TOUGHENING MECHANISMS IN C/C MINICOMPOSITES WITH INTERFACE CONTROL

Beatrice Trouvat, Xavier Bourrat and Roger Naslain
Université de Bordeaux (UMR47 CNRS-SEP-UB1)
Laboratoire des Composites Thermostructuraux 3 Allée La Boëtie 33600 Pessac, France

Introduction.

The objective of this work was to improve the mechanical performances of C/C composites by controlling the fiber/matrix interface in unidirectional model composites.

A strong interface is known to lead to a brittle tensile behavior, whereas a weak interface permits fiber/matrix debonding and sliding. A non-linear stress-strain curve results, and final failure occurs by fiber pull-out. Toughness is achieved by a debond/sliding mechanism [1-8]. These points have been established for some time, but the way to achieve an optimum weak interface in C/C composites remained unclear (i.e. either by heat treating or by gently oxidizing the composite).

Experimental

Three main parameters were assumed to control the interfacial mechanisms : (i) fiber surface state, (ii) residual thermal stresses related to the radial fiber/matrix expansion mismatching and (iii) chemical bonding strength. The first parameter was evaluated by using the same HT PAN-based fiber with two different surface treatments. A surface effect does exist but is not determinant and will not be further commented upon. For the second parameter, fibers were used either as received or after a stabilization annealing pretreament in order to control the **thermal residual stresses**. In that case, transverse coefficient of thermal expansion is higher and the radial residual stress more in tension in the interface. Finally, in order to control the fiber/matrix bonding strength, a carbon **interphase** (or interlayer) with different anisotropy is deposited (CVD) in-between the fiber and the matrix. The lower the interphase anisotropy, the higher the interfacial bonding strength.

By changing the fiber pretreatment and the interphase anisotropy it was possible to switch from brittle to tough behavior and further to point out the existence of two different damaging modes depending on the interlaminar shear strength magnitude.

Results

A strong bonding between the pyrocarbon matrix (smooth laminar pyrocarbon) and the as-received fibers was observed. Those materials failed in a brittle behavior. The tensile load-strain curve is linear up to failure, as shown in figure 1a. Because of the strong interface, the first matrix crack produces a notch effect on the fiber, leading to the catastrophic failure of the material; the fracture surface is flat (SEM micrography Fig. 1a).

In contrast, a weak fiber/matrix bonding is obtained by first thermally stabilizing the fiber. Rearrangement of carbon planes along the fiber axis with HTT, leads to a higher transverse thermal expansion coefficient in stabilized fibers. As a result, the fiber/matrix interface is in higher transverse residual tension : the interface is weak and the tensile tests show a non-brittle tensile behavior (figure 1, curve b). Observations on failed samples show clearly the multiple-matrix-cracking which is responsible for the non-linear behavior. Polished sections show that matrix cracks are deflected at fiber/matrix interface. The tensile curve is characterized by a first linear part followed by a non-linear domain and then a tertiary linear region. Matrix cracking causes the non-linear (plateau) feature with a final high crack spacing (376μm) and large residual crack opening (1.6μm). It results in a low interfacial shear stress τ (1MPa). Toughness is based on the debond/sliding mechanism which is known in CMCs with weak interface and fiber pull-out (SEM Fig.1b).

Addition of a CVD interphase, with a microtexture more disordered than that of the matrix, still leads to a tough behavior (Fig. 1, curve c) but with a different damage mode. Crack deflection occurs at interphase/matrix interface, and the associated interfacial shear stress is higher (τ=6.8MPa). Unlike the case with heat-treated fibers and no interphase (curve b), the tensile curve is characterized by a progressive lowering of the slope after the linearity which means that the matrix remains partially loaded up to failure in this alternative mechanism. For the same elongation, this composite sustains a higher load even if the matrix is one third the stiffness of the fiber (figure 1, curve c). Matrix damaging is characterized by a lower crack spacing (251μm), and a much smaller residual crack opening (0.355μm, Fig. 1, c'). This mechanical behavior should be controlled by the reinforcement but is comparable to that obtained with SiC/SiC (at a much higher τ with SiC matrix). In SiC/SiC, crack deflection occurs in the bulk of the interphase by branching [10], and toughness is due to a matrix-multiple-cracking-based mechanism. Contrary to SiC/SiC, the shift between the two mechanisms in C/C minicomposites covers a much smaller range of interfacial shear strength. The apparently change in mechanical behavior, induced by the interphase, is attributed to an enhanced chemical bonding combined with reduced radial, thermal residual stresses in tension.

Conclusions

By heat treating a HT PAN-based carbon fiber or by adding a controlled-microtexture carbon interphase, toughness of unidirectional C/C minicomposites was successfully controlled with similar energy dissipation processes as for other CMCs : debonding/sliding mechanism

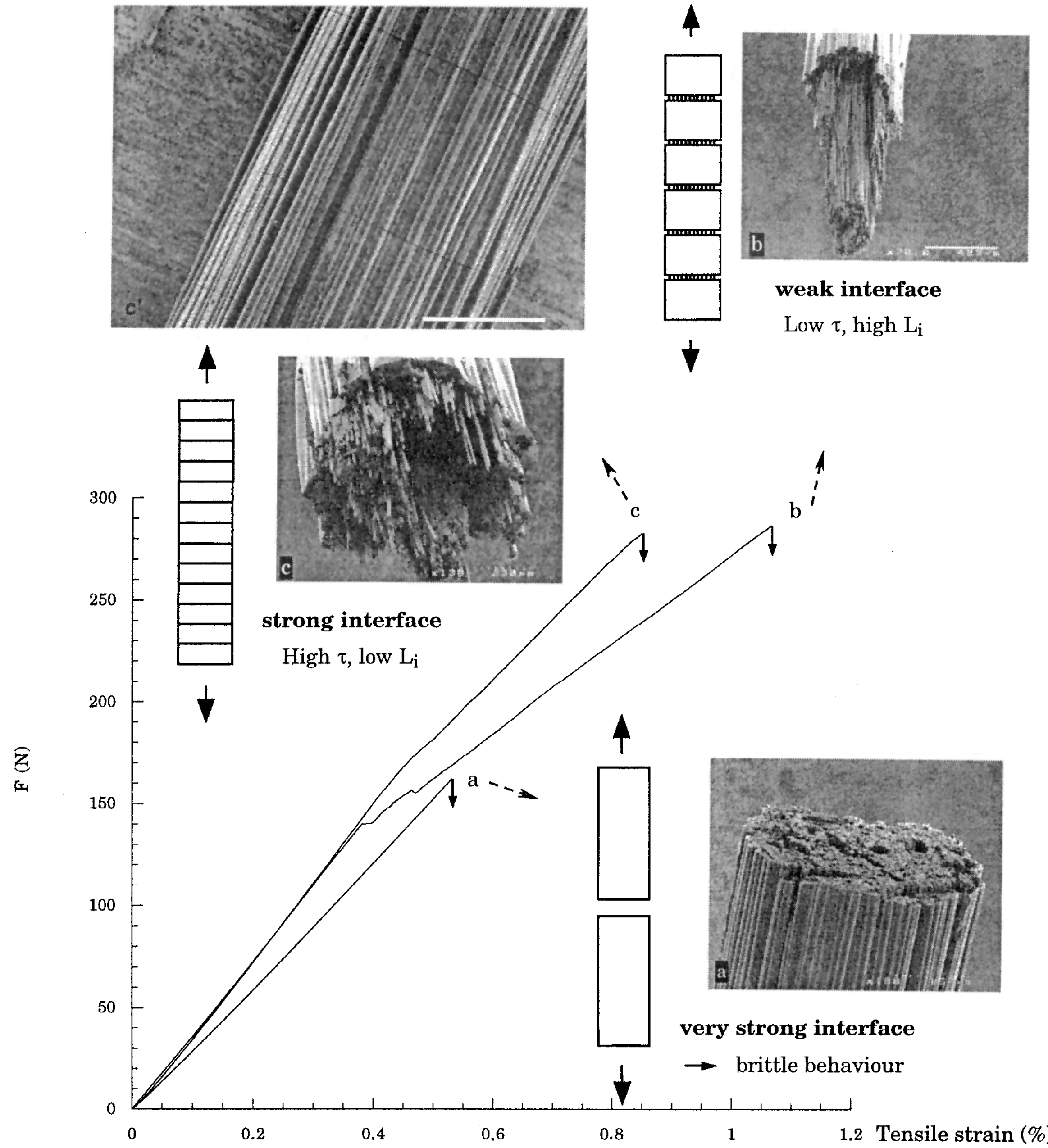

Figure 1: a,b and c : 3 typical tensile curves obtained on UD C/C composites according to interface strength, and schematic of the damaging mode observed on each tensile test specimen.
(SEM micrographies represent in each case a typical fracture surface, c' shows the crack-spacing)

for low τ, and additionally a multiple-matrix-cracking-dissipative mechanism for higher τ.

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