Hybrid composites based on polyethylene and carbon fibres

Part 6: Tensile and fatigue behaviour

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The tensile and fatigue behaviour of unidirectional carbon–high-performance polyethylene/epoxy hybrid composites has been studied, including the effect of hybrid design and surface treatment of the high-performance polyethylene (HP-PE) fibres. Results indicated that the tensile behaviour of carbon–HP-PE hybrids in both monotonic and fatigue testing can be interpreted, adopting the conventional 'constant strain' model for hybrid composites. Deviations from this constant strain model, so-called hybrid effects, were observed in monotonic tensile testing for those hybrid systems with the highest degree of fibre dispersion, incorporating either untreated or treated HP-PE fibres, whereas only the latter displayed synergistic fatigue performance. Hybrid effects under tensile loading conditions were in reasonable agreement with calculations accounting for statistical effects and stress concentrations as determined by finite element analyses.

Key words: composite materials; hybrid; polyethylene fibre; carbon fibre; epoxy; hybrid effect; tension; fatigue; adhesion

Advanced carbon fibre-reinforced composites possess desirable characteristics of high specific strength and stiffness, as well as superior long-term properties including fatigue resistance. However, due to their brittle nature, these materials are rather susceptible to impact damage. Consequently, material design concepts for enhanced damage tolerance of these materials are of great interest. One such concept is hybridization with high-performance polyethylene (HP-PE) fibres. Depending on the volume fractions of the components, hybrid design and interface properties, it is possible to tailor the properties of such hybrids to specific applications requiring either structural performance, damage tolerance or energy absorption.

Besides damage tolerance under impact conditions, hybridization also offers interesting possibilities with respect to crack arrest in composites. It has already been recognized for many years that the incorporation of strips into a laminate, which introduces macroscopic variations in stiffness and fracture toughness, results in a material construction with crack-arrest capability.

The incorporation of a low modulus fibre of high extensibility into a composite with high modulus fibres provides a mechanism of stopping or deflecting cracks at a microscopic level. It is generally recognized that such crack-arresting effects on a microscopic scale are of major importance with respect to the uniaxial tensile behaviour of unidirectional (UD) hybrid composites. Uniaxial tensile failure of hybrids has been studied quite extensively since the mid 1970s. Many of these studies identified a phenomenon called 'synergistic strengthening' or the 'hybrid effect'. In general, hybrid effects are defined as a positive or negative deviation from the rule of mixtures (ROM). However, in the case of uniaxial tensile loadings ROM behaviour is useless with respect to the understanding of the strength of hybrid composites. In such a case the hybrid effect is generally defined as an enhancement of the first failure strain of the low elongation fibre-reinforced component. The first macroscopic fracture in a hybrid occurs in the component with the lowest failure strain, being the carbon fibre composite in hybrid systems based on carbon and glass, aramid or HP-PE fibres. This enhancement of carbon fibre failure strain when tested in a hybrid was first reported by Hayashi for the system carbon–E-glass. Subsequent studies by several other researchers on hybrid composites also reported first failure strain enhancement under uniaxial tensile loading.

One of the early explanations for these hybrid effects was proposed by Bunsell and Harris, who considered...
residual thermal strains as an important parameter. However, although these effects do occur in carbon–glass hybrid systems, in all cases these thermal strains are too small or even negative (in the case of carbon–aramid\textsuperscript{14} and carbon–HP-PE\textsuperscript{3}) to have a significant effect on the enhancement of the carbon failure strain. Another, more important parameter with respect to the degree of strain enhancement is the degree of dispersion of the different fibres. For hybrids with constant fibre volume ratios, the largest hybrid effects have been observed for composites with the highest degree of fibre dispersion\textsuperscript{11,15,16}.

In contrast with the amount of papers published on hybrid effects under monotonic tensile loading, little attention has been paid to the potential benefits of such crack-arresting effects on the performance of UD hybrid composites under fatigue loading conditions. Nevertheless, since a second fibre of higher extensibility can act to arrest a crack in brittle fibre composites, hybridization may also be favourable with respect to improved fatigue life.

Phillips\textsuperscript{17} showed for carbon–glass cloth an increase in tensile fatigue strength which is proportional to the amount of carbon fibres. Hofer\textit{et al.}\textsuperscript{18,19} studied the effect of stacking sequence on fatigue behaviour of (0°), (±45°) and (0°, ±45°) hybrid laminates based on carbon and glass fibres. They showed that interply hybridization with alternating carbon–glass plies was superior to sandwich hybrids. Their work also included the influence of a number of environmental conditioning treatments on the fatigue performance of such hybrids. Grimes\textsuperscript{20} studied fatigue testing of angle-ply hybrids with various compositions. Bunsell and Harris\textsuperscript{21} used acoustic emission (AE) to monitor fatigue cycling of intimately mixed hybrids. More recently a group of researchers from the University of Bath (UK)\textsuperscript{22,24} reported on repeated tension and tension-compression fatigue behaviour of UD and (±45°, 0°) hybrid laminates at various stress ratios. They noted that the fatigue stress for a given lifetime of UD carbon–Kevlar\textsuperscript{®} hybrids varied linearly with composition\textsuperscript{22}. Since the tensile strength of such composites varies with Kevlar\textsuperscript{®} content at a lower rate than predicted by the ROM, the authors stated that the fatigue ratio, being the fatigue stress divided by the monotonic tensile strength, exhibited a positive synergistic or hybrid effect. Another related study on carbon–glass hybrids\textsuperscript{24} stated that in the case of UD carbon–glass hybrids, both fatigue stress and fatigue ratio displayed a positive deviation from the ROM. A study by Marom\textit{et al.}\textsuperscript{25} dealt with flexural fatigue behaviour of UD carbon–aramid hybrid composites. They reported a positive hybrid effect on flexural fatigue stress for hybrids with a carbon fibre-reinforced core and an aramid fibre-reinforced skin. However, the degradation rate of the S/N curve itself for this hybrid composite—i.e., the slope of the S/N curve—was given by a ROM relationship.

Although there is only a limited amount of research devoted to the fatigue behaviour of hybrid composites, even less information is available on the influence of the interface on the fatigue performance of composites. The few studies in this area generally state that improved interfacial bonding results in improved fatigue performance. Sih and Ebett\textsuperscript{26} reported a linear increase in fatigue life and fatigue stress with shear strength for UD glass/epoxy composites subjected to flexural loadings. Improved fatigue performance has also been reported for the system HP-PE/epoxy after fibre surface treatment\textsuperscript{27}.

The objective of the present study was to find conditions for synergistic strengthening in carbon–HP-PE hybrid composites under uniaxial tensile and fatigue loadings, including the influence of hybrid design and surface treatment of the HP-PE fibres.

**THEORY**

**Constant strain model**

The tensile strength of hybrid composites does not obey the ROM since the LE fibres are expected to break when the failure strain is reached. Based on such a ‘constant strain’ assumption, Manders and Bader\textsuperscript{15} and Chou and Kelly\textsuperscript{28} constructed a strength diagram of the type shown in Fig. 1 for the system carbon–HP-PE\textsuperscript{2}. In such a diagram, point A denotes the tensile strength of the LE fibre composite, in this case the carbon composite, and point D that of the high elongation (HE) fibre composite, being the HP-PE composite. The strength of the hybrids is given by the two straight lines AC and CD. The line AE represents the stress in the hybrid at which failure in the LE phase is expected. At HE fibre volume fractions below point C, failure of the LE phase leads to instant failure of the hybrid composite since there is insufficient HE fibre to sustain the load after failure of the LE fibre component. The strength is then given by:

\[
\sigma_{\text{hybrid}} = \sigma_{\text{LEmax}} V_{\text{LE}} + \epsilon_{\text{LEmax}} E_{\text{HE}} V_{\text{HE}}
\]

where \(V_{\text{LE}}\) and \(V_{\text{HE}}\) are the volume fractions of the LE and HE fibres respectively, \(\epsilon\) is the strain and \(E\) is Young’s modulus. With increasing amounts of HE fibres a transition from single fracture to a multiple fracture...
mode occurs when there are sufficient HE fibres to sustain the load after failure of the LE fibres at a stress level given by line CE. The ultimate strength of the hybrid (line CD) is then given by:

$$\sigma_{\text{hybrid}} = \sigma_{\text{HE max}} V_{\text{HE}}$$  \hspace{1cm} (2)

Since the strength of a material and in particular that of fibre-reinforced composites is a 'process' rather than a material property, the constant strain model is an oversimplification, yielding a lower bound of hybrid strength.

**Statistical models**

The requirement for statistical models arises from the variations in fibre strength resulting from a random distribution of flaws. Several theoretical models have been proposed to explain the experimentally observed deviations from the constant strain model—i.e., hybrid effects. Most models invoke a statistical distribution of strength, in which failure of the weakest LE fibre results in a crack that is bridged by the surrounding HE fibres, allowing the stronger LE fibres to reach their full potential. A statistical analysis of the strength of hybrid composites was originally developed by Zweben\textsuperscript{14}. Based on models originally developed by Rosen\textsuperscript{29} and Zweben\textsuperscript{30} for single fibre composites, he presented a statistical analysis of hybrid composites consisting of a two-dimensional array of alternating LE and HE fibres in a matrix. His analysis recognizes a number of important material properties that influence the failure process in hybrid composites; firstly, the statistical failure strain characteristics of fibres and secondly, the stiffness ratio of both components as a result of different fibre moduli and cross-sectional areas. Besides Zweben, several other authors attempted to model the tensile strength of hybrids using statistical models (Manders and Bader\textsuperscript{31}, Fukuda\textsuperscript{32}, Harlow\textsuperscript{33} and Faribotz et al.\textsuperscript{34}). Fukuda\textsuperscript{32} modified the Zweben model, and gives the hybrid failure strain as:

$$\epsilon_{\text{hybrid}} = \left[ \frac{NL\delta p(k_h^N - 1)}{2\delta^N \epsilon^N} \right]^{1/2q}$$ \hspace{1cm} (3)

where \( N \) is the total number of fibres in the composite, \( L \) is the specimen length, \( \delta \) is the so-called ineffective length, \( k_h \) is the stress concentration factor (SCF) in the nearest LE fibre in the hybrid composite and \( p \) and \( q \) are the Weibull parameters of the LE fibre obtained from the Weibull distribution function. Although the Weibull distribution function is not a perfect description of fibre strength, it is mathematically convenient and has been proven to describe the tensile strength of brittle fibres such as glass and carbon fairly well.

In the case of non-hybrid composites the failure strain is calculated as

$$\epsilon_{\text{f}} = \left[ \frac{NL\delta p(k - 1)}{2\delta^N \epsilon^N} \right]^{1/2q}$$ \hspace{1cm} (4)

being the lower bound of the failure strain; \( k \) and \( \delta \) are the SCF and ineffective length for the LE composite respectively. The factor two appears because there are twice as many LE fibres in the all-LE composite than in the hybrid. Since the hybrid effect is the enhancement of the initial failure strain which corresponds to the failure strain of the LE fibre, the hybrid effect may be calculated as:

$$R_c = \frac{\epsilon_{\text{f}}}{\epsilon_{\text{f}}} = \left[ \frac{\delta(k_{\text{h}} - 1)}{2\delta^N \epsilon^N} \right]^{1/2q}$$ \hspace{1cm} (5)

To calculate the hybrid effect \( R_c \), only the values for \( \delta, \delta_h, k, k_h \) and \( q \) have to be known. Thus the hybrid effect \( R_c \) depends on the SCFs, the ineffective lengths and the Weibull parameter of the LE fibre.

**Stress concentrations and ineffective length**

To predict the mechanical properties of hybrid composites, knowledge of the redistribution of stress near an initial fibre breakage is necessary. Hedgepeth\textsuperscript{35} was the first to report such a stress redistribution around fibres in a UD single fibre composite. In his analysis the SCF in the fibre adjacent to a broken fibre was derived using a shear–lag theory. In the case of hybrid composites the effect of a fibre breakage on the stress distribution is even more complicated. Stress concentrations in hybrid composites have been calculated by Zweben\textsuperscript{14} in which the SCF was calculated by modifying his analytical method for conventional composites\textsuperscript{30}. Another study was adopted by Fukuda and Chou who studied the SCF in an intermingled hybrid sheet containing only one discontinous fibre\textsuperscript{36}.

Various models have been proposed to describe the ineffective length in hybrid composites and many differ in the precise definition of \( \delta \). Rosen\textsuperscript{29}, for example, defined the ineffective length as the length from a fibre end in which the average axial stress is smaller than \( \phi \) (\( \approx 0.9 \)) times the stress which would exist in infinite fibres. Zweben\textsuperscript{14,30} and Fukuda\textsuperscript{32} gave approximate solutions for the ineffective length, in which the actual fibre stress distribution is replaced by an equivalent step function.

This paper examines the load redistribution in a hybrid composite sheet using numerical methods. A micromechanical analysis using the finite element program MARC is performed on a hybrid UD laminate having an initial LE fibre fracture. Results of such an analysis can be incorporated into statistical strength models as described above and given by Equations (3), (4) and (5). The hybrid model is shown in Fig. 2 with LE and HE fibres in alternating positions, resembling an intermingled fibre arrangement. Tensile load is applied along the fibre direction and the problem was considered as one of plane strain. Principal directions of the orthotropic plate were aligned in two lines of symmetry.
and therefore only one quadrant of the plate was considered. Fig. 3 shows the mesh used, consisting of a total of 700 elements, of which 420 elements are of the fibre. A special crack-tip element was used to connect the broken fibre end with the matrix, in order to describe the singularity at that point. Perfect adhesion and equal fibre diameters for the LE (carbon) and HE (HP-PE) fibres are assumed. The interfibre distance is chosen such that the total fibre volume equals 50%. Material parameters being considered are listed in Table 1.

Table 1. Material parameters of fibres and matrix

<table>
<thead>
<tr>
<th>Material</th>
<th>$E_x$ (GPa)</th>
<th>$E_y$ (GPa)</th>
<th>$\nu_{xy}$</th>
<th>$\nu_{yz}$</th>
<th>$G_{xy}$ (GPa)</th>
<th>$G_{xz}$ (GPa)</th>
</tr>
</thead>
<tbody>
<tr>
<td>HS-carbon</td>
<td>230</td>
<td>20</td>
<td>0.013</td>
<td>0.23</td>
<td>20</td>
<td>8</td>
</tr>
<tr>
<td>HP-PE</td>
<td>80</td>
<td>2</td>
<td>0.010</td>
<td>0.23</td>
<td>0.8</td>
<td>0.7</td>
</tr>
<tr>
<td>Epoxy</td>
<td>3.2</td>
<td></td>
<td>0.2</td>
<td></td>
<td>1.23</td>
<td></td>
</tr>
</tbody>
</table>

Fig. 4 gives the longitudinal stress in the fibres at the plane of symmetry ($x = 0$) at a strain level of 1.5% for the plain carbon (Fig. 4(a)) and hybrid composite (Fig. 4(b)). In analytical models a constant longitudinal stress over the cross-section of the fibre is assumed. However, the numerical results indicate that for fibres close to a broken fibre this is incorrect. In these fibres the SCF varies along the thickness of the fibre. Therefore in our analysis the SCFs are taken as an average.

The SCFs and ineffective lengths of the broken carbon fibres are listed in Table 2. In our case we used Rosen's definition for the ineffective length, i.e., the load transfer length necessary to reach an arbitrary stress of 0.9 times the undisturbed stress. The numerical results also show that in a hybrid composite of carbon and HP-PE, the SCF in the HP-PE fibre next to a broken carbon fibre is higher than that in a composite consisting of only HP-PE fibres, indicating that, after fracture of the carbon fibres, HP-PE fibres are more sensitive to fracture in a hybrid than in a plain HP-PE composite. The calculated SCFs in the fibres next to a broken fibre in plain carbon and HP-PE composites are in both cases about 1.33 and are similar to the SCF value calculated by Hedgepeth. However, in a hybrid composite we have to focus on the SCFs in the LE fibres, which is 1.08 for the nearest carbon fibre in a carbon–HP-PE hybrid composite with an initial carbon fibre fracture. A good agreement between the numerical results shown here and the analytical SCFs as calculated by Fukuda is obtained if a stiffness ratio ($E_{HE}/A_{HE}/E_{LE}/A_{LE}$) of 0.35 is used, which is a typical value for HP-PE and HS-carbon.

It can be shown that the high peak stress that is observed in the nearest HP-PE fibre (fibre 1 in Fig. 4(b)) is related to the low shear modulus of this highly anisotropic fibre. Similar analyses for a hybrid system based on carbon and (isotropic) glass fibres resulted in a lower peak stress in the nearest glass fibre, although the average SCF in this glass fibre was comparable to the SCF of fibre 1 as listed in Table 2 and calculated for the carbon–HP-PE system.
Using the data of Table 2, we now can calculate the enhancement of the failure strain using Equation (5). A value of $R_e$ of 1.12 and 1.19—i.e., a hybrid effect of 12% and 19%—results from analysis on bundle and single filament level respectively, using different Weibull moduli for a bundle of carbon fibres and for single carbon fibres (2.0 and 7 respectively).

**Fatigue modelling**

The fatigue behaviour of a material is generally represented by an $S/N$ relation. The simplest form of this empirical relation is generally semi-logarithmic:

$$S = a + b \log N$$

where $S$ represents the maximum value of the cyclic stress, $N$ the number of cycles and $a$ and $b$ material constants. Generally, the properties of a hybrid composite are considered to be an average of those of the parent composites. Consequently, the fatigue behaviour of a hybrid is also often weighted by a ROM relation. In virtually all fatigue studies on hybrid composites the appearance of a hybrid effect was related to this ROM behaviour. However, similar to the tensile behaviour of hybrid composites, in tension–tension fatigue ROM behaviour is also less meaningful and thus a 'constant strain' type of behaviour should be adopted rather than ROM. In a hybrid composite, a fatigue damage process will occur in both types of fibres at strain levels close to the failure strain of the LE fibre. Since fibre fracture during fatigue in a composite is strongly governed by the applied strain level, as noted by Talreja, this type of fatigue damage will manifest itself mainly in the LE fibre.

By selecting strain instead of stress as an independent variable, Talreja suggested a basic pattern of the fatigue life diagrams for composites which include the basic fatigue damage mechanisms in UD composites such as fibre fracture, interfacial debonding, matrix cracking and interfacial shear failure.

Based on a constant strain model, both hybrid and LE fibre composites should exhibit the same $e/log N$ curve (Fig. 5(a)). Deviation from such a constant strain fatigue diagram can now be regarded as synergistic or hybrid effects as result of the 'crack arrest' phenomenon as described earlier. This strain/life diagram can easily be converted into the expected stress/life behaviour using
the stiffness of the hybrid, which follows a ROM relation and the linear elasticity theory (Fig. 5(b)). It can be shown that the slopes for the hybrid composites in the constructed S/N curves are less than that for a plain LE composite. When considering such a constant strain fatigue model, the results of Dickson et al. for carbon-glass hybrids can easily be explained. The authors stated that their hybrid composites showed a ‘unexpectedly’ lower slope than that of plain carbon laminates. In fact, the conclusions of Dickson et al. concerning synergistic effects are a misinterpretation of a constant strain fatigue model and the use of an e/log N diagram to elucidate synergistic effects. They interpreted similarities in strain/life behaviour for both plain carbon composite and carbon-glass hybrids as an indication for synergistic effects.

A further implication of such a constant strain fatigue model is that, generally, in a plot of fatigue stress vs. composition, the ROM should not be used to interpret data with respect to synergistic effects but a diagram similar to that of Fig. 1 should be adopted.

EXPERIMENTAL DETAILS

Materials

UD composites were fabricated using a common epoxy resin from Ciba-Geigy (Araldite LY556/HY917/DY070) reinforced with HP-PE fibres supplied by Allied Signal (Spectra* 1000, 650 denier) and a surface-treated high-strength carbon fibre (XAS/3K) from Courtaulds plc. To study the effect of improved adhesion of HP-PE on the performance of the hybrids, composites incorporating untreated and chromic-acid treated fibres were used.

Two hybrid geometries have been studied: layered sandwich constructions with an HP-PE core and carbon skins and tow-to-tow intermingled hybrids with a high degree of dispersion. As a reference plain carbon and HP-PE composites were also prepared. In the HP-PE composites as well as in the hybrid composites, both untreated and chromic-acid treated HP-PE fibres were used. All composites had a volume percentage of fibres of approximately 55% and in the case of hybrids a carbon–HP-PE fibre ratio of 60/40 was selected.

As well as UD laminates, filament-wound (NOL) rings were also manufactured. These rings were prepared by winding HP-PE and/or carbon fibres unidirectionally on a mandrel of stainless steel discs (Fig. 6). To ensure demoulding, the discs were covered with release tape. All rings had an inner diameter of 145 mm and a width of 6 mm. The thickness of the rings was approximately 1.2 mm. Layered sandwich hybrid rings were manufactured by alternately winding carbon and HP-PE yarns. Intermingled hybrids were fabricated by simultaneous winding of carbon and HP-PE yarns. Both types of hybrid had equal numbers of HP-PE and carbon tows.

Consequently, the total cross-sectional area of the fibres in all composites is the same, minimizing the scatter in strength within one type of composite. After winding, the rings were cured for 4 h at 80°C while rotating in an air-circulating oven and subsequently post-cured for 12 h at 110°C.

UD laminates were fabricated by wet winding of fibres on a frame (Fig. 7). After winding, the whole set-up was degassed in a vacuum oven at a pressure of 400 mm Hg at 60°C to obtain a void-free composite. Laminates were
Table 3. Tensile test data for unidirectional composites

<table>
<thead>
<tr>
<th>Material</th>
<th>Modulus, E (GPa)</th>
<th>Tensile strength (MPa)</th>
<th>Failure strain (%)</th>
<th>Hybrid effect (%)</th>
</tr>
</thead>
<tbody>
<tr>
<td><strong>Single fibre composite</strong></td>
<td></td>
<td></td>
<td></td>
<td></td>
</tr>
<tr>
<td>Carbon</td>
<td>126</td>
<td>1750</td>
<td>1.41</td>
<td>–</td>
</tr>
<tr>
<td>HP-PE (untreated)</td>
<td>40</td>
<td>910</td>
<td>3.10</td>
<td>–</td>
</tr>
<tr>
<td>HP-PE (treated)</td>
<td>42</td>
<td>1070</td>
<td>3.30</td>
<td>–</td>
</tr>
<tr>
<td><strong>Hybrid composite</strong></td>
<td></td>
<td></td>
<td></td>
<td></td>
</tr>
<tr>
<td>Sandwich (untreated)</td>
<td>91</td>
<td>1280</td>
<td>1.41</td>
<td>0</td>
</tr>
<tr>
<td>Sandwich (treated)</td>
<td>92</td>
<td>1300</td>
<td>1.43</td>
<td>1.4</td>
</tr>
<tr>
<td>Intermingled (untreated)</td>
<td>88</td>
<td>1360</td>
<td>1.53</td>
<td>8.5</td>
</tr>
<tr>
<td>Intermingled (treated)</td>
<td>87</td>
<td>1460</td>
<td>1.69</td>
<td>19.8</td>
</tr>
</tbody>
</table>

prepared by positioning the frame in the open-ended mould and placing the whole assembly in a hot-press. Again a release film was used between mould and composite. The laminate thickness was controlled by metal stop strips of the required thickness, while during hot-pressing the excess resin was squeezed out through the open ends of the mould. During curing at 80°C for 4 h, a pressure of 4 bar was maintained. After curing, the mould was removed and the laminates were post-cured for 12 h at 110°C.

In the case of sandwich hybrids, the laminates were manufactured in a two-step process. First, the impregnated HP-PE fibres were wound on the frame, subsequently degassed and partly cured for 15 min at 80°C in a hot-press. After B-staging, carbon fibres were wound on this HP-PE 'prepreg laminate' following the same procedure as described above. This two-step process resulted in a three-layer hybrid construction with the HP-PE fibre component sandwiched between two carbon fibre-reinforced plies and a uniform thickness of the HP-PE core with few carbon fibres migrating into the HP-PE laminate. This manufacturing procedure resulted in nearly void-free, high quality, flat UD laminates of dimensions 160 x 300 mm. Laminate thicknesses for the single fibre composites and hybrid composites were 0.4 and 0.8 mm respectively.

Tensile specimens of dimensions 12.5 x 185 mm were cut from these composite plates using a diamond cutting-wheel. Aluminium end-tabs of 30 mm length were adhesively bonded to the specimens, leaving a test length of 125 mm.

**Testing**

Mechanical properties of the hybrids were determined by tensile tests at room temperature. Tests were performed on a Zwick 1474 universal testing machine equipped with an extensometer. Each specimen was loaded monotonically to failure at a cross-head speed of 0.5 mm min⁻¹.

The accumulation of damage during testing was monitored using an acoustic emission system (PAC 8900 Locan AT, Physical Acoustics Corp) including a 40 dB pre-amplifier (PAC 1200A). AE was monitored with a piezoelectrical transducer (PAC R 15) coupled to the specimen using vacuum grease. The AE system allowed the AE event to be analysed with respect to different characteristic parameters such as events, counts, peak amplitude, duration, rise time and energy.

Tension–tension fatigue tests were carried out on a Zwick Rel servo-hydraulic machine at a frequency of 5 Hz and a stress ratio, R(σmin/σmax), of 0.1. The system was operating under load control, applying a harmonic tensile stress with a constant amplitude. To avoid clamping problems, such as shear failure within the tabs in the case of plain HP-PE composites due to their low shear strength, fatigue tests were performed on filament-wound NOL rings using a split-disc loading device (ASTM D 2291). To minimize wear of the composite a PTFE-based lubricant was used. Damage development during fatigue using modulus reduction, ultrasonic C-scan and AE was also evaluated on laminated plates of plain carbon and hybrid compositions.

**RESULTS AND DISCUSSION**

**Tensile behaviour**

The results for the four hybrid systems and their parent composites as obtained from the tensile tests are presented in Table 3 and are an average of at least five test results. The experimental accuracy of all quoted data is within 10%. In all hybrids, no multiple fracture was observed; i.e., the stress/strain curves were all linear up to total fracture of the hybrid. From the failure strain data, it can be seen that an enhanced first failure strain was observed in the case of intermingled hybrids, whereas for sandwich hybrids no strain enhancement was detected. These observations are in good agreement with previous work on carbon–glass. Values for hybrid effects found in the present work are also in good agreement with values quoted in Reference 39 despite the fact that
different materials were used. Furthermore, it should be noted that the experimental values for the hybrid effect are in reasonable agreement with the theoretical predictions based on statistical models including SCF effects. Fig. 8 shows the hybrid effect vs. hybrid composition, including the data from Reference 2. The dotted and full lines are the theoretical predictions based on Fukuda's model using either fibre bundle or single filament Weibull parameters for the carbon fibre.

From this part of the study it can be concluded that, under monotonic tensile loading, hybrid effects are observed for tow-to-tow intermingled hybrids incorporating both untreated and treated HP-PE fibre bundles, whereas no hybrid effects are observed for the sandwich constructions.

**Acoustic emission**

Acoustic emission monitoring was used for two reasons: firstly to gain more insight into the onset of damage in the different types of hybrid compared with plain carbon composite; and secondly, to discriminate between different failure processes in hybrid composites such as carbon fibre breakage, HP-PE fibre breakage, matrix cracking, debonding and delamination. In Fig. 9 the cumulative plots of events vs. strain are shown for both untreated and treated HP-PE composites. The AE events, observed during tensile loading up to failure of the composite, increased in number as the test progressed. For composites incorporating untreated fibres, initial emissions occurred almost immediately upon loading, with a relative gradual increase in event rate with increasing load. Composites incorporating treated fibres showed an onset of emissions starting at about 20% of maximum strain, with a rapid increase at approximately 35% of maximum strain. Obviously, improved adhesion has a positive effect on the onset of damage in HP-PE composites.

Fig. 10 gives the total activity plots for the plain carbon and hybrid composites when loaded up to about 5 kN. From these plots it can be concluded that the onset of damage is approximately the same for the plain carbon composite and the intermingled hybrids. However, in particular, the sandwich hybrid with untreated fibres was consistently noisier than the other composites, presumably due to extensive debonding within the HP-PE component and/or delaminations at the carbon/HP-PE interface. It seems that a reduction of the total carbon/HP-PE interface scale accelerates the initiation of damage processes such as debonding and interlaminar shear failure.

Amplitude distribution analysis of AE signals has shown itself useful in discriminating between different fracture mechanisms in composite materials. Generally, at the initial stages of tensile testing, low-amplitude emissions are associated with matrix crazing or cracking, whereas fibre fracture is characterized by high-amplitude emissions.

Fig. 11 shows the distribution of the peak amplitudes of the AE signals at various load levels for composites incorporating treated HP-PE fibres; at low loads (0-65%) a concave-shaped curve is observed with most amplitude events at approximately the threshold level of 30 dB, indicating debonding. At higher loads (65-100%), a
typical case of a double or even triple-peaked amplitude distribution occurs as would be expected in the normal case of fibre fracture in combination with matrix crazing and/or interface failure. Composites with untreated HP-PE fibres showed even at high load levels an amplitude distribution signature similar to that of treated HP-PE composites at low load levels (0-65%) with no clear double-peak and only a few high amplitude events in the range 70-80 dB. Apparently, the number of fibre fractures in this type of composite is less than that in the case of treated fibres. Presumably, the large debonding length for untreated HP-PE fibres prevents extensive multiple fibre fractures within one filament, since no fibre fracture will occur over this ineffective length. Mercx et al.\textsuperscript{40} reported for a similar chromic acid treatment an increase in pull-out strength from 0.3 MPa to 1.9 MPa. This increase in pull-out strength (with a factor of 6) is indicative of a decrease in ineffective length after surface treatment, enlarging the probability of multiple fibre fractures and leading to higher effectiveness of the fibre with respect to composite strength\textsuperscript{40}.

Amplitude distribution acoustic emission analysis for carbon/epoxy composites is shown in Fig. 12. Here also a typical case of a double-peaked amplitude distribution is observed. However, in contrast to HP-PE signatures, the low-amplitude peak shifts with increasing load completely to higher amplitude levels. Between 60-80% of maximum load a distinct second peak or shoulder is formed at approximately 50 dB. This becomes much stronger at higher loads suggesting that at these load levels a significant amount of fibre fracture occurs. At the higher load levels (80-100%) the low-amplitude peak totally disappears, indicating that fibre fracture is now the dominating fracture process. This peak shifting also indicates that less debonding and delamination is taking place in the carbon fibre composite than in the HP-PE composites, where at high load levels low-amplitude signals were still predominant. This assumption was confirmed by scanning electron micrographs of fracture surfaces, which showed that even in the case of treated HP-PE fibre composites, numerous fibre pull-outs occurred\textsuperscript{2}, whereas the carbon fibre composite failed in a
rather brittle mode with a negligible amount of debonding and pull-out. It is interesting to note that the high-amplitude events associated with carbon fibre fracture (50 dB) are lower than that for HP-PE (70-80 dB), indicating that the latter fibres exhibit higher energy acoustic emissions. This observation is in accordance with the work-to-break of both fibres. Since the tenacity of both fibres is about the same, the fibre fracture energy is expected to relate to the failure strain, which is for HP-PE about twice that of carbon fibre.

Amplitude distribution signatures of the various hybrid configurations were all more or less alike (Fig. 13) and indicated that in all hybrid composites, in the latter part of the test, debonding and/or delamination is a more prominent failure process than in the case of plain carbon composites, since even at high load levels a distinct low-amplitude signal is evident. However, similar to the plain carbon composites there is some indication of the development of a second peak at 50 dB, suggesting carbon fibre fracture at higher load levels.

Fig. 13 Amplitude distribution AE characteristics of carbon-HP-PE intermingled hybrid composite at different load levels

Since there are no significant 70–80 dB signals it seems that no HP-PE fibre fracture is taking place which is not surprising, since these fibres have a failure strain of more than twice that of carbon.

**Fatigue behaviour**

Fig. 14 shows that the $S/\log N$ curves for plain carbon and HP-PE composites incorporating untreated and chromic-acid treated fibres. Although there is some reduction in static composite strength as a result of a reduction in tensile strength of the HP-PE fibres after chromic acid treatment, the $S/\log N$ curves for plain HP-PE composites clearly demonstrate the advantage of improved adhesion on the fatigue performance of these composites. The slopes of the $S/\log N$ curves of untreated and treated HP-PE composites are approximately 230 MPa and 77 MPa per decade respectively. The slope for the treated HP-PE/epoxy composite is even flatter than that for plain HS-carbon composite, illustrating the excellent fatigue resistance of this type of composite. After surface treatment the fatigue performance of HP-PE composites becomes more fibre-dominated rather than interfacial-dominated.

Since the testing of filament-wound rings may have some disadvantages with respect to the non-uniformity of the strain in the ring, the possibility of stress concentrations at the edges of the split discs and frictional wear of the composite, comparative experiments were carried out on UD laminates fabricated from prepreg (Fibredux 913C/XAS-5-34). Fatigue performance of these high-quality laminates was approximately the same as that of the carbon composite rings, with even a small advantage for the rings. Consequently, the rings were considered as representative test samples.

The fatigue data for the various hybrid composites are presented in Figs 15(a) and 15(b). Two important observations can be made from the $S/\log N$ curves of the hybrids. Firstly, it should be noted that the slopes of the $S/\log N$ curves for all the hybrids are lower than that of the plain carbon composite and, secondly, the slopes for the hybrids are all more or less the same, with a positive exception for the intermingled hybrid incorporating...
treated HP-PE fibres. The fatigue parameters $a$ and $b$ for the various composites, adopting the semi-logarithmic $S/N$ relation (Equation (6)), are given in Table 4. Using the experimentally obtained fatigue parameters for a plain carbon composite, Equation (1) for the calculation of the 'constant strain' fatigue stress of the hybrid and ROM for the modulus, one can determine the theoretical values for the fatigue parameters of the hybrid. In the case of a viscoelastic fibre such as HP-PE we have to consider strain-rate effects. Therefore, in our calculations we used a value of 100 GPa for the modulus of HP-PE fibre at 5 Hz, instead of the 80 GPa as listed in Table 1. Predicted values for the slope $b$ of the hybrid composite, expecting constant strain fatigue behaviour, are in good agreement with experimentally obtained values, except for the intermingled hybrid with treated HP-PE which exhibits a significantly lower slope than predicted by the constant strain model. Furthermore, it should be stressed that the degree of scatter in the hybrids, especially in the hybrids with untreated fibres, is smaller than that for the plain carbon composites. These observations are in contrast with earlier work on carbon–Kevlar$^{22}$ and carbon–glass$^{24}$, where more scattering in the fatigue life of the hybrids was observed than in that of the parent composites.

We can also compare the behaviour of the carbon and hybrid composites using a plot similar to that of Fig. 5(b). By extrapolating the data to $\sigma_{\text{max}}=0$ deviations from the constant strain model are observed, since in such a model all lines should have the same intersection with the x-axis ($\sigma_{\text{max}}=0$). From the extrapolated curves in Fig. 16 it can be seen that all curves intersect the x-axis at about log $N=15$, except the curve for the intermingled hybrid with treated HP-PE which intersects at approximately log $N=28$, illustrating the synergistic fatigue life for this type of composite. The authors hasten to add that extrapolation of the fatigue data up to $\sigma_{\text{max}}=0$ and ignoring the existence of a fatigue limit has no physical relevance. However, in this context the point of intersection is only used as a convenient way to compare the slopes of the $S/\log N$ curves.

When we convert the stress data to strain data, by using the relevant moduli, and plot an $\varepsilon/\log N$ diagram, the slopes for the hybrids can directly be compared with that of the plain carbon composite (Fig. 17). With respect to the interpretation of such diagrams, it should be noted...
that only small changes in slope result in large effects in fatigue life since a logarithmic scale is used.

From the data listed in Table 4 and Fig. 16 as well as Fig. 17 it can be concluded that all hybrids follow the constant strain fatigue model, except the intermingled hybrid with treated HP-PE fibres which exhibits synergistic or hybrid effects with respect to fatigue performance.

**Fatigue damage characterization**

To investigate the effect of the HP-PE fibres and the surface treatment of these fibres on the crack propagation in the hybrids, damage development was studied using modulus reduction measurements, ultrasonic C-scan and AE. For experimental convenience, in preference to the filament-wound rings, damage characterization was performed on flat laminates, similar to the test specimens used in the tensile test. Since we are particularly interested in the damage evolution of the hybrids relative to that of plain carbon composites, we only considered these composites. When comparing the damage state in the hybrid composites with that in a plain carbon composite, all tests have to be carried out at equal strain levels. In these experiments an initial maximum strain level ($\varepsilon_{\text{max}}$) of 1.10% and $R = 0.1$ were used.

The essential first step in fatigue characterization of composites is to measure changes in the stiffness properties of the laminate. However, similar to what is generally observed for UD composites, all materials show excellent fatigue resistance since the damage which occurs during fatigue (such as debonding, delamination and fibre fracture) has no or little influence on the stiffness. Consequently, differences in intermittently measured moduli are within experimental error and moduli remain essentially constant over their fatigue life.

To study the effect of improved adhesion of the HP-PE fibres on the accumulation of debonding and delamination in the hybrid composites, ultrasonic C-scan was used. Fig. 18(a)-(c) shows some of the extreme results; plain carbon, sandwich hybrids with untreated HP-PE and intermingled hybrids incorporating treated HP-PE respectively. All samples were fatigued at the same initial strain level under load control. The plain carbon laminate showed little detectable damage in the form of delaminations. Only after 400 cycles is a small edge delamination observed. In all hybrids, especially in the sandwich construction with untreated HP-PE (Fig. 18(b)), there is clear evidence of extensive debonding and delamination. However, in the case of intermingled hybrids with treated fibres there is no sign of damage, even at higher cycle numbers (Fig. 18(c)).

Table 5 gives the number of acoustic emission events until fracture for the plain carbon and hybrid composites using a threshold level of 40 dB. The threshold was chosen at 40 dB to reduce background noise and to concentrate on the high amplitude signals, which are indicative of fibre fracture. The lowest number of 60 dB AE signals was measured for the intermingled hybrid with treated HP-PE, suggesting relatively little severe fatigue damage in this type of composite. The highest number of low-amplitude (40 dB) signals manifested itself in the hybrids with untreated HP-PE fibres, presumably due to extensive debonding and delamination. These findings are in accordance with the fatigue life diagrams, which showed the best fatigue
performance for the intermingled hybrids with treated HP-PE.

CONCLUSIONS

The mechanical properties of UD hybrid composites based on carbon and HP-PE fibres have been studied under both monotonic and cyclic tensile loading. Two different hybrid configurations have been studied; core-shell sandwich hybrids and tow-to-tow intermingled hybrids. Both hybrid types were manufactured incorporating either untreated or chromic-acid treated HP-PE fibres, resulting in a total of four hybrid systems evaluated. Results indicated that the existence of synergistic or hybrid effects depends on both hybrid design and the interfacial bond strength of the HP-PE fibres. Under monotonic tensile loading, positive deviations from the constant strain behaviour were observed for both types of intermingled hybrid, with the highest hybrid effect for the intermingled system with surface-treated HP-PE fibres. No hybrid effects were observed for the sandwich hybrids. Obviously, the possibility of crack arrest due to the presence of HP-PE fibres, preventing rapid crack extension from initially failed carbon fibres, diminishes when these fibres are not highly dispersed throughout the carbon fibre composite.

AE studies on hybrid and parent composites during monotonic tensile loading also indicated better performance for the intermingled hybrids with respect to the onset of damage and showed no evidence of HP-PE fibre breakage in the hybrid.

Finite element micromechanical calculations were performed on a hybrid model to calculate stress concentrations in carbon-HP-PE hybrid composites, which can be used to calculate hybrid effects using statistical models. Using such a model, hybrid effects could be described reasonably well.

A fatigue model, based on the conventional ‘constant strain’ model for the tensile strength of hybrid composites, is adopted and describes the fatigue behaviour of all hybrids fairly well. Synergistic effects under fatigue loading conditions—i.e., positive deviations from the expected fatigue behaviour based on such a model—were observed for the intermingled hybrids with treated HP-PE. These hybrids also showed the lowest level of fatigue damage as characterized by ultrasonic C-scan and acoustic emission.

Table 5. Number of AE events during fatigue

<table>
<thead>
<tr>
<th>Material</th>
<th>40 dB</th>
<th>60 dB</th>
</tr>
</thead>
<tbody>
<tr>
<td>Carbon (untreated)</td>
<td>1300</td>
<td>600</td>
</tr>
<tr>
<td>Sandwich hybrid (treated)</td>
<td>6600</td>
<td>800</td>
</tr>
<tr>
<td>Intermingled hybrid (untreated)</td>
<td>2400</td>
<td>500</td>
</tr>
<tr>
<td>Intermingled hybrid (treated)</td>
<td>4600</td>
<td>1400</td>
</tr>
<tr>
<td>Intermingled hybrid (treated)</td>
<td>1100</td>
<td>120</td>
</tr>
</tbody>
</table>

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