MIXED-MODE FIBRE-MATRIX DEBONDING IN TRANSVERSE CYCLIC LOADING OF UNIDIRECTIONAL COMPOSITE PLIES

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Abstract

Fatigue of composite materials is of great concern in load-carrying structures. The first type of damage to form is generally transverse cracks in off-axis plies. These cracks form when fibre-matrix debonds coalesce. The underlying mechanism is hence fatigue growth of debonds at the fibre-matrix interfaces. Since the fibres are cylindrical, the debonds grow necessarily under mixed-mode conditions. In the present study, debond growth has been characterized under tensile and compressive cyclic loading of single glass fibres embedded in polymer matrix. The debond length was determined by in-situ microscopy with transmitted polarized light. Boundary element calculations have been used to determine the energy release rate, in modes I and II, for the growing debonds. These results may be used to formulate a fatigue growth law at a local microscopic level, at a stage prior to the formation of any visible damage, i.e. transverse cracks. The main result is that it is possible to measure debond growth at the fibre level, and that a quantitative analysis can be used to predict growth. Ideas of how to develop this methodology further are also discussed. The goal is to be able to predict failure, and to develop a tool for materials design of more fatigue resistant composites.

Introduction

Most failures of composite structures can be attributed to fatigue. Due to the heterogeneity of composite material at different scales, a large variety of interacting mechanisms contribute to fatigue failure. If the incipient mechanisms at the onset of damage accumulation could be better understood, measures could be taken to suppress these in order to extend the lifetime of the material. The first observable type of damage in fatigue of composite laminates is generally transverse cracking in plies with oblique fibre orientation to the direction of the largest principal ply strain. Microscopic investigations have shown that the transverse cracks are initiated from coalescence of fibre-matrix debonds, both in static and cyclic loading [1]. A schematic illustration is shown in Fig. 1. This study concerns experimental characterization of transverse debonding under cyclic loading in a single-fibre composite. The boundary element method has been used to determine the mixed-mode energy release rate of the debonds.

Another aspect of fatigue in composites is that these materials are generally more sensitive to compressive excursions during cyclic loading compared with monolithic materials. For composite materials, tension-compression fatigue has shown to be more deleterious than tension-tension fatigue. The *S-N* curves plotted with the peak maximum stress or strain along the ordinate show significantly steeper slopes for load wave forms containing compressive load excursions. This has been shown to be the case for multidirectional laminates containing off-axis plies [2, 3] and in particular for those containing transverse plies [4-6]. Detrimental effects of compressive load excursions have also been observed for pure unidirectional transverse composites [7, 8]. The physical reason for the adverse effect of tension-compression loading of composite materials has not been clarified entirely. For multidirectional laminates or unidirectional laminates with distinct resin-rich inter-ply regions, delamination followed by out-of-plane ply buckling would certainly play an important role on the ply-level scale. On the scale of an individual fibre or a group of fibres, the underlying mechanics remains more unclear. If this could be clarified, further progress in the development of more predictive tools and constructive materials design is to be anticipated. For materials with a more homogeneous microstructure fatigue crack growth is characterized by self-similar propagation of single large cracks until the largest one of them achieves a critical size and catastrophic failure takes place. This propagation is primarily governed by crack opening under remote tensile loading. A compressive load excursion would close the crack tip and effectively result in negligible propagation of the crack in comparison with corresponding tensile load fluctuations. However, for a heterogeneous material like a composite that consists of phases with different elastic properties, distributed microcracks of various sizes and shapes would form at an early stage of loading. The interfaces between the constituents would be particularly prone to cracking due to the inherent stress concentration and residual stresses from processing. Considering the cracks between media of dissimilar elastic properties, it is likely that some of these microcracks would open under global compression, especially those oriented along the load direction. In a multidirectional composite laminate with continuous fibre plies, cracks with an effective opening under compression could be incipient delamination on the ply level and large fibre-matrix debonds on the microscopic level. For notched composite specimens, it has been reported that cracks grow parallel or perpendicular to the load direction depending on whether the load is on the compressive or tensile side, respectively [9].

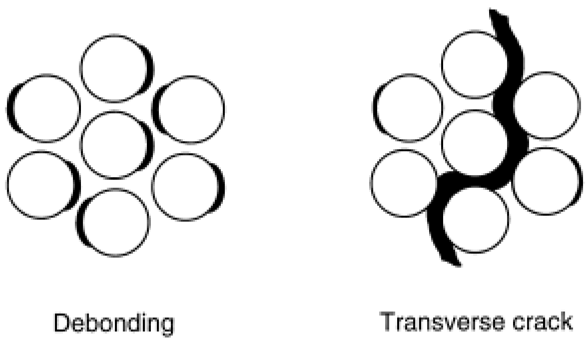

FIGURE 1. Drawing of the formation of debonds and subsequent link-up and transverse cracking (under uniaxial transverse horizontal loading in the figure).

The formation of transverse cracks can be initiated by microcracks which grow under compressive loading. It is not far-fetched to suggest that this type of microcracks would be interfacial debonds with a fairly large circumferential size. Debonding is known to be the micromechanism incipient to transverse cracking. As the debonds at the interface between the fibres and matrix coalesce and grow unstably, a transverse crack forms [10]. Even if it is assumed that transverse cracking is initiated in the matrix through yielding or cavitation, it has been shown that this will take place infinitesimally close to the fibre-matrix interface for uniaxial transverse loading [11]. This crack initiation will subsequently immediately cause a partial debond.

Experimental procedures

The matrix was a flexible epoxy vinyl-ester (VE) Norpol Cor Ve 8515 from Jotun Polymer AS, with a styrene content of 32-36 wt%. The glass fibre was a 2400 Tex fibre from Owens Corning of type R25H. The mean diameter of the fibre was 23 µm. The vinyl-ester resin was cured with 2wt% methylethylketone peroxide as catalyst and 2 wt% cobalt-octoate (1% Co) as accelerator. For injection of the resin, 0.2 wt% inhibitor NLC-10 was used to obtain a sufficiently long time to gelation.

The single-fibre transverse specimens were manufactured in a similar way to that described in a study by Zhang et al. [12]. The specimens were prepared by mounting the fibres on a 1 mm thick steel frame with double-sided adhesive tape. The frame was then placed between two flat Teflon-coated aluminium mould plates, separated by spacers of 2 mm thickness and provided with a silicon tube sealing. Positioning in the thickness direction was achieved by 1 mm thick steel slivers on the tape. After closing the mould, the resin was carefully poured into the vertically positioned mould and cured. To reduce the amount of entrapped air from the mixing operations, the resin was vacuum treated for a minimum of 5 minutes before injection or pouring into the moulds. Post-curing of the laminates and plates was performed for 2 h at 80°C.

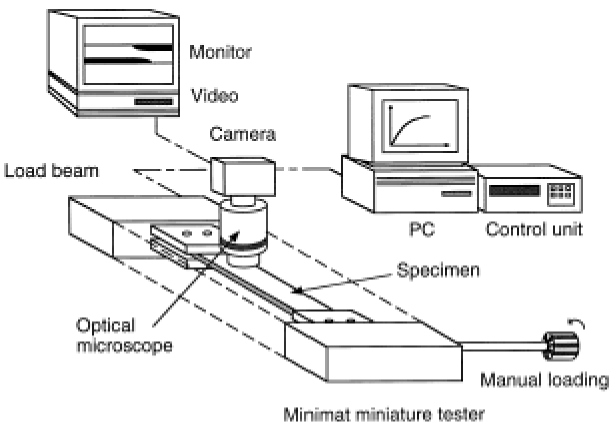

FIGURE 2. Schematic illustration of the experimental set-up for monitoring debond growth in a single transverse fibre specimen.

The single-fibre composite specimens were tested in a horizontal miniature mechanical tensile stage (Minimat) from Rheometric Scientific Ltd. The experimental set-up is schematically depicted in Fig. 2. The test machine was mounted on the *x*-*y* table of a microscope. Interfacial debonding between fibre and matrix was observed in situ by optical microscopy, where light in the transmission mode was used. The microscope was equipped with a video camera, as well as with a conventional photographic camera. The miniature tensile machine was manually controlled, but the load and the displacement were continuously recorded by and stored in a computer. The maximum stress was 50 MPa which is lower than the yield stress of the vinyl ester. Interfacial debonding was observed during step-wise loading. After each load cycle, the fibre-matrix debonds were photographed, video recorded and stored on a hard disk. After the tests, the debond size and geometry were quantified by an image analysis program.

Results and discussion

In Fig. 3, schematic drawings of the half plane of a single fibre composite with partial interfacial debonds are shown. The extent of the debond length can be expressed in terms of the total debond angle $\theta = \theta_o + \theta_c$, which is the sum of the partial angle of debond opening and closure, respectively. For a sufficiently large debond in conventional material systems with a stiffer fibre in a more compliant matrix, a contact zone will develop at the debond tip under tensile loading, and an opening zone will form under compressive loading [13].

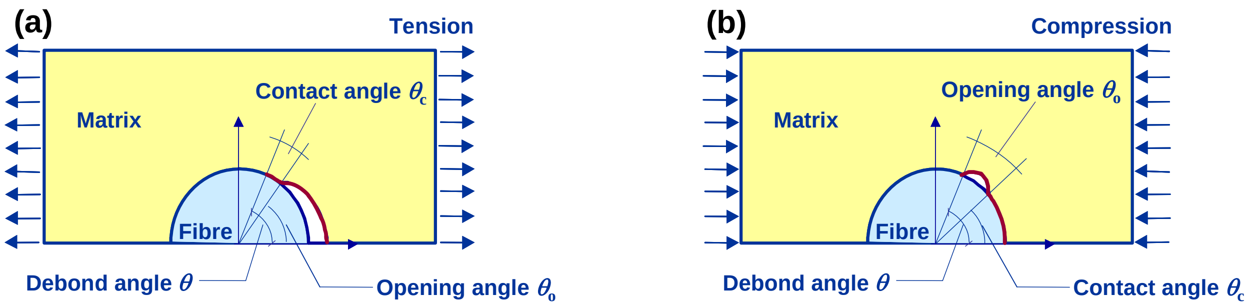

FIGURE 3. Contact and opening zones for large transverse debonds subject to (a) tensile loading and (b) compressive loading.

A schematic illustration of the experimental set-up is shown in Fig. 4(a). By transmission optical microscopy the debond zones could be measured intermittently, Fig. 4(b), and plotted during low-cycle fatigue, Fig. 4(c). Details are explained in Ref. 1. The propensity to debond growth under fully reversed cyclic loading as compared to purely tensile loading can be understood by the crack tip opening and closure phenomena portrayed in Figs. 3(a) and (b).

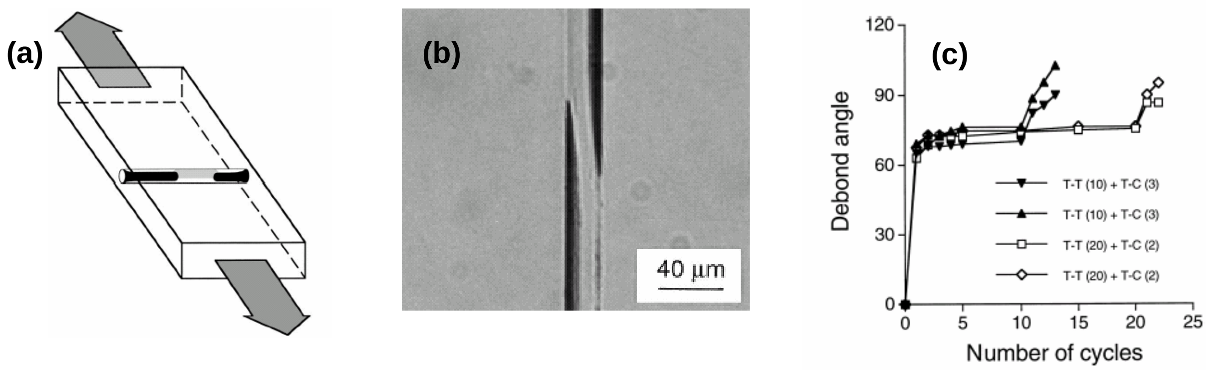

FIGURE 4. (a) Single-fibre composite test, wherein the extent of transverse debonding is measured by (b) transmission optical microscopy, and (c) plotted during cyclic loading.

A boundary element model (cf. Ref. 14) was used to determine the mixed-mode energy release rates for the tested material configurations. A crack-closure technique is used, where the debond propagates from a certain angle α to $\alpha + \delta$, where $\delta << \alpha$. The energy release rate is expressed as [15]

$$G_\delta(\alpha) = \frac{1}{2\delta}\int_{\alpha}^{\alpha+\delta}\left\{(\sigma_{rr})_\alpha (u_r)_{\alpha+\delta} + (\sigma_{r\theta})_\alpha (u_\theta)_{\alpha+\delta}\right\} d\theta \tag{1}$$

where σ_{rr} and $\sigma_{r\theta}$ represent radial and shear stresses along the interface, respectively, and u_r and u_θ their associated displacements. The two modes of fracture, I (associated to σ_{rr}) and II (associated to $\sigma_{r\theta}$), are obviously considered in this expression.

The results are shown in Fig. 5(a) for a tensile unit load and in Fig. 5(b) for a compressive unit load. From the growth data and energy release rate calculations, a propagation law can be determined. The growth parameters in such laws can be interpreted as interfacial properties, and ultimately used for comparisons of various fibre coatings to identify procedures in fibre surface treatment and manufacturing of composites for improved fatigue resistance. These parameters can also be used in simulations to predict transverse cracking in fatigue loading.

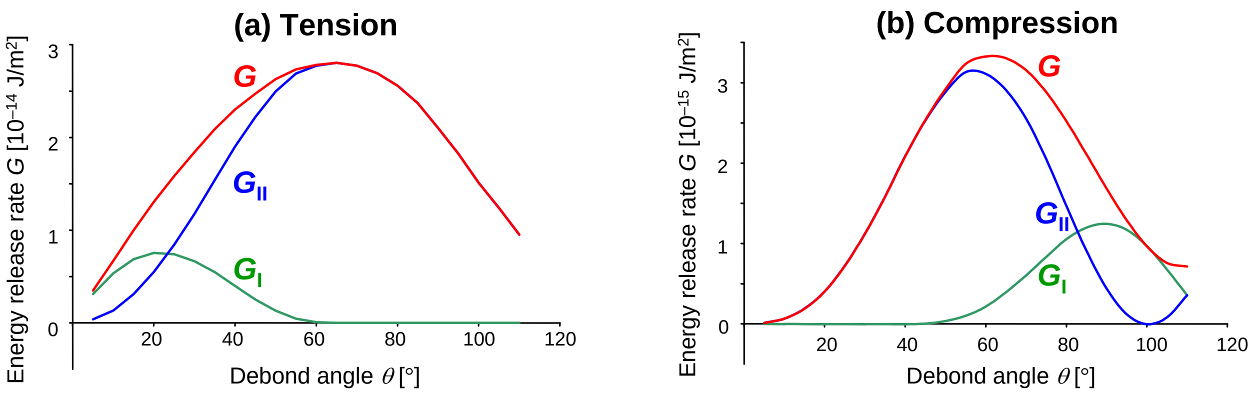

FIGURE 5. Energy release rate with mode I and II contributions with a far-field stress of 1 Pa in (a) tension and (b) compression.

As mentioned above, the next natural step to take would be to use the calculated energy release rates together with the experimental data, as those shown in Fig. 4(b), and determine numerical values of propagation parameters in a mixed-mode Paris-type of growth equation. These parameters could e.g. be used to rank different material combinations, interfacial coatings, processing procedures etc. with regard to the resistance to interfacial fatigue crack growth. The parameters could also be used as input in more complex models to predict failure of transverse plies, as individual debonds kink out into the matrix and coalesce to form contiguous and catastrophic cracks, as shown in Fig. 1. Such detailed analysis would require information on the spatial distribution of fibres [16], and criteria for debond coalescence [17].

There is a plethora of different formulations of fatigue growth laws for mixed-mode conditions. A summary of some of them has been presented in a table by Blanco *et al.* [18]. One straightforward example is to use

$$\frac{\mathrm{d}a}{\mathrm{d}N} = D\left[\Delta G(a)\right]^r \tag{2}$$

where the debond length *a* may be integrated. Both the coefficient *D* and exponent *r* can be considered as material parameters, and depend on the mode mixity, G_{II}/G_{tot}, where G_{tot} is the total energy release rate, i.e. $G_{tot} = G_I + G_{II}$. The mode mixity, G_{II}/G_{tot}, also depends on the crack length *a*, or debond angle α. A monotonic linear function between the parameters and the mode mixity can be assumed as a first approximation [17]. In some cases, this might not be enough, and a hyperbolic non-monotonic function can be employed [18]. Once the parameters have been identified by integration of the differential equation (2) and fitting to experimental growth data, these values can be tabulated for comparison of e.g. different fibre surface treatments aiming for enhanced resistance to fatigue degradation and failure.

Conclusions

Mechanisms have been identified that accounts for the inferior fatigue behaviour in tension-compression compared with tension-tension of laminates containing transverse plies. The mechanism is debond crack propagation under compression for sufficiently large debond angles. Crack opening zones form at the crack tips under global compression. The interfacial crack can now propagate more easily than if there were contact zones at the crack tips for equal tensile far-field loading. The validity of this mechanism hypothesis has been investigated by a number of experiments. On the macroscopic level, fatigue testing of multidirectional composite laminates shows enhanced transverse crack formation in tension-compression loading and shorter fatigue lives. On the microscopic level, model specimens with single transverse fibres were tested to examine the debond growth under low cycle tension-compression fatigue loading. It was shown that debond propagation occurred also during the compressive load excursion, due to the mismatch in elastic properties and curved crack along the interfacial debond. Boundary element calculations was used to determine the energy release rate in modes I and II. These may be used to quantify a growth law on the microlevel to predict onset of transverse cracking.

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