
Some Factors Affecting the Strengths of Carbon Fibres [and Discussion]

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Some factors affecting the strengths of carbon fibres

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The strength-limiting flaws that develop in carbon fibres from impurity particles in the polyacrylonitrile (PAN) precursor fibres have been studied using a fibre-spinning apparatus installed in a clean room. Carbon fibres prepared from PAN fibres spun from a filtered spinning solution in the clean room had tensile strengths of about 3 GPa after heat treatments to both 1400 and 2500 °C. Flaws due to impurities reduced the strengths of the carbon fibres, particularly after heat treatments to 2500 °C, and caused the strengths to decrease as the length tested was increased.

Possible mechanisms are discussed for the formation of flaws and a failure criterion, based on the strain energy developed in a misaligned crystallite, is used to predict carbon fibre strengths. The ranges of strength predicted from the model show satisfactory agreement with the ranges of strength obtained experimentally for fibres heat-treated to various temperatures.

1. INTRODUCTION

The Young modulus of carbon fibres derived from polyacrylonitrile (PAN) fibres increases continuously with heat-treatment temperature (Moreton *et al.* 1967) owing to improvements in the preferred orientation of the graphitic layer planes in the fibres (Watt & Johnson 1969). The tensile strength, however, normally reaches a maximum and then decreases as the heat-treatment temperature is increased (Moreton *et al.* 1967; Johnson 1969). Thus current commercial carbon fibres of maximum strength are obtained after heat treatments in the range 1200–1500 °C and typical values for the strength and breaking strain are then 2.7 GPa and 1.3% respectively (for single fibre tests on 5 cm gauge length).

Suggested fracture mechanisms have included such factors as dislocation pile-up at grain boundaries, the unbending of curved ribbons present in the structure, density fluctuations and yield processes involving local shear deformation and slippage. These ideas have been discussed in review articles by Bacon (1973) and Reynolds (1973).

Tests on different lengths of carbon fibres have shown that the strength increases as the gauge length decreases (Moreton 1969) and the mode of variation suggests a random distribution of flaws. Johnson (1969) and Johnson & Thorne (1969) investigated the fracture surfaces of carbon fibres with a scanning electron microscope and they concluded that impurities in the PAN precursor fibre caused flaws in carbon fibres. Sharp & Burnay (1971) used a high voltage electron microscope to examine carbon fibres in transmission and found inclusions and cavities in the fibres. They suggested that the cavities were formed mainly at high temperatures by the volatilization of inclusions and that this was the cause of the decreases in strength after heat treatments above 1200–1500 °C. Flaws caused by the reaction of impurities with the fibres during processing provide a more satisfactory explanation of the strength–gauge length

[43]

effect than the microstructural mechanisms, which have not yielded quantitative predictions of general significance.

PAN fibres have been spun at the R.A.E. from carefully filtered spinning solutions under clean-room conditions and used to obtain carbon fibres with significantly higher strengths, particularly after heat treatments to 2500 °C. This work has included an assessment of the relative importance of filtering the spinning solution and of using a clean air environment. It has also involved experiments on the deliberate contamination of samples of clean PAN fibre with impurities that may be present near commercial spinning plants.

The results obtained have been combined with other published data to give a consistent explanation of fibre mechanical properties in terms of established structural features.

2. EARLIER EXPERIMENTS

(a) Influence of precursor fibre properties

Experiments have been described in which different hot stretch ratios and stretching temperatures were used to obtain PAN fibres with different mechanical properties for conversion to carbon fibre (Moreton 1971). This work showed that the Young moduli of the carbon fibres were related to the corresponding moduli in the PAN precursor fibres, and indicated that

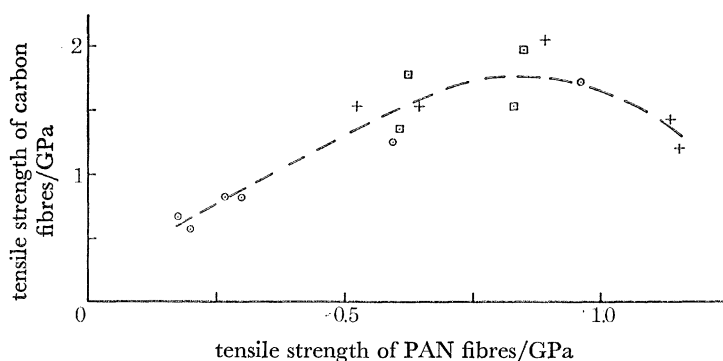


FIGURE 1. Relation between the tensile strengths of PAN precursor fibres and the tensile strengths of carbon fibres heat-treated to 2500 °C. \circ , Polymer A; +, polymer B; \square , polymer C. Gauge length 5 cm.

improvements in the preferred orientation of the PAN fibres were reflected in the preferred orientation that developed in the carbon fibres. A similar link was obtained between the strengths of the carbon and PAN fibres, as can be seen from figure 1, which gives values for carbon fibres heat-treated to 2500 °C. Figure 1, however, shows that PAN fibres with strengths > 1 GPa did not yield correspondingly high carbon fibre strengths, and this was due to surface damage introduced during the spinning of the PAN fibre (Moreton 1976). The use of too high a stretch ratio caused ribbons of polymer to pull away from the fibre surface and this resulted in flaws in the carbon fibres. The low strength PAN fibres probably resulted in low carbon fibre modulus and strength because there was less preferred orientation in the structure of the fibres.

(b) Other factors

A number of additional factors can affect carbon fibre strengths. Thus Watt & Johnson (1970) found that PAN fibres that had been incompletely oxidized at 220 °C before

carbonization gave carbon fibres of low strength due to the development of a sheath/core structure and, in some cases, cavities within the fibres. Cavities in carbon fibres can also be the result of cavities in the precursor fibre and Thorne (1970) suggested that these could be due to gas dissolved in the fibre spinning solution or to irregular flow of the spinning solution through the spinneret nozzles.

The strengths of carbon fibres can be increased by modifying or removing surface flaws by means of oxidative etching treatments (Johnson & Thorne 1969) or by the application of carbon coatings (Thorne 1972). On the other hand, strengths can be reduced by more severe oxidation treatments due to the development of pits in the fibre surfaces.

It has also been reported that carbon fibre strengths can be increased by neutron irradiation (Cooper & Mayer 1971; Jones & Peggs 1972). Both tensile modulus and strain to failure can be significantly increased and the contributing mechanisms are thought to include both the pinning of dislocations and the closing-up of structural porosity (see Reynolds & Sharp 1974).

3. EFFECTS OF IMPURITY PARTICLES

(a) *Experimental work*

The effects of impurities in the PAN precursor fibres on the strengths of carbon fibres have been investigated using a laboratory wet-spinning apparatus (figure 2) installed in a clean room (Moreton & Watt 1974). The apparatus was positioned in front of a bank of five laminar

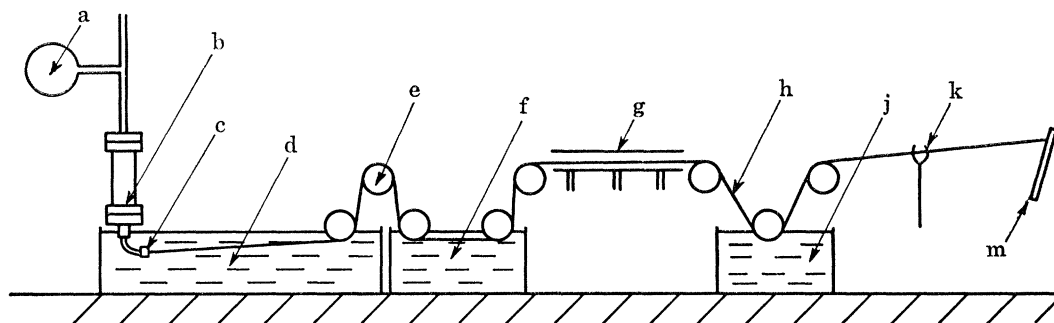


FIGURE 2. Apparatus for spinning PAN fibres: a, argon pressure gauge; b, PAN spinning solution; c, stainless steel spinneret; d, coagulation bath; e, glass roller; f, wash bath at 50 °C; g, steam stretch tube; h, PAN fibre; j, wash bath at 30 °C; k, traversing device; m, collecting frame.

flow air filters fitted along the 5 m length of the room. These units provided a class 100 clean air environment for the fibre spinning, i.e. less than 100 particles/ft³ ($3.5 \times 10^3/\text{m}^3$) of size $\geq 0.5 \mu\text{m}$ with none greater than 5 μm . Control experiments were also carried out by using a duplicate apparatus under normal laboratory conditions where the particle count was about $10^7/\text{m}^3$, four orders of magnitude greater than in the clean room. The following samples of PAN precursor fibre, stretched $\times 14$ in steam, were evaluated: (a) filtered-clean: spun from a filtered (1.5 μm) spinning solution in the clean room; (b) filtered-normal: spun from a filtered (1.5 μm) spinning solution under normal laboratory conditions; (c) unfiltered-clean: spun from unfiltered spinning solution in the clean room; (d) unfiltered-normal: spun from unfiltered spinning solution under normal laboratory conditions.

Figure 3 shows the effect of heat-treatment temperature on the average single fibre strengths of the samples, after they had been converted to carbon fibre by using a 5 h constant length oxidation treatment at 220 °C and carbonization in nitrogen. It is clear that impurity particles in the PAN fibre have a major effect on the strengths of the carbon fibres, and that the strength decrease normally found after high temperature heat treatments can be avoided by the use of clean precursor fibre. It is also worth noting that the unfiltered-clean sample gave carbon fibres with higher strengths than the filtered-normal sample after carbonizing to 1000 °C and after heat treatment to 1400 °C, but not after treatment to 2500 °C. This suggests that the strengths are largely limited by surface flaws after heat treatments to intermediate temperatures with internal flaws becoming more important after heat treatments to high temperatures. Thus the samples spun under clean conditions showed somewhat less pronounced gauge

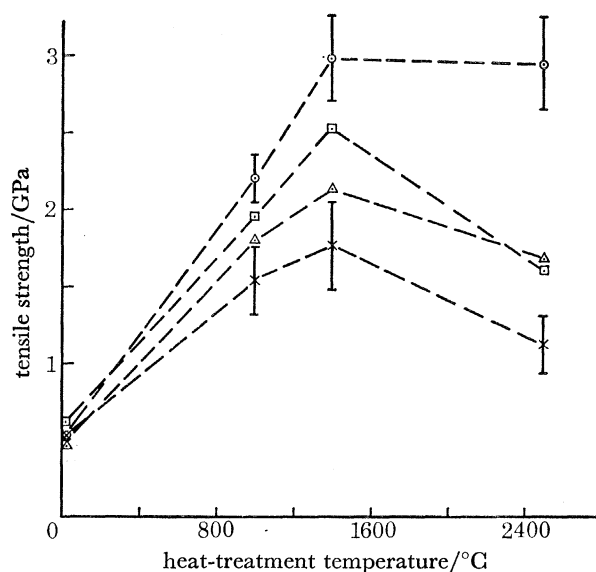


FIGURE 3. Effect of heat-treatment temperature on the tensile strengths of fibre samples. \circ , Filtered-clean; Δ , filtered-normal; \square , unfiltered-clean; \times , unfiltered-normal; gauge length, 5 cm; $\bar{\text{I}}$, 95% confidence limits.

length effects than the samples spun under normal laboratory conditions after heat treatment to 1400 °C (figure 4) while the samples spun from filtered solutions showed smaller gauge length effects after heat treatment to 2500 °C (figure 5).

The effects of particles of carbon black, silica and ferric oxide, as surface impurities, have also been studied by using samples of PAN fibre spun from a filtered spinning solution in the clean room. The samples were deliberately contaminated at the final stage of the spinning process by allowing 0.1% (by mass) suspensions of the impurities in water to drip onto the fibres as they were being collected on the frames (see figure 2). Carbon black and silica were chosen since analysis of airborne dust samples collected near the spinning equipment had indicated that particles of soot, silica and silicates were common constituents (together with various types of organic material). Ferric oxide was included since compounds containing iron are common dust particles that may be present near commercial spinning plants. Figure 6 shows that the contaminants caused decreases in the average strengths of the carbon fibres with the reductions relative to the clean fibre sample becoming more pronounced with increasing

STRENGTHS OF CARBON FIBRES

455

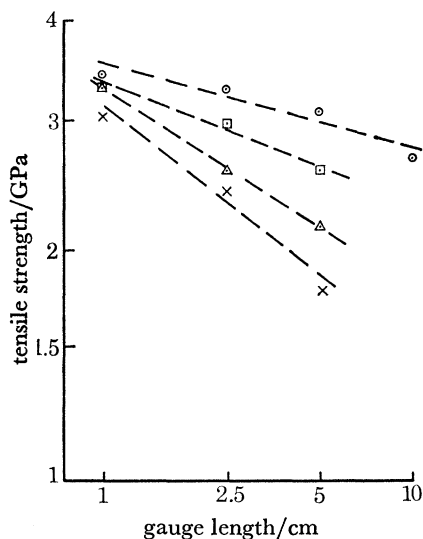


FIGURE 4

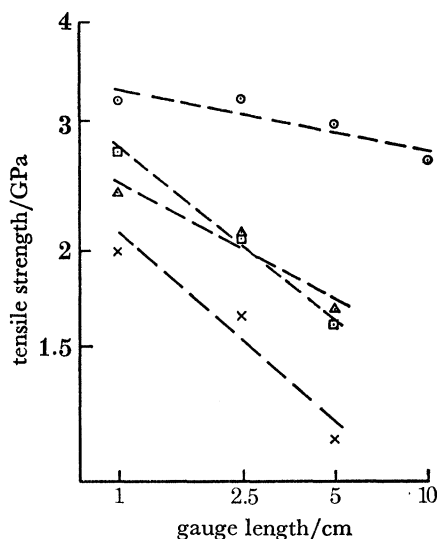


FIGURE 5

FIGURE 4. Effect of length on the tensile strengths of fibre samples heat-treated to 1400 °C. \circ , Filtered-clean; Δ , filtered-normal; \square , unfiltered-clean; \times , unfiltered-normal.

FIGURE 5. Effect of length on the tensile strengths of fibre samples heat-treated to 2500 °C. \circ , Filtered-clean; Δ , filtered-normal; \square , unfiltered-clean; \times , unfiltered-normal.

heat-treatment temperature. Thus the flaws due to the impurities continued to develop in severity over a wide temperature range. The individual fibre strengths for the contaminated samples showed more scatter than was found with the clean sample. For example, after the heat treatment to 2500 °C the coefficients of variation for the samples contaminated with silica and ferric oxide were 53 % and 44 % respectively, whereas the corresponding figure for the clean fibres was only 22 %. The Young moduli of the samples discussed here showed only small differences at each heat-treatment stage, and so variations in the preferred orientation

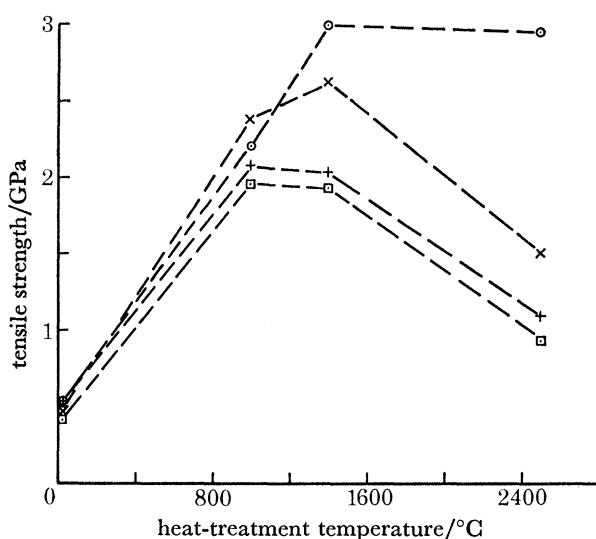


FIGURE 6. Effect of heat-treatment temperature on the tensile strengths of deliberately contaminated samples of fibre. \circ , Filtered-clean fibre; \times , contaminated with carbon black; +, contaminated with silica; \square , contaminated with ferric oxide; gauge length, 5 cm.

of the samples are unlikely to have affected the tensile strengths. For example, after the heat treatment to 2500 °C the Young moduli of the samples were all within 10% of the overall average value, 412 GPa. Mechanisms of flaw development from impurity particles are discussed in § 3(c).

(b) *Fractography*

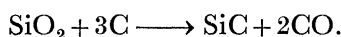
The fracture surfaces of individual carbon fibres broken in tension have been investigated with a scanning electron microscope (Moreton 1976). Generally carbon fibres derived from PAN fibres that had been spun from filtered spinning solutions failed at surface flaws which often appeared to be pits in the fibre surfaces. Some broken fibres showed a pit on one fracture surface with a matching surface impurity particle visible on the second fracture surface. When the PAN precursor fibre was spun in the clean room, fractures were still initiated by surface flaws, and it seems likely that these were also caused by impurities. Even when stringent precautions were taken to avoid contamination the complete elimination of impurities < 1 µm in diameter could not be achieved.

Carbon fibres derived from PAN fibres spun from unfiltered spinning solutions failed either at internal flaws (cavities) or at surface flaws after heat treatment to 1400 °C. After heat treatment to 2500 °C, however, the fibres almost invariably failed at cavities within the fibres. These were probably due to the volatilization of impurities as suggested by Sharp & Burnay (1971).

Examinations of the fracture surfaces of deliberately contaminated fibres confirmed that particles of silica and iron oxide can give rise to surface flaws. Fractures in fibres contaminated with carbon black, on the other hand, did not appear to be associated with the carbon black, which must have contained other impurities since it had an ash content of 0.2%.

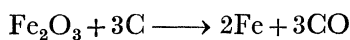
(c) *Development of flaws from impurities*

It is clear from the results described in § 3(a) that flaws due to impurity particles, such as inorganic oxides, are the main factor limiting carbon fibre strengths. The first stage of flaw development from an oxide particle is probably a chemical reduction reaction that takes place during the carbonization and heat-treatment stages. For silica particles, for example, the overall reaction is as follows:

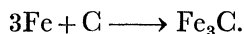


Humphrey *et al.* (1952) have studied this reaction and they calculated that the equilibrium pressure of CO was about 60 kPa (0.6 atm) at 1400 °C rising to atmospheric pressure at 1472 °C. Thus the heat treatment to 1400 °C of the silica-contaminated sample must have produced a reaction giving rise to small surface pits with either SiC or partly reduced SiO₂ particles within them.

Ferric oxide may have produced flaws by the reactions



and



However, Fe₂O₃ can be reduced at temperatures below 1000 °C (Dannatt & Ellingham 1948) and it is known (Watt *et al.* 1974) that gases such as hydrogen, carbon monoxide and methane are evolved during carbonization. Thus some reduction of the Fe₂O₃ particles by the pyrolysis gases is also likely to have occurred.

The strengths of carbon fibres from the samples contaminated with silica and ferric oxide both decreased over a wide range of heat-treatment temperature (see figure 6) and it is unlikely that this was entirely due to chemical reactions, particularly in the case of the ferric oxide particles where the reduction was probably completed below 1000 °C. A second stage of flaw development probably occurred because the impurities promoted the formation of localized randomly oriented three dimensional graphite within the existing fibre structure. It has been shown elsewhere that the walls of some cavities in carbon fibres consisted of lamellar three dimensional graphite (Sharp & Burnay 1971). It is also known that the bulk graphitization of carbon can be accelerated by additions of metals such as nickel, cobalt, iron and silicon (Ishikawa & Yoshizawa 1963; Baranieski *et al.* 1969). The mechanism is thought to involve the dissolution of amorphous carbon into the impurity particle, followed by the precipitation of graphite at temperatures where the metal carbide is unstable. It is evident that such processes will cause a large change in the local population of crystal defects, such as dislocations, tilt and twist boundaries, stacking faults and, possibly, microcracks, which could well contribute to the observed decrease of tensile strength with heat treatment temperature. The development of flaws due to both chemical reaction and localized graphitization probably occurs with a number of oxides, in view of the number of metals that can promote graphitization. Other types of impurity in addition to oxides are also likely to cause flaws; oxides have been investigated since they were frequently found when laboratory dust samples were analysed. Impurity particles within the fibres also result in cavities after heat treatments to high temperatures (Sharp & Burnay 1971). In some cases the gas pressure that develops during the heat treatment is sufficient to cause plastic deformation and localized swelling of the fibres. Some flaws have also been found where the outer surface of the fibre had collapsed into the cavity; this may have been due to a drop in gas pressure within the cavity as the fibre was cooling after the heat treatment.

4. A FRACTURE CRITERION

Since fractures in carbon fibres are initiated at isolated flaws it is possible that the fibre strengths might be described in terms of the Griffith brittle fracture equation. Sharp & Burnay (1971), the first to adopt this approach, calculated that the critical crack lengths in fibres with a Young modulus of 350 GPa should be about 0.1–0.3 μm , assuming a value of 4.2 J m^{-2} for the surface energy of graphite. These crack lengths were about $\frac{1}{10}$ of the size of the internal flaws (cavities) found in their fibres so they concluded that the failures must have been initiated at cracks in the surrounding carbon. Sharp & Burnay supported this conclusion by showing that the failures did not necessarily occur at the largest cavity in the fibre, and also by showing that the cavity walls were made up of lamellar three dimensional graphite markedly different from the turbostratic carbon fibre structure. Whitney & Kimmel (1972) on the other hand used the Griffith equation to calculate apparent surface energy values for batches of fibre heat-treated to 1500 °C. They estimated flaw sizes by examining fracture surfaces and they used the maximum dimension as the critical crack length in all cases. This led to high surface energy values, ranging between 14 and 55 J m^{-2} , as might have been expected from the work of Sharp & Burnay. Whitney & Kimmel, however, took the high surface energy values to indicate that plastic deformation had occurred at the crack tips during fracture. The surface flaws on some of the fracture surfaces of carbon fibres prepared under clean

conditions were in the 0.1–0.3 μm size range predicted by the Griffith equation. On the other hand the internal flaws (cavities) in some fibres heat-treated to 2500 $^{\circ}\text{C}$ were significantly larger than the Griffith values, as had been found by Sharp & Burnay.

Further electron microscopy failed to produce evidence of cracks, and a crystallite shear limit model of fibre fracture was therefore proposed by Reynolds & Sharp (1974). This attempted to relate intrinsic strengths to the strengths found in the presence of observable internal flaws, for fibres of all types ranging from isotropic carbon fibres to graphite whiskers. The model was based on the idea that the crystallites are weakest in shear on the basal planes and that failure occurred when the shear strain energy in a misaligned crystallite became sufficient to cause basal plane rupture. It was thought, however, that an initial flaw would only propagate to reach a critical size for fibre fracture if the local structure was sufficiently coherent with the misaligned crystallite. A crystal shear strain limit of about 20% was proposed to provide a consistent failure criterion for flaw-free fibres of different types and to give a general indication of the strengths likely to be found in fibres containing flaws.

Now that more experimental data are available for carbon fibres containing flaws, and for relatively flaw-free fibres, a further attempt can be made to develop a failure criterion based on the concentration of strain energy in misaligned crystallites. It is recognized that this approach may not account for all carbon fibre fractures (for example, surface pits due to oxidation can act as flaws) but nevertheless there is evidence that misaligned crystallites are commonly associated with fibre fractures, particularly after heat treatments to high temperatures.

(a) *Fibre modulus*

Before the fibre strength data can be analysed it is necessary to understand the effects of crystallite orientation on fibre modulus. Previous discussions of this relation have relied on graphite crystal data for the crystallite elastic constants, but this is not satisfactory for the fibres discussed here as they consist of ungraphitized turbostratic carbon, for which the elastic constants are not generally known. Much the most significant of these constants for the purposes of the present discussion is the shear compliance on the basal planes, s_{44} , which varies widely with the carbonization temperature or the interlayer spacing. Sufficiently accurate values of s_{44} can, however, be obtained from the measurements of fibre shear compliance S_{44} obtained by Dean & Turner (1973).

For an essentially linear structure such as a fibre, the longitudinal Young modulus E and the longitudinal shear modulus C_{44} are obtained to a good approximation by calculating the average compliance of a crystallite with basal planes at an angle ϕ to the fibre axis. With lower case letters applying to crystallites and upper case letters to fibres the appropriate expressions may be written:

$$1/E = S_{33} = s_{33} + (2s_{13} - 2s_{33} + s_{44}) I_3/I_1 + (s_{11} - 2s_{13} + s_{33} - s_{44}) I_5/I_1 \quad (1)$$

and

$$1/C_{44} = S_{44} = s_{44} + (3s_{11} + 2s_{33} - s_{12} - 4s_{13} - \frac{5}{2}s_{44}) I_3/I_1 - (2s_{11} + 2s_{33} - 4s_{13} - 2s_{44}) I_5/I_1, \quad (2)$$

where $I_n = \int_0^{\frac{1}{2}\pi} I(\phi) \sin^n \phi \, d\phi$ and $I(\phi)$ is the orientation function of the crystallites with $n = 1, 3$ or 5 .

The crystallite shear compliance s_{44} is much larger than any of the other compliances and for a highly oriented fibre $I_3/I_1 \approx I_5/I_1 \approx 1$. Hence from (2) to a first approximation

$$S_{44} \approx \frac{1}{2}s_{44}. \quad (3)$$

In the absence of information about the other constants the values adopted are those for the graphite crystal given in table 1.

TABLE 1. CRYSTALLITE ELASTIC COMPLIANCE CONSTANTS/GPa⁻¹

s_{11}	0.0011
s_{33}	0.033
s_{13}	-0.0025
s_{12}	-0.0005
s_{44}	-depends on heat-treatment temperature

If $I(\phi)$ is measured by X-ray diffraction techniques, calculated Young moduli can be compared with the values determined experimentally. The alternative approach used here however assumes that $I(\phi)$ can be put in the integrable form $I(\phi) = \sin^m \phi$ and involves finding values for m that give the correct values for E . The values of ϕ for which $I(\phi) = 0.5$ can then be compared with published values of ϕ_m , the measured half-width of the (002) X-ray diffraction peak. The results obtained are shown in table 2 and it is clear that the calculated ϕ_m

TABLE 2. CALCULATED AND OBSERVED PREFERRED ORIENTATION PARAMETERS FOR CARBON FIBRES

heat-treatment temperature/°C	Young† modulus/GPa	torsional‡ modulus/GPa	m	calculated ϕ	observed ϕ †		
					(1)	(2)	(3)
1000	197	26	13	18°	19°	14°	—
1500	245	24	21	15°	17°	12°	—
2500	428	14	107	7°	8°	8°	5°

† Taken from a survey by Hawthorne *et al.* (1970) with data from (1) Watt *et al.* (1966), (2) Blakelock & Lovell (1969) and (3) Johnson *et al.* (1969).

‡ Taken from Dean & Turner (1973).

values lie within the range of observed values for PAN-based carbon fibres of the appropriate Young moduli. As a consistency check the values of m in table 2 were used in equation (1) to recalculate the crystallite shear moduli. It was found that this produced no significant differences from the original estimates. The fact that equation (1) can be used to account for fibre modulus suggests the adoption of the uniform stress approximation to explain the fracture of fibres at misaligned crystallites.

(b) Fibre strength

To make estimates of the intrinsic strengths of carbon fibres with different Young moduli, in the absence of isolated flaws, it is necessary to know the intrinsic strength of one particular sample. This cannot yet be established with certainty, but the sample with the highest stored strain energy provides a minimum estimate of the required quantity, and hence a quantitative criterion for the failure of other fibres assuming that

$$\frac{1}{2}\sigma^2/E = K \text{ (a constant),} \quad (4)$$

where σ is the fibre strength.

The maximum value of $\frac{1}{2}\sigma^2/E$ found in this work was 38 MJ m^{-3} for a fibre with a strength of 4.0 GPa and a modulus of 210 GPa from the filtered-clean sample described in § 3. From this figure, maximum values of strength and strain to failure have been calculated for fibres heat-treated to different temperatures. These values can be compared with the observed strengths obtained for the clean sample in table 3. If fibres capable of sustaining a significantly higher strain energy are produced the calculated values will have to be increased accordingly.

TABLE 3. CALCULATED AND OBSERVED† CARBON FIBRE STRENGTHS (σ) AND BREAKING STRAINS (ϵ)

heat-treatment temperature/°C		σ/GPa		$\epsilon \%$	
		max	min	max	min
1000	calculated	3.90	1.50	2.00	0.75
	observed	2.45	1.54	1.36	0.86
1400	calculated	4.30	1.50	1.80	0.60
	observed	3.41	1.77	1.42	0.89
2500	calculated	5.70	1.30	1.30	0.30
	observed	3.13	1.13	0.83	0.29

† Observed average values for the filtered-clean and the unfiltered-normal samples (see § 3).

As discussed in § 3(c), impurity particles of various types produce local crystallite misorientations in carbon fibres and hence regions with a lower local modulus. The effective modulus E' of a crystallite with basal planes at an angle ϕ to the fibre axis can be calculated from the equation:

$$1/E' = s_{11} \cos^4 \phi + s_{33} \sin^4 \phi + (s_{44} + 2s_{13}) \sin^2 \phi \cos^2 \phi. \quad (5)$$

This expression has been used to compute minimum local modulus values for fibres heat-treated to different temperatures and these values are shown in table 4. Carbon fibres are likely to fracture at a misoriented crystallite since the combination of a low local modulus with a uniform fibre stress produces a localized strain energy concentration. On this basis minimum

TABLE 4. MINIMUM LOCAL MODULI AND THE CORRESPONDING MISORIENTATION ANGLES FOR CRYSTALLITES IN CARBON FIBRES

heat-treatment temperature/°C	$E'_{\text{min}}/\text{GPa}$	angle to fibre axis/deg
1000	30	74
1400	29	59
2500	22	53

values for carbon fibre strengths and breaking strains have been calculated, from equation (4), for fibres containing grossly misaligned crystallites. These values are included in table 3. Overall the results show that the observed range of carbon fibre strengths generally falls between the limits arrived at using the strain energy criterion except that for the high modulus fibres the observed limits are appreciably lower than those calculated. The calculations show that the minimum strength is only reached at high misorientation angles as compared with the average value characteristic of the fibre structure. Thus, since large flaws are rarely able to develop during heat treatments to 1000 and 1400 °C, fibres of more consistently high strength are produced. For high modulus fibres the shortfall in fibre strength may be associated with

a higher structural coherence in the fibres that enables flaws to propagate more easily. The constant stress approach used assumes that the stress on misaligned crystallites is equal to the fibre stress and this implies that large localized strains can occur prior to fracture. Such strains may be possible, however, since the fibres are known to contain elongated pores running parallel to the basal plane structure. Although there is encouraging agreement between the predicted strengths and those obtained experimentally, the detailed mechanism of failure and the significance of a stored strain energy of 38 MJ m^{-3} , have yet to be elucidated in terms of crystallite structure and dimensions and the possible roles of dislocations and grain boundaries or microcracks in the fracture process.

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Discussion

P. J. WORTHINGTON (*C.E.R.L., Kelvin Avenue, Leatherhead, Surrey*). What determines the heat-treatment temperature at which the strength of normally produced fibres begins to fall, and the heat treatment temperature at which no further increase in strength is observed in fibres produced from filtered spinning solutions under clean conditions?

W. N. REYNOLDS. According to our model, fibre strength is limited by the development of defects containing misoriented crystallites which appear to grow from particles of foreign matter. The critical temperature for maximum strength would therefore be determined by the associated chemical reaction rates.