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EXTRINSIC FRACTURE MECHANISMS IN TWO LAMINATED METAL COMPOSITES

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Abstract

The crack growth behavior and fracture toughness of two laminated metal composites (6090/SiC/25p laminated with 5182 and ultrahigh-carbon steel laminated with brass) have been studied in both "crack arrester" and "crack divider" orientations. The mechanisms of crack growth were analyzed and extrinsic toughening mechanisms were found to contribute significantly to the toughness. The influence of laminate architecture (layer thickness and component volume fraction), component material properties and residual stress on these mechanisms and the resulting crack growth resistance are discussed.

Introduction

It is generally recognized that fracture-related properties in composites (such as fracture toughness, fatigue crack growth behavior and impact behavior) can be improved by both intrinsic and extrinsic mechanisms (1). As previously reviewed (2) intrinsic mechanisms act to improve these properties by increasing the inherent microstructural resistance to crack growth through control of microstructural parameters such as grain size, particle size, particle spacing etc. Extrinsic mechanisms, on the other hand, improve these properties by reducing the driving force for crack growth through various processes that shield the crack tip from the applied stress intensity. These processes include crack deflection, crack bridging and crack trapping. Laminated metal composites (LMCs), which consist of alternating metal or metal matrix composite (MMC) layers, are especially effective at improving fracture-related properties in materials by various extrinsic mechanisms (3,4,5).

In this paper we report on studies that have been done of extrinsic toughening in LMCs containing MMCs. Specifically we have studied LMCs consisting of alternating layers of a MMC, 6090/SiC/25p, and a monolithic aluminum alloy, 5182. We have also studied LMCs consisting of a spheroidized ultrahigh-carbon steel (UHCS) and brass. The UHCS in this system can be considered an in-situ processed MMC since the material consists of a ductile matrix (ferrite) that is reinforced with a discontinuous, hard, brittle phase (spheroidized carbide particles). The details of crack growth have been analyzed and mechanisms of extrinsic toughening in these systems are discussed. These mechanisms are related to the influence of macrostructure (such as layer thickness, component volume fracture and interface strength), properties of the component layers and residual stress.

Materials, Processing and Testing

Materials and Processing

All laminates in this study were made by hot pressing alternating layers of the component materials. For the UHCS/Brass system the press bonding temperature was 750°C. The UHCS used in this
investigation had a nominal composition of 1.8% C-1.6% Al-1.5% Cr-0.5% Mn and was processed to a 3 mm thickness by hot and warm working (6) to achieve a fine ferrite grain size of 0.5 μm containing spheroidized carbide particles. The brass of a nominal composition of 70 wt.% Cu and 30 wt.% Zn and with 3.2 mm thickness was obtained in the fully annealed condition. All UHCS/Brass laminates in this study contained an equal volume percentage of the two component materials and laminates of two different layer thicknesses (200 μm and 750 μm) were produced.

The Al/Al-SiC laminates were press bonded at 450°C. Prior to press bonding the materials were chemically cleaned to remove surface oxides. For these laminates the influence of volume fraction of the component materials was studied. The variations in volume fraction were obtained by press bonding material with different starting thicknesses. The volume fraction of the 6090/SiC/25p component in the laminate was varied from 50% to 97% which produced a variation in the global concentration of silicon carbide in the laminate from 12.5% to 24.3%. The 6090/SiC/25p MMC was obtained from a commercial source in the form of 2.6, 6, and 10 mm thick warm or hot rolled sheets. The 5182 was obtained from a different commercial source in the form of warm or hot rolled sheets of 0.25, 2.1, and 2.6 mm thicknesses. After press bonding, the Al/Al-SiC laminates were heat treated by soaking the laminate at 530°C for 75 minutes and then aging at 160°C for 16 hours. This procedure provided a T6 heat treatment to the 6090/SiC/25p layers and had virtually no effect on the microstructure of the 5182 layers. For all the laminates in this study sharp interfaces were maintained between the component layers. Additional details concerning the materials and processing procedures are available in references 7 and 8.

Testing

Two types of fracture toughness tests were performed. The UHCS/Brass laminates were tested using chevron notch short bars according to ASTM Standard Test Method E1304-89. Samples were tested in two layer thicknesses (200 μm and 750 μm) and in both the "crack arrester" and "crack divider" orientations as shown in Fig. 1.

![Fig. 1. Sketches of fracture toughness samples in (a) crack arrester and (b) crack divider orientations. The loading direction and the intended fracture plane (shaded region) are shown in (c).](image_url)

The toughness and work of fracture were evaluated in the Al/Al-SiC laminates using chevron notch three-point bend bars, as sketched in Fig. 2. As with the UHCS/Brass laminates tests were done in both the "crack arrester" and "crack divider" orientations. The depth, thickness, and span length of the chevron-notch bend bars were 10.2 mm, 15.2 mm, and 61.0 mm, respectively. Measurements were made of load, load point displacement and crack opening displacement. Testing was done according to the procedures described by Wu (9). The procedure does not require fatigue pre-cracking of the test sample and the toughness was calculated from the maximum load in the load-crack opening displacement record. The work of fracture ($\gamma_f$) was calculated from the area under the load-load point displacement curve up to the point of fracture ($W_f$) according to the following expression (10, 11).

$$\gamma_f = \frac{W_f}{2A}$$

where A is the area of the triangular ligament shown in Fig. 2.
Results and Analysis

UHCS/Brass Laminates.

Fracture Toughness. Fracture toughness values for the UHCS/Brass and Al/Al-SiC laminates are summarized in Tables I and II. The toughness values for both these materials are given in terms of $K_I$ rather than $K_{IC}$ because the size requirement, especially the thickness requirement, specified in the standard test method could not be met. Additional details of data analysis are available in reference 7. Laminates in the "crack divider" orientation had higher toughness than those in the "crack arrester" orientation. In the "crack divider" orientation similar toughness values were obtained for both layer thicknesses suggesting that, over the range of thicknesses studied, a layer thickness effect does not exist. It is interesting to note that Shaw and Abbaschian (12) studied the influence of layer thickness on toughening in MoSi$_2$/Nb laminates containing 20% Nb. In this study a layer thickness effect was observed over a Nb thickness range from 125 μm to 1000 μm. However over the range in thicknesses comparable to the ones in this study the influence of layer thickness on fracture toughness was virtually nil.

Table I

<table>
<thead>
<tr>
<th>Layer Thickness (μm)</th>
<th>Orientation</th>
<th>Number of layers in the notch</th>
<th>$K_I$ (MPa m$^{-1/2}$)</th>
</tr>
</thead>
<tbody>
<tr>
<td>750</td>
<td>Crack Arrester</td>
<td>15</td>
<td>53.2</td>
</tr>
<tr>
<td>200</td>
<td>Crack Arrester</td>
<td>45</td>
<td>45.0*</td>
</tr>
<tr>
<td>750</td>
<td>Crack Divider</td>
<td>17</td>
<td>76.0</td>
</tr>
<tr>
<td>200</td>
<td>Crack Divider</td>
<td>51</td>
<td>75.1</td>
</tr>
</tbody>
</table>

* Value underestimated. Specimen arms broke off.

Crack Growth Characteristics. The fracture surfaces of the UHCS/Brass laminates tested in the "crack arrester" and the "crack divider" orientation are shown in Figs. 3 and 4 respectively. For both orientations the 750 μm layer thickness samples show extensive delamination whereas the 200 μm layer thickness samples show little delamination. The "crack arrester" orientation shows delaminations at every other interface whereas the "crack divider" orientation shows delaminations at every interface. The change in delamination characteristics between thick and thin layer systems is related to the higher interlaminar residual shear stress in the thick layer system. This point will be discussed further in another section. Ridges can be observed in the brass layers, indicating significant plastic flow, whereas the UHCS surfaces are relatively flat. The ridges in the brass layers are especially pronounced in the 750 μm layer thickness samples.

From the fracture surfaces in Fig. 3 it is apparent that crack growth through the laminate in the "crack arrester" orientation occurs in a discontinuous, jerky manner with rapid, catastrophic crack growth in the UHCS and slow, ductile tearing in the brass. The micrographs show that severe delamination occurs on the entering side of the UHCS-brass interface. Clearly these interfaces indicate regions where the advancing crack was arrested and subsequent crack growth required re-nucleation.

From the fracture surfaces in Fig. 4 it is apparent that the crack growth through the laminate in the "crack divider" orientation does not occur with a planar crack front. Cracks in the UHCS layers lead cracks in the brass layers. In the 750μm layer thickness system these non-planar crack fronts are connected through delaminations at the interfaces. The shape of the advancing crack front is shown schematically in Fig. 5 for thick and thin layer laminates. The growth in the crack front is retarded by the plastic tearing required for crack growth in the brass layer. Clearly the high ductility and high work
hardening rate in the brass contribute substantially to the overall crack growth resistance in these laminates. Fig. 5(b) shows the crack front for a thin layer laminate in which limited delamination occurs. Clearly the convolutions in the crack front for this thin layer system are not as deep as the convolutions for the thick layer system in which delamination occurs. However, because of the thinner layers, there are more convolutions in the crack front. These results can explain the lack of a layer thickness effect in this orientation. The toughness benefits of deep convolutions and local delamination in the thick layer system are offset by the larger number of convolutions in the thin layer system.

Fig. 3. Fracture surfaces of UHCS / brass laminates tested in the "crack arrester" orientation having (a) 750 µm and (b) 200 µm layers.

Fig. 4. Fracture surfaces of UHCS / brass laminates tested in the "crack divider" orientation having (a) 750 µm and (b) 200 µm layers.
Advancing crack front (a)

Local delamination

Direction of crack propagation

(b)

Fig. 5. Schematic drawing showing the shape of the advancing crack front for UHCS/Brass laminates tested in the "crack divider" orientation for a thick layer system (a) and a thin layer system (b). The thick layer system shows local delaminations at the interfaces.

Al/Al-SiC laminates.

Fracture Toughness. The load-load point displacement records for the Al/Al-SiC laminates are shown in Fig. 6 for several global SiC contents. Results are shown for both the "crack arrester" (6(a)) and "crack divider" (6(b)) orientations. The load - load point displacement data for the crack divider orientation show a much more gradual loading of the sample than the data for the "crack arrester" orientation. The data for the "crack arrester" orientation show distinct load drops that are associated with delaminations at 5182 - 6090/SiC/25p interfaces. These delaminations are a very effective source of extrinsic toughening since they arrest the crack. Subsequent crack growth in this orientation requires re-nucleation of the crack. By contrast in the "crack divider" orientation crack growth occurs in a relatively smooth and continuous fashion.

The fracture toughness for the Al/Al-SiC laminates is shown in Table II and is plotted as a function of global percentage SiC in Fig. 7. The composition of the component layers is the same for all data points in Fig. 7; the increase in global volume percentage SiC was obtained by increasing the relative percentage of the 6090/SiC/25p-T6 layer. The component materials of the laminate differ significantly in fracture toughness (with 5182 having a higher toughness than 6090/SiC/25p-T6). Remarkably, however, the fracture toughness of the laminate actually increases with increasing global volume percentage of SiC up to a component volume fraction in which the LMC is almost 100% MMC. These results highlight the strong influence that interfaces and controlled delamination have on fracture toughness. Clearly in these laminates extrinsic toughening mechanisms dominate the fracture toughness. The intrinsic toughening mechanisms associated with the individual layers are having less influence on the overall toughness of the laminate.

The work of fracture is also shown in Table II. In contrast to the toughness, the work of fracture decreases with increasing global % SiC. Also the work of fracture shows a much greater variation with global percentage SiC than the toughness.

Crack Growth Characteristics. These conclusions regarding extrinsic toughening are supported by the features observed on the fracture surfaces. In both orientations local delamination was observed in the fracture surfaces. Results are shown in Fig. 8 for the "crack divider" orientation with three different volume percentages of the MMC component - 50%, 76% and 97%. Every interface between layers shows delamination. In addition every 5182 layer and 6090 MMC layer failed in shear. The 5182 component has higher inherent toughness than the 6090 MMC component. Thus, as in the case of the UHCS/Brass laminates, the growth of the crack through the laminate in the "crack divider" orientation does not occur with a planar crack front. The cracks in the MMC layer lead the cracks in the 5182 layer. As with the UHCS/Brass laminates these non-planar crack fronts are connected through delaminations at the interfaces. This highly convoluted crack front contributes significantly to the extrinsic toughness of these laminates.
Fig. 6. Load - load point displacement records for Al/Al-SiCp laminates with three different global SiCp carbide contents. Samples were tested in the "crack arrester" (a) and "crack divider" orientations (b) using the chevron-notch three point bend bar. Global percentage of SiCp is indicated.

Table II

<table>
<thead>
<tr>
<th>Global Volume Percentage SiC</th>
<th>Percentage MMC Component</th>
<th>Orientation</th>
<th>Number of Layers in Notch</th>
<th>$K_I$ (MPa m$^\text{1/2}$)</th>
<th>$W_f$ (Joules/m$^2$ x 10$^3$)</th>
</tr>
</thead>
<tbody>
<tr>
<td>12.5</td>
<td>50</td>
<td>Crack Arrester</td>
<td>13</td>
<td>32</td>
<td>15.4*</td>
</tr>
<tr>
<td>19.8</td>
<td>79</td>
<td>Crack Arrester</td>
<td>9</td>
<td>37</td>
<td>14.6</td>
</tr>
<tr>
<td>24.3</td>
<td>97</td>
<td>Crack Arrester</td>
<td>6</td>
<td>37</td>
<td>11.8</td>
</tr>
<tr>
<td>12.5</td>
<td>50</td>
<td>Crack Divider</td>
<td>15</td>
<td>28</td>
<td>12.2*</td>
</tr>
<tr>
<td>18.5</td>
<td>74</td>
<td>Crack Divider</td>
<td>11</td>
<td>30</td>
<td>9.75</td>
</tr>
<tr>
<td>19.8</td>
<td>79</td>
<td>Crack Divider</td>
<td>10</td>
<td>31</td>
<td>6.67</td>
</tr>
<tr>
<td>22.8</td>
<td>91</td>
<td>Crack Divider</td>
<td>22</td>
<td>34</td>
<td>5.70</td>
</tr>
<tr>
<td>24.0</td>
<td>96</td>
<td>Crack Divider</td>
<td>13</td>
<td>36</td>
<td>6.21</td>
</tr>
<tr>
<td>24.3</td>
<td>97</td>
<td>Crack Divider</td>
<td>9</td>
<td>33</td>
<td>5.82</td>
</tr>
</tbody>
</table>

*Values underestimated. Test stopped before fracture.

Fig. 7 Fracture toughness versus global volume percent silicon carbide for a laminate containing 6090/SiC/25p-T6 and 5182 layers in the crack arrester and crack divider orientations. The fracture toughness of 6090/SiC/25p-T6 is shown in the figure.
Fig. 8. Fracture surfaces of "crack divider" samples containing three different volume percentages of the component materials. Dark layers are 6090/SiC/25p and the light layers are 5182. The volume percentage of the MMC component is (a) 50\%, (b) 76\% and (c) 97\%.

**Influence of Residual Stress.**

Results in the previous sections have shown that delamination is an important characteristic of laminate fracture. Residual shear stresses developed at the interfaces can have a significant influence on delamination (13). These residual shear stresses can be produced by the thermal expansion mismatch between the component materials that occurs during cool-down from the processing temperature. A qualitative estimation of the residual shear stress distribution near the interface in a laminated metal composite is available by analogy with similar stress distributions in a microelectronic thin film deposited on a thick substrate (14, 15). In this study, which was done using finite element analysis, the residual shear stress was found to exist near the edges of the structure over a characteristic distance approximately equal to the thickness of the thin film. By analogy one can hypothesize that for laminated metal composites the residual shear stress distribution near the interface can be represented as shown in Fig. 9. The shear stress distribution over most of the interface is zero and reaches non-zero stress values near the edges over a distance approximately equal to the layer thickness as shown in the inset in Fig. 9. The maximum value in the stress distribution is taken equal to the yield strength in shear of the ductile component (brass in the UHCS-Brass laminate). Thus, one would expect failure to originate from the...
edges of a laminate, which is a common failure origin for polymer matrix laminated composites (16, 17). For changes in layer thickness the distribution shown in Fig. 9 is expected to have the same maximum value of shear stress and a characteristic width (over which the residual stress is non-zero) that is approximately equal to the layer thickness. The characteristic width will increase with an increase in layer thickness increasing the likelihood and extent of delamination.

The exact nature of the residual shear stress distribution, shown in Fig. 9, however, will depend on the magnitude of the mismatch strain that is produced between the component materials upon cooling to room temperature from the processing temperature. A one-dimensional analysis of this strain, assuming isostress behavior, can be taken as $\Delta\alpha\Delta T$, where $\Delta\alpha$ is the difference in thermal expansion coefficients between the component materials and $\Delta T$ is the temperature change upon cooling to room temperature. These mismatch strains are given in Table III for the two systems studied here. The strains are quite large with the UHCS/Brass system having greater mismatch strain than the Al/Al-SiC system. This mismatch strain will influence the stress distribution shown in Fig. 9. However the resulting influence of this mismatch strain on delamination is much more complex with delamination depending on interface strength and the plastic flow behavior of the component materials.

![Fig. 9. Residual shear stress distribution near the interface for a laminated metal composite based on ultrahigh-carbon steel and brass.](image)

$\sigma_0$ is the yield strength in shear of the ductile brass component. Non-zero values of shear stress exist over a characteristic distance near the edge that is approximately equal to the layer thickness. An increase in layer thickness will increase this characteristic distance with a resulting increase in the extent of delamination during crack growth.

<table>
<thead>
<tr>
<th>Laminate System</th>
<th>Processing Temperature ($^\circ$C)</th>
<th>CTE (Component 1) $\alpha_1$ [x10^-6 $^\circ$C^-1]</th>
<th>CTE (Component 2) $\alpha_2$ [x10^-6 $^\circ$C^-1]</th>
<th>Mismatch strain ($\Delta\alpha$)$\Delta T$</th>
</tr>
</thead>
<tbody>
<tr>
<td>UHCS / Brass</td>
<td>750</td>
<td>(UHCS) 10.1</td>
<td>(Brass) 19.9</td>
<td>.0071</td>
</tr>
<tr>
<td></td>
<td></td>
<td>(6090/SiC/25p)</td>
<td>(5182)</td>
<td>.0045</td>
</tr>
<tr>
<td>Al / Al-SiC</td>
<td>450</td>
<td></td>
<td></td>
<td></td>
</tr>
</tbody>
</table>

CTE = Coefficient of Thermal Expansion
As can be seen from the preceding sections one of the most significant advantages of LMCs is the large improvements in fracture toughness that are possible. For weight critical applications these improvements are particularly dramatic; comparisons between materials for such lightweight applications can best be made if critical properties are normalized by density. For applications requiring lightweight materials strength, stiffness and toughness can be very important. In Figs. 10 and 11 these density normalized properties, namely the specific strength, specific stiffness and specific toughness, are compared for a number of structural materials including two ceramics and two laminates. One laminate consists of 50% 6090/SiC/25p-T6 and 50% 5182 while the other consists of 97% 6090/SiC/25p-T6 and 3% 5182. For reference the data for the quench and tempered 4340 steel in Figs 10 and 11 was derived from a UTS of 1465 MPa and a fracture toughness of 85 MPa m\(^{\frac{1}{2}}\) (18). The figures show that the laminates have very attractive properties for weight critical structures. The laminate containing 97% MMC (24.3% global percentage SiC) is particularly impressive since it has a specific toughness greater than any of the other materials in the figures. Figure 7 shows that the laminates reported here can be particularly effective for applications requiring lightweight materials with high stiffness, strength and fracture toughness.

Fig. 10. Specific toughness versus specific strength for a number of structural materials including a laminate containing 50% 6090/SiC/25p-T6 and 50% 5182 and a laminate containing 97% 6090/SiC/25p-T6 and 3% 5182.

Fig. 11. Specific toughness versus specific stiffness for a number of structural materials including a laminate containing 50% 6090/SiC/25p-T6 and 50% 5182 and a laminate containing 97% 6090/SiC/25p-T6 and 3% 5182.

Conclusions

1. Lamination can significantly improve the fracture toughness of MMCs.
2. In the "crack arrester" orientation, increases in fracture resistance are achieved through crack arrest and re-nucleation.
3. In the "crack divider" orientation increases in fracture resistance are achieved from a non-planar crack front. Both the depth and frequency of convolutions in the crack front are contributing factors to toughness.
4. Residual shear stresses produced by thermal expansion mismatch between the component materials can promote delamination during crack growth. These residual stresses exist over a characteristic distance approximately equal to the layer thickness.
5. Laminates can have very attractive properties relative to other structural materials for lightweight critical applications requiring high stiffness, strength and fracture toughness.
References


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