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# High strain rate testing of a unidirectionally reinforced graphite epoxy composite

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Since accurate, reproducible methods of testing polymer composites are not very well developed or standardized, this research forms part of a program to gain a better understanding of the mechanical properties and failure mechanisms of polymer composites at high strain rates. Since failure modes differ markedly depending on fiber architecture, orientation, fiber/matrix combination and so forth, these initial tests were carried out using a simple unidirectionally reinforced composite. Beginning with testing in the longitudinal and transverse directions, reported here, future experiments are being carried out to determine how the high strain rate properties vary with angle of testing, and then move on to other simple fiber lay-ups, ±90°, ±45°, etc.

Plates 40 mm × 40 mm × 10.75 mm, consisting of 75 plies of unidirectional IM7/8551-7 graphite-epoxy, were fabricated by hand-lay-up, vacuum bagging and curing. The graphite fibers have a tensile modulus of 303 GPa and tensile strength of 5520 MPa. Cylindrical specimens 10.7 mm long and 8.3 mm in diameter were core drilled from the finished plates. Compression tests were conducted in the longitudinal direction (1) and the two transverse directions, namely, perpendicular to the plate surface (2) and parallel to it (3). High strain rate tests were conducted using a compression Split Hopkinson Pressure Bar (SHPB) with 19 mm diameter Inconel-718 bars. Details of the specific SHPB technique and its data reduction method are given elsewhere [1]. Two different striker bars were used with stress pulse time-windows of 150 and 290 $\mu$s.

A major concern for composites is the time for stress homogenization in the specimen. Ravichandran and Subhash [2] showed that, for isotropic high impedance ceramic materials, stress homogenization is achieved after approximately $4t_o$ where $t_o$ is the transit time for the leading edge of the pulse traveling through the specimen. Li and Lambros [3] showed that, even for non-axisymmetric loading of unidirectionally reinforced composites, stress homogenization is still achieved after 3 or 4 reflections. In order to be even more certain of stress homogenization in the present work, some of the experiments were performed using pulse shapers to produce a gradually increasing compressive stress pulse. In addition, some samples were strain gaged in order to compare the stress/strain curves with those produced by conventional SHPB data reduction routines [4]. Fig. 1 shows comparison stress/strain curves from such a transverse sample and it is seen that the two curves are essentially identical up to the point where the strain gage detached; this gives great confidence in the accuracy of subsequent results.

Another problem of SHPB testing is crushing of the tested sample by further movement of the bars after failure. Instead of using a momentum trap to prevent crushing after failure, in the present work loosely-fitting high strength steel collars were placed around the samples. The collars were prepared ∼5–7% shorter than the sample and permitted recovery of tested material by preventing further deformation after fracture. Fig. 2a and b show the raw and reduced data respectively, clearly showing where the sample failed and where the bars subsequently came into contact with the steel collar to prevent further crushing or damage to the composite.

Fig. 3 shows that, in the transverse direction, the fracture strength increased noticeably from ∼215 MPa at quasi-static strain rate to an approximately constant value of ∼360 MPa at high strain rate. The fracture strain was almost constant at $5 \pm 0.3\%$, and no significant change was noted in Young's modulus over the strain rate range investigated. In the longitudinal direction there was considerably more scatter in the data, Fig. 4, but only a slight indication of a strain rate dependence as the fracture strength increased from ∼850 MPa to ∼950 MPa. The variability in the data for the longitudinal data is ascribed [5] to the effect of slight misalignment of the fibers in and between the prepreg sheets. The modulus showed a slight decline from ∼90 GPa down to ∼75 GPa over the strain rate range

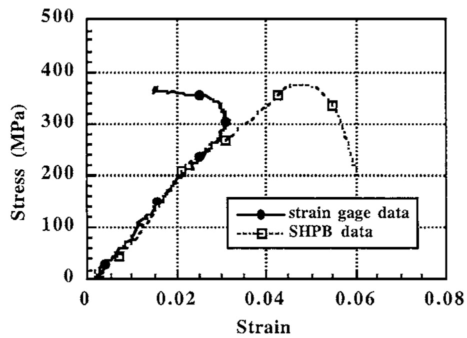
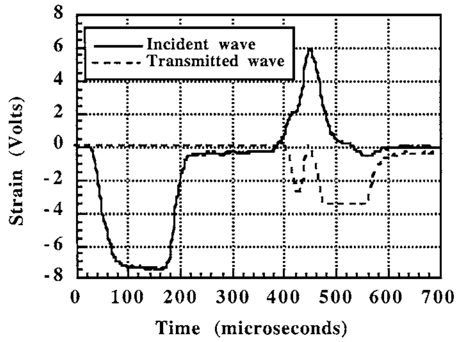
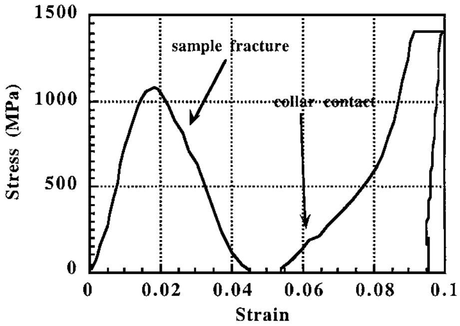



*Figure 1* Stress/strain curves from transverse sample generated from i) SHPB data alone and ii) strain gage data showing close correspondence between the two.




(a)




(b)

*Figure 2* (a) Raw and (b) reduced data from SHPB showing the effect of collars in ensuring preservation of fractured material.

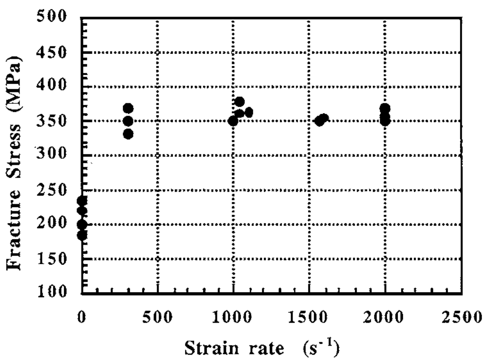



*Figure 3* Fracture strength vs. strain rate for transverse samples.

but the fracture strain remained constant at $2 \pm 0.3\%$. The decline in modulus is unusual insofar as increases in moduli have been reported elsewhere [6]. The reasons for this discrepancy are still under investigation.

Fractographic examination of transverse samples showed that failure occurred by shear on planes at an

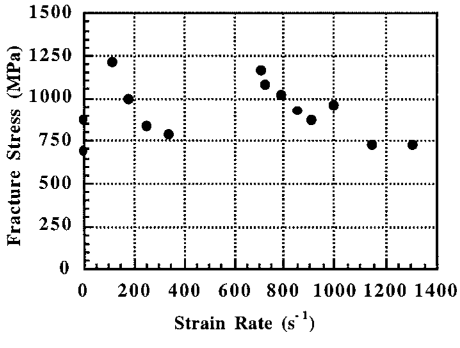
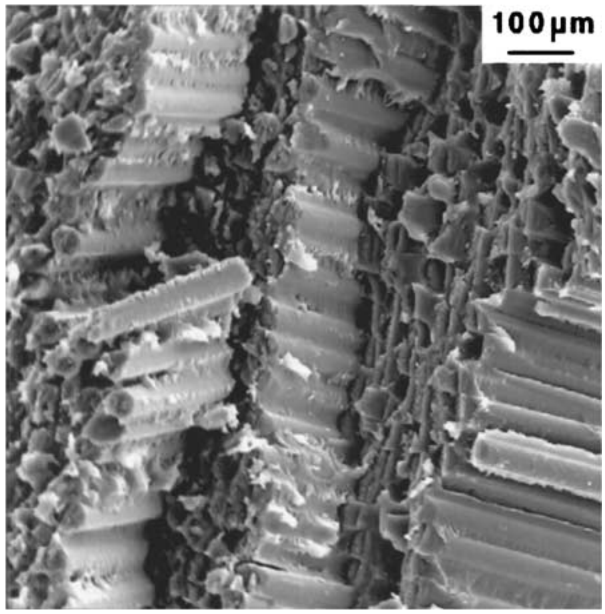
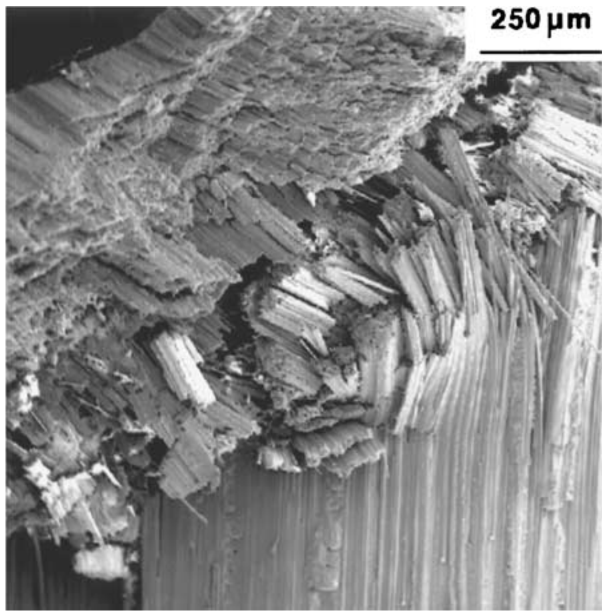



*Figure 4* Fracture strength vs. strain rate for longitudinal samples.




(a)




(b)

*Figure 5* (a) Formation of kink band on fracture surface of longitudinal sample tested at $\sim 1500\ s^{-1}$, and (b) initiation of longitudinal splitting following kink band formation.

angle to the compression axis, forming two major fragments at quasi-static strain rate and considerably more smaller fragments at high strain rates. In the longitudinal direction, failure was by microbuckling that led to kinking, Fig. 5a, and ultimately longitudinal splitting occurred, Fig. 5b. The thickness of the kink bands decreased with increasing strain rates but their frequency increased and samples split into many slivers. Interply interfaces, where the ratio of matrix : fiber is locally increased, were clearly visible on these surfaces, Fig. 6.

The results indicate strong strain rate dependence of the strength properties in the transverse direction but no similar dependence longitudinally and the reasons for this behavior may be deduced from the fracture surfaces. Since the only significant event occurring in the transverse direction is failure of the epoxy matrix, this

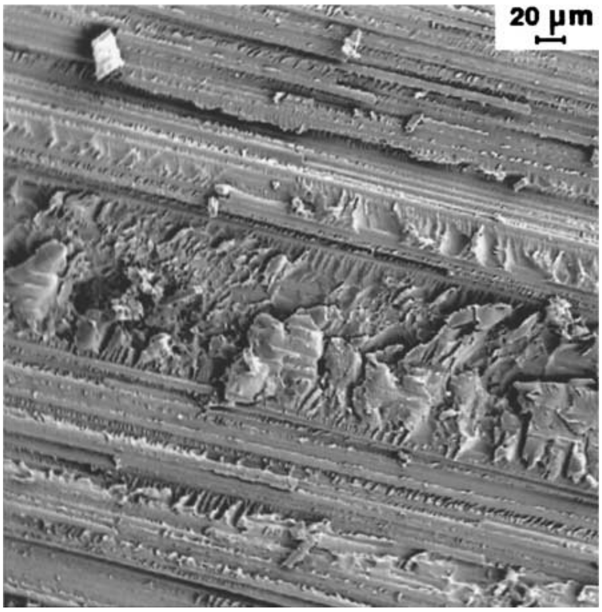



*Figure 6* Surface of split fragment of longitudinal sample fractured at $\sim$1000 $s^{-1}$ showing inter-ply interface.

must be the source of the strain rate sensitivity noted. Longitudinally, the fibers themselves undergo extensive buckling, followed by larger-scale kinking as a precursor to failure of the fiber/matrix interface. Buckling and kinking require, of course, failure at the fiber/matrix interface but the stress required must be of the order of that required to cause failure at this location in transverse samples, i.e. $\sim$360 MPa. Since this is well below the intrinsic compressive strength of the fiber bundles and it does not contribute to the strain rate sensitivity in this case.

It is concluded that high strain rate mechanical properties of polymer matrix composites can be reliably and reproducibly measured using the SHPB provided that care is taken to ensure stress homogenization. Furthermore, this particular composite exhibits only strain rate sensitivity of the fracture strength, and this only in the transverse direction.

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