JOURNAL OF MATERIALS SCIENCE **39** (2004) 2855–2859

Interfacial fracture toughness of polyester-based fiber-metal laminates with primary contact and secondary adhesive bonding

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Fiber-metal laminates (FMLs) consist of alternate thin layers of aluminum alloy and fiber-reinforced polymer-matrix composite. They combine the toughness and machinability of aluminum with the excellent specific and fatigue properties of composites [1, 2]. FMLs also exhibit excellent impact performance. Vlot *et al.* [3] reported that a series of thermoset-based FMLs exhibit superior energy absorption for damage initiation and perforation under low velocity impact, when compared to aluminum on the basis of areal weight. The major failure modes reported by Vlot *et al.* were fracture of the aluminum, fiber and matrix. Further work by Reyes and Cantwell [4] showed that delamination at the bi-material interface does not occur under low velocity impact if a tough bond between the composite and aluminum is achieved. The absence of delamination is significant as residual properties will be improved.

To date, FMLs have been used almost exclusively for aerospace applications, where the expense of thermoset or thermoplastic-based prepreg and metals such as titanium and aerospace grade aluminium can be justified. The superior performance of FMLs is relevant for other industries such as marine and transport, but high performance metals and composite prepreg would not be desirable due to cost. Instead, FMLs would be attractive if manufacturers could use current materials and associated manufacturing techniques, such as wet-lay up with glass-fiber/polyester composites or aluminum in a high volume stamping process. One of the important first steps in developing new FML combinations for non-aerospace applications is to achieve a satisfactory bond between the layers, which will ensure load transfer and provide delamination resistance. This letter presents preliminary results on the characterization of interfacial fracture toughness for two FMLs based on materials and manufacturing techniques suitable for non-aerospace and relatively low cost applications. The first FML is based on primary contact bonding at the bi-material interface; that is, simple wet-lay-up of glass/polyester on to aluminum. The second is based on secondary adhesive bonding where the aluminium and pre-cured laminates are manufactured in a simple stamping process with a hot-melt thermoplastic adhesive at the interface.

The primary bonded FML was made by laminating 10-plies of 0/90° woven roving E-glass fiber (Colan AR106; areal weight 638 g/m^2) onto a 2 mm thick sheet of 5005-H34 aluminum alloy using a room temperature curing isophthalic polyester (Nuplex F61042)

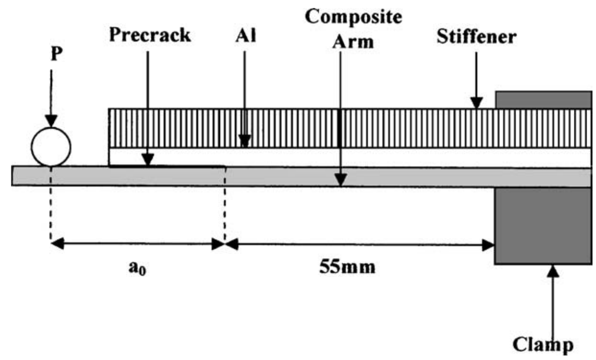

Figure 1 Single cantilever beam specimen geometry for interfacial fracture testing.

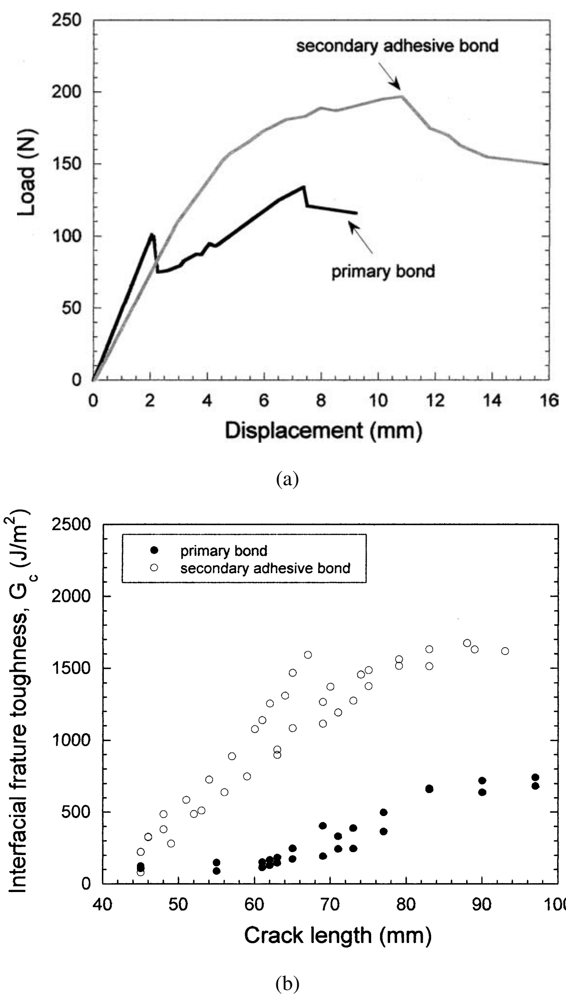

Figure 2 (a) Load-displacement plots and (b) resistance (*R*) curves from the interfacial fracture tests.

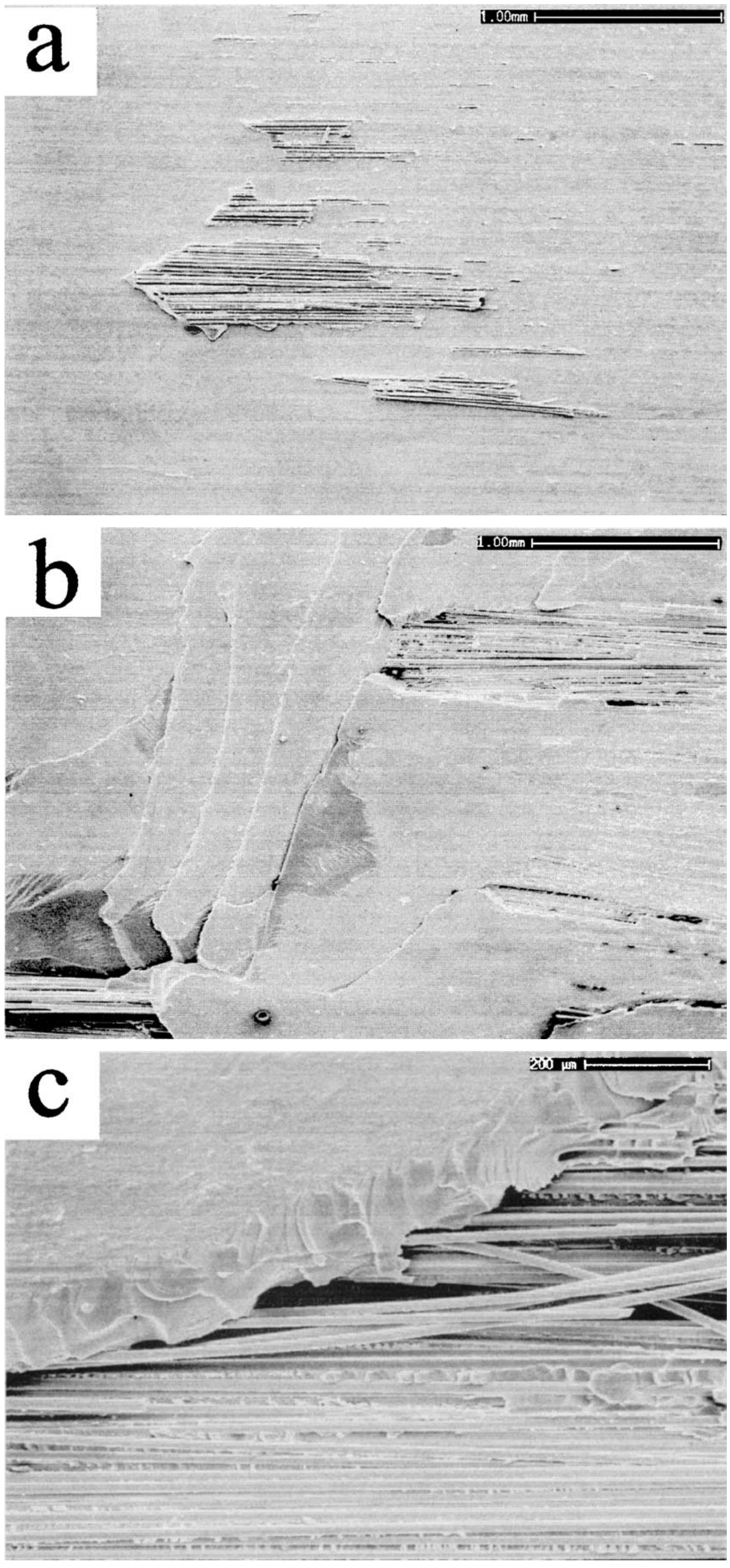

Figure 3 Interfacial fracture surface micrographs of the primary bonded FML: (a) aluminum arm, (b) composite arm and (c) higher magnification of delamination within the composite arm.

resin. The aluminum alloy was given a proprietary anodizing treatment prior to lamination. A piece of 15 μm thick aluminum foil, coated with a Teflon-based release agent, was placed at mid-thickness to act as precrack for fracture testing. A caul plate was placed on the laminate, which was then briefly evacuated to remove entrapped air and left to cure at room temperature for a minimum of 24 h. The secondary bonded FML was made

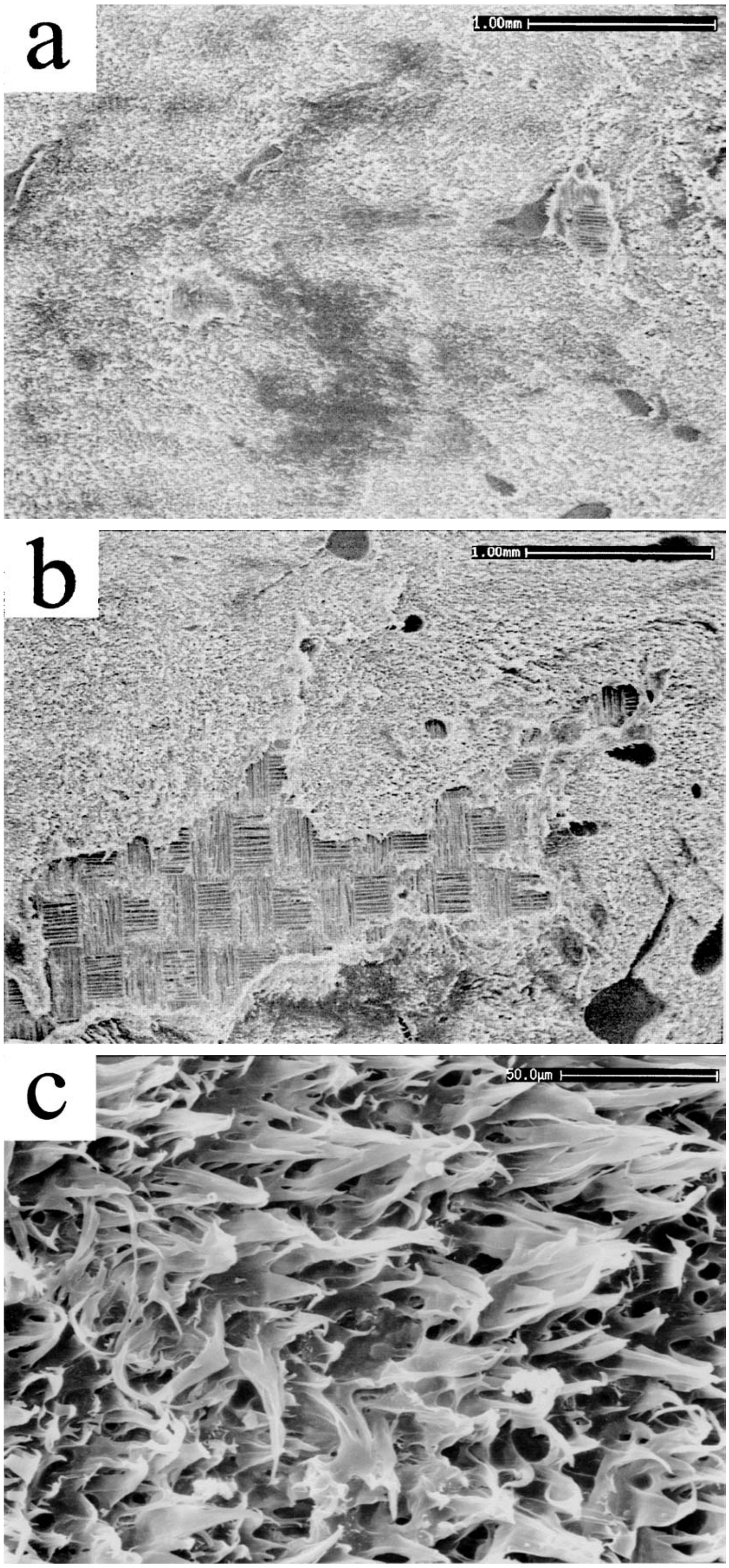

Figure 4 Interfacial fracture surface micrographs of the secondary adhesive bonded FML: (a) aluminum arm, (b) composite arm and (c) higher magnification of ductile deformation in the ethylene-based adhesive.

by adhering a cured 10-ply composite laminate to a 2 mm thick aluminum sheet in a heating-stamping procedure. The adhesive was a 50 μm thick ethylene-based hot-melt thermoplastic film (XAF 22.400, Collano Xiro AG) recommended by the manufacturer for bonding polyester with aluminum. A layer of the film adhesive was placed on the aluminum sheet, followed by the composite. The starter crack was placed between the aluminum and adhesive layer. The laminate was heated to 160 °C in an oven then removed and immediately stamped at a pressure of approximately 65 kPa.

The interfacial fracture toughness characterization was performed using the single cantilever beam (SCB) specimen geometry [4, 5] as shown in Fig. 1. A nomex honeycomb stiffener with carbon-fiber/epoxy skins was bonded to the aluminum arm to ensure it remained rigid during the test. The nominal specimen width was 20 mm and the initial crack length, a_o, was 45 mm. The average composite arm thickness was 5.75 mm. The edges of all specimens were polished and covered with a layer of white correction fluid and marked at 5 mm intervals from the crack tip. This allowed identification and monitoring of the crack growth, using a traveling microscope, during the test. In addition, the aluminum was removed from one end of the specimen to allow the load to be applied to the composite arm. The other end of the specimen was clamped in a steel fixture, leaving approximately 55 mm available for crack growth. A minimum of three specimens of each FML was tested in displacement control at 1 mm/min on an Instron (model 4505) universal testing machine (UTM). The interfacial fracture energy, G_c, was calculated using a compliance calibration method of the form:

$$G_c = \frac{3P^2ma^2}{2B}$$

where m is the slope of compliance versus a^3. A resistance (R) curve (G_c versus a) was obtained for each specimen.

Typical load-displacement plots from the SCB tests are shown in Fig. 2a. Fracture behavior is predominantly stable in the primary bonded laminate, but there are also short instances of unstable crack growth. Crack growth is completely stable in the secondary bonded laminate, and there is significantly greater load and displacement. The R curves in Fig. 2b show an initial rise in G_c with crack growth followed by a plateau region that reflects a steady-state between energy input and absorption during fracture. The plateau for both FMLs develops at a crack length of approximately 80 mm, and the average value of G_c in this region for each specimen was used to determine an average interfacial G_c value for each FML. The average (standard deviation) interfacial G_c for the primary bonded FML is 693 (26) J/m^2, which is a reasonably high value given that negligible pressure was applied across the contact surfaces of the layers during fabrication. In comparison, the interfacial G_c for the secondary bonded FML is significantly higher at 1545 (83) J/m^2.

Fracture surface micrographs from the SCB specimens elucidate the major failure modes and explain the difference in interfacial G_c. Large areas of bare aluminum are still visible on the aluminum arm of the primary bonded FML in Fig. 3a, and the composite arm in Fig. 3b shows large areas of smooth undeformed matrix-resin. These features are clear evidence of adhesive failure at the bi-material interface. Small areas of composite can still be seen bonded to the aluminum arm, which correspond to areas on the composite arm where fibers can be seen. A higher magnification view of one of these areas, Fig. 3c, shows further failure modes when the crack propagates into the composite. The matrix is deformed with evidence of hackle marks, which are a result of the mode II shear loading component introduced by the SCB geometry, and there is also some fiber debonding. The reason for crack propagation into the composite is not clear, however it is likely that a defect such as a void or poor fiber wet-out in the composite close to the interface produced a stress concentration that briefly caused the crack to propagate through the composite. It is also possible that the crack growth through the composite corresponds to the unstable crack growth shown by the load-displacement plots, however this point requires further investigation. While the failure modes in the composite can contribute significantly to energy absorption, it is concluded that the dominant interfacial failure is responsible for the comparatively low G_c value.

The improved G_c for the secondary bonded sample is due to excellent bonding of the ethylene-based adhesive to the aluminum and composite arm as shown in Fig. 4a and b, respectively. The adhesive covers most of both arms after fracture, indicating predominantly cohesive failure. There are areas showing some interfacial fracture, but unlike the primary bonded specimen there was no clear evidence of crack propagation into the composite. A higher magnification micrograph of the adhesive, Fig. 4c, (on the composite arm) shows extensive ductile deformation of the adhesive. This observation indicates that deformation and fracture of the adhesive layer is the major energy absorbing mechanism responsible for higher interfacial G_c value. The interfacial G_c for this secondary bonded FML compares well with results from other thermoplastic-based FMLs characterized using the SCB test. Polyamide and polyetherimide-based FMLs bonded with a 30 μm thick ethylene-based hot-melt adhesive (Surlyn, Du Pont de Nemours) produced results of 1200 and 800 J/m^2 respectively [6, 7].

The results presented here show that a simple heating-stamping manufacturing process has potential for high volume production of relatively low cost thermoset-based FMLs with a tough ethylene-based thermoplastic film adhesive at the bi-material interface to ensure excellent interfacial fracture toughness. Further work will investigate the impact performance and post-impact structural integrity of this secondary bonded polyester-based FML.

Acknowledgments

Electron microscopy was conducted in the Electron Microscope Unit at the Australian National University. The supply of the Collano Xiro ethylene-based film adhesive from Amanda Smith of ADS Projex Pty Ltd. (Australia) is gratefully acknowledged.

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Received 12 August
and accepted 13 November 2003