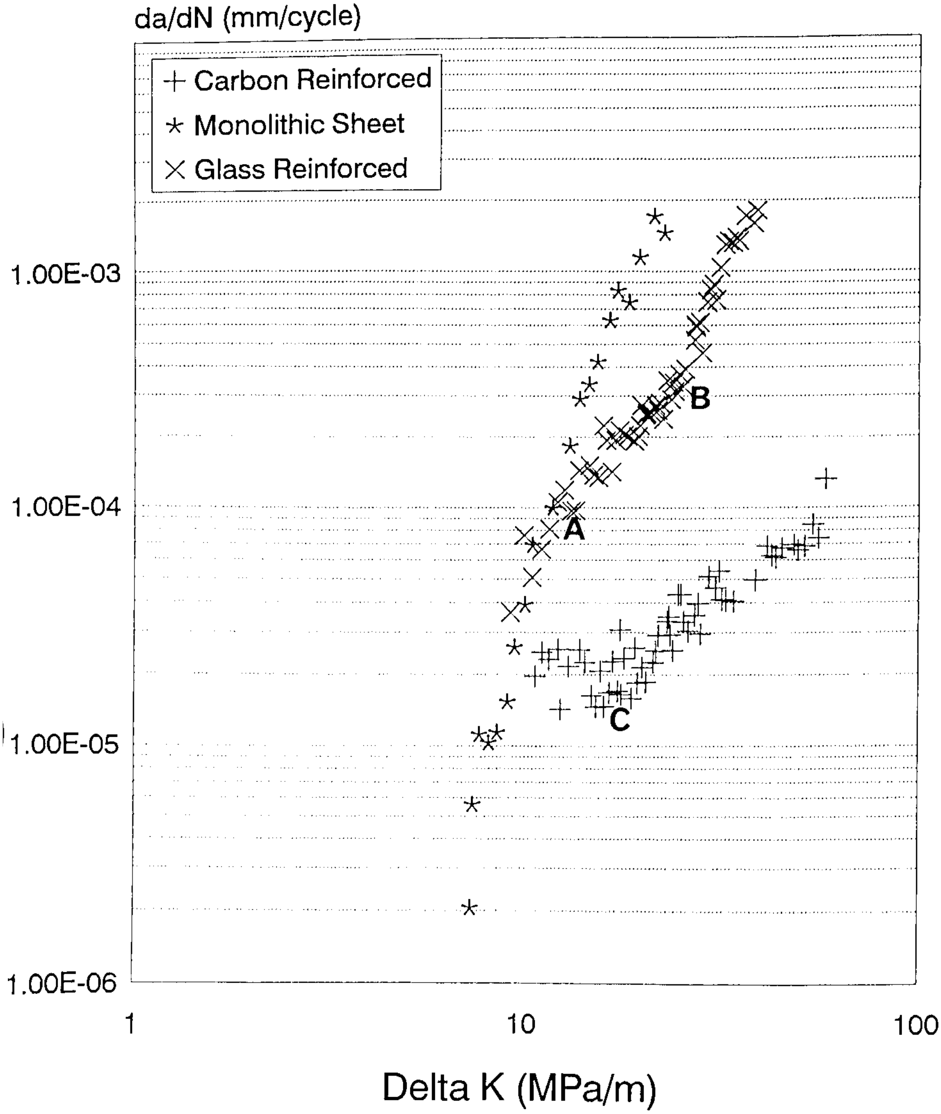

Figure 7: Fatigue crack growth rates for carbon and glass reinforced laminates and monolithic sheet tested at R=0.1, initial $\Delta K \approx 9$ MPa$\sqrt{m}$, f = 20Hz.

SEDIMENTATION AND DIRECTIONAL SOLIDIFICATION OF A PARTICULATE Al-Al_2O_3 METAL MATRIX COMPOSITE

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Abstract

An understanding of the mechanisms controlling the distribution of particles in particulate reinforced metal matrix composites (MMCs) is crucial for the optimisation of their mechanical properties. Sedimentation and directional solidification experiments have been conducted in order to investigate microscopic effects on the distribution of particles in an MMC consisting of a hypoeutectic Al-Si alloy reinforced with Al_2O_3-based ceramic microspheres. It has been found that the sedimentation of reinforcement microspheres proceeds at a much higher rate than that predicted by Stokes' theory for the settling of single spheres. Three different modes of settling were observed, with the mode dependent upon the volume fraction of reinforcement in the sample. It was found that the presence of ceramic spheres had a profound effect on the microstructure of the directionally solidified matrix alloy, probably due to the low thermal conductivity of the microspheres.

Introduction

The material used for this work was Comral 90F™ composite, an MMC developed by Comalco Research Centre [1,2]. This composite consists of an Al-Si alloy, AA603 (Al-7wt%Si-0.5wt%Mg), as the matrix reinforced with 19 vol% MICRAL™-20 microspheres. These microspheres are alumina-based spherical particles of mean diameter 20 μm. A significant density difference exists between the matrix and reinforcement: the density of the alloy is approximately 2.7 gcm^{-3} while the alumina-based microspheres have a density of 3.4 gcm^{-3}. The effects of this difference become an important consideration when employing a melt processing route for the production of MMC components.

The distribution of particles in MMC castings is frequently inhomogeneous [3]. It is considered that three factors influence the formation of the final particle distribution. The velocity of the advancing solid/liquid interface is of importance in the determination of the type of interaction between the interface and the particles, which may be engulfed or pushed by the solidification front, thereby rearranging the particle distribution [4,5]. Fluid flow during mould filling can cause the migration of particles either towards the mould