JOURNAL OF MATERIALS SCIENCE 15 (1980) 2523–2538

Compression strength of carbon, glass and Kevlar-49 fibre reinforced polyester resins

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The compression behaviour of a series of polyester resins of various compositions and in different states of cure has been investigated. Their mechanical characteristics having been established, the same range of resins was then used as a matrix material for a series of composites reinforced with carbon, glass and aromatic polyamide fibres. The composites were unidirectionally reinforced, having been manufactured by pultrusion, and were compression tested in the fibre direction after a series of experiments to assess the validity of a simple testing procedure. "Rule of Mixtures" behaviour occurred in glass–polyester composites up to limiting volume fractions (V_f) of 0.31 for strength and 0.46 for elastic modulus, the compression modulus being equal to the tensile modulus, and the apparent fibre strength being in the range 1.3 to 1.6 GPa at this limiting V_f. At a V_f of 0.31 the strengths of reinforced polyesters were proportional to the matrix yield strength, σ_{my}, and their moduli were an inverse exponential function of σ_{my}. For the same matrix yield strength a composite with an epoxy resin matrix was stronger than polyester based composites. At $V_f = 0.30$, Kevlar fibre composites behaved as though their compression modulus and strength were much smaller than their tensile modulus and strength, while carbon fibre composites were only slightly less stiff and weaker in compression than in tension. The compression strengths of the polyester resins were found to be proportional to their elastic moduli.

1. Introduction

The compression properties of fibre composites are imperfectly understood. Compression strength is difficult to measure because of the tendency for unconfined or unwaisted specimens to split at one end at low stress [1]. Few data are available for the elastic moduli in compression of composite materials.

Experimental data strongly indicate that the compression strength is a linear function of volume fraction of fibres, V_f, both for reinforced metals [2] and reinforced polymers with steel [3, 4], boron or glass [5], or carbon fibres [6]. The most widely quoted theory predicts that strength should be equal to the shear modulus divided by $(1 - V_f)$ [7], but this gives results which are too high, and the variation with V_f does not agree with the experimental evidence.

The effect of matrix properties has not been investigated in detail, although Hayashi and Koyama [5] found that the assumption that the composite fails at the matrix yield strength was compatible with their results. The correct volume fraction effect was obtained, and the magnitude of the strengths predicted was close to the experimental results.

The effect of fibre properties has not been widely investigated. It is known that fibres with poor compression properties give composites which are also weak in compression, typical

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examples being Kevlar [2] and soft steel wire [3] reinforced polymers. Steel–epoxy composites have been shown to have strengths which are linearly related to the flexural strength of the steel. The strength of the fibre–matrix bond, or mechanical keying, seem to be important since untreated carbon fibres make weaker epoxy-based composites than do the surface treated fibres [6].

This investigation was carried out with the object of determining the influence of matrix and fibre properties on the compressive strengths and moduli of aligned fibre composites made by a simple, but standardized production method which was a modified form of pultrusion.

2. Experimental methods

A common isophthalic polyester resin, Crystic 199, manufactured by Scott Bader, was used as the principal matrix material for this work. A detailed study has been made of the properties of composites of this resin with Pilkingtons E-glass fibres, including the effect of fibre volume fraction, matrix state of cure and some testing variables. The effects of fibre character on composite properties were studied on samples of this resin reinforced with Courtauld's high strength and high modulus carbon fibres (both surface treated), Dupont's Kevlar-49 aromatic polyamide fibre, and Silenka E-glass fibres. Finally, the effects of matrix character were studied on a wide range of polyester resins, with varying proportions of several different dibasic alcohols and diols, all supplied by Scott Bader. One set of epoxy–glass composites were also made for comparison with the polyesters. Table I summarizes the composite materials tested in this work.

2.1. Specimen preparation

Short skeins of fibre were made by winding tows of fibre on a jig with two 6 mm diameter pins 0.3 m apart. When the required number of turns was in place, string was threaded through the loops and tied so that the skeins could be removed from the jig and easily manipulated. The skeins were soaked in a bath of catalysed resin and then pulled by the string into glass tubes (nominally 6 mm internal diameter) with flared ends. This pultrusion process squeezed out excess resin and trapped air, and when glass fibres were used the pultruded rods appeared to be clear and free from voids and other faults. The presence of the fibres was revealed by a slight iridescence.

After curing, the rods were cut into 30 mm lengths, while still inside the glass, with a high speed diamond saw. The composite rods were then extracted, and their ends were polished in a jig to ensure that they were flat and normal to the rod axis. In the case of the carbon–polyester composites the glass tubes were removed prior to cutting, because they splintered during cure.

Specimens of each resin used for the impregnation were cast into 25 mm long, 8.7 mm internal diameter tubes. These were cured simultaneously and under the same conditions as the pultruded rods, and the ends of these samples were also polished.

The fibre distribution was as uniform as would be expected in composites of $V_f = 0.31$ manufactured from separate fibre tows, and there were no

TABLE I Summary of composites tested

Composite	V_f	Effects investigated
Polyester–E-glass fibre (Pilkingtons)	0 to 0.55	Effects of fibre volume fraction; matrix yield strength; sample test length; testing speed.
Polyester–E-glass (Silenka) Polyester–carbon (type HMS) Polyester–carbon (type HTS) Polyester–Kevlar-49	0.31	Effect of fibre properties
Polyester–E-glass (Pilkingtons) (18 compositions of resins as detailed in Table IV)	0.31	Effect of matrix type
Epoxy–E-glass (Pilkingtons) (CIBA-Geigy Araldite VY 219)	0.31	For comparison with polyester polyester composites

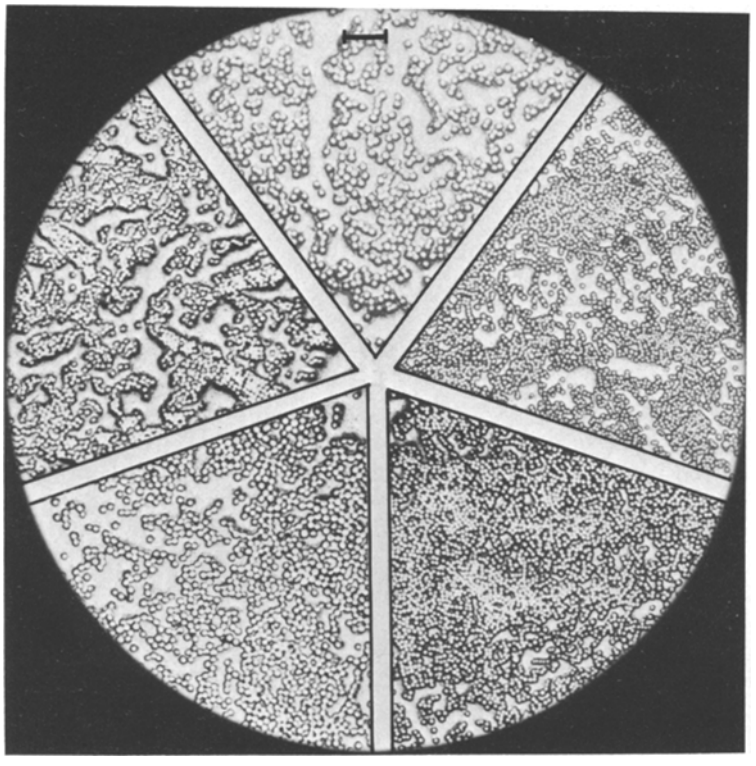

Figure 1 Fibre distributions in composites containing various fibres. Clockwise from top centre, the fibres are Kevlar-49, type I carbon, type II carbon, Silenka-glass, and Pilkington's glass (scale ≡ 0.1 mm).

substantial differences in this uniformity of distribution between composites containing different types of fibre (Fig. 1). At lower V_f the distribution was much less uniform, as would be expected. (See Fig. 2).

2.2. Compression testing

The compression testing method was chosen after preliminary tests on commercial pultruded E-glass–epoxy rods, which showed that simple end confinement gave about 12% higher strength values than were obtained with waisted specimens when carefully aligned and supported. The end confinement method also had the advantage that preparing and testing the specimens was simple and rapid, and the approximate modulus of the specimen can be easily determined.

The experimental specimens all had circular cross sections, and before testing they were inserted into close fitting end pieces. A choice of hole sizes was available in these end pieces, as shown in Fig. 3 because the specimen diameter varied by as much as 0.1 mm as a result of variation in tube diameter and cure shrinkage. The end pieces fitted snugly inside recessed discs. The assembly was placed so that the specimen was centrally located between the platens of an Instron machine, and compressed at a rate of 0.5 mm min^{-1}.

The tightness of fit of the specimens in the end fittings did not appear to be important. Loose specimens, and specimens that were forced into undersized holes with a force as great as 15 kN gave similar results. Reducing stress concentrations at the ends by using an oversize hole, and wrapping the specimen ends with masking tape also had no effect. The edges of the holes in the end fittings were lightly bevelled.

Standard 30 mm long specimens had an effective test length (i.e. the length between the end pieces) of 16 mm, or about 2.7 diameters.

2.3. Analysis of data

The results varied a great deal from rod to rod when brittle fibres were used, and at least four tests were carried out for each data point. The stresses and volume fractions were calculated on the basis of the average composite diameter, 5.9 mm. The results of individual tests were averaged, and standard deviations were calculated using the Bessel correction for small sample size. The error bars in the graph and the numbers following the ± sign in the tables all represent

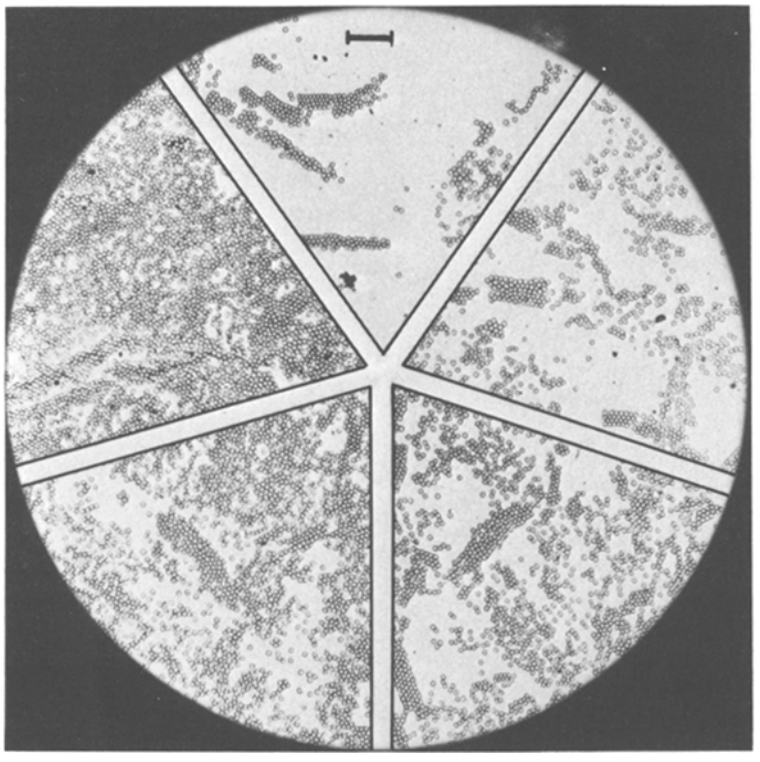

Figure 2 Fibre distributions in glass polyester composites of varying fibre volume fraction. Clockwise from the top, volume fractions are 0.09, 0.18, 0.31, 0.46 and 0.55 (scale ≡ 0.1 mm).

one standard deviation on either side of the mean.

The modulus of the glass-reinforced polymers was estimated from the slope of the force–distance curve after making a correction for end effects and machine compliance. Compliance was determined by operating the machine without a specimen present. It was 19.8 nm N^{-1} for loads of 10 to 100 kN, but was larger at low loads. Corrections were unnecessary at low loads, however, since the specimens that withstood only small loads were themselves very compliant.

The magnitude of the end effects was determined from tests on a set of 30 fully cured glass–polyester rods having different lengths. The apparent compliance of each specimen, calculated from the slope of the straight region of the force–distance curve, is plotted as a function of specimen length in Fig. 4. The slope of this line, 1.46 μm N^{-1}, gives the composite modulus as 24.9 GPa, slightly higher than the rule of mixtures value of 24.2 GPa for the appropriate volume fraction of 0.31. This slope was calculated using a regression analysis, including all the results except for those from 6 cm long specimens which buckled slightly before failure. Inclusion of these reduced the estimate of the modulus to 23.5 GPa, which is still within 3% of the rule of mixtures value; in both cases the correlation coefficient was 0.99. The intercept of the regression line (Fig. 4) with the compliance axis gives the machine plus grip compliance as 26.5 nm N^{-1}, thus the grip compliance is 7.7 nm N^{-1}, which indicates that these specimens depress the steel plates by 38 μm for a load of 1 tonne. Load transfer also takes place along the sides of the grips and this requires a further correction of just under 5%. Thus the 3 cm specimens behaved as though their length was 2.87 cm.

The moduli of higher modulus composites containing carbon fibres, were measured directly using specimens with different lengths.

3. Experimental results

3.1. Volume fraction

Up to about $V_f = 0.31$, the strength of glass–polyester rods is a linear function of V_f, as shown in Fig. 5, and this linear region extrapolates to 1.6 GPa at $V_f = 1.0$. Above about $V_f = 0.45$ the strength falls as V_f is increased further. The Young's modulus is similarly a linear function of V_f, up to $V_f = 0.45$ (Fig. 6). Since $V_f = 0.31$ limits the linear region for strength, most subsequent tests were carried out with specimens

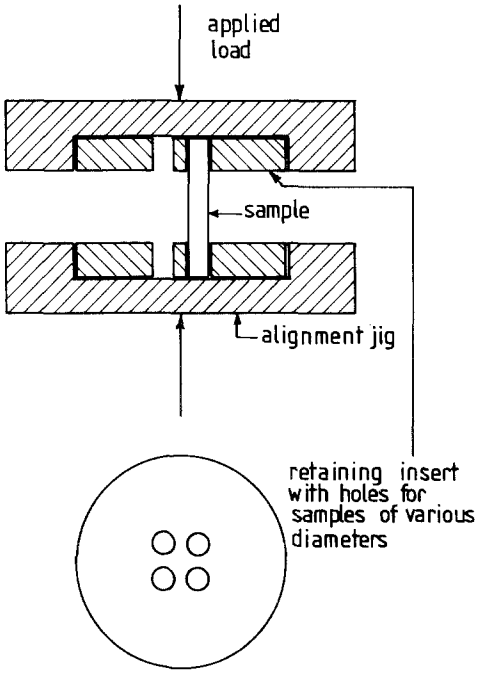

Figure 3 Compressive testing arrangement. Top; section of supports with a specimen in place: bottom; upper side of lower specimen grip.

having this volume fraction. Longitudinally sectioned and polished samples of the tested glass–polyester composites showed distinctive changes in failure mode as V_f increased (Fig. 7). At the lowest volume fractions failure occurred by multiple transverse cracking, the cracks being at an angle to the transverse plane (Fig. 7a). At the highest volume fractions there was

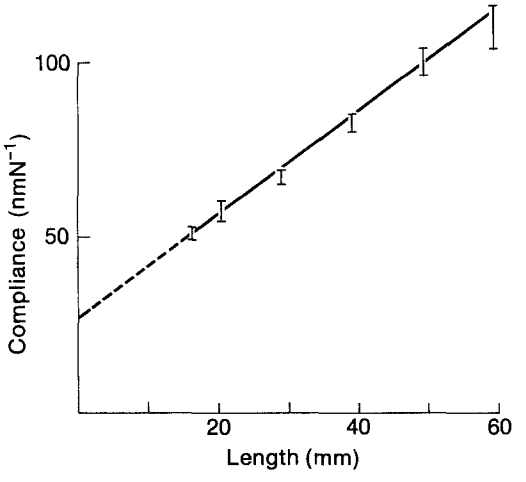

Figure 4 Reciprocal of force required for a 0.5 mm indicated displacement for glass–polyester pultruded rods of various lengths, $V_f = 0.31$.

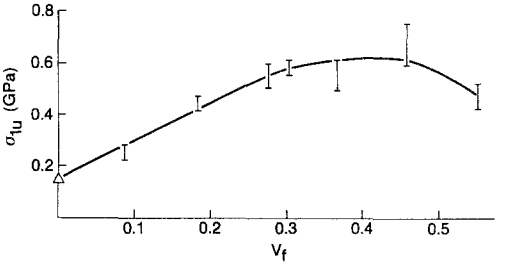

Figure 5 Strength of glass–polyester composites as a function of fibre volume fraction, V_f.

little or no gross distortion of the sample, and again brittle cracks traversed the section at various inclined angles. There were some classical kink bands rather like narrow twins in metals, (Fig. 7c) and there was evidence of longitudinal debonding at glass–resin interfaces (or splitting). At intermediate V_f failure was accompanied by gross distortion of the sample, with large scale bending of the fibres into broad kinks (Fig. 7b), together with substantial cracking and splitting.

3.2. Testing variables

The effect of varying the machine testing speed from 0.05 to 50 mm min^{-1} is shown in Table II. The most notable effects were the increase in matrix yield strength, σ_{my}, and the apparent increase in composite modulus at the highest speed.

As Fig. 8 shows, the specimen length has little effect on composite strength. Samples with gauge lengths as short as 1.5 mm (i.e. $d/4$), failed in the same manner and at the same stress as samples as long as 36 mm. However, buckling occurred when the gauge length was 46 mm, and the failure stress was reduced.

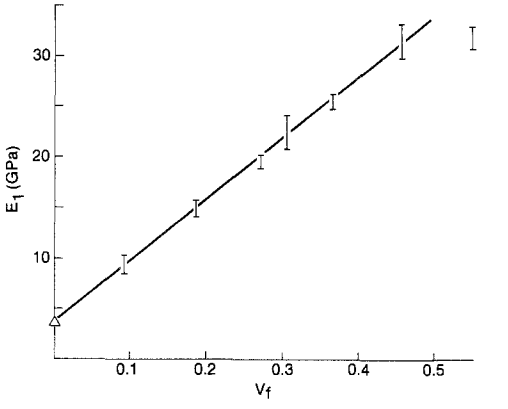

Figure 6 Modulus of glass–polyester composites estimated from the force–distance curves of 30 mm specimens.

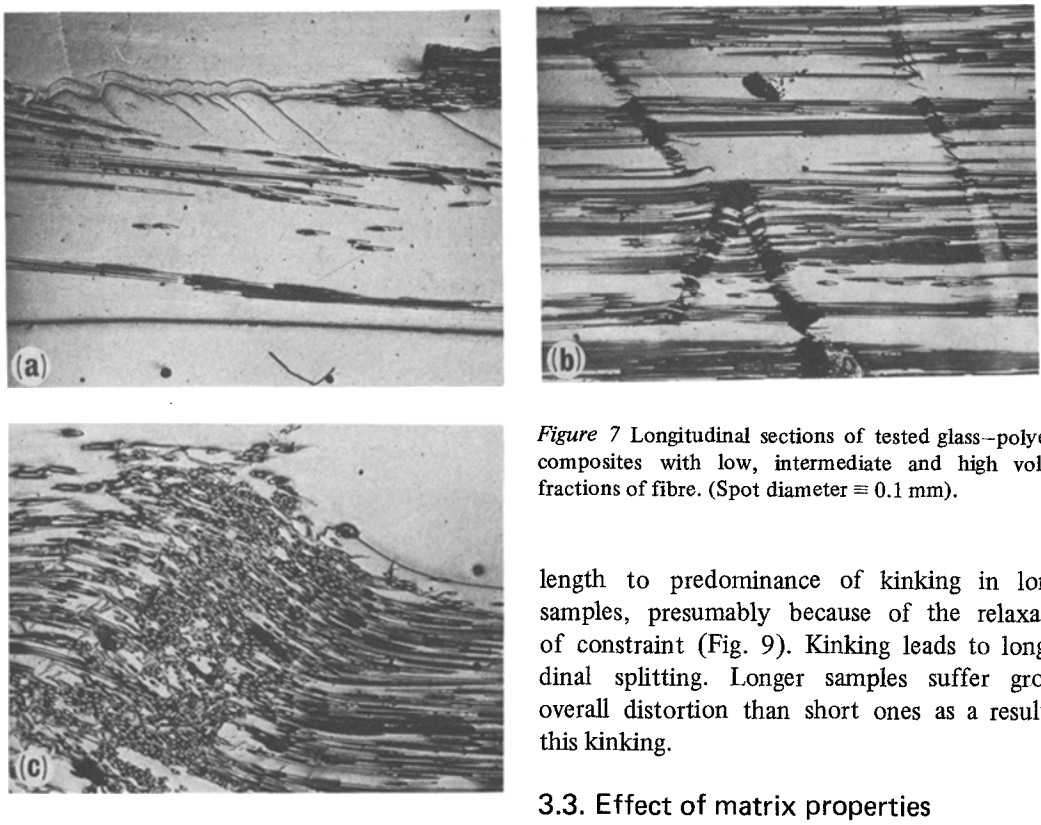

Figure 7 Longitudinal sections of tested glass–polyester composites with low, intermediate and high volume fractions of fibre. (Spot diameter ≡ 0.1 mm).

The specimen length tests were carried out on a set of five rods, made as a single batch, using the same polyester resin mixture, and cured together. They thus give an indication of between-rod variability. Table III shows how the average strength and modulus varied from sample to sample. (The results from the longest samples have been excluded). Variations of ± 10% of the overall means occur in both modulus and strength, and the *t*-test indicates that these are highly significant. The strongest samples had the highest moduli.

As the test length of samples was increased, there was a marked change in the mode of failure from predominance of cracking at the shortest length to predominance of kinking in longer samples, presumably because of the relaxation of constraint (Fig. 9). Kinking leads to longitudinal splitting. Longer samples suffer grosser overall distortion than short ones as a result of this kinking.

3.3. Effect of matrix properties

Tests were carried out on glass–polyester composites ($V_f = 0.31$) in which the state of matrix cure was varied from just gelled to fully cured. The yield stresses of these resins changed from 0.4 to 84 MPa. The composite strength was approximately equal to ten times the matrix yield strength, up to about $\sigma_{my} = 60$ MPa as shown in Fig. 10. At matrix yield stresses above 60 MPa the strength levelled off or decreased, however. For composites of $V_f = 0.55$, which is beyond the linear region of the strength–V_f plot, composite strength (σ_{1u}) was not proportional to matrix yield strength but a log–log plot (Fig. 11) shows that up to about $\sigma_{my} = 70$ MPa it is given by the power law relation $\sigma_{1u} = 50(\sigma_{my})^{0.6}$. At higher matrix yield stresses the strength decreases

TABLE II Effect of testing speed on matrix and composite properties

Speed (mm min^{-1})	σ_{my} (MPa)	E_m (GPa)	Composite Strength (GPa)	Composite Modulus (GPa)
0.05	74	3.3	0.45 ± 0.07	19.3 ± 0.8
0.50	94	3.3	0.47 ± 0.07	19.5 ± 0.2
5.0	96	3.6	0.53 ± 0.06	20.2 ± 0.5
50.0	117	3.7	0.52 ± 0.10	26.2 ± 1.0

σ_{my} = yield strength and E_m = modulus of the matrix resin. The strength and modulus of the composite are measured in the fibre direction. These are glass–polyester composites with $V_f = 0.31$.

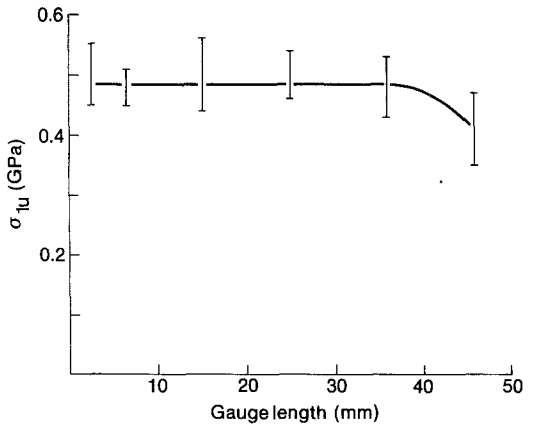

Figure 8 Effect of gauge length on strength of glass–polyester composites ($V_f = 0.31$).

again. Fig. 11 also shows the replotted results for $V_f = 0.31$ but in order to separate the two sets of results, 0.5 has been added to $\log \sigma_{my}$ for all the $V_f = 0.31$ results. The linear relation can be seen to fit the lower volume fraction results quite well.

Composite strength is similarly related to matrix modulus, E_m, because $E_m \simeq 43\,\sigma_{my}$, as shown in Fig. 12. Regression analysis gives a slope for the log–log plot of 1.06, indicating that the relationship is very nearly linear. Some of the resins were made with half the normal catalyst content, and it can be seen that results for these samples tended to fall below the regression line.

The incompletely hardened resins did not appear to have an ultimate compressive strength. The stress–strain curve rose monotonically to yield and then increased linearly to 1.5 to 2 times the yield stress. Fig. 13 shows the stress–strain curves for a polyester resin cured at 20° C for 24 h (labelled 0) and for the same resin cured 1 h and 3 h at 80° C. The fully cured resin is the only one to show a maximum stress. The stresses were calculated assuming that the specimen remained approximately cylindrical, and that the volume remained constant during yielding.

These polyesters were fully ductile in the strain rate range tested, i.e. 4×10^{-5} to $4 \times 10^{-2}\,\text{sec}^{-1}$. After deforming them so as to halve their length they recovered their original shape in a few minutes when released. Below the yield stress their behaviour was completely reversible, but at normal testing speeds (i.e. strain rates of about $4\,\text{m sec}^{-1}$) hysteresis occurred as soon as the yield stress was exceeded. The composite modulus was strongly affected by the matrix yield strength. Fig. 14 shows that the modulus is very low when σ_{my} is small but approaches an approximate Rule of Mixtures modulus asymptotically when σ_{my} exceeds about 40 MPa.

Fig. 15 shows typical stress–strain curves for glass–polyester composites in different states of cure. The softest resins gave composites with some apparent ductility, but after 5 h at 20° C, when the resin was still quite soft, brittle processes appear to be taking place, since the curves display sharp load drops. These load drops occurred both before and after the maximum load. Curing at 80° C resulted in completely brittle behaviour, each load drop being accompanied by an audible sound. Again the load drops occurred before and after the maximum load, and some specimens supported loads close to the maximum (up to strains of 6% or more) after as many as seven load drops.

With the softest resins, visible indications of failure were very localized, in the form of moderately sharp fibre kinking adjacent to one grip (Fig. 16). These external signs only manifested themselves after the maximum load. With harder resins the external surface of the specimens remained smooth, round and straight-sided until after the maximum load, though occasionally slight splintering occurred after a large load drop. Final failure was accompanied by opening up of the specimens at one of the grips, as is also shown in Fig. 16. The samples in this figure are arranged on an inked line. It can be clearly seen, therefore, that translucency in the deformed

TABLE III Variation of mean strength and modulus between consecutive pultruded rod samples

Rod number	1	2	3	4	5
Mean strength (GPa)	0.53	0.48	0.48	0.44	0.53
Standard deviation (GPa)	0.02	0.01	0.03	0.03	0.03
Modulus (GPa)	27.2	23.8	25.5	22.5	26.4
Correlation coefficient for modulus	0.998	0.998	0.997	0.999	0.999

The modulus is calculated from experiments in which the specimen length was varied. Glass–polyester, $V_f = 0.31$.

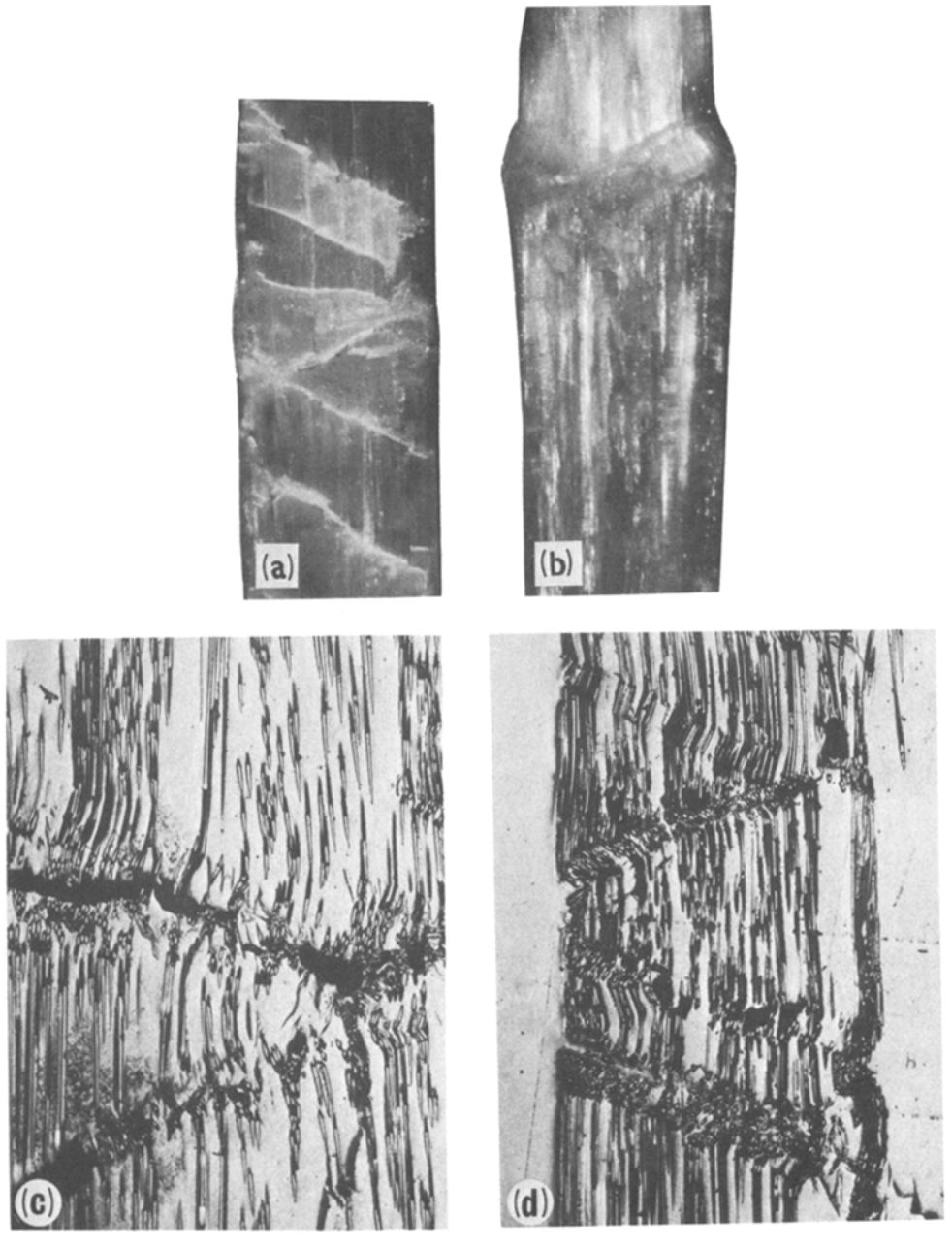

Figure 9 Effect of sample length on deformation mode, changing from multiple cracking in the shortest sample (a and c) to kinking in the longest (b and d). Glass polyester composites, $V_f = 0.31$ (spot diameter ≡ 0.1 mm).

composites is increasingly reduced as the matrix resin gets harder.

Nineteen different matrix types were tested at $V_f = 0.31$ with glass reinforcement. These were mainly polyester resins with different degrees of unsaturation, and with different glycols, some containing added fire retardants. For comparison, sets of samples containing a plain epoxy resin and an epoxy modified polyester (designated S3) were also made. Information on resin composition and resin and composite mechanical properties is given in Table IV.

Fig. 17 shows that the resins fall quite close to, but generally above, the line $E_m = 43\,\sigma_{my}$ obtained from the experiments on resins in various stages of cure. The filled resins are well above the line, as might be expected, since the moduli of the filler particles is greater than that of the

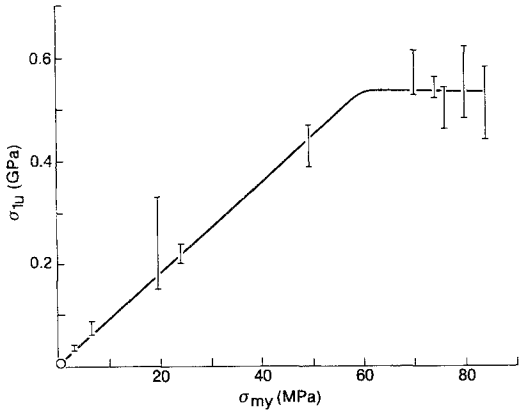

Figure 10 Variation of composite strength with matrix yield strength. Glass–polyester composites ($V_f = 0.31$).

polyester resin alone. The ultimate compressive strengths, σ_{mu} were about 70% higher than the yield stresses and, as Fig. 18 shows, σ_{mu} is approximately proportional to σ_{my}.

No distinct pattern emerges from Table IV, although resin I_1 has an unusually high strength and resin O_2, the only one containing ethylene glycol, is unusually low. When the results are plotted versus matrix yield strength (Fig. 19) these deviations are more marked. The strengths generally follow the rule of being controlled by the matrix yield strength, the correlation being significant only at the 2% level, though the epoxy resin composite deviates from the regression line shown on the graph by an even greater amount than either I_1 or O_2. Results for filled polymers tend to fall somewhat below the regression line. There is almost no correlation

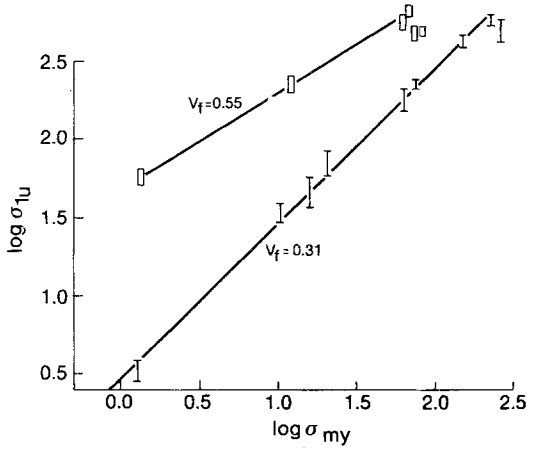

Figure 11 Log–log plot of composite strength versus matrix yield strength. To separate the sets of results, 0.5 has been added to log σ_{my} for all the $V_f = 0.31$ results. Glass–polyester composites.

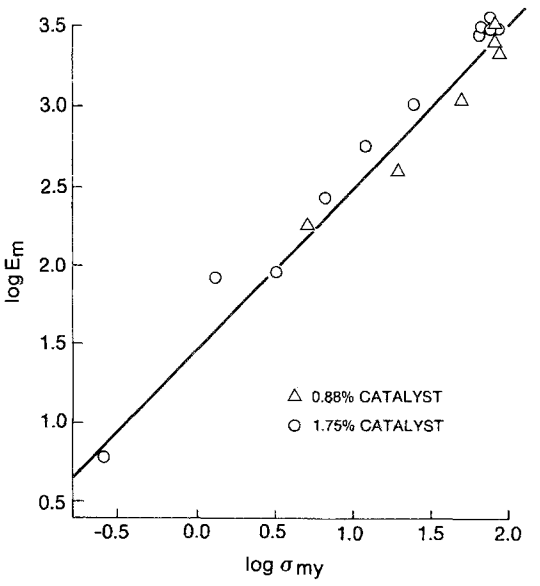

Figure 12 Log–log plot of matrix modulus versus matrix yield strength for a polyester resin in various stages of cure.

between the moduli of composites and resins. The filled resin composites have moduli close to the average for all the composites (21.8 GPa), despite the higher moduli of two of the filled polymers.

Microscopic examination of longitudinal sections suggested that the matrix strength affected the failure mode. In the samples with the lowest two matrix yield strengths failure was very localized with narrow kinks occurring perpendicular to the fibres. There was no gross distortion of the sample and only very localized resin cracking. In the samples with the highest matrix yield strengths, however, conventional kink/crack/split behaviour

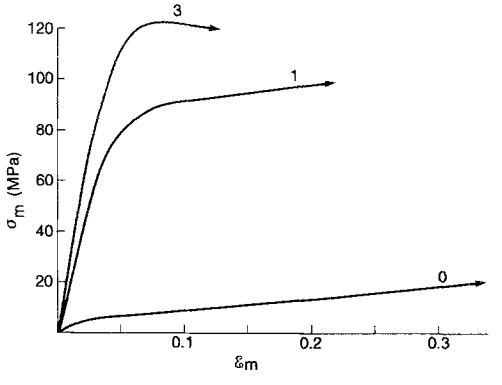

Figure 13 Stress–strain curves for a polyester resin in various stages of cure. Curves 1 and 3 were obtained from specimens cured at 80° C for 1 and 3 h, respectively, and curve 0 was obtained from a specimen curved at 20° C for 24 h. The strain in this case is the logarithmic strain.

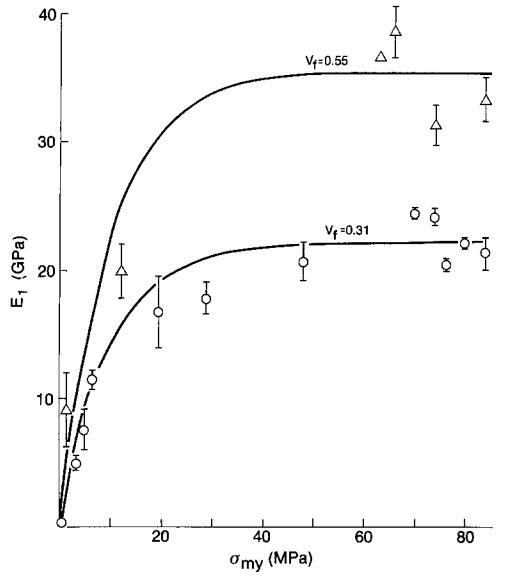

Figure 14 The variation of composite Young's modulus with matrix yield stress for polyester resin matrices in various stages of cure.

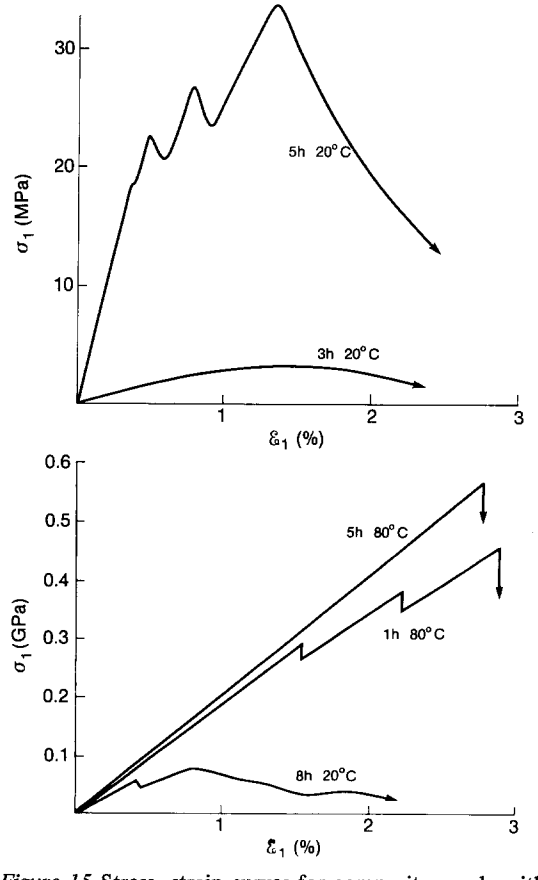

Figure 15 Stress–strain curves for composites made with polyester resins cured as indicated on the curve.

occurred with large scale longitudinal resin cracking.

3.4. Effect of fibre character

Strength and modulus values for composites containing several different kinds of fibre, all with $V_f = 0.30 \pm 0.01$, are given in Table V. The polyester resin matrix had a yield strength of 87 MPa, an ultimate strength of 127 MPa, and a modulus of 3.7 GPa in compression.

The moduli for the Kevlar-49 and carbon fibre composites were obtained from a set of tests in which a length of rod, initially 50 mm long, was tested and then retested after successive reductions in lengths, its final length being 1.5 cm. Fig. 20 shows the results obtained, together with those for glass fibre composites from Fig. 4. The two sets of carbon composites give good straight lines with correlation coefficients of 0.96 for type I and 0.99 for type II, but the behaviour of the Kevlar composites was quite different. Two consecutive tests on the same sample at 5 cm length gave very different results, the second giving a much higher compliance than the first. Further increases in compliance occurred in successive tests as the length was shortened, and the slope of the compliance versus length plot (Fig. 20) gave a rather higher modulus (25.5 GPa) than indicated by the slope of force–distance curves on specimens of fixed length. The results in Table V for Kevlar and the glass fibres were therefore calculated from the slope of the force–displacement curve. As Table V shows, the experimental moduli for Kevlar and carbon composites are all substantially less than rule of mixture values calculated from fibre tensile moduli.

The compression strengths of the glass and carbon composites are not significantly different from each other, but the Kevlar composite is relatively weak. Its stress–strain curve and failure mode were also different from those of the other composites. Fig. 21 shows some typical stress–strain curves. All five glass samples failed abruptly, while a few of the carbon samples showed load drops before failure, the rest failing abruptly. The Kevlar composites appeared to be quite ductile, with a distinct yield point. When re-tested after yielding the Kevlar composites were more compliant than they had been originally.

No external damage could be seen on any of the specimens prior to failure. In the case of the glass and carbon, the increase of diameter

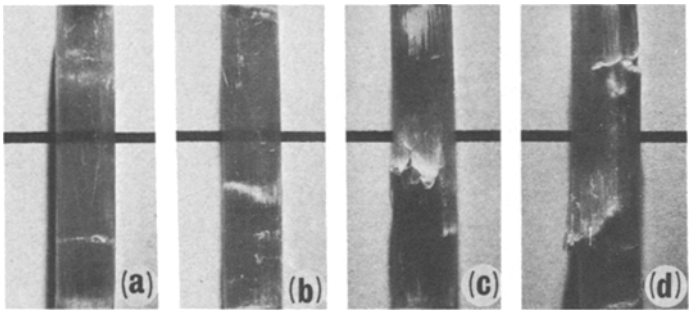

Figure 16 Changes in appearance of tested polyester–glass composites with increasing strength of the matrix polymer, as affected by its state of cure $V_f = 0.31$. (a) $\sigma_{my} = 0.1$ MPa, (b) $\sigma_{my} = 3.2$ MPa, (c) $\sigma_{my} = 19$ MPa, (d) $\sigma_{my} = 84$ MPa.

at one of the grips occurred suddenly, at the failure point, whereas with the Kevlar, kinking occurred after the load drop at a strain of 6%, and became progressively greater as the strain was further increased. After deforming to a strain of 5.3%, samples of glass fibre composites appeared slightly whiter than untested samples. When the carbon specimens were removed from the grips they usually separated into two parts, whereas all other samples retained their integrity.

Longitudinal polished sections again revealed characteristic differences between composite types (Fig. 22). Kevlar fibre composites deform by the kinking mechanism only, often with a single inclined kink band, little general distortion of the sample and with no evidence of cracking or splitting (Fig. 22d). The sample retains its integrity after quite severe kinking. Carbon fibre composites, on the other hand, failed in a brittle manner with a single transverse crack which, unlike those in any other material, was perpendicular to the fibre axis (Fig. 22b and c). The failure was not accompanied by any longitudinal splitting or debonding, whereas the transverse cracks in type I carbon composites were very clearly initiated by kinks, albeit highly localized ones. There was no visible evidence of any kinking failures in type II carbon composites.

4. Discussion and conclusions

4.1. Test method used

In compression testing there is a serious danger that the measured strength will be affected by the test method. The end confinement method used here resulted in failures which were initiated in the specimen interior, and external signs of failure only appeared when the load carrying capacity of the rod had fallen to a low value. It seems likely therefore, that stress concentrations due to the end confinement do not initiate sample failure, since they would clearly manifest their effects as surface features coinciding with indications of failure given by the load–deflection curve. Supporting evidence for this is given by the negative results of tests in which the tightness of fit into the end pieces was varied. Further confidence in the test method may be gained from a comparison of the results with those of other workers. The glass fibre composite results agree with those of Greenwood and Rose

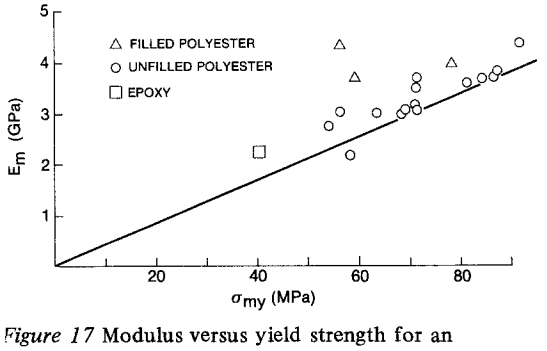

Figure 17 Modulus versus yield strength for an epoxy resin and various polyester resins. The line represents the relationship $E_m = 43\,\sigma_{my}$, as found with plain polyester resin in various stages of cure.

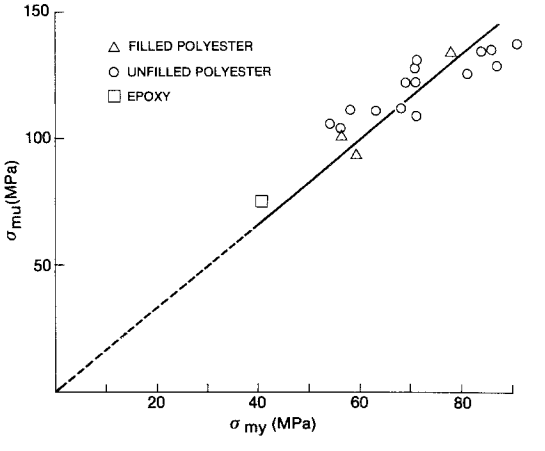

Figure 18 Ultimate compressive strength versus yield strength for an epoxy resin and various polyester resins.

TABLE IV Compositions, strengths and moduli of resins, and the strengths and moduli of resin–glass composites ($V_f = 0.31$). Relative molar proportions of acids and principal glycols are given

Resin Designation		I_1	I_2	I_3	I_4	I_5	O_1	O_2	O_3	O_4	F1	F2	F3[a]	F4[a]	F5[a]	S_1	S_2	S_3[b]	S_4[c]	Epoxy[d]
Acid	Fumaric	75	–	50	–	–	–	–	–	–	–	–	–	–	–	–	–	–	–	–
	HET[e]	–	–	–	–	–	–	–	–	–	60	25	–	–	60	–	–	–	–	
	Isophthalic	25	50	50	50	33	–	–	–	–	–	25	–	–	–	50	33	–	–	–
	Maleic	–	50	–	50	67	75	50	40	50	40	50	50	67	40	50	67	50	100	–
	Orthophthalic	–	–	–	–	–	25	50	60	50	–	–	50	33	–	–	–	–	–	–
Principal glycol	Diethylene	–	100	–	–	–	–	–	50	–	50	–	–	–	50	–	–	–	–	–
	Ethylene	–	–	–	50	50	–	100	50	–	50	–	–	–	50	–	–	–	–	–
	Neopenta	–	–	–	–	–	–				–	–	–	–	–	100	100	–	–	
	Propylene	100	–	100	50	50	100	–	–	100	–	100	100	100	–	–	–	–	–	–
Resin properties	σ_{my} (MPa)	69	71	56	71	86	71	71	91	84	87	54	78	59	56	81	58	63	68	40
	σ_{mu} (MPa)	122	109	104	122	135	130	127	137	134	129	106	134	94	101	125	111	111	112	76
	E_m (GPa)	3.2	3.1	3.1	3.5	3.7	3.7	3.1	4.4	3.7	3.8	2.7	4.0	3.7	4.4	3.6	2.2	3.0	3.0	2.2
Composite properties	σ_c (GPa)	0.55	0.46	0.42	0.46	0.51	0.44	0.35	0.47	0.50	0.48	0.40	0.40	0.36	0.40	0.41	0.44	0.41	0.44	0.45
	Standard deviation	0.11	0.05	0.08	0.07	0.01	0.06	0.04	0.02	0.06	0.06	0.02	0.03	0.07	0.05	0.03	0.04	0.05	0.10	0.01
	E_c (GPa)	22	22	22	20	23	21	20	22	21	24	21	19	20	21	17	20	21	20	19
	Standard deviation	0.5	0.3	0.4	1.3	0.2	0.7	1.0	0.6	0.9	0.8	0.3	1.8	1.5	0.3	1.2	0.6	0.9	1.0	1.3

[a]Contained fire retardants; [b]Other acid; tetrahydrophthalic anhydride; [c]Epoxy modified; [d]CIBA-GEIGY Araldite VY219; [e]Hexachloroendomathylene tetrahydrophthalic acid.

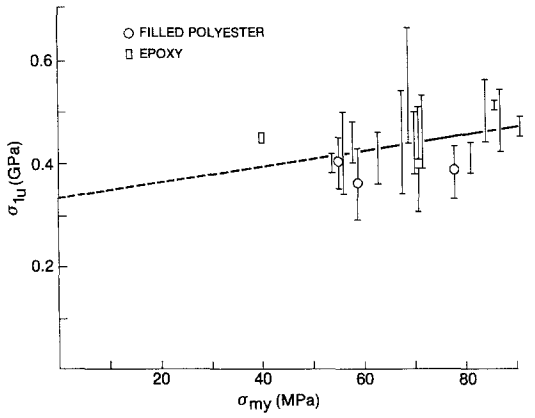

Figure 19 Composite strength versus matrix yield strength for an epoxy resin and various polyester resin matrices. $V_f = 0.31$.

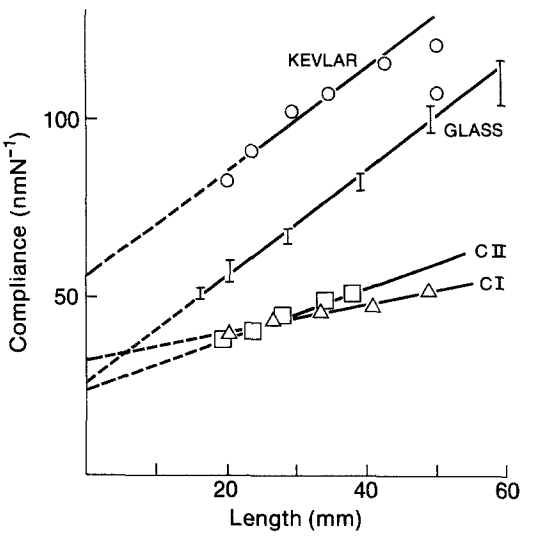

Figure 20 Reciprocal of force required for a fixed indicated displacement for a polyester resin reinforced with various fibres, $V_f = 0.30$.

[8] (0.46 GPa at $V_f = 0.50$ for glass–epoxy and 0.23 GPa at $V_f = 0.53$ for Kevlar–epoxy). The result for type I carbon composites is higher than that obtained by Hancox [6] (0.45 GPa at $V_f = 0.3$), although that for type II carbon is much lower (0.45 as compared with 0.76 GPa). Hancox used epoxy matrices rather than polyester, however. The results for glass fibre composites are markedly lower than those that Chaplin [9] obtained with extremely carefully made specimens. His strength for glass–epoxy with $V_f = 0.6$ was 0.95 GPa. This result is very close to the value obtained when the commercial, pultruded glass–epoxy rod, with end confinement, was tested.

Pultruded rods do not appear to have well-controlled compressive properties. The manufacturer's data indicate that the commercial rod on which preliminary experiments were carried out has a very variable modulus. With the laboratory pultrusions, adjacent samples from the same rod had different strengths and moduli, and different rods, prepared by the same method with the same ingredients and at the same time, gave significantly different strengths and moduli. In these rods high strength was accompanied by high modulus, suggesting that a fibre alignment effect may be responsible. Testing rate does not seem to be important. Most of the increases in strength and modulus of the resin with increasing testing speed are too small to affect the composite significantly.

TABLE V Compression strength and Young's modulus of composites made with different fibres but the same polyester resin ($V_f = 0.30 \pm 0.01$)

Fibre	Strength (GPa)	Modulus (GPa)	ROM[a] Modulus (GPa)
E-glass (Pilkingtons)	0.46 ± 0.06	23 ± 0.7	24
E-glass (Silenka)	0.50 ± 0.03	24 ± 0.8	25
Kevlar-49	0.173 ± 0.006	21 ± 2.6	38
Carbon-type I[b]	0.50 ± 0.05	92	109
Carbon-type II[c]	0.45 ± 0.02	52	68

[a]Rule of mixtures modulus, calculated from the fibre *tensile* modulus.
[b]High modulus, surface treated.
[c]High strength, surface treated.

The standard specimen length used for the tests (30 mm, giving a gauge length of 16 mm) is close to the centre of the constant region of the length–strength plot (Fig. 8). With the 45 mm gauge length buckling occurred, and this is almost certainly the cause of the decreased strength of these specimens. Such buckling would not be observed in isotropic rods of the same dimensions. This would require a force of 27 kN for buckling in the mode observed, whilst composite buckling was clearly visible at loads below 9 kN. Thus specimen lengths must be kept well below the minimum for Euler buckling of isotropic materials.

4.2. Volume fraction

The results indicate that for moderate volume fractions both strength and modulus are a linear function of volume fractions. It is interesting to note that deviations from the Rule of Mixtures type behaviour increase as the fibre distribution improves. Fig. 2 shows sections of glass–polyester pultruded rods at various volume fractions. The

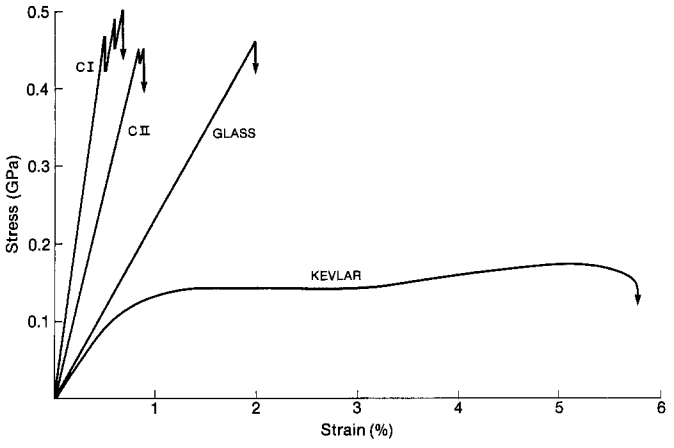

Figure 21 Stress–strain curves for different reinforced polyester composites, $V_f = 0.30 \pm 0.01$.

lowest volume fractions, which fit the mixture rule best, have the most extensive resin rich areas.

4.3. Composite strength

The composite strength depends both on the type of fibre used and on the matrix. The differences between the strengths of composites made with carbon and glass fibres appear to be quite small, but Kevlar makes composites which are much weaker in compression. Assuming that the Rule of Mixtures can be used for compressive strength, the effective fibre "compression strengths" are as shown in Table VI. However these figures should be treated with great circumspection. The strengths of the pultruded rods are very variable. For example, a different set of tests in which V_f was varied, indicated that the "compressive strength" of glass fibres was about 1.6 GPa (see Fig. 5).

The matrix has a profound effect on the strength. For soft matrices the compressive strength was given by the power law relationship

$$\sigma_{1u} = A(\sigma_{my})^n \quad (1)$$

where $n = 1$ and $A = 10$ for $V_f = 0.31$, and $n = 0.6$ and $A = 50$ for $V_f = 0.55$. The lines drawn in Fig. 11 are given by Equation 1 with the appropriate values for A and n.

When the matrix yield stress exceeds about 60 MPa, the behaviour appears to change and the composite strength either remains constant ($V_f = 0.31$) or declines ($V_f = 0.55$) with further increase in σ_{my}. This may indicate a change from matrix control to interface control of strength. Hancox [6] has shown that adhesion has a large effect on the compressive strength of fully cured carbon–epoxy composites.

The chemical composition of the resin has little or no effect on the composite strength, except as a result of its effect on the matrix yield strength. Epoxy resin, however, gave a greater strength than expected, while a polyester containing ethylene glycol gave a lower strength.

4.4. Composite modulus

The modulus appears to be given by the expression

$$E_1 = B(V_f E_f + V_m E_m)(1 - e^{-b\sigma_{my}}) \quad (2)$$

where $b = 0.1\,\mu m\,Pa^{-1}$. The lines shown in Fig. 14 were plotted using this equation with $B = 0.95$ for $V_f = 0.31$ and $B = 0.85$ for $V_f = 0.55$. When the matrix yield strength exceeds about 60 MPa, the exponential term becomes insignificant.

When carbon fibres were used the moduli of the composites were significantly lower than would have been predicted by the Rule of Mixtures. This could be due to fibre misalignment, and this may also be true of the commercial pultruded rod which had, according to the manufacturer's data, a compressive modulus which varied from 36% to 50% of the tensile modulus.

The low modulus of the Kevlar composites is probably due to the poor resistance of the Kevlar

TABLE VI Compression strengths of fibres estimated from composites strengths using the Rule of Mixtures

Fibre	Strength (GPa)
E-glass	1.3 ± 0.2
Silenka-glass	1.3 ± 0.1
Kevlar	0.28 ± 0.01
C I	1.4 ± 0.1
C II	1.2 ± 0.1

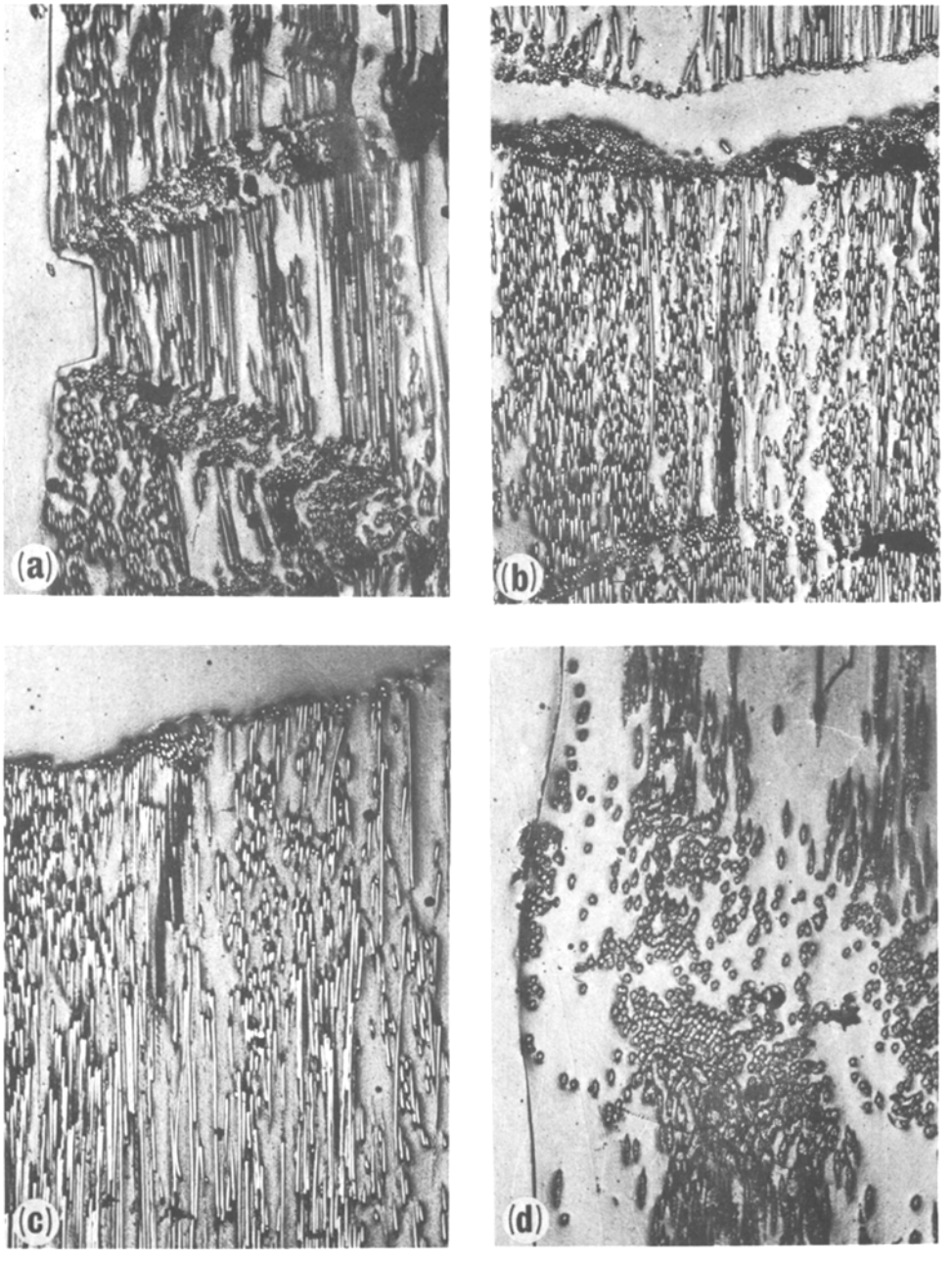

Figure 22 Sections of failed polyester composites reinforced with Pilkington's glass fibre (a); type I carbon (b); type II carbon (c); and Kevlar-49 (d); (spot diameter ≡ 0.1 mm).

fibre itself to stresses other than tensile. This material also showed a continuing deterioration of rigidity during loading.

4.5. Theories for the compression strength

These results lend no support to existing theories. The linear relation between strength and volume fraction is incompatible with the Rosen [7] theory. Many other results show the same linear relation, as discussed in Section 1. The Rosen theory predicts no difference in compression strength between composites reinforced with different fibres whereas significant differences have been found both in this study and in other work. In addition, levels of strength predicted by the Rosen model are too high. It seems timely, therefore, to suggest abandoning this theory.

The Hayashi and Koyama theory [5] also

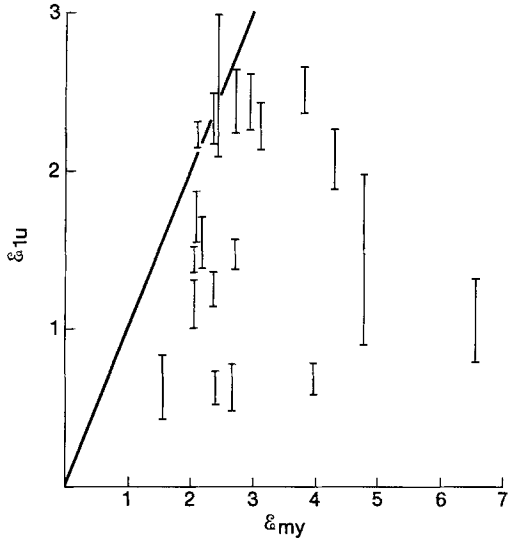

Figure 23 Composite failure strain, ϵ_{1u} for glass–polyester composites as a function of matrix yield strain. The line indicates $\epsilon_{1u} = \epsilon_{my}$.

appears inapplicable, since failure is supposed to occur at the matrix yield stress. Fig. 23 shows that for the resins in various states of cure this is not the case. The composite normally fails while the matrix is still elastic. With carbon fibres the failure process starts at even lower strains than in glass–polyester composites. Incidentally, this also invalidates attempts to explain the high values predicted by the Rosen theory as due to yielding of the resin before failure.

Acknowledgements

The assistance of Scott Bader in the provision of resin samples is gratefully acknowledged. In addition, one of the authors (MRP) received a Senior Visiting Fellowship from the Science Research Council, (Polymer Engineering Directorate).

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Received 25 January and accepted 26 March 1980.