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Influence of Matrix and Interface on the Mechanical Properties of Unidirectional Carbon/Carbon Composites

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Abstract

The mechanical properties and microstructure of unidirectional carbon/carbon (UD C/C) were investigated. The strength of one type of UD C/C, produced with an intermediate modulus fibre treated to four different levels of an oxidative surface treatment, was determined after each step of the production cycle (of loose, impregnated and carbonized impregnated fibre bundles). The impregnated bundle had a strength 1·9–4·3 times the strength of the loose bundle, whereas after carbonization the strength of the bundle dropped below the strength of the loose bundle. It is suggested that this is mainly caused by the formation of defects in the fibres due to the shrinkage process during carbonization. These defects are larger if a good fibre/matrix bond strength in the green material exists. The possibility that the low strength of carbon/carbon could be caused by stress–concentration effects was excluded with the aid of TEM investigations. They showed that the carbon matrix mainly consisted of vitreous carbon, the modulus of which (E = 35 GPa) does not become effective due to micro- and macrocracks.

Die mechanischen Eigenschaften und die Mikrostruktur von unidirektionalem kohlenstoffaserverstärktem Kohlenstoff (UD C/C) wurden untersucht. Die Festigkeit von einem Type UD C/C, hergestellt mit Hilfe einer mittel-moduligen Faser mit unterschiedlich starker Faseroberflächenbehandlung (vier Stufen), wurden nach jedem Schrit des Herstellprozesses ermittelt (von losen, imprägnierten und carbonisiert imprägnierten Bündeln). Das imprägnierte Bündel erreichte das 1·9- bis 4·3-fache der Festigkeit des losen Bündels während nach der Carbonisierung die Bündelfestigkeit unterhalb der des losen Bündels absank. Dies wird voraussichtlich hauptsächlich durch die Bildung von Fehlern in den Fasern als Folge der Schrumpfung während der Carbonisierung verursacht. Diese Fehler sind größer, wenn die Faser/Matrix Haftung im Grünkörper gut ist. Die Möglichkeit, daß die niedrige Festigkeit von C/C durch Spannungskonzentrationseffekte verursacht wird, konnte mit Hilfe von TEM-Untersuchungen ausgeschlossen werden. Sie zeigten, daß die Kohlenstoffmatrix hauptsächlich aus Glaskohlenstoff bestand, dessen E-Modul (E = 35 GPa) jedoch wegen der Mikro- und Makrorisse nicht wirksam wird.

On a étudié les propriétés mécaniques et la microstructure de carbone renforcé par des fibres de carbone (UD C/C). On a déterminé aprés chaque étape du cycle de production (du faisceaux de fibres libres, à imprégnés, et imprégnés et carbonisés) la résistance mécanique d'un type de UD C/C, produit avec une fibre de module moyen ayant subit quatre stades différents d'un traitement d'oxydation de surface. Le faisceau imprégné a une résistance mécanique de 1·9 à 4·3 fois supérieure à celle d'un faisceau libre, alors qu'aprés carbonisation elle est inférieure. Il est suggéré que ceci est surtout dû à la *formation de défauts dans les fibres lors du retrait linéaire lors de la carbonisation. Ces défauts sont plus importants s' il existe dans le matériau initial une forte liaison fibre/matrice. La possibilité que la mauvaise résistance mécanique soit due à des effets de concentration tension a été exclus par les observations au MET. Elles montrent que la matrice de carbone est essentiellement constituée d'une phase amorphe dont le module d'élasticité, qui est de 35 GPa, est rendu inefficace par les micro- et macro-fissures.*

1 Introduction

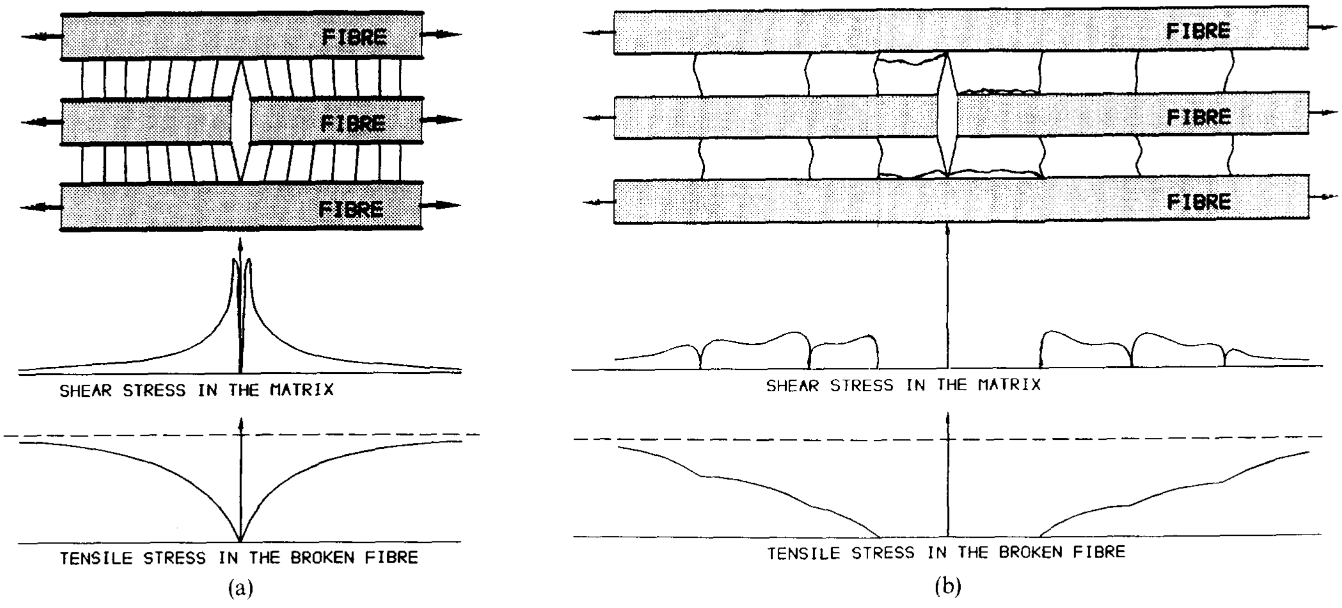
The influence of matrix and interphase on the strength of CFRPs (carbon fibre-reinforced plastics) has been investigated recently. One way to change the strength of the interphase is the application of a fibre surface treatment.[1,2] Due to a fibre surface treatment the fibre/matrix bond strength generally increases with a positive effect on many mechanical properties. Although not clarified in detail it can be concluded in general that CFRP has a high strength in fibre direction if the fibre/matrix bond strength is high. The dependency of the strength of CFRP as a function of the bond strength is usually complicated, due to the fact that the positive effect of the bond strength increase (realized through a surface treatment) is accompanied by a negative effect of the same surface treatment, which is a reduction of fibre strength. The influence of matrix and interphase (or interface) on the strength of carbon fibre-reinforced carbon (C/C) is even less clear. One of the reasons is that in most cases properties of 2-D or 3-D C/C are investigated which are additionally influenced by the damage pattern which develops in this material during processing.

In order to be able to study the influence of the constituents on the properties of C/C more easily, it is worthwhile to investigate the mechanical behaviour of 1-D; (or UD; unidirectional) C/C. This has been done with PAN-based[3] as well as pitch-based fibres.[4]

The development of the tensile[4] or flexural[3] properties as influenced by the different process steps generally shows a large drop in strength comparing the strength of the C/C material after the first carbonization cycle with the strength of the green (i.e. CFRP) material. This is believed to be caused by a certain state of damage in the carbon matrix, which reduces the ability to transfer stresses. This damage state causes the ineffective length (= the length at which the stress in a broken fibre is *re*introduced) for C/C to be larger than for CFRP (see Fig. 1). Assuming the load of broken fibres is equally shared by the remaining fibres (equal or global load sharing), Rosen[5] has shown that in this case the strength of bundles with a length of the ineffective length δ is decisive for failure. Due to the weakest-link type of failure, shorter fibres (and bundles) are stronger, so that a higher strength can be expected the shorter the ineffective length is.

This explanation is also given for the difference in strength of different C/C materials.[4] Low-modulus types of C/C develop their strength at a large strain, whereas high-modulus fibre C/C only require a small applied strain. Thus, the state of damage is larger in the matrix of the low-modulus C/C, giving rise to a lower strength (due to the larger ineffective length).

The influence of the fibre/matrix interphase on



(a) (b)

Fig. 1. (a) Efficient stress transfer near a broken fibre in CFRP and (b) reduced stress transfer in C/C due to matrix cracking and splitting.

the strength of C/C has not been investigated thoroughly. This is the aim of the present investigation, in which use is made of an intermediate-modulus fibre treated to four different levels of an oxidative surface treatment. Further, another type of unidirectional C/C material is also investigated, in the direction transverse to the fibres. The microstructure of this material is analysed by TEM to explain the influence of the matrix on the mechanical properties.

2 Materials and Experiments

Two types of unidirectional material were investigated, impregnated bundles and specimens cut from two unidirectional plates, with a thickness of 1·65 mm and 3·4 mm, respectively. Carbon/carbon bundles were produced with the aid of an intermediate-modulus fibre CG 43-750 produced by Courtaulds. This fibre was made available by Courtaulds with four different levels of surface treatment, designated 0% (untreated), 10%, 50% and 100% (the commercial treatment level). These 12 k fibre bundles were originally supplied in the scheme of a EURAM programme with the aim to investigate the influence of a surface treatment on fibre/matrix bonding and on the mechanical properties of CFRP. Results on fibre-surface analysis and the fibre/matrix bond strength in CFRP can be found in Ref. 6 and Refs 7 and 8, respectively. For the present investigation these differently treated fibres were impregnated with a phenolic resin. The impregnated bundles were cured in a vacuum bag under a pressure of 700 kPa and at stepwise increased temperatures with the maximum temperature measuring 180°C. Some of the cured bundles were subsequently carbonized in vacuum at a temperature of 1100°C. Tensile experiments were performed on the loose bundles, on the phenolic resin impregnated and on the carbon/carbon bundles.

The unidirectional carbon/carbon plates were produced with the aid of HTA fibres (produced by AKZO) and the same phenolic resin system as used for impregnation of the bundles. The filament-wound plates were cured according to the cure cycle applied for the impregnated bundles. The unidirectional plates were carbonized at 1100°C, reimpregnated with the phenolic resin and recarbonized (in total 4 times) and finally heat-treated at 2100°C. A cross-section of one of these plates (Fig. 2) reveals the damage present in these materials. The cracks and pores are partly filled during the reimpregnation process. From these unidirectional plates tensile specimens (dimensions 200 × 16 × 1·65 mm) and three-point bending specimens were cut, out of the thin plate with dimensions 80 × 10 × 1·65 mm and out of the thick plate with different lengths from 13·6 to 102 × 10 × 3·4 mm. Transverse to the fibre direction three-point bending specimens (dimensions 50 × 10 × 3·4 mm) and compression specimens (dimensions 17 × 10 × 3·4 mm) were carefully cut out of the thick plate with a diamond wire. Tensile experiments were performed in an Instron testing machine at a crosshead speed of 2 mm/min and the strain was measured with an extensometer with a gauge length of 50 mm. Three-point bending experiments were performed measuring the displacement at mid span with the aid of an LVDT. Transverse compression tests were performed between two parallel plates at a crosshead displacement of 0·5 mm/min. The compression

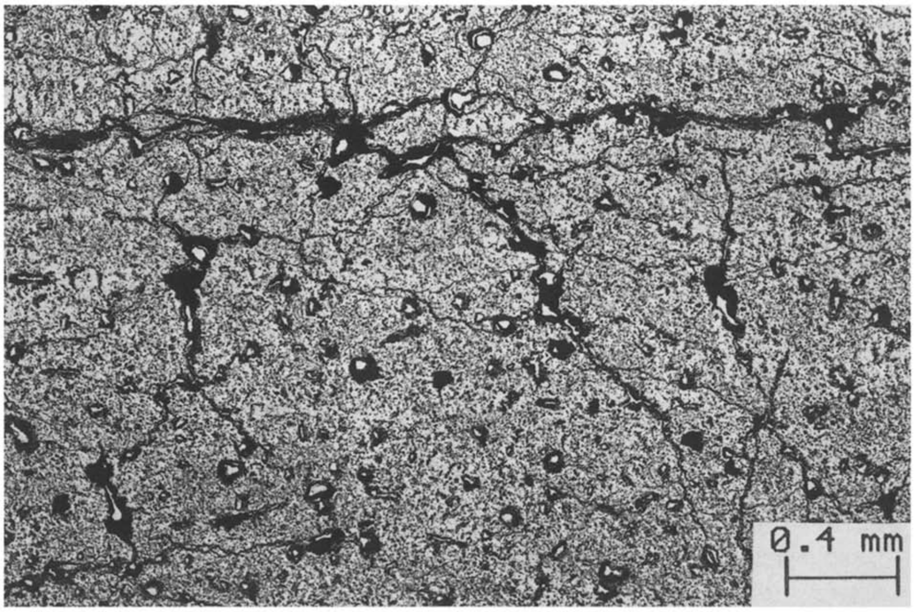


Fig. 2. Cross-section of unidirectionally reinforced carbon/carbon reimpregnated (4 times) and heat treated at 2100°C.

surfaces of the specimens were made parallel to each other by surface grinding, thus assuring the introduction of homogeneous compressive stresses. The strain in these specimens was measured with an extensometer with a gauge length of 10 mm.

The microstructure of the graphitized unidirectional C/C was investigated by transmission electron microscopy (TEM). A Philips EM 430 T electron microscope operating at 300 kV equipped with a LaB_6-cathode was utilized for this study. TEM specimens of C/C were prepared by a combination of grinding, dimpling and ion-thinning techniques resulting in transparent areas sufficiently large for the TEM work.

The negatives taken with the TEM were analysed quantitatively by a digital image processing system (Semper image-processing software, Synoptics Ltd, Cambridge UK; resolution of the system used: 512 × 512 pixels) to obtain the distribution of the different carbon variants.

3 Results

3.1 Loose, impregnated and carbonized bundles

The loose bundles were loaded in tension during which the displacements of the grips were registered on averaging 2 LDTV signals (one LDTV on the left, one on the right-hand side of the bundle). Typical load–strain curves for the differently treated fibres are given in Fig. 3. The fibre surface treatment obviously reduces the fibre strength, leading to a decrease of the maximum load at increasing level of surface treatment. It has been shown[9,10] that, with the aid of the load–displacement curve of a loose bundle tensile test, the single-fibre strength distribution can be determined. The procedure followed is that at any displacement, the ratio y/y_1, with y for the actual load carried by the fibre bundle and y_1 for the load carried if all fibres are unbroken (y_1 on extrapolation of initial linear load–displacement curve), can be described by

$$y/y_1 = 1 - F = \exp -(\sigma/\sigma_0)^a \qquad (1)$$

assuming the fibre strength distribution can be described with the aid of a two-parameter Weibull distribution. In eqn (1) F is the probability of failure at a stress of σ, while the two Weibull parameters are the shape parameter a and the scale parameter σ_0. Thus, from the load–displacement curve the two Weibull parameters can be determined, making use of the data points transferred to values of $\ln(-\ln(1 - F))$ and $\ln \sigma$, applying linear regression and least-squares analysis. The resulting deduced data (values of a and σ_0 in Table 1) for the bundle experiments indicated in Fig. 3 are depicted in Fig 4. Further the mean stress at maximum load (mean value of five experiments) for the differently treated fibre bundles are given in Table 1.

Table 1. Fibre strength distribution and mean strength of five loose, 14 impregnated and six carbonized impregnated bundles as a function of level of fibre surface treatment

	0% Treated	*10% Treated*	*50% Treated*	*100% Treated*
Fibre strength distribution				
σ_0, MPa	3 417	3 028	2 356	2 183
a	6·2	5·0	2·2	2·6
UTS, MPa				
Loose bundle	2 226	1 580	1 136	1 009
Impregnated bundle	4 333	4 530	4 798	4 320
Carbonized impregnated bundle	1 595	1 170	728	613

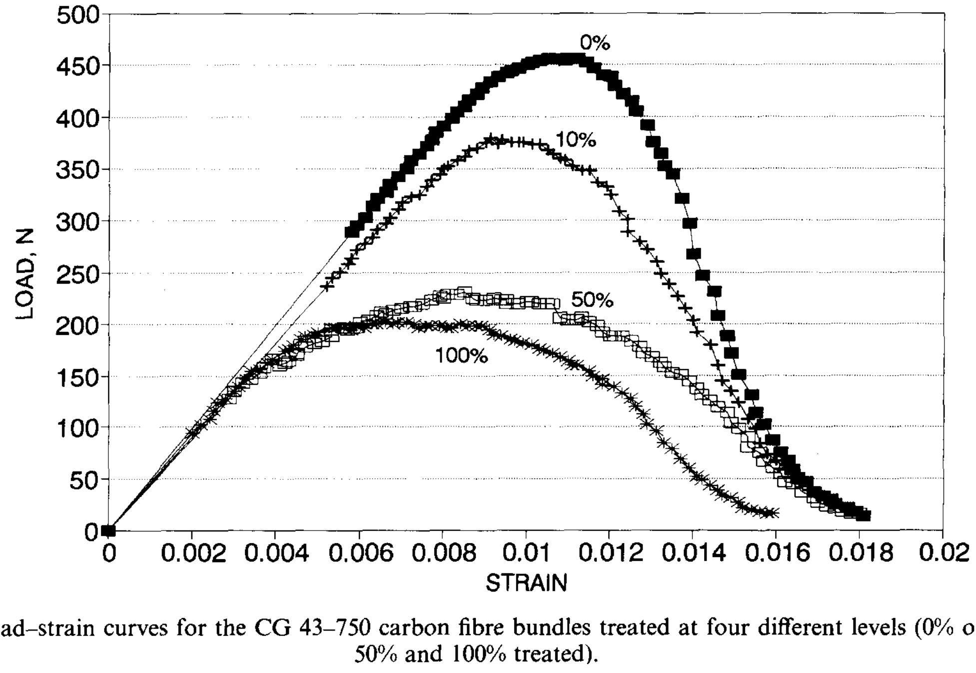


Fig. 3. Typical load–strain curves for the CG 43–750 carbon fibre bundles treated at four different levels (0% or untreated, 10%, 50% and 100% treated).

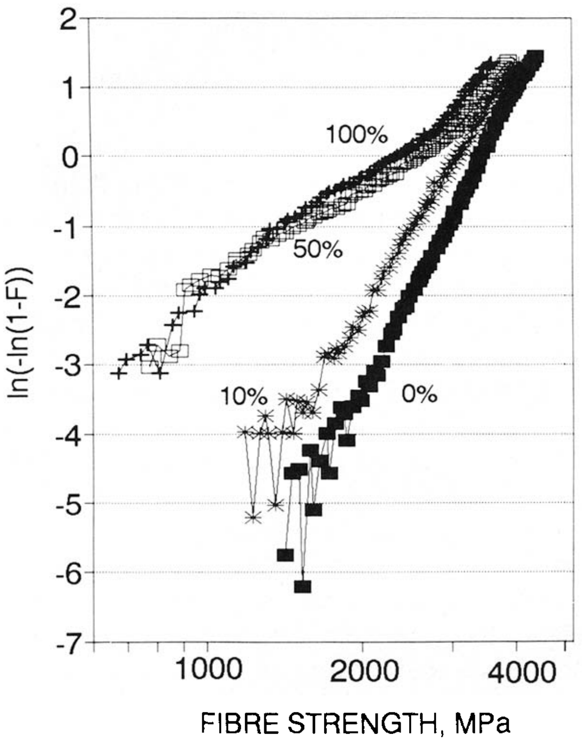


Fig. 4. Weibull distributions for the (single) fibre strength deduced from the load–strain curves of the differently treated fibres.

The strength of the impregnated bundles (= maximum load divided by fibre cross-sectional area of 0·2062 mm^2) is clearly higher than those of the loose bundles. The mean strength of 14 tested impregnated bundles is given in Fig. 5 and Table 1. Figure 5 shows that, although the strength of the loose bundles decreases at increasing level of surface treatment, the strength of the impregnated bundle initially increases up to a level of surface treatment of 50% before finally a decrease takes place.

Six impregnated bundles per level of surface treatment were carbonized and after this tested in tension. The mean strengths (= maximum load divided by fibre cross-sectional area) are indicated in Table 1 and Fig. 5. Figure 5 shows that the strength of C/C bundles drops below the loose bundle strength and that the dependency of its strength on the level of surface treatment closely follows those of the loose bundles. Fracture of the C/C bundles with untreated and 10% treated fibres takes place accompanied by substantial splitting (brush-type failure), whereas the fracture surface of the C/C bundles with 50% and 100% treated fibres is flat (no splitting). A SEM graph of typical fracture surface of a C/C bundle with 100% treated fibres is indicated in Fig 6. Fracture of the fibres appears to be initiated near points of contact between the different fibres. The fracture mirror indicates that fibre fracture occurs starting from a site (probably a defect) at the fibre surface.

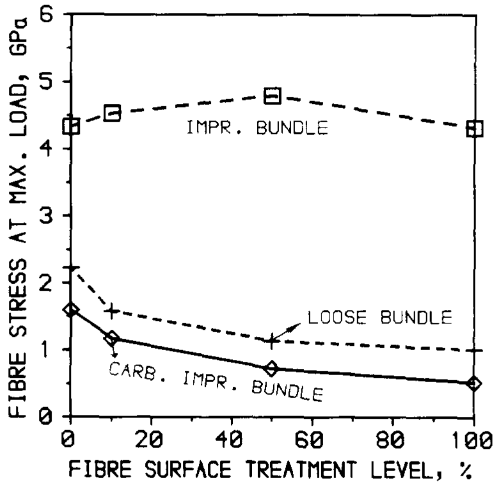


Fig. 5. The mean strength of loose bundles, phenolic resin impregnated bundles and carbonized phenolic resin impregnated bundles as a function of the level of surface treatment.

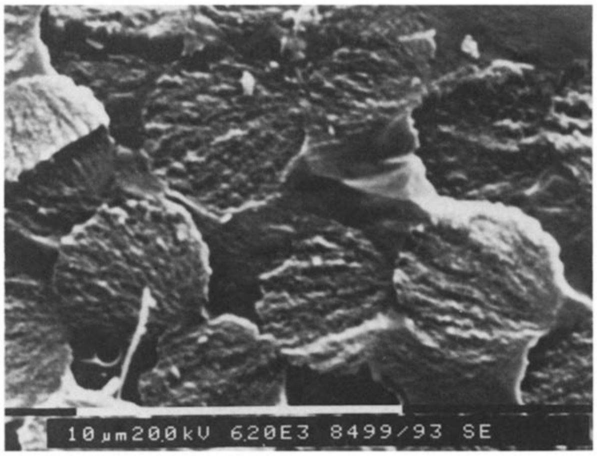


Fig. 6. Typical fracture surface of a C/C bundle with the 100% treated fibre.

3.2 Tensile and three-point bending testing of unidirectional C/C

Tensile experiments on the 1·65-mm thick unidirectional material were performed in the fibre direction. Because the material failed in the gripping range no reliable values concerning the tensile strength can be given. In terms of tensile modulus it was found that the modulus of unidirectional material increased from E_{11} = 140 GPa for the green (phenolic resin impregnated) material to E_{11} = 251 GPa. If this modulus is completely assigned to the fibre, the fibre modulus thus increased during processing from the original E_f = 248 GPa to the value of E_f = 480 GPa (assuming a fibre volume of V_f = 60 %).

Three-point bend specimens cut in the fibre direction were loaded up to failure. Specimens with a different span L were tested. The specimens with a short span (L = 13·6 mm, thickness 3·4 mm) showed shear failure, whereas the specimens with a large span (L = 80 mm, thickness 1·65 mm) showed fibre failure. For each specimen the maximum shear stress was calculated with

$$\tau_{\max} = \frac{3P_{\max}}{4BW} \tag{2}$$

and the maximum tensile stress $\sigma_{\max}$ was calculated with:

$$\sigma_{\max} = \frac{3P_{\max}L}{2B^2W} \quad (3)$$

where $P_{\max}$ is the maximum load and B and W are the specimen thickness and width respectively. The values of $\tau_{\max}$ and $\sigma_{\max}$ are given in Fig. 7 as a function of the span to thickness ratio. The shear stress for the shortest span measured as much as 22 MPa, whereas the maximum tensile stress reached a rather high value of 1039 MPa for $L/B = 49$.

Three-point bend experiments performed transverse to the fibre direction reflected the low strength of the carbon matrix. The strain at failure (at maximum load) was calculated with the aid of:

$$\varepsilon_{\max} = \frac{6DB}{L^2} \quad (4)$$

where D, B and L are the deflection of the specimen at mid span, the width and span respectively.

The modulus transverse to the fibre direction was calculated with the aid of:

$$E = \frac{L^3m}{4WB^3} \quad (5)$$

where m is the slope of the load–displacement curve. In all cases the transverse fracture strain measured between 0·1 and 0·15%, whereas the modulus was 3–3·5 GPa. The modulus was expected to be influenced by the (micro) cracks which were present in the material (see Fig. 2). For this reason some additional transverse compression tests were performed on 17-mm long specimens. The specimens were loaded between two flat parallel surfaces as described before. Now transverse moduli with a maximum value of $E_{22} =$ 4·7 GPa were measured.

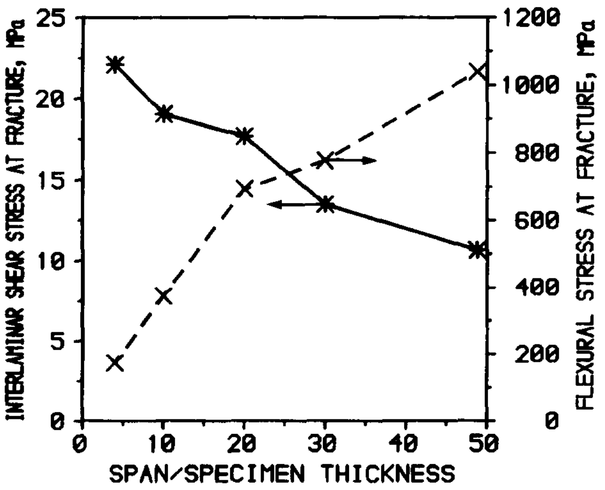


Fig. 7. The three-point-bend interlaminar shear stress and the flexural stress at failure of unidirectional C/C as a function of the relation specimen span to specimen thickness.

3.3 Microscopical structure of unidirectional graphitized C/C

In contrast to other composite materials the matrix of C/C is formed *in situ* during processing. So one can expect a strong interaction between the carbon fibres and the matrix. More than about 1 μm away from the fibre the matrix has a structure similar to vitreous carbon. This is the normal type of carbon for a carbonized 3-D interlaced thermoset resin like phenolic resin. At the fibre/matrix interface, however, the matrix is transformed to graphite. This is caused by shrinkage stresses in the matrix during pyrolysis which are more important if a strong fibre/matrix bond exists. During pyrolysis a stress of only about 20 MPa is sufficient for a structural change to graphite.[11]

Besides vitreous carbon and stress-oriented graphite a third phase is found in the matrix carbon. Its structure is similar to vitreous carbon, but it has a mean pore size which is one order of magnitude higher (40 nm instead of 4 nm). The conditions of development of this foam-like vitreous carbon are not exactly known yet.

Table 2 gives a view of the quantitative composition of unidirectional C/C. Vitreous carbon is the dominant phase in the matrix. Further, less than 10% of the matrix volume fraction is transformed into graphite. The stress-oriented graphite is not distributed evenly but located at the interface between fibre and matrix. Figure 8a shows the importance of the arrangement. Due to the low shear strength of graphite microcracks are formed under critical stresses as a result of either the production cycle or specimen preparation. TEM investigations on this graphitized C/C further showed practically touching fibres with topographic identical surfaces over a larger region (see Fig. 8b).

The thickness of the intermediate graphite layer is nearly constant. These are indications that locally, due to a high pressure, the surface area of the fibres can be deformed and transformed into graphite.

The cause of this structural transformation again must be sought in the shrinkage stresses. During processing the pure matrix shows a shrinkage of about 50%.[12] This shrinkage is hindered by

Table 2. Quantitative composition of unidirectional C/C

Phase	*Volume fraction (%)*
Fibre	55–65
Pores	3–7
Vitreous carbon	30–35
Stress-oriented graphite	1–3
Foam-like vitreous carbon	0·3–0·5

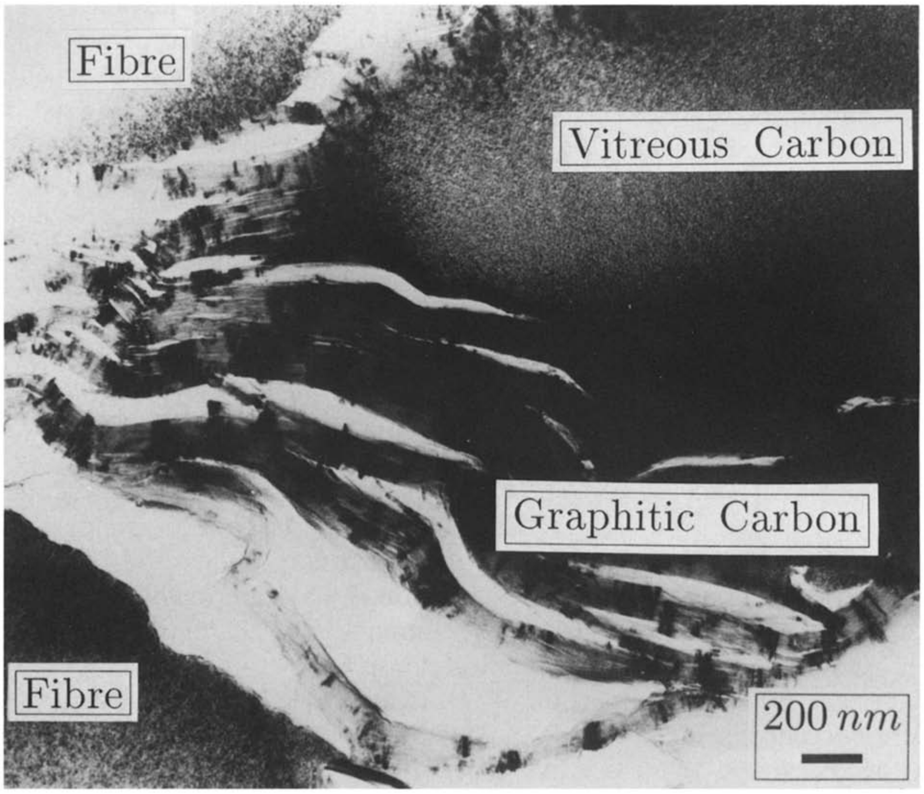


Fig. 8a. Transverse cross section of unidirectional C/C. The fibres are surrounded by stress-oriented graphite which split into smaller crystallites easily. Thus cracks are deflected around fibres and vitreous carbon.

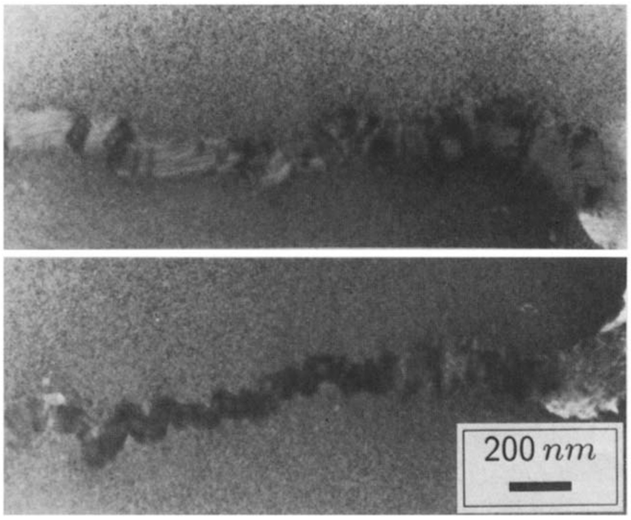


Fig. 8b. Two examples of closely packed fibres with equidistant surfaces.

the presence of the fibres, leading to shrinkage stresses in fibres and matrix. Unlike the resin the strong fibres cannot reduce the load by plastic deformation if they are pressed on each other. A pressure higher than 500 MPa can be built up this way. This is sufficient for the transformation of carbonized vitreous carbon into graphite.[13]

4 Discussion

Figure 5 makes clear that the strength of the impregnated bundle is considerably larger than that of the loose bundle. If use is made of another polymer, an even larger increase can be found, e.g. by impregnation with bisphenol-A.[14] The strength increase is based on the phenomenon that, in contradiction to the loose bundle where a broken fibre is completely lost, the stress in broken fibres of an impregnated bundle is reintroduced by shear stresses in the matrix. The ratio of impregnated bundle strength to loose bundle strength rising from 1·9 for the untreated fibre bundles to 4·3 for the 100% treated fibre bundles reflects the increasing bond strength (leading to a reduced ineffective length) at an increasing level of surface treatment. Because the stress in the broken fibre approaches the undisturbed stress asymptotically at infinity, the ineffective length is usually chosen to be that length at which 90% of the undisturbed stress is reached. Assuming the load of broken fibres is equally distributed among the remaining fibres (equal or global load sharing) the strength of unidirectional composites can be predicted, based on the strength of an element with a length twice the ineffective length.[6] In this approach the ineffective length δ is calculated with a shear lag analysis (assuming the fibre/matrix bond is ideal):

$$\frac{\delta}{d_f} = \frac{1}{2}\left[\frac{(1 - V_f^{1/2})}{V_f^{1/2}}\left(\frac{E_f}{G_m}\right)\right]^{1/2} \ln\left(\frac{1}{1 - \varphi}\right) \tag{6}$$

where d_f is the fibre diameter ($d_f = 5$ μm), V_f the fibre volume ($V_f = 0{\cdot}60$), E_f the tensile modulus of the fibre (here $E_f = 305$ GPa) and G_m the shear modulus of the matrix (estimated to be $G_m = 1{\cdot}3$ GPa). The factor φ is the proportion of the undisturbed stress reached at the border of the ineffective length (here $\varphi = 0{\cdot}9$, whereas $\varphi = 1{\cdot}0$ for $\delta = \infty$). For the impregnated bundles the ineffective length according to eqn (6) measures $\delta = 0{\cdot}1$ mm. Due to non-elastic effects (e.g. debonding occurring around a single or groups of broken fibres) the ineffective length in CFRP is of a length of $\delta \sim 0{\cdot}5$ mm. This short ineffective length leads to the theoretical maximum strength of CFRP.[15] A prediction of the strength of the impregnated bundles based on this short ineffective length was found to be inaccurate.[16] This is probably based on the assumption made, that the strength for this short fibre length could be estimated with the aid of the Weibull distribution determined on fibre bundles with a length of 80 mm.

The strength of C/C bundles can be expected to be smaller than that of impregnated CFRP bundles. This is the result of the ineffective length for C/C being larger than for CFRP (see Fig. 1). According to the theory of equal load sharing, the strength of the loose bundle should, however, be a lower bound for the strength of C/C. For all surface-treatment levels, however, the strength of C/C bundles is lower than that of the loose bundles. This can be caused, in principle, by two effects:

- —a stress concentration effect (making the theory of equal load sharing invalid),
- —a degradation of the fibre due to the carbonization process.

It is believed that larger stress concentrations are not effective in the C/C bundles. Larger stress concentrations are possibly due to the use of a high modulus matrix, high in comparison with the matrix modulus of CFRP which measures $E = 3$–4 GPa. The present material with the carbon matrix was shown to consist mainly of vitreous carbon with, in principle, a higher modulus ($E = 35$ GPa). The low transverse modulus of the unidirectional C/C is, however, an indication that the matrix is not effectively loaded. This is caused by the micro- and macrocracks and the probably loose bond between the different types of carbon in the matrix. This inability to transfer stresses also excludes the existence of larger stress concentrations.

The remaining factor responsible for the low strength of the C/C bundles thus must be the degradation of the fibre strength during carbonization. Interesting in this respect is the ratio of the strength of C/C bundles to the strength of the loose bundles as a function of the level of surface treatment. This ratio measures 0·75 for the material with untreated fibres and decreases to 0·60 for the material with treated fibres. Further, it was reported that the fracture surface of C/C with 0% and 10% surface-treated fibres showed substantial splitting in fibre direction, whereas C/C with 50% and 100% surface-treated fibres showed a flat fracture surface (Fig. 6). From this fracture surface one could conclude that the strength of the interface is high. This conclusion is, however, wrong, due to the fact that the matrix poorly transfers stresses. For this reason the cause of the flat fracture surface must be explained differently. The fracture surface of an unidirectional fibre-reinforced material tends to be flat in the case where:

- —The bond strength is high,
- —The defects in the fibre do not differ strongly in size (in other words, the shape parameter is large).

This last aspect now appears to govern the appearance of the fracture surface. Interestingly, the lowest strength is found for those C/C bundles which have an improved bond strength before carbonization. Thus a good bond strength in the green (CFRP) state seems to reduce the strength of the resulting C/C material. A possible explanation for this behaviour can be found in the stresses locally exerted on the fibres, due to the strong shrinkage of the matrix during the carbonization process. If the fibre/matrix bond strength is high, these shrinkage stresses are more rigorously transferred locally to the fibre causing formation of defects. These defects seem to be formed especially in areas where there is practically complete contact between the fibres. This is visible from Fig. 6

which shows the fibres to fracture starting from surface defects in areas of close fibre/fibre distance.

The properties of the unidirectional C/C material, however, show that during further processing steps (reimpregnation, recarbonization and graphitization) a high strength in C/C in the fibre direction can be realized. The main reason for this high strength is the change of the structure of the carbon fibre from a less orientated (HT-type) to a more oriented one (HM-type). The possibility that the defects produced during carbonization (partly) heal out during the graphitization process will be investigated in future programmes.

5 Conclusions

Microscopical investigation of unidirectional C/C graphitized material showed that the main part of the carbon matrix consisted of vitreous carbon (E = 35 GPa). Nevertheless the matrix dominant transverse modulus of this material measured only 3–4 GPa. This low value can be explained by the fact that the vitreous carbon does not effectively transfer the stresses. This is caused by micro- and macrocracks and a loose bond between the different types of carbon in the carbon matrix.

This poor ability to transfer stresses was one of the reasons for the relatively poor strength of (only carbonized) C/C bundles. The main reason, however, is a degradation of the fibre strength during carbonization, which seems to be most severe if the fibre/matrix bond strength in the green state (with polymer matrix) is high. This is probably caused by shrinkage stresses in the matrix which, due to the good bond, are transferred to the fibre, leading to defects. The unidirectional graphitized C/C showed a high flexural strength, indicating that due to the further processing steps (reimpregnation, recarbonization and graphitization) the strength increases again. The main contribution is expected to arise from the graphitization process, which improves the properties of the fibre. The possible healing effect on the defects produced during carbonization is under investigation.

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