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Transverse (interlaminar) cracking under tensile loading in pultruded CFRP

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The examination of microstructure of tensile specimens of pultruded 60% V_f carbon fibre-reinforced epoxide of up to 6 mm unreduced diameter shows that transverse cracking precedes the tensile failure of groups of fibres. In material whose strength is $\sim 2\,GN\,m^{-2}$, the process can commence in waisted specimens at stresses as low as $1\,GN\,m^{-2}$; in those of unreduced section it was not detected below $1.5\,GN\,m^{-2}$. This failure initiation stage can be associated with the decrease in the slope of the load–extension curve. With increasing load the inter-tow cracks were observed to grow and some surface fibre bundles detached. It is suggested that misaligned fibres in these surface bundles were straightened out and contributed to the load-carrying capacity of the rod. Only following detachment of numerous bundles (for the specimens with unreduced section) or growth of interlaminar cracks into the specimen shoulders (for those with a reduced gauge diameter) did tensile failure of fibre bundles lead to catastrophic fracture. It is to this last propagation stage that statistical models of failure of bundles at different cross-sections should refer.

1. Introduction

A major, if not the most important, failure mode in laminated composite materials is delamination between the composite layers [1]. It is recognized that delaminations may develop during manufacture or during loading and that they may grow during cyclic loading. If laminate theory, which assumes identical unidirectional laminae, is used for a balanced symmetric angle-ply laminate with $\theta = 0°$, it obviously predicts elastic constants of the unidirectional composite to be those of the laminae [2] and thus no interlaminar stresses. Hence transverse cracks in unidirectional composites tend to be associated with broken fibres [3] or sub-bundles containing only a few fibres [4] and, when such cracks are found in broken specimens, their origin is invariably interpreted as the final shear failures which link neighbouring bundle failures [3–14].

This assumption, made earlier also by one of us [10], should be questioned. Fibres in unidirectional composites are not straight and not well-aligned, particularly in hot-pressed laminates, as examination of Figs. 11 and 12 of [4], for example, clearly demonstrates. Consistent with this "metallographic" evidence, Fuwa *et al.* [4] found, on dissolving the epoxide matrix of their CFRP, carbon fibre-reinforced plastic, and testing the "fibre bundle specimens", some *slackness* in the fibre bundles. They reported that the rigid matrix of the cured material inhibits major fibre readjustments and thus that the strain–stress curves for these latter samples are linear from the origin. Further interesting observations on their intermediate volume fraction, V_f, of 0.35 composite and its fibres include the suggestion that fibres do not fail in isolation, but rather that a progressive failure mechanism such as related fibre breakage or crack propagation could be occurring or alternatively random fibre failure may be occurring throughout the sample and the ultimate failure of the specimens to be a statistically determined occurrence. They concluded, however, that although microscopic examination revealed crack growth in sub-bundles of fibres, this was limited, at least initially, by shearing between the sub-bundles and that eventual failure was attributable to the statistically determined

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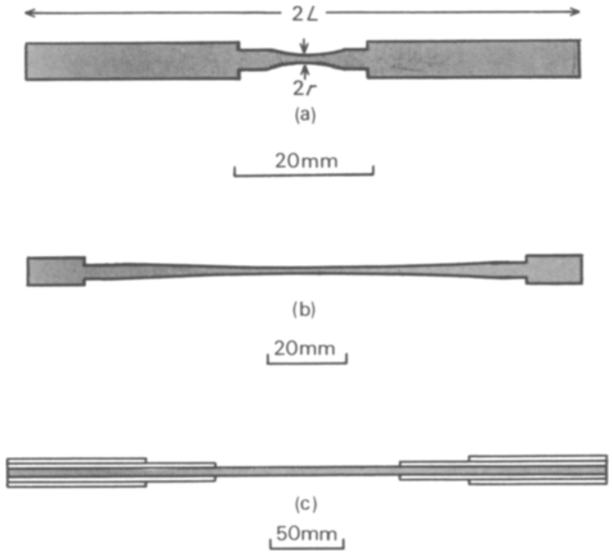

Figure 1 Round tensile specimen designs used in the investigation. Note that (a) and (b) are machined from 5 to 6 mm diameter CFRP rod and for (c) metal end tabs are adhesively bonded.

accumulation of fracture regions across a particular cross-section.

Transverse cracks in unidirectional CFRP tend to follow resin-rich regions between laminae or tows. These observations have been almost always made on broken specimens; microscopic examinations of loaded, unbroken samples have been few and the best-known, that of Rosen [3] was made on $\sim 6\%$ glass-reinforced plastic, GRP, hardly a high-strength composite. Nevertheless, a number of developments in the statistical theory of composite fracture rely substantially on his observations [5, 7, 9]. These, though very elegant, more realistically relate to a low V_f lamina (of only ~ 100 fibres).

Interlaminar cracks in high ($\sim 65\%$) V_f CFRP specimens tested in tension have been referred to in the context of "pull-through" the shoulders [11, 15] of the "gauge-section slug" or "central core". The relevance of those observations to simple tensile stressing may be questioned; it should be pointed out, however, that the shoulder radius of some of the specimens examined was in excess of 600 mm [15]. These latter results have not been reported previously in the refereed, scientific literature and will be described in Section 3 because of the similarities of observations with other specimen geometries, including one of no reduced composite gauge section.

The importance of "groups of fibres behaving in unison" has been recognized in the context of uniaxial compressive properties [16, 17] and, for example, our recent model considers bundle buckling [17]. The curvature of the groups of fibres could be caused by axial compression [17], or could already be present [16], as "slackness" reported by Fuwa *et al.* [4] indicates. It is suggested that this slackness, present before curing, and the resin-rich regions occurring between tows or laminae should be considered also when discussing uniaxial tensile properties. It should be added that in the literature "fibre bundle" has usually referred to *all* the fibres in a sample; our use of the term in this communication is for a microstructural entity: those fibres in a lamina or tow which are postulated by us to "act together" when the composite is loaded.

2. Experimental procedure

The experiments reported here were performed on samples machined from ~ 6 mm diameter pultruded rod produced by AERE, Harwell, Courtaulds or their associates. Fibres were carbon type A, V_f approximately 60% and the matrix epoxide. Tensile testing was performed on a range of specimen designs illustrated in Fig. 1. Samples were strained on Hedeby, Denison or Hayes (at RARDE, Fort Halstead) machines at rates close to 1 mm min^{-1} using split collets to grip the smaller diameter samples and wedge grips in combination with aluminium end tabs for those of ~ 6 mm diameter. Some of the specimens had miniature resistance strain gauges bonded to their gauge lengths. In one series of experiments linear variable differential transformers, LVDTs, monitored the displacement

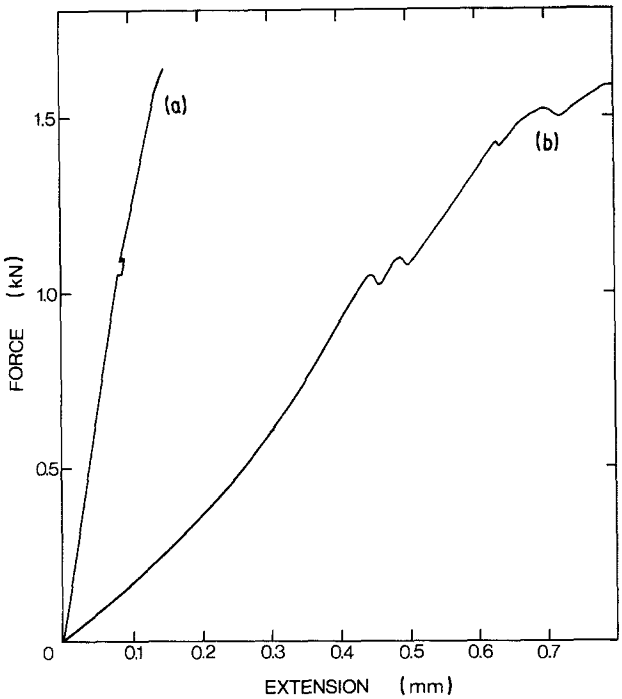

Figure 2 Initial parts of load–extension curves for a 10 mm gauge length specimen, of design similar to Fig. 1a. The extension was recorded using (a) strain gauges and (b) LVDTs between the collets holding the specimen. The specimen "yielded" at a stress of 1.34 GN m^{-2} and fractured at 2.09 GN m^{-2}.

between the specimen grips whilst resistance strain gauges measured the gauge length tensile strain.

Specimen gauge and failure surfaces were examined using scanning electron and optical microscopes. Some failed samples and others which were strained and unloaded prior to failure were machined parallel to the fibre axis, mounted in polyester resin and further sectioned and polished for examination by reflection optical microscopy.

3. Results

The tensile strength of the specimens, i.e. those which failed consequently to the fracture of the fibres, ranged from ~ 1.5 to above 2 GN m^{-2}. Fig. 2 presents for a type (a) specimen, which failed at a stress of ~ 2.1 GN m^{-2}, the (initial parts of) plots of both load–displacement, monitored by LVDTs between the specimen grips, and load–gauge length extension, measured by resistance strain gauges. Tensile failure of such samples occurred only when the length-to-diameter ($L/2r$ of Fig. 1) ratio exceeded ~ 40. For ratios below this critical value failure was by "pull-out" of the central core of the specimen, as illustrated in Fig. 3. It should be noted that the slugs that pulled through were not cylindrical, but approximated to parallelepipeds containing apparently a discrete number of tows. The apparent interlaminar shear strength, calculated on the prospective area of the slug at failure, evaluated to 12 ± 2 MN m^{-2} for a material which showed a ILSS of 61 ± 4 MN m^{-2} when measured in three-point bending.

The regions of pull-out within the specimen shoulders were delineated by "tramlines" on sectioned and polished samples that had failed by this "pull-through" mechanism (e.g. Fig. 4). Similar features, however, were also observed in sections of specimens that had apparently failed by a tensile mechanism, as shown in Fig. 5, but then the "tramlines" did not extend to the ends of the specimens' shoulders.

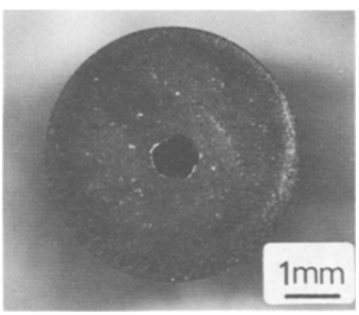

Figure 3 End-on view of the shoulder section of a Fig. 1a type specimen which, on tensile straining, failed through pull-out of a central "core" or "slug".

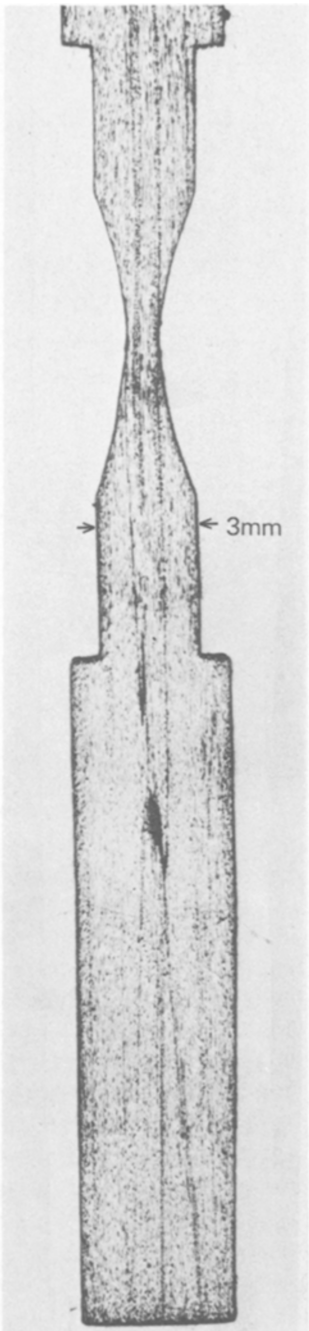

Figure 4 Longitudinal section of a specimen which was subjected to a tensile load but failed by transverse cracking, evidenced by "tram-lines" extending to the end of the unreduced section (composite micrograph).

Figure 5 Longitudinal section of a specimen which failed in a conventional tensile mode. Note the "tram-lines", terminating within the specimen shoulders (composite micrograph).

In most tensile tests the load–crosshead displacement curves exhibited a notable reduction in slope at a stress of $\sim 1\,\mathrm{GN\,m^{-2}}$, which was occasionally preceded by a drop in load. It would appear that this stress corresponds to the "elastic limit" detected when LVDTs monitored the displacement between the specimen grips and not that of the gauge length alone, see Fig. 2. It seems

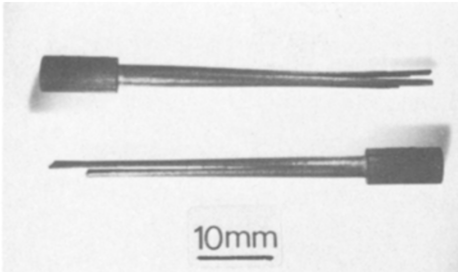

Figure 6 A failed pultruded CFRP tensile specimen of Fig. 1b design, i.e. with radius of curvature of the necked region of 613 mm. Note the transverse cracks.

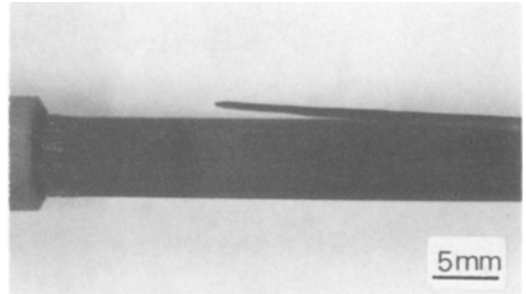

Figure 8 Detached surface bundle in a Fig. 1c type specimen loaded to a tensile stress of 1.74 GN m^{-2}.

that extensions so monitored, and also machine crosshead displacements, above stresses of ~ 1 GN m^{-2} include appreciable specimen deformations from outside the gauge lengths.

It has been argued that such behaviour may be associated with severe stress concentrators and numerous fibre ends (in the shoulder extremities) and accordingly some specimens were carefully machined with a very gradual change in profile, type (b) in Fig. 1, the radius of curvature, according to the recommendations of Ewins [18], being 613 mm. The tensile behaviour, including the mode of failure, of these specimens was essentially similar to that of specimens with much sharper shoulders, type (a). In particular, shear cracking was clearly visible (e.g. Fig. 6) and SEM examination confirmed it to be between the tows [15].

Tensile failure of type (c) (Fig. 4) specimens of (unreduced) 6 mm diameter was not achieved due to the failure of the adhesive bonds between the pultruded rod and the copper or aluminium tubes encapsulating the CFRP at loads of up to 53 kN, producing tensile stresses up to ~ 1.9 GN m^{-2} in the composite. Acoustic emission, monitored solely by the human ear, appeared to commence at stresses as low as 0.6 GN m^{-2} and interlaminar cracking at ~ 1.5 GN m^{-2}. Most of these cracks, but not all, e.g. Fig. 7, were associated with detachment of surface bundles, e.g. Fig. 8. Some of these unfractured specimens were sectioned and Fig. 9 clearly shows transverse cracking associated with a surface bundle. A random two-dimensional section will, of course, cut detached bundles at various diameters. Maximum values, however, postulated as bundle diameters, were up to 1 mm, consistent with the maximum thickness of completely detached surface bundles. These sections containing delaminated fibre bundles should be compared with similar sections of unloaded material, which shows (e.g. Fig. 10) resin-rich regions, but no inter-tow cracks. It should be added that in poor pultrusions non-bonded regions can exist, e.g. Fig. 11, and our work on pre-failure interlaminar cracking has concentrated on sound material in which transverse cracks result from the application of tensile loads. At stress levels of up to 85% of the composite strength (evaluated with smaller samples) individual fibre fractures were

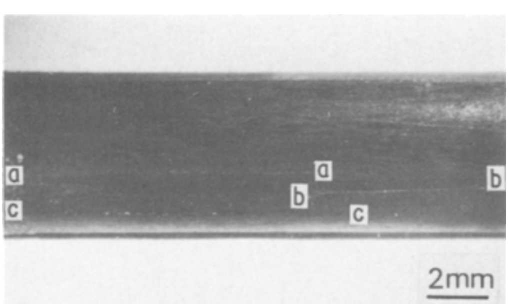

Figure 7 Transverse cracks, labelled aa, bb and cc, on the surface of a Fig. 1c type specimen loaded to a tensile stress of 1.6 GN m^{-2}. Note that the debonding is *not* associated with fibre bundle detachment.

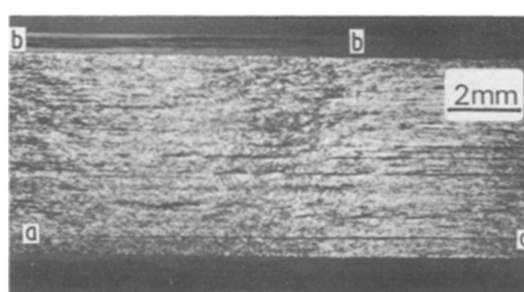

Figure 9 Transverse cracking, aa, and bundle detachment, bb, in a Fig. 1c type specimen sectioned after tensile loading to a stress of 1.69 GN m^{-2}.

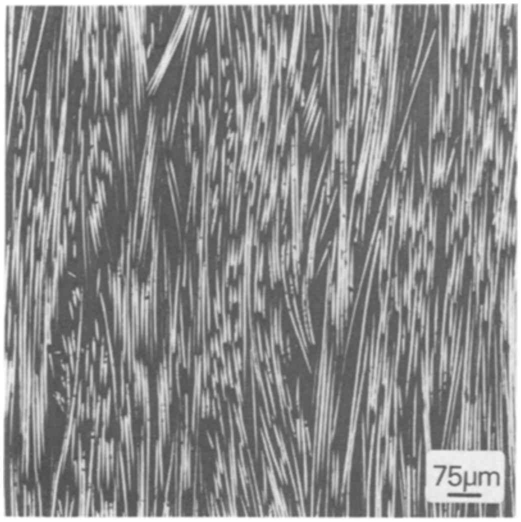

Figure 10 Resin-rich regions present in as-received CFRP pultrusion. Note fibre curvature and absence of debonding.

not detected; detachment of surface fibre bundles, however, involved fracture of a group of fibres as a result of loading to stresses greater than $\sim 1.5\,\mathrm{GN\,m^{-2}}$. It seems, nevertheless, that delaminations initiated at lower stresses.

4. Discussion

Introducing their paper on the tensile failure mechanisms in CFRP, Fuwa *et al.* [4] distinguish between models which treat such fibrous materials as anisotropic but homogeneous [19–22] and those that tackle the problem from a microscopic point of view, considering mechanisms relating to single fibres and fibre bundles [3, 5, 6]. These latter models are based on analyses of matrix-free groups of fibres (commonly called bundles), of which the first classical, statistical treatment appears to have been carried out by Daniels [23]. Coleman [24] went on to consider a Weibull fibre strength distribution and Gucer and Gurland [25] incorporated both the "weakest link" and "bundle" statistics into their model of a "fibrous aggregate". It consisted of an assembly of layers in series, each layer containing a set of elements loaded in parallel, and material failure was predicted to take place when sufficient numbers of volume elements (which fractured independently) in a layer have cracked so that the remaining elements can no longer support the applied load. It was this "chain of bundles of elements" model which Zweben and Rosen [7] modified by considering increases in average stress in the adjacent elements in the layer (rather than a uniform stress redistribution in the layer) in developing their statistical theory of strength of composite materials which, for simplicity, were assumed to fail in a *single* cross-sectional layer. The effect of load concentrations resulting from broken fibres has been taken account of by e.g. Zweben [5], Zweben and Rosen [7] and Barry [26], but only in the work of Pheonix [27] did we find a study of the random fibre slack effects resulting, in his case of matrix-free fibres, from uneven clamping and variations in fibre length. This was shown to result in an additional reduction in fibre bundle tensile strength and it was suggested that degree of non-linearity of the fibre bundle stress–strain behaviour at low strains correlated with this loss of strength.

Having established the existence of fibre slackness in a composite in which the resin has been dissolved away [4] and the importance of this slack in bundle properties [27], it remains to consider its effect on the composite tensile strength. A statistical treatment will not be attempted, nor should it be until a statistical model will allow for composite failure outside an unique cross-section and for the formation of cracks transverse to the tensile axis.

It is now suggested that the misalignment of fibres in the composite, present even in pultruded

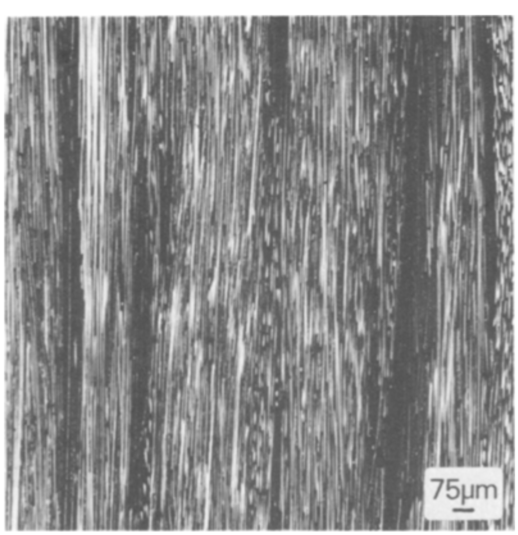

Figure 11 Resin-rich regions, some non-bonded, in poor CFRP pultrusion which has *not* been loaded. These are to be contrasted with debonding and detachment of surface bundles, Fig. 9, caused by tensile straining.

materials, is a major contributory factor in the formation of transverse cracks in uniaxial composites subjected to tensile strain. Our concern will be with cracks formed between tows, bundles or laminae (e.g. Figs. 7, 9) rather than interfacial cracks between individual fibres and the matrix. It is possible that the latter behaviour is well-modelled by Rosen [3]. It is further suggested that the formation of inter-tow cracks does not necessarily impair the immediate load-carrying capacity of the composite. The debonding of a bundle (containing prior curvature) could lead to that bundle subsequently carrying a higher tensile load, straightening out being followed by tensile loading.

It is thus postulated that initial loading of the composite need not lead to failure of curved or misaligned bundles, but can activate a process of straightening the fibres making up the bundle. Such straightening can only occur against the transverse support provided by the matrix. It is suggested that, as load is increased, the transverse tensile strength of the matrix, σ_t (typically $80\,\mathrm{MN\,m^{-2}}$ for epoxides [28]) or that of the fibre/matrix bond can locally be eventually exceeded. This mechanism of debonding is, therefore, in some ways analogous to that postulated by us when considering buckling of a surface bundle in axial compression against the support of the matrix. It is further suggested that the debonded "straightened" bundle can then carry some of the applied load. As the load is increased the bundle curvature decreases, its crack travels in the axial direction and new inter-tow cracks form adjacent to bundles with less prior curvature. This mechanism is expected to operate more easily in specimens where there is a change of section (and free fibre ends) and thus would account for subcritical crack growth (e.g. Fig. 5) and its initiation at lower stresses ($1\,\mathrm{GN\,m^{-2}}$ for our type (a) specimens) compared to the unwaisted, type (c), in which cracks did not appear to initiate below $1.5\,\mathrm{GN\,m^{-2}}$. We would add that previously we have observed detachment of surface "slivers" when CFRP rods were tested in three-point bending with a span-to-depth ratio of 80, sufficient to allow initiation of failure at the tensile surface [29].

It is thus postulated that the limit of proportionality in the load–displacement response (Fig. 2) corresponds to the initation of this mechanism of debonding, followed by interlaminar cracking

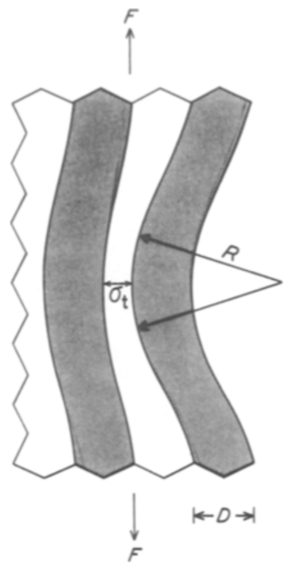

Figure 12 Profile of a surface fibre bundle of diameter, D, and curvature $1/R$, restrained by the composite matrix of yield strength, σ_t, when tensile force F is applied to the composite.

in tension or buckling/kinking in compression. This could account for any similarities (often commented upon, e.g. [30] of uniaxial tensile and compressive strengths of CFRP, but these are to be related to the matrix strength in restraining bundle curvature, we would suggest, and not the shear strength of the (brittle) fibres.

If it is assumed that the conditions for debond initiation we suggested for axial compression [16, 17, 31] are applicable also to tension, then the initiation stress would be given by:

$$E = \frac{8R\sigma_t}{\pi D} \tag{1}$$

where R is the radius of curvature of a bundle of diameter, D. This assumes the curved bundle is adjacent only to those that are straight and parallel (Fig. 12) to the tensile axis; in reality there are various degrees of misalignment and curvature.

Typical values of R, estimated from micrographs such as Figs. 7, 9 to 11, are generally above 12.5 mm. Values of D observed for detached surface bundles ranged up to 1 mm and putting these D and R values into Equation 1 gives σ_E of $\sim 2.5\,\mathrm{GN\,m^{-2}}$, much in excess of observed.

It is to be recalled that typical carbon fibre tows contain 10^4 fibres, approximately of $9\,\mu\mathrm{m}$ diameter. If contained in a $0.6V_f$ CFRP cylinder its diameter would be ~ 2.1 mm. So for an unwaisted specimen and R of 12.5 mm, Equation 1 evaluates to $\sim 1.2\,\mathrm{GN\,m^{-2}}$, close to the observed values for debonding. We would add, however, that three-dimensional and more detailed consideration of curvatures of adjacent bundles is recommended. The simple case of two adjacent bundles of opposite curvature halves σ_E. Another reduction of σ_E would result if only locally or rarely the curvature were greater. For a 6 mm diameter rod there may be up to 40 bundles on the circumference and never have we examined metallographically more than about 10.

It is, therefore, suggested that *qualitatively* transverse cracking can be accounted for in terms of curvature of bundles (tows) and matrix restraint when tensile straining results in the straightening out of such bundles.

Our model of failure in tension of CFRP would, therefore, consist of the three classical stages: initiation, growth and propagation. The *initiation* stage is the one we have discussed: for misaligned surface bundles of fibres formation of inter-tow or inter-laminae cracks: *debonding*. The next stage, we would suggest, is the axial *growth* of these transverse cracks; as the detached bundles straighten, they can carry more load, but the load–extension characteristics are no longer Hookean. Some of the bundles can become detached (e.g. Fig. 8), which would throw more load on the neighbouring groups of fibres. When these fracture as a consequence, we have the *propagation* stage at various cross-sections in the detached bundles of fibres. We would accept that statistical theories may well be then applicable, but note should be taken that axial splitting precedes tensile failure.

It is possible that the intiation and propagation of transverse cracks is responsible for the variability of tensile strength of unidirectional composites [11] above a lower bound corresponding to failure on a given transverse section (as modelled by some statistical theories), and for the variability being greater than that of the constituent fibres. Our analysis also indicates that, above a lower bound, the composite tensile strength should be influenced by matrix sterength, as indeed the compressive can be [16, 17]. Interesting size effects are also possible, not all consistent with a simple Weibull-type hypothesis that the larger the volume of stressed material the lower its strength. If alignment of fibre bundles takes place and these are then able to carry load, larger sections and longer lengths may give rise to higher composite strengths. It should be added, however, that it is not claimed that interfacial cracks are desirable; their deleterious effect is anticipated in dynamic and adverse environmental loadings.

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