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The role of matrix cracks and fibre/matrix debonding on the stress transfer between fibre and matrix in a single fibre fragmentation test

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ABSTRACT

The single fibre fragmentation test is commonly used to characterise the fibre/matrix interface. During fragmentation, the stored energy is released resulting in matrix cracking and/or fibre/matrix debonding.

Axisymmetric finite element models were formulated to study the impact of matrix cracks and fibre/matrix debonding on the effective stress transfer efficiency (EST) and stress transfer length (STL). At high strains, plastic deformation in the matrix dominated the stress transfer mechanism. The combination of matrix cracking and plasticity reduced the EST and increased STL.

For experimental validation, three resins were formulated and the fragmentation of an unsized and uncoupled E-glass fibre examined as a function of matrix properties. Fibre failure was always accompanied by matrix cracking and debonding. With the stiff resin, debonding, transverse matrix cracking and conical crack initiation were observed. With a lower modulus and lower yield strength resin the transverse matrix crack length decreased while that of the conical crack increased.

1. Introduction

The fragmentation test is a commonly used test method for the characterization of the fibre/matrix interface. It displays the basic phenomena that are present in multi-fibre composites such as fibre fracture, interfacial debonding, matrix yielding and matrix cracking [1]. In the fragmentation test a continuous fibre is embedded in a resin coupon and subjected to axial tension. Load transfer from matrix to fibre occurs through interfacial shear. As the load increases the fibre breaks into smaller and smaller fragments until they are too short to be further loaded to fracture. Kelly and Tyson [2] related the interfacial shear stress τ to the critical fibre length l_c in the following Eq. (1),

$$\tau = \frac{r\sigma_{uf}}{l_c} \tag{1}$$

where r is the radius of the fibre and σ_{uf} is the strength of the fibre at a length l_c.

Eq. (1) assumes that a constant shear stress exists at the interface of the fragment. This assumption is valid when fragmentation has saturated, either as a result of complete debonding or matrix yielding at the interface.

In the fragmentation test a fibre-break is often accompanied by debonding and/or matrix cracking. Three failure modes are encountered; a transverse matrix crack, an inclined conical matrix crack and/or fibre/matrix debond [3,4]. The failure mode depends on the quality of interface [4] as indicated below;

(a) If the interface is strong in an elastic matrix, a fibre-break is accompanied by a transverse matrix crack.
(b) If the interface bond strength is good but the matrix has a low shear strength an inclined conical crack will form.
(c) Where the interface bond strength is low, debonding is initiated at the interface.

The matrix mechanical properties also have a significant influence on the formation of the matrix cracks. However, the reason for the formation of transverse or conical matrix cracks is a subject of further study. According to Huang and Talreja [5] plastic deformation precedes the fracture and the predictive finite element model uses critical plastic strain as a criterion for facture initiation of conical matrix crack.

During a fibre-break the stored elastic energy is released to the surrounding matrix causing either interface failure and/or matrix cracking. Studies by various authors [3,4,6,7] have shown that matrix cracks propagate with applied strain, perpendicular to the fibre fracture. Matrix cracks which form at lower applied strains propagate a shorter distance into the polymer than those which form at high strains, because of a lower stored elastic energy [7]. It was also shown that transverse matrix cracks propagate faster than conical matrix cracks which are usually at 45° to the fibre axis

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[4]. The existence of a matrix crack changes the stress field surrounding the fibre-break and influences the stress transfer across the interface. An analytical and finite element study by Liu et al. [8,9] has shown that a matrix crack significantly affects the rate of stress transfer. A matrix crack emanating from a fibre-break can cause premature failure of a high volume fraction composite because the stress concentration on adjacent fibres and matrix leads to the formation a crack of critical dimensions which propagates through the composite. Thus the fibres do not reach their full load carrying potential [4]. Recently Behzadi et al. [10] have studied the role of multiple fibre-breaks on the failure of a 0° fibre composite. It has been reported that the formation of a matrix crack is influenced by cure temperature, matrix modulus, surface treatment of the fibres, and strain rate [3,4,11–13].

An energy based approach is frequently used to characterise the interface by balancing the stored energy in the system to that released in the formation of new debonded surfaces. Energy employed in the formation of a matrix crack should also be included to correctly characterise the interface. Strain energy released by interfacial debonding was estimated to be 57–342 J/m^2 and for conical and transverse matrix cracks 58–103 J/m^2 [3]. Therefore the force required to initiate debonding appears to be larger than that for matrix cracking.

In the present work, a systematic experimental study was performed to understand the influence of matrix modulus and yield strength on the formation of a matrix crack and/or fibre/matrix debonding. Following this finite element analysis was used to study the impact of transverse matrix cracks, conical matrix cracks and fibre/matrix debonding on the stress transfer efficiency at the fibre/matrix interface since there is no established method for quantifying the stress in glass and other fibres where direct methods such as laser Raman spectroscopy, which can be used for high modulus carbon and ceramic (eg. alumina) fibres, cannot be employed.

2. Experimental

2.1. Fibres

The unsized E-glass fibres used in this work were obtained from Owens Corning Fibreglass, with a fibre diameter of 15.32 ± 0.46 μm.

2.2. Matrix resin

The epoxy resin used was Araldite LY 5052 and GY 298 (Ciba-Geigy). These were mixed in proportions of 100/0 (PHY), 70/30 (PMY) and 60/40 (PLY) with the calculated amount of hardener Aradur 5052 to give three different resin types. This ensured that the resins were chemically similar but with varying mechanical properties. The proportions used are given in Table 1. Properties of the elastic resin used were taken from Ref. [17]. 'P' denotes plastic, 'E' denotes elastic, H, M, L stands for high, medium, low and 'Y' denotes yield strength. The resins were blended using a mechanical stirrer for 5 min and degassed in a vacuum chamber.

For the preparation of fragmentation test specimens, single untreated-unsized E-glass fibres were separated from the fibre tow and mounted across a metal frame. The mounted fibres were then carefully placed in the dog-bone mould and the degassed resin was poured into the dog-bone moulds and cured in an oven at 80 °C for 8 h. The resin properties are referred here as,

PHY = High modulus and high yield strength epoxy resin.
PMY = Medium modulus and medium yield strength epoxy resin.
PLY = Low modulus and low yield strength epoxy resin.
EH = Elastic high modulus resin.
EL = Elastic low modulus resin.

Table 1
Resin formulations used in the study.

Resin (pbw)	PH	PM	PL
Araldite LY 5052 (epoxy 1)	100	70	60
Araldite GY 298 (epoxy 2)	0	30	40
Aradur 5052 (hardener)	38	30.57	28.09

pbw – parts by weight.

2.3. Specimen testing (single fibre fragmentation test)

Testing of the single fibre dog-bone specimens was performed using a specially designed mini-tensile testing machine supplied by Micromaterials Ltd., Wrexham, UK. The set-up is illustrated schematically in Fig. 1a. Testing of the single fibre dog bone samples were carried out by clamping the specimens within the grips of the testing machine, and loading the samples at a crosshead speed of 0.13 mm min^{-1}, with the use of the stepper motor. The output from the load cell, and the LVDT (linear voltage displacement transducer) were monitored using a commercial data-logging package (PICOLOG), which enabled direct observation of stress and strain applied to the sample. The fragmentation test was also monitored using a transmitted optical light microscope interfaced to a JVC CCD camera (the CCD was connected to an external monitor and the computer). A polarising plate was affixed to the end of the microscope, and another polarising plate (out of phase) was affixed on top of the light source. This enabled any stress birefringence phenomena occurring within the sample to be observed simultaneously on the screen (Fig. 1b). The strain applied to the specimens was determined by measuring the distance between the lines before and after the application of load. The monitored length was 20 mm along the gauge length. This was achieved by measuring the sample using a vernier caliper fitted to the microscope. Once the required length was obtained the test was interrupted and the specimen was removed for further study. A custom-written computer program (IMAGE) windows based C++ program was used to quantify the images captured during the single fibre fragmentation test. For each specific study three specimens were examined.

3. FEA modelling

The axisymmetric finite element model given in Fig. 2, was used to study five representative micromechanical events which were observed during fragmentation of a single embedded filament. The material properties used for modelling were extracted from the experimental data. Five cases were modelled:

(1) A perfectly bonded fibre and matrix (PB).
(2) Perfectly bonded fibre with a two fibre diameter-transverse matrix crack (PBTC).
(3) Perfectly bonded fibre with two fibre diameter inclined-conical matrix crack (PBCC).
(4) Two fibre diameter debond (DB).
(5) Two fibre diameter debond with transverse matrix crack (DBTC).

The model was 0.5 mm in length and 0.3 mm in width from the centre of the fibre. The dimensions were selected in order to reduce the edge effects. The fibre was assumed to be elastic and the matrix elastic or elasto-plastic. Fibre and matrix were assumed to be two