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Compressive failure and kinking in uniaxially aligned glass–resin composite under superposed hydrostatic pressure

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The failure process in uniaxially-aligned 60% fibre volume fraction glass fibre–epoxide compressive specimens strained parallel to the fibre axis was investigated at atmospheric and superposed hydrostatic pressures up to 300 MN m^{-2}. The atmospheric strength was about 1.15 GN m^{-2} (about 20% less than the tensile) and strongly pressure dependent, rising to over 2.2 GN m^{-2} at 300 MN m^{-2} pressure, i.e. by about 30% per 100 MN m^{-2} of superposed pressure. The corresponding figure is 22% if the maximum shear stress and not the maximum principal compressive stress is considered. This is incompatible with atmospheric compressive failure mechanisms controlled by weakly dependent or pressure independent processes, e.g. shear of the fibres. The results also could not be satisfactorily interpreted in terms of microbuckling of individual fibres. Kinking, involving buckling of fibre bundles was proposed as the mechanism of failure propagation, but the critical stage (for this glass reinforced plastic) is suggested as being yielding of the matrix, which initially restrains surface bundles from buckling. A strong pressure dependent failure criterion, about 25% increase per 100 MN m^{-2}, was derived by modifying the Swift–Piggott analysis of deformation of initially curved fibres. It is postulated that it is the axial compression that causes bundle curvature. Other systems, particularly carbon fibre-reinforced plastic, in which there appears to be a transition in the critical stage of failure from bundle buckling to matrix yielding with increasing superposed pressure, are also considered.

1. Introduction

It has been suggested that composite compressive strength can be limited by the strength of the fibres either in compression (acting individually as columns) or, to account for similarities between compressive and tensile strengths, in shear [1, 2]. Compressive failure has also been associated with the matrix; either as "shear instability" [3] controlled by the shear modulus or by the stress in it reaching a critical value, as postulated initially by Hayashi and Koyama [4]. The former theory overestimates the strengths of glass and carbon/resin systems as also, more seriously, of metal matrix composites; the latter requires that the composite should fail at the matrix yield strain, which is not generally observed [5].

Compressive strength has also been reported to be influenced by factors such as fibre–matrix adhesion (controlling interfacial splitting), fibre linearity and local misalignment [6]. Piggott [6] considered fibres possessing initial curvature and adapted the sinusoidal fibre model of Swift [7] to compressive loading. For a short length of fibre of diameter, d, the compressive stress in it was postulated as being:

$$\sigma_f = \frac{2\lambda^2 \sigma_t'}{\pi^3 ad} = \frac{8R\sigma_t'}{\pi d} \quad (1)$$

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where R is the minimum radius of curvature of the fibre, σ_t', the transverse stress exerted on the fibres by the matrix, and a and λ the amplitude and wavelength of the sine curve, respectively. Piggott argued that, as σ_f increases, unless some other failure process intervenes (e.g. fibre yielding and failure), σ_t' will become large enough to cause fibre/matrix separation or matrix yielding, σ_t.

In Piggott's [6] terminology glass fibres are "non-yielding" and he postulated that the fibre stress at composite failure is close to the fibre tensile strength for $V_f < 0.4$, but that with increasing fibre volume fraction, V_f, the composite strength falls below a modified rule of mixtures prediction and failure by debonding or splitting is governed by adhesive strength or compressive or yield strength of the matrix. In the discussion of the glass fibre composites Piggott [6] concentrated on the polyester matrix system he studied with Harris [5]. In their glass reinforced plastics (GRP) the compressive strengths did not exceed $0.6\,\mathrm{GN\,m^{-2}}$; this is, as they note, markedly lower than Chaplin's [8] $0.6\,V_f$ epoxy matrix and commercial pultruded epoxy matrix composites, about $0.95\,\mathrm{GN\,m^{-2}}$. It should be added that for both glass and carbon fibre composites tested in uniaxial compression which fail at the higher levels of applied stress [9], the most important fractographic feature appears to be kinking. These kinks have also been described as "compressive creases" and have been attributed to local "shear instability" or microbuckling [8, 10]. Kinking has been observed in GRP compressed or bent at atmospheric pressure, particularly in specimens designed to give stable propagation [8]. Similarly for carbon fibre-reinforced plastic (CFRP) micrographic evidence of kink-associated compressive failure has been presented [9] in samples tested in compression and flexure [11] at atmospheric (and also under superposed hydrostatic) pressure, where propagation can be stable as a result of the stress gradient.

In axial testing of CFRP the superposition of pressure has been useful in determining whether the failure mechanism was controlled by shear stresses, as these are unaffected by hydrostatic pressure. Further, as Chaplin [8] argued with reference to his elastic instability model, in the presence of a hydrostatic stress component or reinforcement in a direction that restricts lateral expansion, the effect of this extra constraint would be an increase in the compressive strength. In high V_f CFRP failure was shown to be strongly pressure dependent; mechanisms controlled by shear stresses alone should thus be discounted for CFRP. That work, [9], however, did not result in identification of a critical failure condition; at pressures above $150\,\mathrm{MN\,m^{-2}}$ strength appeared linearly dependent on pressure with a slope of 3.2, not predicted by any critical stress or strain criterion then considered, even for anisotropic materials. In particular we were unable to apply a model of buckling and fracture of individual fibres, developed by Weaver and Williams [1] to interpret the compressive behaviour of $0.36\,V_f$ CFRP under superposed pressure and suggested buckling of the fibre bundles should be considered [9].

It is against a background with a lack of experimental support for established theories that Piggott [6] has proposed a theoretical framework within which to consider composite compressive strength. In it criteria relating to the operating of six (that he thought most likely) mechanisms are presented and the one satisfied at the lowest stress for a composite of given fibre and matrix properties is predicted to cause failure. Piggott considered the influence on composite strength of fibre strength and linearity, prior fibre curvature, the properties of the matrix, the adhesion between matrix and fibres as well as any fibre–fibre interactions. He restricted his analysis to simple compression and therefore did not consider the possible effects of superposing hydrostatic pressure – which affects different meachanisms differently – thus enabling, at least, some to be excluded from further consideration. Our objective was to study the failure of high V_f GRP using pressure as the experimental variable.

2. Experimental procedure

All experiments were performed on 6 mm diameter pultruded rod containing 60% of S glass fibres in an epoxy resin matrix made available by AERE, Harwell. Longitudinal compression (gauge diameter about 1.5 mm) and in-plane shear (gauge diameter about 6 mm with an overlap of about 8 mm) specimens were fabricated to the same design as reported previously in an investigation of CFRP [9]. These samples were also fitted with aluminium end rings to aid alignment within the pressure testing rig, which was attached to a Hedeby Universal tester on which all testing was performed. Specimens of both designs were strained in axial compression, at a rate of $0.1\,\mathrm{mm\,min^{-1}}$. The pressurizing medium was "Plexol", a synthetic diester, and load, which included frictional forces on the loading and dummy rods, were monitored on a Baldwin–Lima–Hamilton semiconductor load cell. These forces could be determined before and after a sample was tested. The interpolated value at failure was subtracted from the total force to give the superposed compressive load.

The failure surfaces of specimens of interest were examined on an ISI Super III scanning electron microscope. Other samples were mounted in polyester resin and sectioned and polished parallel to the fibre axis. Photomicrographs were taken on a Zeiss Ultraphot II optical microscope.

3. Results

The atmospheric compressive strength of our GRP material was found to be $1.15 \pm 0.1\,\mathrm{GN\,m^{-2}}$, somewhat lower than the tensile, about $1.4\,\mathrm{GN\,m^{-2}}$. The compressive strength (maximum principal compressive stress) increased linearly with pressure, with a slope of nearly 3.5 (Fig. 1). Failures were catastrophic and separation occurred at an angle of approximately $30°$ to the fibre axis. Evidence of kinking, e.g. Fig. 2 (for a sample tested at atmospheric pressure), was obtained by sectioning failed samples parallel to the fibres. In this case (untypically) the micrograph reveals a fully developed kink before specimen separation had occurred. More usually separation took place along one of the boundaries, as shown in Fig. 3, which is of a specimen tested at $300\,\mathrm{MN\,m^{-2}}$ superposed pressure. The same failure mode persisted over the entire range of pressures investigated which were up to $300\,\mathrm{MN\,m^{-2}}$.

Kink band widths, distances between boundaries where failure propagated, were 250 to $500\,\mu\mathrm{m}$ and apparently unaffected by pressure (see Figs. 2 and 3). As the failures were catastrophic, it was difficult to estimate the laterial size of the microstructural unit (bundle of fibres) which initially kinked. Metallographic evidence is not inconsistent with its diameter being about 0.4 mm, as found in CFRP [9].

Shear strength, calculated from the prospective failure area for the in-plane shear specimens, was evaluated to be $42 \pm 5\,\mathrm{MN\,m^{-2}}$ at atmospheric pressure. This value, as noted previously for CFRP, did not vary appreciably with "overlap length". It is lower than that obtained, $59 \pm 1\,\mathrm{MN\,m^{-2}}$, for the same material using the short (10 mm) span three-point bend testing geometry with loading rollers of 6 mm in diameter. This "interlaminar shear strength" parameter, however, was found to vary with the roller size, as has also been reported for CFRP [11], decreasing to about $48\,\mathrm{MN\,m^{-2}}$ as the roller diameter was reduced to 2 mm.

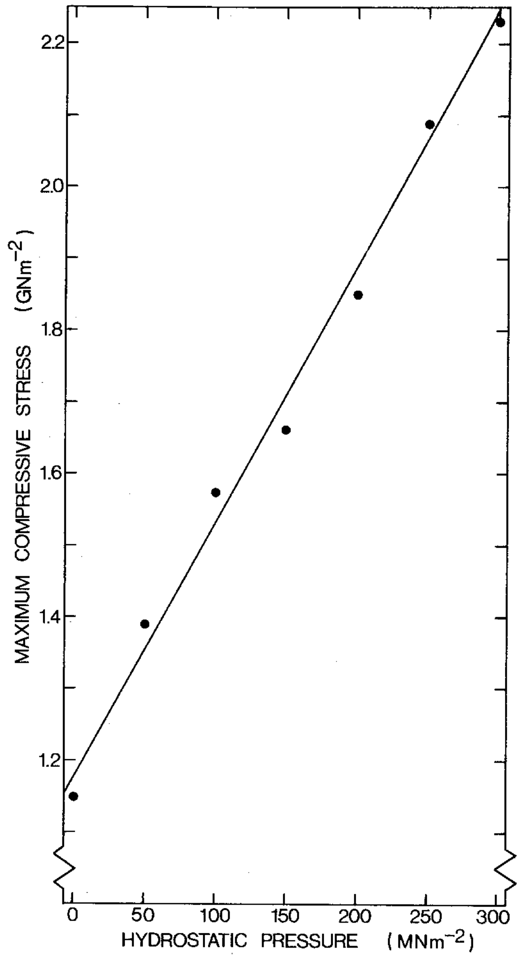

Figure 1 Maximum compressive stress for GRP specimens tested in axial compression under superposed hydrostatic pressure.

The shear strength (evaluated in the in-plane geometry using the minimum prospective failure area) increased with increasing superposed pressure approximately linearly with a slope of about 0.2 (Fig. 4). The shear mode of failure along the midplane operated throughout the range of pressures investigated in this GRP (unlike CFRP previously studied, when kinking from the tip of the notches was observed above $150\,\mathrm{MN\,m^{-2}}$).

4. Discussion

Evidence quoted in support of the fibre shear mechanism of compressive (and tensile) failure is

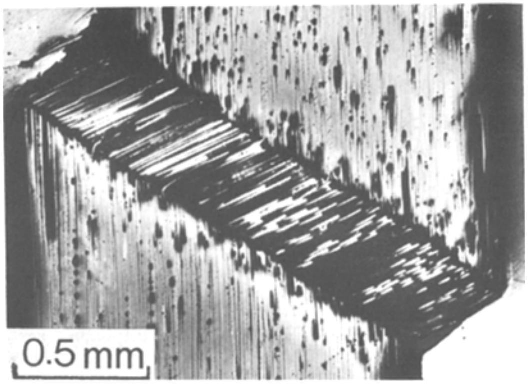

Figure 2 A fully developed kink band in a failed, sectioned compressive GRP specimen strained at atmospheric pressure.

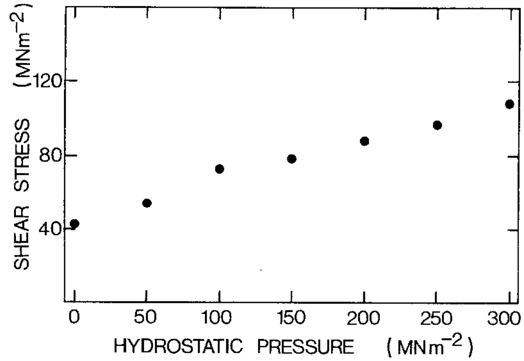

Figure 4 Maximum (interlaminar) shear stress for in-plane shear specimens of GRP tested under superposed hydrostatic pressure.

the inclination of the fracture surfaces: approximately at 45° to the fibre axis [12]. This feature was observed also (approximately) in this investigation; closer examination, however, revealed this angle to be nearer 30° and, by sectioning (see for example Fig. 3) the mode of failure could be identified as kinking. The pressure dependence of the strength is also inconsistent with a critical shear stress for failure, which predicts a slope of unity for Fig. 1 in place of the observed value of about 3.5. Models based on the shear strength of the fibres [2, 12], thus, can be discounted.

It should be noted that the strengths of our CFRP [9] and our GRP of similar V_f of 0.6 and matrix are nearly equal at superposed pressures above 150 MN m^{-2}. It is therefore tempting to try to relate kink associated compressive failure more with the properties of their common matrix rather than the dissimilar fibres of these composites.

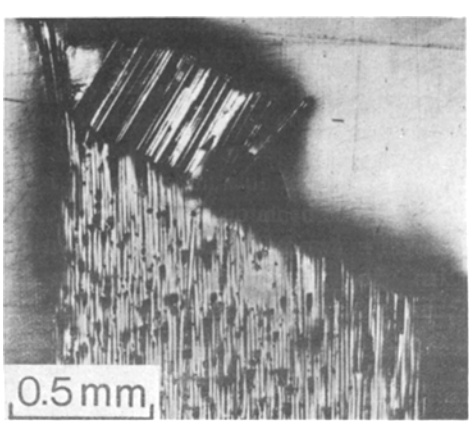

Figure 3 Fracture profile and associated kink band in a sectioned compressive GRP specimen strained under a superposed pressure of 300 MN m^{-2}.

Rosen's [3] model of fibre buckling, dependent on matrix shear modulus, is not applicable as the variation of composite strength with pressure would then be that of the modulus, which is only weakly pressure dependent for epoxies [1]. The slope predicted for Fig. 1 would thus be < 2 rather than > 3, which was observed. This model has also been criticized on other grounds, in particular the predicted values of composite strength, σ_c, are grossly overestimated for CFRP and GRP and it fails to predict the observed variation with V_f [5].

In discussing the CFRP data [9] we have already discounted Eulerian buckling of individual fibres, but found the kink band widths and other microstructural features are not inconsistent with buckling of groups of fibres. The fibre bundle size was then found to be about 0.4 mm in diameter; interestingly Piggott, when analysing results on the compressive strength of a composite containing curved fibres, concluded that, to be consistent with his model [6] of failure by debonding or matrix yielding, groups of about 2000 fibres "must behave in unison". This model, based on an earlier analysis of Swift [7], was developed for composites containing fibres initially with sinusoidal curvature. Even for initially (nominally) straight fibres, however, the role of the matrix cannot be ignored when considering microbuckling of fibre bundles. This buckling, which causes lateral displacement before the critical load is attained, occurs against the support of the resin matrix. If failure is initiated when this support is lost (locally), i.e. when the matrix yields, continued straining will cause gross deformation in this area (from which failure would propagate) and resin yielding or splitting spreading along the bundle boundary. Our interpretation of d in Equation 1 is thus the bundle diameter, D (see Fig. 5), not the fibre, and thus the bundle and composite compressive strength is:

$$\sigma_{c(my)} = \frac{8R\sigma_t}{\pi D} \qquad (2)$$

The estimate of the bundle diameter, D, consistent with the observation on CFRP [9] and Piggott's 2000 fibres behaving in unison [6], is about 0.4 mm. Piggott's analysis concentrated on initial curvature and suggested the relevant mechanism to be the overcoming of the compressive strength of the resin. We would postulate, however, that it is the axial compression which causes bending of the fibres until buckling of a surface bundle can overcome the restraint of the matrix, which is exerting a tensile stress, σ_t', to cause kinking. This interpretation appears equally consistent with Piggott's [6] for the data of Piggott and Harris [5] on a series of glass/polyester composites in which the composite strength, $\sigma_c \approx 9\sigma_t$. It should be added, however, that Piggott's estimate of amplitude, a, in Equation 1, as 4 "fibre" diameters, if these are in bundles, seems excessively large.

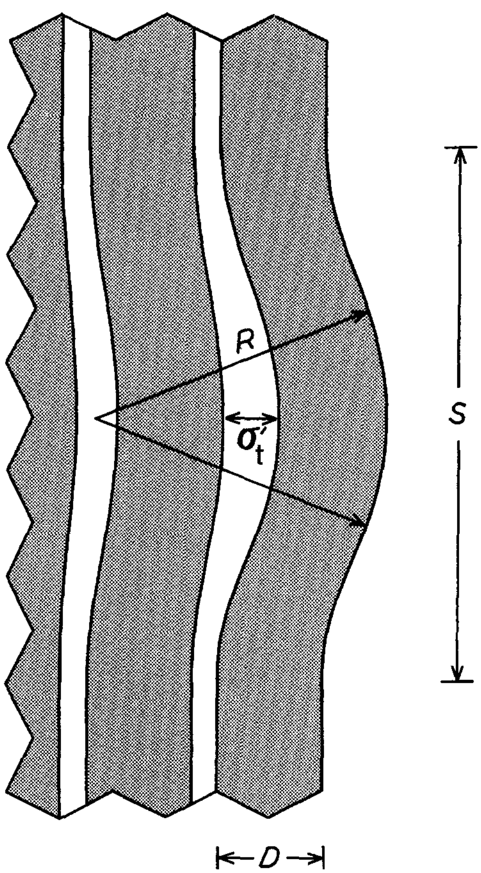

Figure 5 Profile of an outer fibre bundle of diameter, D, bent to a radius of curvature, R, by an axial compressive load and being restrained by the composite matrix of yield strength, σ_t. It is postulated that the width of the resultant kink band is associated with one-half of the segment length, S.

As catastrophic failure of our GRP prevented examination of buckled fibres (prior to kink propagation), there was no simple way of measuring a or R, the radius of curvature of bent fibre bundles. If Equation 2 is assumed to hold, however, and, consistent with our data for 0.6 V_f GRP:

$\sigma_{c(my)}$	$= 1.15\,\mathrm{GN\,m^{-2}}$	(compressive strength of the bundle and the composite)
σ_t	$= 80\,\mathrm{MN\,m^{-2}}$	(tensile strength of matrix)
D	$= 0.4\,\mathrm{mm}$	(bundle diameter)

R evaluates to 2.3 mm. A value of $80\,\mathrm{MN\,m^{-2}}$ was used for σ_t as in a previous study of two epoxy resins Wronski and Pick [13] found the tensile yield strengths to be 67 and $88\,\mathrm{MN\,m^{-2}}$ and the compressive strengths to be 90 and $119\,\mathrm{MN\,m^{-2}}$, respectively.

It should be recalled that the yield strength of polymers (unlike that of metals) is pressure dependent. For the two epoxides the superposed tensile stress for yield increased with pressure, H, by -0.19H [13]. The fractional increases in yield strength, per $100\,\mathrm{MN\,m^{-2}}$ of pressure, were thus 0.28 and 0.22, respectively. (The corresponding figures for the superposed compressive stress for yield per $100\,\mathrm{MN\,m^{-2}}$ pressure, are 0.27 and 0.20 [13].) These values are similar to those of the slope of the GRP compressive strength versus pressure plot (Fig. 1), a 0.30 increase in strength per $100\,\mathrm{MN\,m^{-2}}$ superposed pressure. The behaviour of GRP and CFRP in axial compression beyond $150\,\mathrm{MN\,m^{-2}}$ pressure was similar to that observed by Parry and Wronski [9]. Therefore, the pressure strengthening observed in these brittle fibre-reinforced resins can be interpreted as caused by the pressure dependence of yielding of the matrix. Strengthening of the matrix may thus be expected to inhibit failure initiation in simple compressive loading. In CFRP below the critical pressure, matrix yielding and resultant splitting off of a surface bundle are suggested as being easier than bundle buckling. Accordingly fibre bundle buckling, $\sigma_{c(bb)}$, with a smaller pressure dependence, controls failure (Fig. 6). The weak pressure dependence of $\sigma_{c(bb)}$ shown in Fig. 6 can be interpreted along the lines suggested by Chaplin [8].

It would appear that our GRP data are consistent with the matrix yielding-controlled initiation stage being critical in the composite failure process in the entire pressure range investigated. Let us now consider the apparently easier propagation stage in GRP which we associate with bundle kinking. It is suggested that, when restraint of the matrix is locally overcome, a yielded or debonded zone propagates along the bundle boundary until the buckled bundle can undergo kinking. In the absence of criteria for kink initiation and propagation, we take the upper bound, the load/lateral extension criterion of Euler [14], which transformed into a bundle (and composite) compressive stress is:

$$\sigma_{c(bb)} = \frac{\pi^2 E_c}{(l/K)^2} \tag{3}$$

where E_c is the composite modulus, about $48\,\mathrm{GN\,m^{-2}}$, K, the radius of gyration and l, the buckling length. With no delamination (as in CFRP) it was not possible to measure l but (consistent with CFRP data) if it is assumed to be about 3 mm, for K^2 of 0.01 mm^2, $\sigma_{c(bb)}$ evaluates to about $0.5\,\mathrm{GN\,m^{-2}}$, i.e. less than the observed compressive strength of our GRP. This result implies that the bundle buckling condition, Equation 3, is easier to satisfy in this GRP than the matrix yielding criterion, Equation 2, which controls accordingly the composite compressive strength, σ_c, in a strong matrix high V_f glass fibre composite.

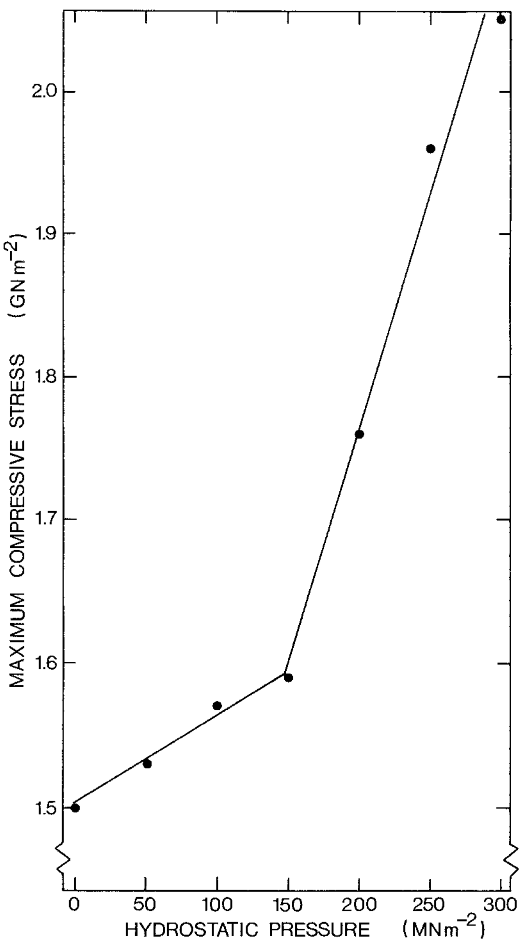

Figure 6 Compressive strength of 0.6 V_f CFRP under superposed hydrostatic pressure [9]. It is postulated that in the first segment, $< 150\,\mathrm{MN\,m^{-2}}$ superposed pressure, failure is governed by fibre bundle buckling, $\sigma_{c\,(bb)}$, and in the second, $> 150\,\mathrm{MN\,m^{-2}}$ superposed pressure, matrix yielding, $\sigma_{c\,(my)}$ criteria.

For weak matrices, e.g. the polyester studied by Piggott and Harris [5], $\sigma_{c(bb)}$ could be more difficult to attain than $\sigma_{c(my)}$ and thus equal σ_c. In CFRP, on the other hand, E_c is appreciably greater and, if l and K remain unaltered in Equation 3 and $E_c = 110\,\mathrm{GN\,m^{-2}}$, then $\sigma_{c(bb)}$ becomes $1.2\,\mathrm{GN\,m^{-2}}$, i.e. greater than $\sigma_{c(my)}$ of Equation 2. Thus, unless the matrix strength is very high, $\sigma_{c(bb)}$ should control σ_c. In our study [9] of CFRP, σ_c was in fact $1.5\,\mathrm{GN\,m^{-2}}$ and was associated with bundle buckling. This atmospheric pressure value rose only to $1.6\,\mathrm{GN\,m^{-2}}$ at $150\,\mathrm{MN\,m^{-2}}$ superposed pressure, when, it is proposed, bundle buckling became easier than splitting (which accordingly no longer occurred), failure being governed by the matrix yielding criterion, Equation 2. Thus only above the transition pressure did the CFRP behaviour closely resemble that of GRP reported now.

Failure of the in-plane shear specimens over the entire pressure range was by shear cracking. This contrasts with the behaviour of CFRP where a transition from shear cracking to kinking from the notch tips was observed. Kinking was probably not observed in these GRP samples because, with the reduced failure loads (as a result of lower values of shear strength), the concentrated compressive stresses at the notch tips were insufficient to initiate kinking. The atmospheric GRP value of shear strength, $42 \pm 5\,\mathrm{MN\,m^{-2}}$, is in the range of yield stresses of most epoxy resins and also shows the similar fractional pressure dependence, about 0.2 per $100\,\mathrm{MN\,m^{-2}}$ of superposed pressure. This composite property therefore can also be interpreted in terms of the matrix strength.

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