these tests, which were conducted with a specimen pre-deformed in compression at room temperature by about 2%. This specimen had a gauge length of 5 mm only, a fact that emphasizes the high sensitivity of the measuring system.

The results show very clearly features which can be predicted and interpreted from simple arguments of microdeformation [8, 9]. For example, at low stresses the loops may appear closed or open depending, in a predictible manner, on the sense of the previous loading cycle. The loops in the tensile cycle appear open if this cycle has been preceded by one in the opposite sense (compression) whilst they appear closed if it has been preceded by one in the same sense (tension). It follows therefore that a loop in tension–compression cannot have a "butterfly" shape, a fact which was already anticipated by Brown [10] and is clearly shown in Fig. 3b.

The loops are not of purely lenticular shape and they exhibit clearly an initial region with higher slope. The value of this initial slope is 1.7×10^4 kg mm^{-2} or about half the dynamic Young's modulus of molybdenum crystals with this ⟨1 0 0⟩ axial orientation. A change in slope is observed at a stress of about 50 g mm^{-2}. The reason why this initial slope is less than the unrelaxed elastic modulus of the material must be found in the fact that no correction has been made for the elastic shear of the collets where the transducer is attached. In those cases where long specimens are used and where the transducer can be attached to the gauge length of the specimen itself, true values of Young's modulus can probably be obtained. In all other cases, including compression, correction by calibration is required [11]. It is however worth noting that such correction is not always essential because the shape and width of the loops and the residual strains at zero load are not affected by elastic deflection.

A systematic study of microstrain in molybdenum single crystals is now under way and the results reported here are intended only to show the potential of the experimental method described.

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Some factors controlling transverse cracking in cross-plied composites

When a resin based cross-plied fibrous composite is strained in tension beyond a (low) critical strain, a series of cracks form in those plies of fibres aligned with a substantial normal component to the applied stress. Although the cause has been identified as the strain concentration effects of the relatively stiff fibres [1, 2], the factors controlling the spacing of the cracks have not previously been defined. In this note we show that the relationship between stress and crack spacing can be explained in terms of shear-stress transfer from the adjacent longitudinal plies.

When orthogonal cross-plied laminate is stressed along one of the material axes, the strain increases linearly with the stress until the failure strain of the weakest section of the transverse ply is reached. The load carried by the transverse ply is obviously zero at the fracture and the total load must be carried by the intact longitudinal plies. The stress

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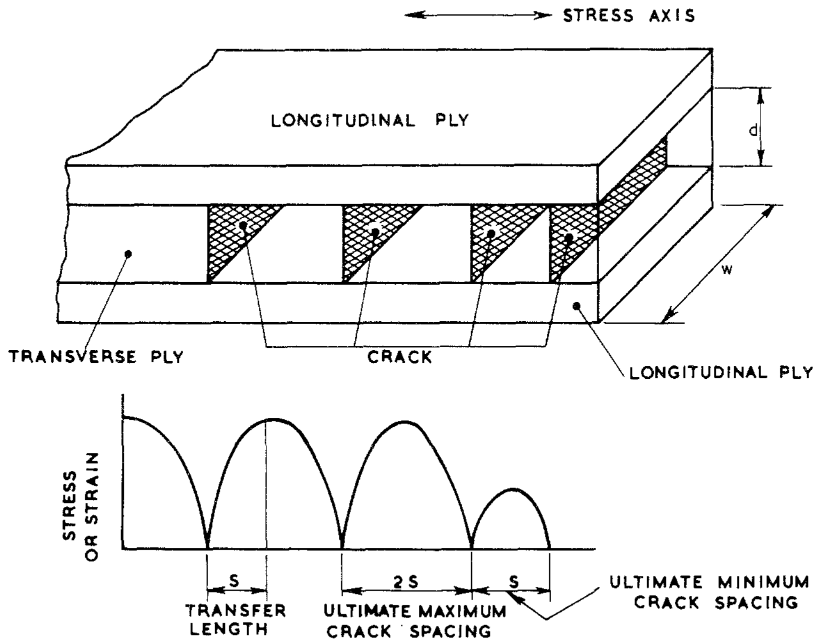

Figure 1 Schematic illustration of transverse cracking.

is transferred from the longitudinal plies by shear stresses and, as shown in Fig. 1, the stress in the transverse ply builds up over a short distance (the stress transfer length S) and approaches the pre-fracture level. The excess load in the intact plies decreases to zero over the same distance. Further straining takes place at reduced apparent modulus because the actual stress in the longitudinal plies around the fracture is higher than the nominal stress. Note that although there is a weakest section of the transverse ply, the numerous stress concentrators (i.e. the fibres) ensure that the strength of the transverse ply is relatively uniform on a macroscopic scale.

After the first crack, the transverse ply can be regarded as being made up of two regions. Immediately each side of the crack for length S, the strain is significantly reduced, while in the remainder of the ply the strain is reduced to slightly below the previous level. Each subsequent crack gives rise to a similar pair of regions (Fig. 1). Assuming as a first approximation, that the reduction in strain η, which occurs in that region of the transverse ply that is furthest from the crack, is inversely proportional to the length of those regions, we have

$$\eta = \frac{k}{L - 2NS} \tag{1}$$

where $k =$ a constant which includes the effects of specimen geometry, $L =$ gauge length and $N =$ number of cracks in gauge length.

The stress σ, necessary to bring the strain in these regions back to the fracture strain is then

$$\sigma = E\eta = \frac{kE}{L - 2NS} \tag{2}$$

where $E =$ elastic modulus of the composite and, on rearranging

$$N = \frac{L}{2S} - \frac{kE}{2S} \cdot \frac{1}{\sigma}. \tag{3}$$

Equation 3 predicts that the number of cracks per unit length is inversely related to the stress level. As indicated in Fig. 1, the average final crack spacing will lie between $2S$ and S.

The numbers of cracks observed in the 50.8 mm gauge length of a glass-fibre laminate test coupon are plotted against the reciprocal of the stress level in Fig. 2. Full experimental details are given elsewhere [3]. It can be seen that Equation 3 provides a satisfactory fit to the data. Furthermore, the simple theory outlined above can be applied to other systems e.g. the crack density data in "stress coat" studies [4].

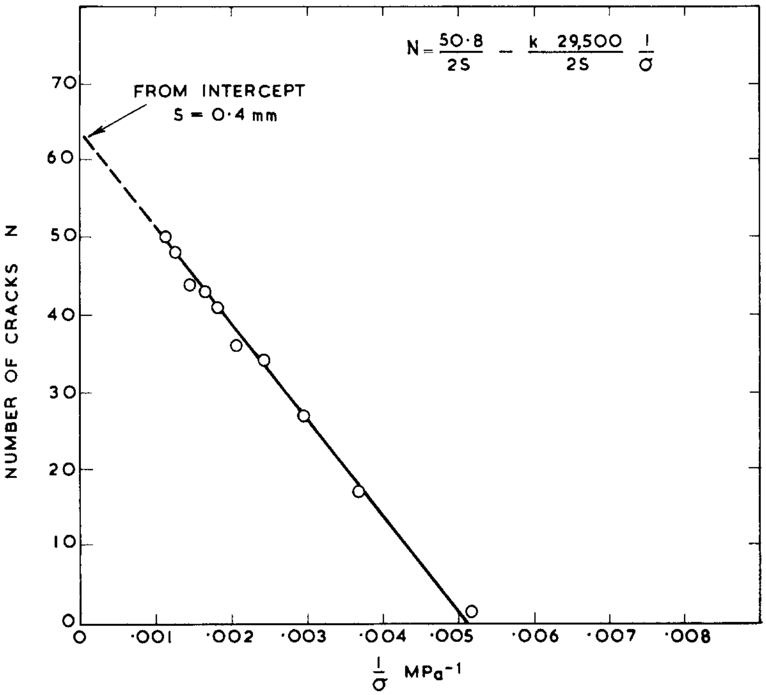

Figure 2 Relationship between number of cracks and stress level in cross-plied fibre glass/epoxide laminate [3].

The transfer length is related to material properties. The distance over which the load builds up in the transverse ply depends on the shear forces at the upper and lower ply interfaces. The maximum shear force is limited by the shear strength of the interface and the stress in the ply will reach its fracture stress at the transfer length, S.

Hence, with reference to Fig. 1, we have

$$dw\sigma_{tf} = 2Sw\tau$$

where d = transverse ply thickness, w = ply width, σ_{tf} = fracture stress of transverse ply, τ = interply shear strength. Therefore

$$S = \frac{d\sigma_{tf}}{2\tau} \tag{4}$$

Using experimental data [3], a value of $S = 0.5$ mm was obtained for the specimen involved in Fig. 2 (cf $S = 0.4$ mm via Equation 3).

These analyses indicate that the number of transverse cracks per unit length at a given stress level may be reduced by (i) increasing the transverse ply thickness, (ii) increasing the transverse ply fracture stress, (ii) increasing the elastic modulus of the laminate, and (iv) decreasing the interply shear strength.

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