The intralaminar crack growth resistance of polymer-based composite materials

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To characterize the intralaminar fracture behavior of a unidirectional carbon fiber/epoxy resin composite material, two distinct sets of experiments were performed. First, the tensile mechanical properties of the lamina were determined by testing specimens with different fiber orientation at varying rates of deformation. Then, the double-torsion technique was applied to characterize the fracture resistance of the material. The strain energy release rate vs. crack speed curve so derived is in agreement with measurements conducted by using the Double Cantilever Beam and Compact Tension techniques.

INTRODUCTION

The fracture processes in fiber composite materials have been extensively studied in the past 20 years, following different approaches: yet, the development of a unifying theory to account for crack orientation, loading conditions, microscale damage, temperature, moisture, and other environmental parameters presents a formidable challenge. When composite matrices are polymeric materials, which display viscoelastic behaviour even under normal service conditions, it is also necessary to account for the time-dependent characteristic of the fracture process. The present work will show the applicability of Double Torsion (DT) testing for characterizing intralaminar fracture resistance of unidirectional laminates.

EXPERIMENTAL DETAILS

Unidirectional composite laminates were prepared from ready-made prepregs of unidirectional, continuous carbon fibers embedded in an epoxy resin matrix (F-155 Hexcel). Tensile specimens were obtained from 8, 16, and 24-ply laminates (ply thickness = .15 mm). The ASTM recommendation D-3039 was used as the guideline for specimen design. Specimens were instrumented with resistance strain-gauges bonded on both sides, so as to monitor the actual deformations during the tests.

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Fracture specimens were obtained from 24 and 32-ply laminates. ASTM recommendation E-399 was adopted for Compact Tension (CT) specimens. DT specimens of rectangular (64x152 mm) form, with fibers oriented parallel to the longer axis of the specimen, were edge notched at a 40 mm depth and V-grooved along the bottom face to a depth of 15% of the specimen's thickness, along the centerline path parallel to the longer side of the specimen. The starting notch was sharpened by pressing a razor blade into the material, so as to generate a natural crack in it. For testing, specimens were supported on two parallel rollers spaced 49 mm apart and loaded by pressing two small hemispheres symmetrically located over the notch,15 mm apart. All tests were conducted at constant displacement rate.

RESULTS

Basic Material Properties. In this work, the following material properties were measured in simple uniaxial tensile tests: E_1, E_2, ν_{12}, ν_{21}, G_{12}. In this notation, as usual, E indicates Young's modulus, ν Poisson's ratio, and G shear modulus, 1 denoting the fiber direction and 2 the in-plane transverse direction. To assess the time-dependence of mechanical properties, tensile tests at varying rates of deformation were performed. Tab. 1 shows the values of the initial tangent moduli (i.e., measured at the very beginning of the stress-strain plot) obtained from 0° and 90° tests. Neither one of the two moduli exhibited significant variation with the strain rate applied. The secant modulus for the 90° laminate, however, decreased as testing time increased, as expected for any viscoelastic polymeric material. Poisson's ratios were also calculated, based on transverse strain measurements: the values obtained were unaffected by strain rate, and the average values are also shown in Tab. 1. Shear modulus was finally measured on 10° off-axis specimens, directly from strain gauge readings. In Fig. 1 shear moduli values vs. applied shear strain rate are plotted: a minimum was found within the range explored, probably indicating that microscale damage also occurs during loading, which would compete with viscoelastic load relaxation.

DT test analysis. If unidirectional fibers are oriented along the specimen axis and crack propagation takes place along the same axis, conventional LEFM analysis can be applied [1] and the linear elastic strain energy release rate, G_{IC}, for transverse fracture is given by:

$$G_{IC} = \frac{P_c^2}{2B_c}\left(\frac{dC}{da}\right) \tag{1}$$

where P_c is the critical load at fracture, B_c is the specimen thickness at groove

TABLE 1

Property	Average value	Standard deviation
E_1	122.5 GPa	3.0
E_2	7.9 GPa	0.2
ν_{12}	0.34	0.01
ν_{21}	0.025	0.001

and (dC/da) is the derivative of the specimen compliance, C, with respect to the nominal crack length, a. For linear elastic and isotropic materials, the DT specimen compliance is a linear function of crack length [2].Then, if the test is carried out at constant cross-head displacement rate, $\dot{x}$, and P_c is constant during the test, the crack speed, $\dot{a}$, is constant too. Its expression is given by:

$$\dot{a} = \frac{\dot{x}}{P_c \left(\frac{dC}{da}\right)} \tag{2}$$

From both Eq.s 1 and 2, it appears possible to derive the $G_{IC}(\dot{a})$ characteristic of the material, with no need to measure a or $\dot{a}$ during the test.
For linear elastic orthotropic materials,the analytical expression for the DT specimen compliance can easily be derived, based on the analysis of the torsional behaviour of a rod, with a rectangular cross-section and sides parallel to the planes of elastic symmetry [3]:

$$C = \frac{h^2}{\beta G_{12} W B^3}\, a + C_o \tag{3}$$

where C_o is a constant, h is the distance between the loading points for a single beam, B and W are, respectively, the thickness and the width of the specimen, and β is a function of the aspect ratio of the rod section, and of material's torsional rigidities G_{13} and G_{12}. From the above equation, it turns out that specimen compliance is linearly dependent on crack length, as it is for isotropic materials, and the conventional DT analysis appears to be applicable to unidirectional composites with no substantial modifications.

Fracture results. Experimental measurements of specimen compliance, C, for different crack lengths, a, were performed to calibrate the test. Measurements on 24 ply specimens at varying displacement rates did not show rate dependence. Results obtained for 24 and 32 ply specimens are shown in Fig. 2.

The calculated* (via Eq. 3) and the experimental values of the compliance derivative, (dC/da), are given in Tab. 2: the agreement is not completely satisfactory, suggesting that direct experimental calibration of the test, according to the procedure described, should precede DT fracture testing of laminates.

TABLE 2

Type of laminate	**Experimental compliance derivative**	**Theoretical compliance derivative**	**Variance**
24 ply	$7.7\ 10^{-5}\ N^{-1}$	$8.5\ 10^{-5}\ N^{-1}$	9.4 %
32 ply	$5.1\ 10^{-5}\ N^{-1}$	$4.0\ 10^{-5}\ N^{-1}$	21 %

Fracture tests were then performed at varying displacement rates: as expected, a constant load was observed during crack propagation, which is a necessary condition for crack speed being constant during the test [2].The values of G_{IC} were also corrected, according to Leevers [4], to account for the crack front curvature. Fig. 3 shows the results obtained for 24 and 32-ply laminates (not distinguished on the plot), compared with interlaminar fracture results obtained on the same material by DCB tests. The fracture toughness values were expected to be different in the two geometries, because fracture occurs on a more resin-rich path in the interlaminar case; nevertheless, the two slopes agree closely, suggesting that the viscoelastic matrix is dominating crack propagation in both cases.

In order to ascertain the lack of geometry-dependence in the results obtained in DT testing, a series of CT tests, with fibers oriented perpendicular to the direction of the applied load, was carried out at different displacement rates. Since CT configuration leads, typically, to unstable crack propagation, the more general form:

$$\dot{a} = \frac{\dot{x}}{P_c \dfrac{dC}{da} + C \dfrac{dP_c}{da}} \tag{4}$$

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* Based on some pure-torsion tests, with fibers oriented along the axis of the applied torque, a ratio $G_{13}/G_{12} = 1.118$ was assumed for calculations.

has to be used in place of Eq. 2. As for the strain energy release rate, G_{IC}, it can be derived from the stress intensity factor, K_{IC} , calculated at the load maximum, by using the following expression [1]:

$$G_{IC} = K_{IC}^2 \left(\frac{1}{2E_1E_2}\right)^{\frac{1}{2}} \left\{ \left(\frac{E_1}{E_2}\right)^{\frac{1}{2}} + \frac{1}{2}\left[\frac{1}{G_{12}} - 2\left(\frac{\nu_{21}}{E_2}\right)\right] E_1 \right\}^{\frac{1}{2}} \qquad (5)$$

By assuming the fracture onset to be coincident with the maximum of the load-displacement curve, the values of G_{IC} and $\dot{a}$ were determined from each test.

The experimental results are shown in Fig. 4, where the points obtained in DT tests are also reported, for comparison. The magnitude of the two sets of data appears in fairly close agreement, although no clear rate-dependence can be observed in CT results. The latter outcome remains open to further investigation.

CONCLUSIONS

The Double Torsion test has been found to be applicable to characterize the time dependence of intralaminar fracture resistance in unidirectional composite laminates. The strain energy release rate vs. crack speed correlation has shown the same dependence as the interlaminar fracture data obtained in the DCB test. Measurements of intralaminar fracture resistance on CT specimens validate the DT results in terms of magnitude, although rate effects do not appear on the crack speed range explored.

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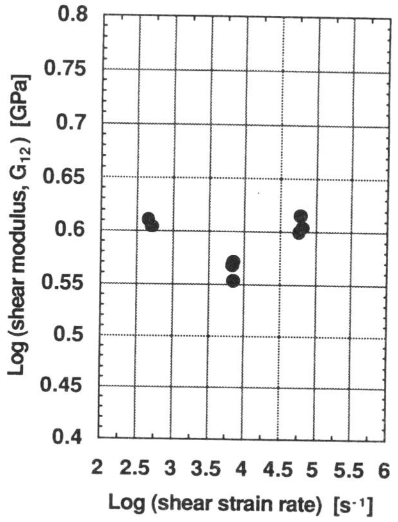

Figure 1. Strain rate effect for the in-plane shear modulus, G_{12}.

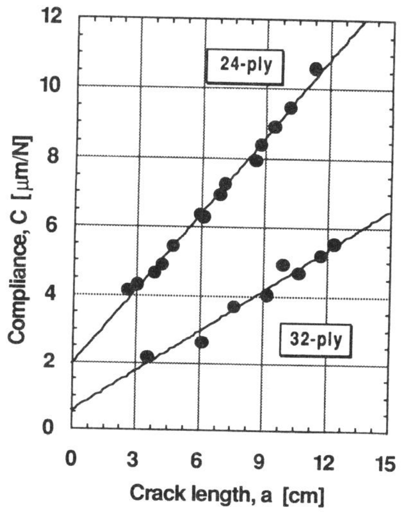

Figure 2. Experimental DT compliance vs.crack length for 24 and 32-ply laminates.

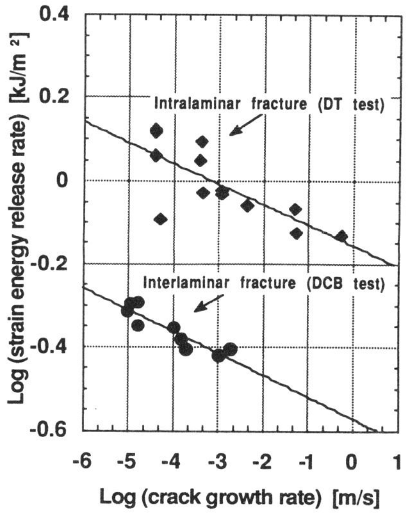

Figure 3. Strain energy release rate vs. crack speed for intralaminar (DT) and interlaminar (DCB, Ref.[5]) fracture.

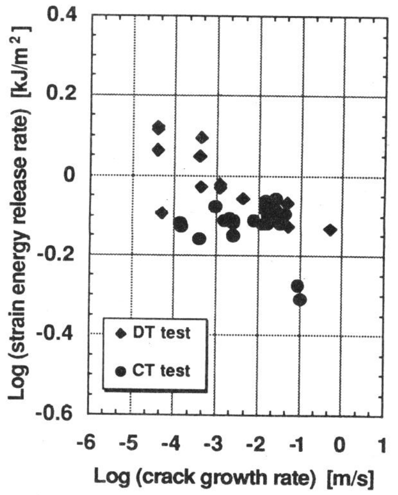

Figure 4. Strain energy release rate vs. crack speed results obtained in DT and CT testing geometries.