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Characteristics of several carbon fibre-reinforced aluminium composites prepared by a hybridization method

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The properties and microstructures of several high-strength and high-modulus carbon fibre-reinforced aluminium or aluminium alloy matrix composites (abbreviated as HSCF/Al and HMCF/Al, respectively, for the two types of fibre) have been characterized. The composites evaluated were fabricated by pressure casting based on a hybridization method. It was found that the strength degradation of high-modulus carbon fibres after infiltration of aluminium matrices was not marked and depended upon the type of aluminium matrix. However, the strength of high-strength carbon fibres was greatly degraded by aluminium infiltration and the degradation seemed to be independent of the type of aluminium matrix. The longitudinal tensile strength (LTS) of CF/Al composites was very different between HMCF/Al and HSCF/Al composites. The HMCF/Al composites had LTS values above 800 MPa, but the HSCF/Al composites had only about 400 MPa. In contrast, the transverse tensile strength of the HSCF/Al composites, above 60 MPa, was much higher than that of the HMCF/Al composites, about 16 MPa. Chemical reactions were evident to the interface of high-strength carbon fibres and aluminium matrices. There was no evidence of chemical products arising between high-modulus carbon fibres and Al–Si alloy and 6061 alloy matrices. However, it was considered that some interfacial reactions took place in pure aluminium matrix composites. Fracture morphology observation indicated that the good LTS of CF/Al composites corresponded to an intermediate fibre pull-out, whereas a planar fracture pattern related to a very poor LTS and fibre strength transfer. The results obtained suggested that interfacial bonding between carbon fibres and aluminium matrices had an important bearing on the mechanical properties of CF/Al composites. An intermediate interfacial bonding is expected to achieve good longitudinal and transverse tensile strengths of CF/Al composites.

1. Introduction

Carbon fibres can be classified into two main categories according to their mechanical properties, i.e. the high-strength type and the high-modulus type. The mechanical properties of carbon fibres depend upon their microstructure and the degree of crystallite orientation. High-modulus carbon fibres (HMCFs) are highly graphitized at temperatures of above 2273 K and are characterized by a high modulus. They have the long-distance order of graphite crystallites, mainly parallel to the fibre axis. High-strength carbon fibres (HSCFs), which are heat-treated to temperatures lower than 1773 K, have a low modulus but very high strength. Although HSCFs also have crystallite orientation mainly parallel to the fibre axis, the size of the crystallites is very small compared to that of HMCFs [1]. The two types of the carbon fibre have been applied in aluminium matrix composites. Generally, it has been reported that HMCFs are more suitable for reinforcing aluminium matrices than HSCFs because HMCFs have a lower degree of reactivity between the fibres and aluminium matrices at elevated temperatures [2, 3]. However, Towata *et al.* [4] have also reported that HSCF/Al alloy composites with SiC whiskers and particulate had a longitudinal flexural strength of about 1000 MPa.

On the other hand, the aluminium alloy matrix is also very important for achieving successful composite. This is not only because the matrix is usually the major constituent of the composite, but also because alloying elements in the aluminium matrix have obvious effects on the wettability and reactivity between the carbon fibres and the matrix [5–8]. For instance, it was considered that one possible way to

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limit carbide formation in carbon fibre–aluminium systems is to alloy aluminium matrices with elements very similar to carbon, such as silicon, which changes the interaction character of the fibre–matrix interface.

The present work is an approach to study the behaviour of HSCFs and HMCFs as reinforcements for several aluminium alloy matrices, and to discuss the effects of different matrices on the properties of the composites obtained. Composites of high-strength carbon fibre and high-modulus carbon fibre with matrices of pure aluminium, Al–12 wt % Si alloy and Al–Mg–Si alloy (6061) (abbreviated as HSCF/P–Al and HMCF/P–Al, HSCF/Al–Si and HMCF/Al–Si, HSCF/6061 and HMCF/6061 composites, respectively) were fabricated by pressure casting using the hybridization method described elsewhere [9]. The characteristics of these composites were clarified and the influences of HSCFs, HMCFs and different matrices on the properties of the composites were investigated.

2. Experimental procedure

The carbon fibres used in the present investigation was PAN-based high-modulus and high-strength carbon fibres without a sizing treatment. Some of their properties, along with those of the aluminium matrices used in the experiment, are listed in Table I.

The basis of the fabrication technique has been described in detail elsewhere [9]. Essentially, carbon fibres were impregnated into an aqueous slurry of SiC particles consisting of a polymer as a binding agent and an organometallic compound as a dispersing agent. A unidirectional preform was then prepared from the partially dried impregnated fibres. The fibre preform was set in a mould and preheated to 648 K, and infiltrated with molten aluminium matrix under pressure. The casting pressure was 49 MN m^{-2}, the melt temperature 1053 K and the pressure holding time 60 s. Both preheating and casting were carried out in air without any protection.

All the composites prepared were hybridized with SiC particles with a fraction of about 1.0 vol %. The fibre volume fraction of all the composites was measured according to the systematic point-count method. The points counted were 360 in a measuring area of 96 mm × 60 mm. For every specimen 12 to 20 regions of optical cross-section photographs (× 1000) were counted and an average value was calculated as the fibre volume fraction of the composite. The pure aluminium and Al–Si matrix composites and some of the 6061 alloy matrix composites were examined in the as-cast condition (AC). The others of 6061 alloy matrix composites were investigated after being subjected to a T6 heat treatment (T6). The T6 heat treatment was directly conducted on the as-cast composite billets so that the material in the centre of the billets could be protected from oxidation during solution treatment. The conditions of the T6 treatment were 793 K for 1 h followed by an immediate water quenching for solution treatment and 413 K for 18 h for artificial ageing. These conditions were determined by measuring the Vickers hardness of a set of unreinforced 6061 alloy samples T6-treated in various conditions in the present study.

TABLE I Typical properties of the raw materials used

Material	Diameter (μm)	Tensile strength[a] (MPa)	Tensile modulus (GPa)
Reinforcement			
HMCF	6.7	2331	359
HSCF	7.0	3160	236
Additive			
α-SiC particles	0.6	–	–
Matrix			
99.99% aluminium	–	38	–
Al–12 wt % Si alloy	–	172	–
6061 Al alloy	–	158	–
6061 Al alloy-T6	–	294	–

[a] Values measured in the present work.

"Dog-bone" shaped longitudinal tensile specimens of CF/Al composites were prepared having a 25 mm gauge length, a width of about 4 mm and a thickness of 1–2 mm. The transverse tensile strength (TTS) was also measured with a sample 30 mm in length, 8 mm in width and 1–2 mm in thickness. All the longitudinal and transverse tensile specimens were polished with SiC papers to 1000 grit before tensile testing. The tensile testing was performed at room temperature with an Instron-type testing machine at a crosshead speed of 8.3 μm s^{-1}. The fracture morphology of both longitudinal and transverse tensile specimens was observed using scanning electron microscopy (SEM).

All types of CF/Al composite were observed with an optical microscopy in order to clarify their infiltration characteristics and the interfaces in these composites were investigated with transmission electron microscopy (TEM). The specimens for TEM observation were mechanically polished to 30–50 μm thick and then thinned by ion-thinning with the following sequence: 18° and 5 kV for 3 h, 15° and 4.5 kV for 10 h and 10° and 4 kV for 4–5 h (The angle is the included angle between ion-thinning beam and the sample.)

In order to understand the strength change of the carbon fibres after composite fabrication, single-fibre tensile tests were performed on the fibres after preheating and extraction from CF/Al composites. The extraction of carbon fibres from CF/Al composites was accomplished with a 10 wt % NaOH aqueous solution. The number of fibres tested was 50 with a gauge length of 25 mm. Tensile testing was carried out at a crosshead speed of 8.3 μm s^{-1}. The results obtained were evaluated via the Weibull distribution theory.

3. Results and discussion

3.1. Infiltration of aluminium matrices

Fig. 1 shows optical micrographs of the cross-sections of HMCF/P–Al (Fig. 1a), HMCF/6061 (Fig. 1b), HSCF/Al–Si (Fig. 1c) and HSCF/6061 composite (Fig. 1d). It is apparent from these micrographs that satisfactory infiltration of aluminium matrices into carbon fibres was achieved. Negligible porosity and few unwetted regions could be observed in the com-

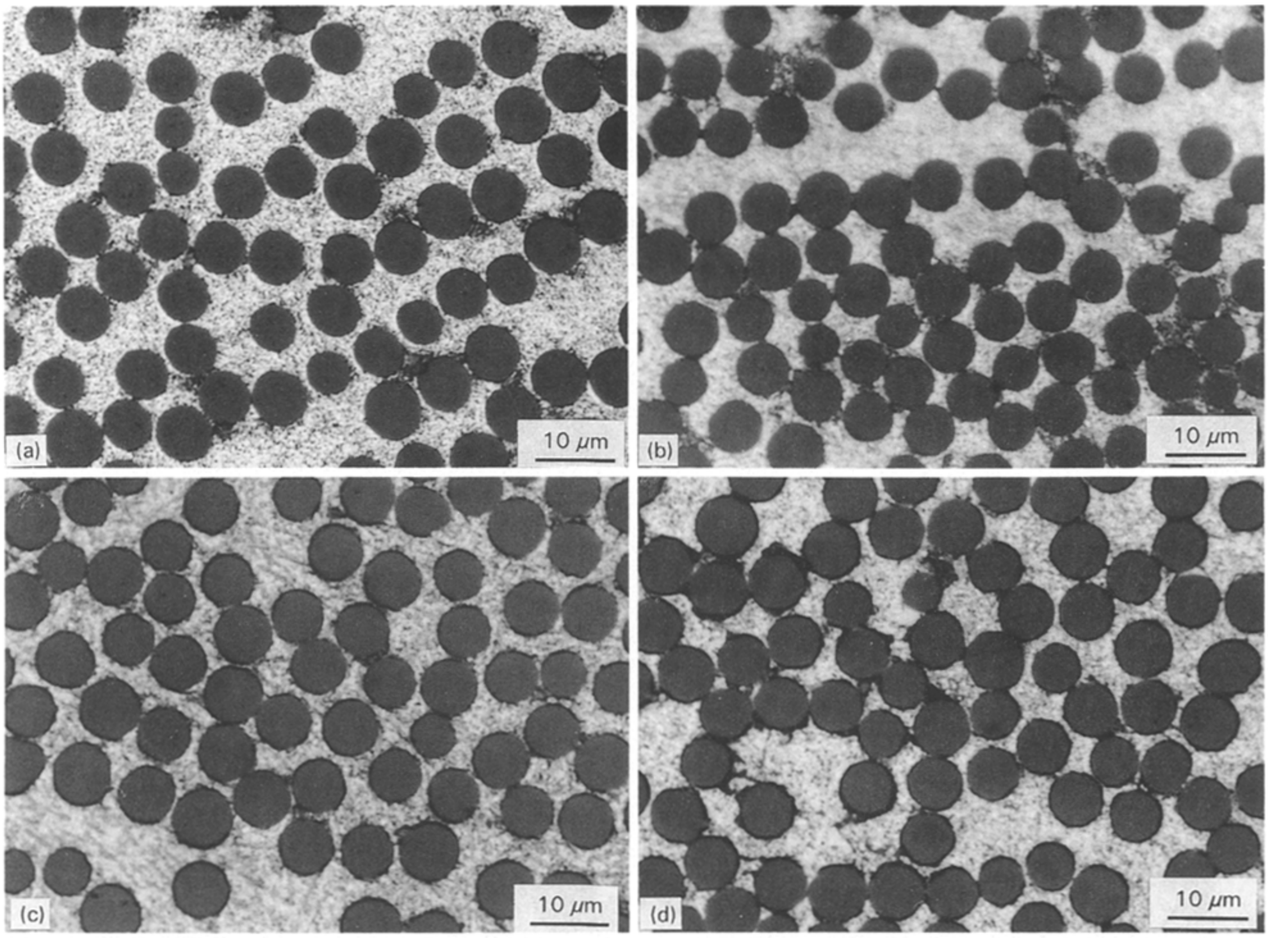


Figure 1 Optical micrographs of cross-sections of composites: (a) HMCF/P–Al, (b) HMCF/6061, (c) HSCF/Al–Si, (d) HSCF/6061.

posites. In all cases carbon fibres were distributed relatively uniformly in the aluminium matrix and only a few of them made contact with one another, because the hybridization of SiC particulate among the carbon fibres results in good impregnation of aluminium into fibre preforms [9]. There seemed to exist an interfacial zone between high-strength carbon fibres and the matrix in the case of HSCF/Al alloy composites (Fig. 1c and d). Nevertheless, there was no evidence of an interface in HMCF/Al composites (Fig. 1a and b), although the preparation conditions were completely the same. Furthermore, it was observed in the experiments that pure aluminium was the worst matrix to infiltrate the carbon fibre preforms because of unsuccessful casting, which rarely occurred with the other matrices.

3.2. Fibre strength degradation of CF/Al composites

In order to investigate the strength degradation of carbon fibres during fabrication, single-fibre tensile tests were performed on the as-received carbon fibres as well as on those taken from the preforms after preheat treatment and those extracted from the fabricated composites. The results obtained are illustrated in Fig. 2, along with the computed degradation rate which is the ratio of the strength decrease of the treated carbon fibres to the strength of the as-received carbon fibres.

In the case of high-modulus carbon fibres, as shown in Fig. 2a, the fibre strength was degraded only marginally after preheating. This decrease of fibre strength was considered to be caused by mechanical damage during preparation of fibre preforms by hand. In Al–Si and 6061 matrix composites the carbon fibres maintained a strength of about 85% of the as-received counterpart. However, those in the pure aluminium matrix composite and in the T6-treated 6061 matrix composite were deteriorated with degradation rates of 26 and 22%, respectively. Because chemical interactions at the interface directly affect the strength of carbon fibres, the strength change of carbon fibres in those composites implies that aluminium matrices have an important influence on the interfacial characteristics. There exists a sequence of matrix activity, i.e. from the highest, pure aluminium, to 6061 alloy, and to the lowest, Al–Si alloy. The strength of carbon fibres in the T6-treated 6061 matrix composite was further degraded by the T6 treatment, perhaps because interfacial reactions occurred during the solution treatment at the temperature of 793 K. However, whether reaction products are formed could not be observed by either SEM or TEM. Fig. 3 illustrates the etched cross-section (Fig. 3a) and the interface (Fig. 3b) of HMCF/6061 composite after the T6 treatment. On the circumference of the carbon fibres it could be observed that no foreign substance formed. This fact indicates that the chemical interactions be-

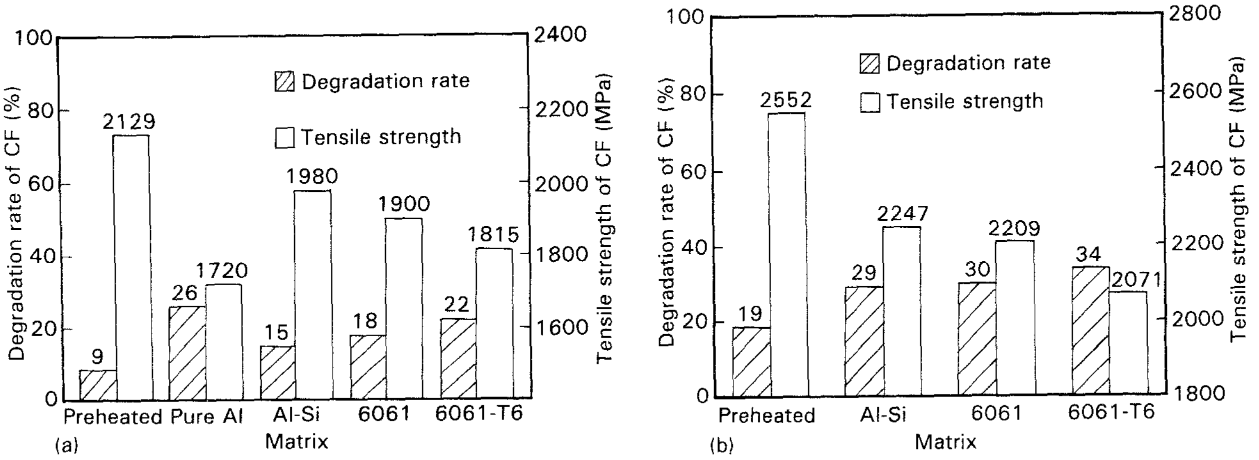


Figure 2 Tensile strength and degradation rate of carbon fibres in CF/Al composites: (a) high-modulus carbon fibres, (b) high-strength carbon fibres.

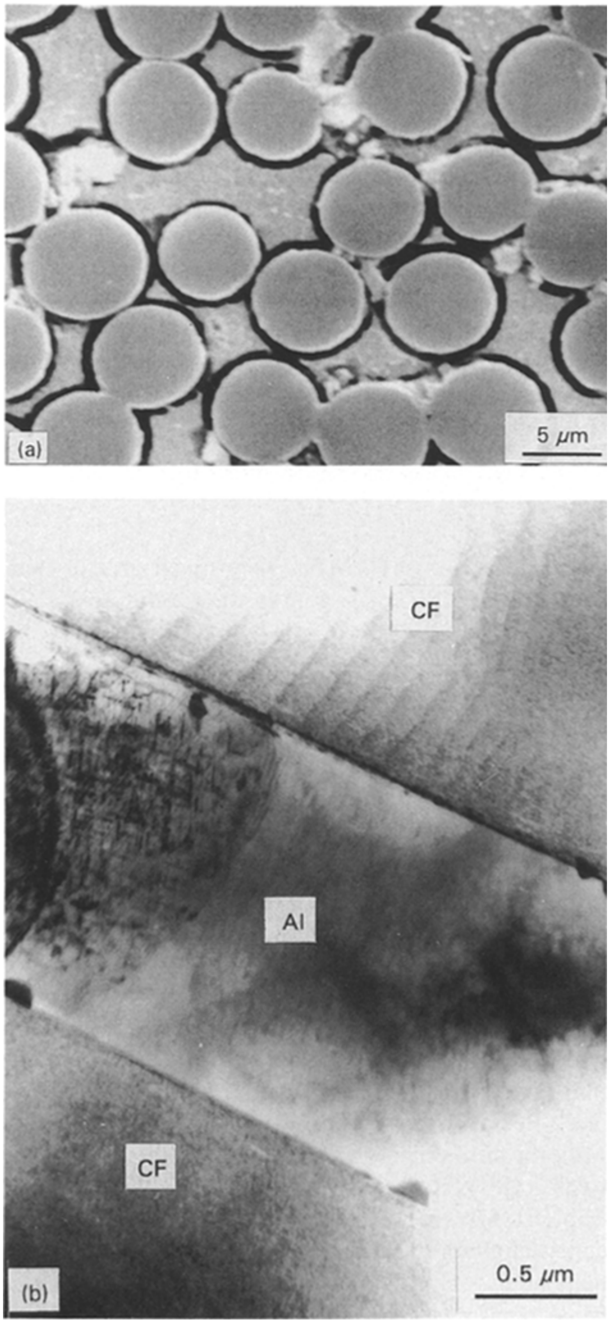


Figure 3 SEM and TEM micrographs of the interface in the HMCF/6061-T6 composite: (a) SEM observation, (b) TEM observation.

tween the carbon fibres and the 6061 matrix were not greatly accelerated, although the fibre strength was decreased, by T6 treatment. In the pure aluminium matrix composite, although the fibre strength was decreased greatly, interfacial interactions could not be directly observed by either SEM, TEM or even electron probe microanalysis (EPMA) in the experiment.

On the other hand, from Fig. 2b it can be seen that in the case of high-strength carbon fibres, the fibre strength was decreased by preheating treatment and the degradation rate was as high as 19%. Obviously, this behaviour cannot be explained by mechanical damage during processing. However, other reasons for the unexpectedly rapid decrease of the strength were not definitely identified. The most probable factor considered may be the oxidation of high-strength carbon fibres during preheating treatment in air. It is known that high-strength carbon fibres begin to oxidize at much lower temperatures than high-modulus carbon fibres. Thus, oxidation of the high-strength carbon fibres might take place during preheating which lasted about 3 h from 573 to 648 K, and consequently cause the degradation of fibre strength. Because of the high reactivity of high-strength carbon fibres with aluminium, the strength degradation of the fibres after casting and after the T6 treatment was very marked. The degradation rate was about 30% for the fibres both in the 6061 matrix composite and in the Al–Si matrix composite, and seemed not to depend upon the type of aluminium matrix.

The fact of the light degradation of high-modulus carbon fibres and the heavy degradation of high-strength carbon fibres in the Al–Si and 6061 alloy matrix composites is consistent with the results reported by other researchers [3, 10, 11]. This difference in degradation behaviour according to fibre type can be generally attributed to the differences in crystal structure and surface properties which determine the reactivity of carbon fibres. Fig. 4 shows an etched surface of the HSCF/6061 composite after T6 treatment. Comparing it with that of the HMCF/6061-T6 composite shown in Fig. 3, it clearly indicates the difference in interfacial morphology between the two types of carbon fibre: obviously the interfacial interactions were much more active in the former than in

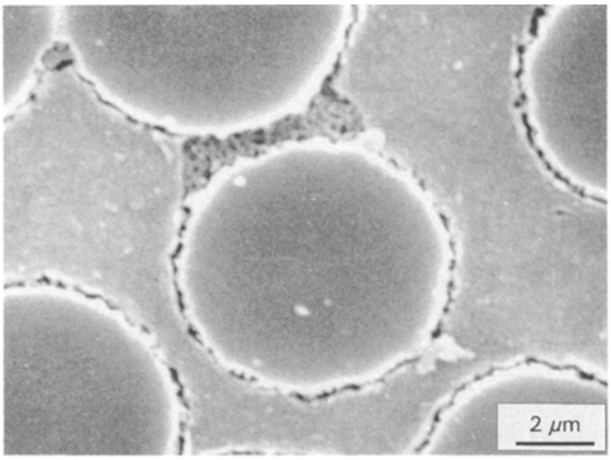


Figure 4 An interface between the fibre and the matrix in HSCF/6061-T6 composite.

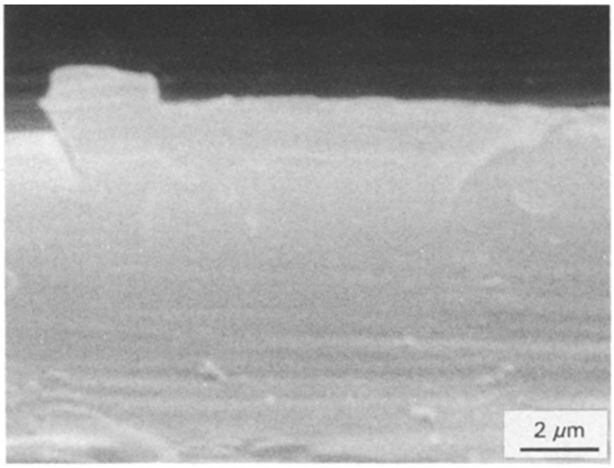


Figure 5 SEM micrograph of a fibre on the transverse fracture surface of HSCF/Al–Si composite.

the latter. As a result, the fibre strength in the former case was greatly degraded.

Interfacial chemical interactions also occurred in the HSCF/Al–Si composite, as shown in Fig. 5. It is found from Fig. 5 that the high-strength carbon fibre has reacted in some degree with the molten aluminium and formed a discontinuous layer on the surfaces of carbon fibres which was probably the brittle aluminium carbide.

3.3. Tensile strength of CF/Al composites

Table II presents the properties of the CF/Al composites obtained from tensile tests. ATEFS and RTEFS in the table, which stand for the apparent and real transfer efficiency of fibre strength, respectively, are defined elsewhere by the authors [12]. They were calculated using the strengths of the as-received carbon fibres and of fibres extracted from the as-fabricated composites. In the calculation, the contribution of SiC particulate (about 1 vol %) was ignored because of its small volume fraction and ineffectiveness as a reinforcement compared to carbon fibres.

As shown in Table II, all the composites had a similar fibre volume fraction of about 41 vol %. The longitudinal tensile strength (LTS) values of these composites, however, varied considerably from the greatest value of 1020 MPa to the lowest of 407 MPa. In particular, the values for the HMCF/Al composites were much greater than those of the HSCF/Al composites, although the tensile strength of the high-strength carbon fibres was much higher than that of the high-modulus carbon fibres. On the basis of the transfer efficiency of fibre strength, the difference between HMCF/Al and HSCF/Al composites was even more outstanding from 0.19, the lowest, for the HSCF/6061-T6 composites to 0.89, the highest, for the HMCF/6061-T6 composites. These results suggest that the high-modulus carbon fibres have effectively strengthened the aluminium matrices, but the high-strength carbon fibres have hardly transferred their excellent strength to the composites. Nevertheless, in contrast to the LTS, the TTS values of the HSCF/Al composites, above 60 MPa, were much greater than those of the HMCF/Al composites, about 16 MPa.

TABLE II Properties of CF/Al composites obtained[a]

Composite	V_f (vol %)	LTS (MPa)	TTS (MPa)	ATEFS	RTEFS
HMCF/P–Al	40	575	–	0.60	0.81
HMCF/Al–Si	40	807	16	0.76	0.90
HMCF/6061-AC	40	811	16	0.77	0.94
HMCF/6061-T6	41	1020	20	0.89	1.14
HSCF/Al–Si	41	412	61	0.24	0.34
HSCF/6061-AC	42	407	63	0.24	0.34
HSCF/6061-T6	41	425	83	0.19	0.30

[a] LTS = longitudinal tensile strength, TTS = transverse tensile strength, ATEFS = apparent transfer efficiency of fibre strength, RTEFS = real transfer efficiency of fibre strength (see text).

3.4. Effect of interfacial characteristics on properties of CF/Al composites

It is worth noting that the carbon fibres in HMCF/6061-T6 composite have more efficiently reinforced the matrix, with an ATEFS value of 0.89, than those fibres in the HMCF/6061-AC composite with an ATEFS of 0.77. In contrast, the apparent strength transfer efficiency of carbon fibres was 0.19 and 0.24, respectively, for HSCF/6061 in the T6-treated state and in the as-cast state. In the high-modulus carbon fibre–aluminium system, the interfacial interactions between carbon fibres and aluminium matrices could hardly happen in the as-cast composites. This means that almost no chemical bonding exists in the interface. Because the surfaces of the fibres are quite smooth, mechanical bonding is also very weak. The weak transverse tensile strength, 16 MPa given in Table II for the as-cast HMCF/Al composites, is consistent with this implication because the transverse behaviour is closely and directly related to the interfacial bonding properties of a composite [13]. Thus, the interface bonding of HMCF/Al composites is very weak between the fibres and the matrix, and this results in a lower fibre strength transfer.

However, since it is known that above 773 K chemical interactions occur between high-modulus carbon fibres and the aluminium matrix, some chemical bonding might be formed in HMCF/6061-T6 composites during the solution treatment at 793 K for 1 h. This would improve the bonding strength of the fibres to the matrix. The increase of the transverse tensile strength of the T6-treated HMCF/6061 composites compared to the as-cast value shown in Table II agrees with the above discussion. Therefore, these facts mean that it is important to be able to control the degree of bonding between aluminium matrices and carbon fibres so that the applied load can be effectively transferred from the matrix to the fibres through fibre–matrix interfaces.

In the case of the high-strength carbon fibre–aluminium system, because of the stronger interfacial interactions between the fibres and aluminium matrices occurring during casting, the TTS values of the as-cast HSCF/Al composites were 61 and 63 MPa, respectively, for the HSCF/Al–Si and HSCF/6061 composites, which were much higher than those of the HMCF/Al composites. This implies directly that a very strong interface existed in this system. Fig. 6 is a comparison of the transverse fracture surfaces of the HMCF/6061-T6 and HSCF/6061-T6 composites. It can be seen that the transverse failure of the HMCF/6061-T6 composite was through interfacial debonding at the interface of carbon fibres and aluminium matrix. However, the failure of HSCF/6061–T6 composite was caused by fibre shear and interlamilar rupture because the strong interface bonding could not allow debonding failure to occur at the interface.

Therefore, from the above-mentioned observations on the 6061 matrix composites it can be concluded that some controlled amount of reaction at the matrix–reinforcement interfaces may even be desirable for achieving good transverse properties and obtaining a bonding between the reinforcements and the matrix strong enough to effectively transfer the applied stress from the matrix to the fibres; however, too strong an interfacial bond will adversely degrade the composite longitudinal properties.

3.5. Relationship between longitudinal fractography and properties of CF/Al composites

It can be summarized so far from the above results that the reasons for the variation in the strengthening behaviour with fibre type appear to be the following: (i) high-strength carbon fibres were degraded by oxidation during the preheat treatment of fibre preforms, and (ii) the high-strength fibres were heavily attacked by aluminium matrices during casting and T6 treatment. The single-fibre tensile tests showed that the strengths of the fibres extracted from the cast and T6-treated composites were considerably decreased as shown in Fig. 2b. The two reasons given above are consistent with the single-fibre tensile tests and the observations on the interfacial chemical interactions between carbon fibres and aluminium matrices. However, they cannot sufficiently explain the changes from the standpoint of apparent and real strength transfer efficiencies of carbon fibres. Thus, there must be some other important factors to be considered.

Fig. 7 shows the longitudinal tensile fracture surfaces of HMCF/Al composites. It can be seen that there were various fractographic patterns, indicating that these composites have different interfacial bonding. On the fracture surfaces of HMCF/Al–Si and HMCF/6061-AC composites (Fig. 7a and b), long and numerous fibre pull-outs were observed, which is a characteristic of a weak interface. This long fibre-pull-out fracture pattern relates to a good transfer efficiency of fibre strength into the composites, as could be seen from Table II.

Nevertheless, as Fig. 7c shows, the fibre pull-out on the fracture surfaces of HMCF/6061-T6 composite was much shorter than that of the composites in the as-cast state (Fig. 7b), which corresponds to a very high transfer efficiency of fibre strength. This observation proves again that the bonding has been enhanced somewhat after the T6 treatment, and that the interface bonding of the HMCF/6061-T6 composite was stronger than that of HMCF/6061-AC. Therefore, it can be concluded that the improvement in bonding can to some extent make a more efficient strength

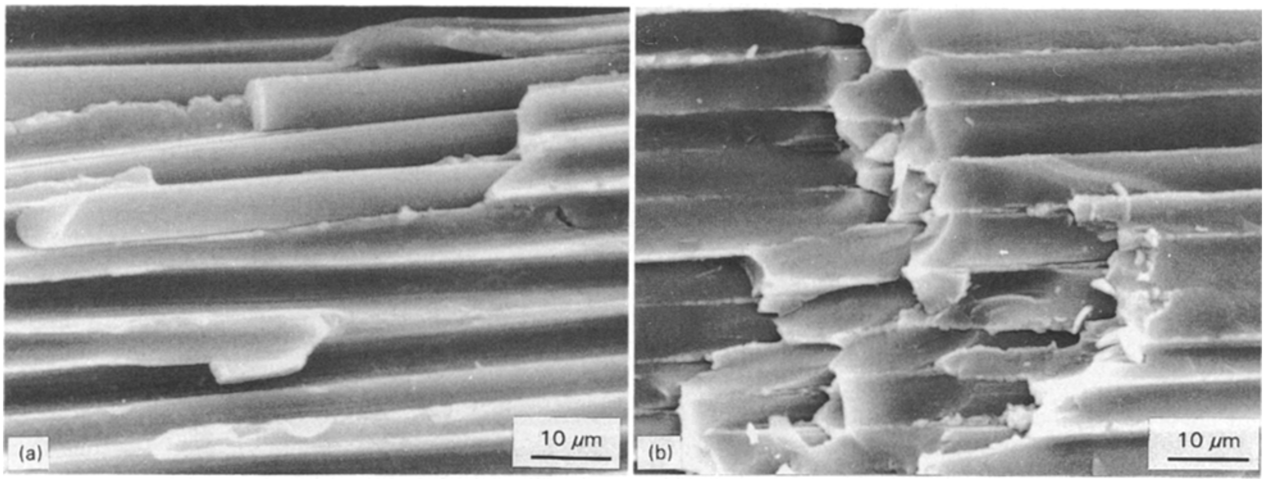


Figure 6 SEM micrographs of the transverse fracture surfaces of composites: (a) HMCF/6061-T6, (b) HSCF/6061-T6.

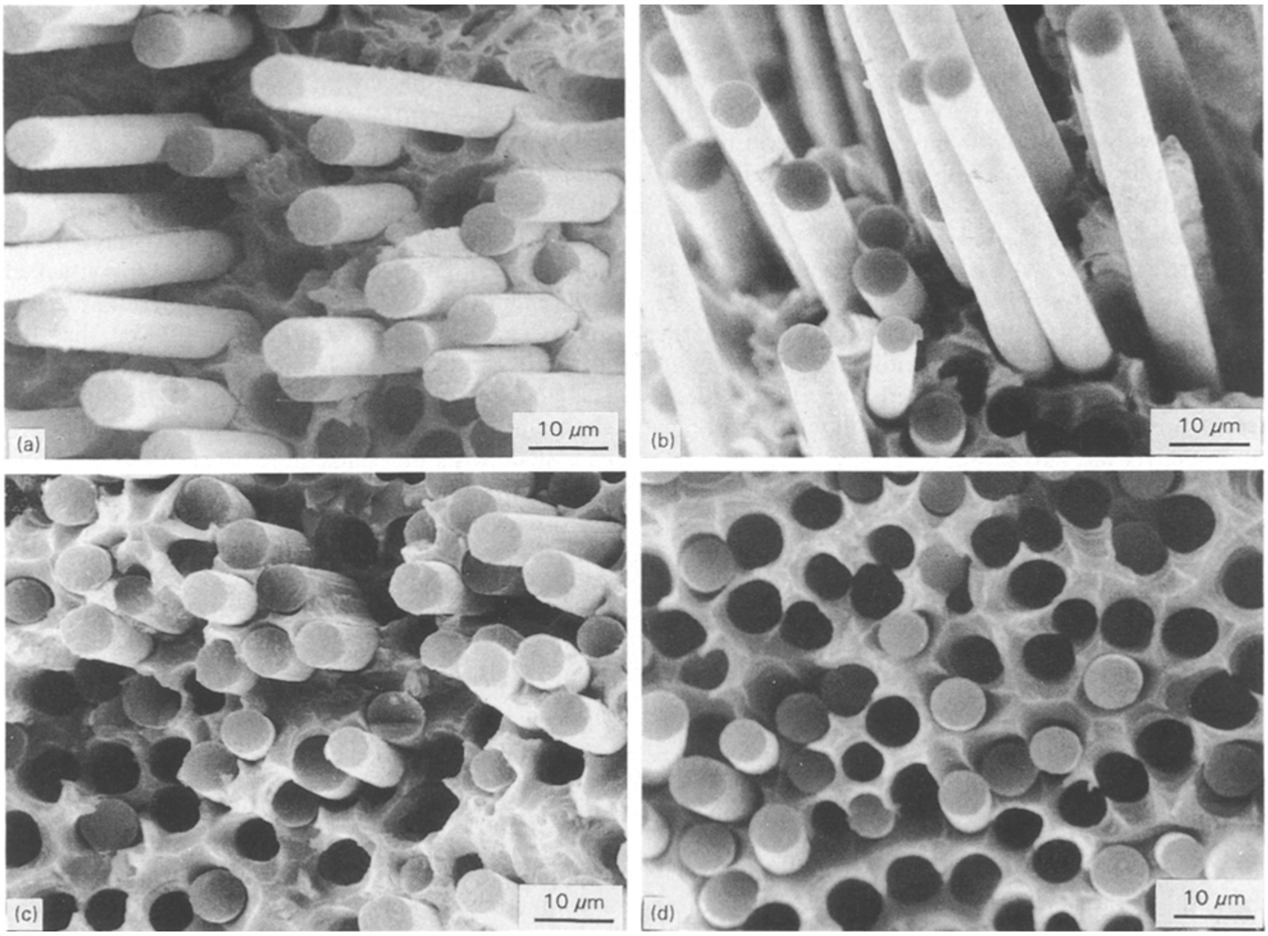


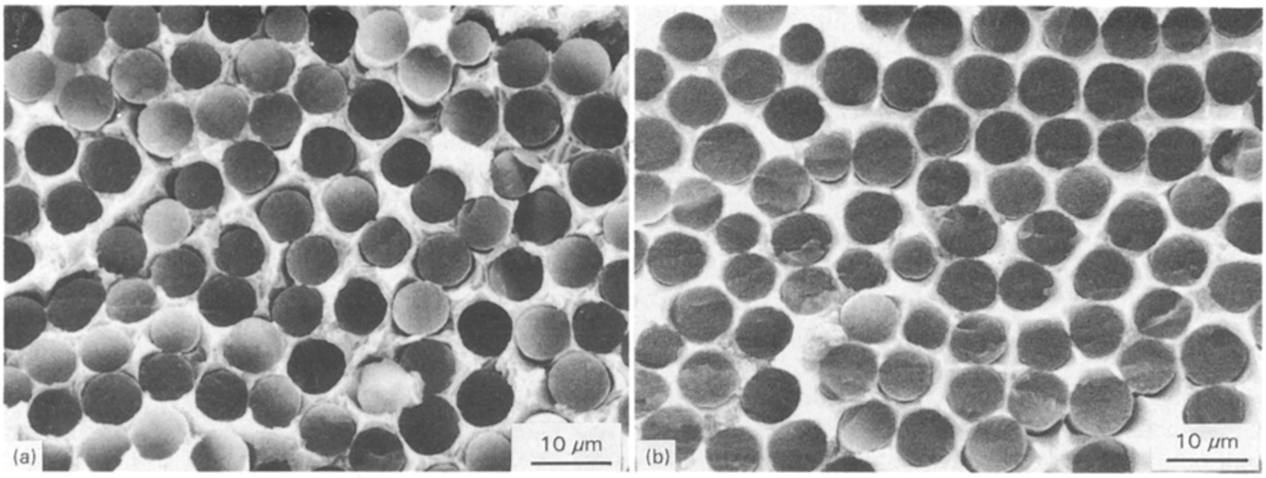
Figure 7 Longitudinal tensile fracture morphologies of composites: (a) HMCF/Al–Si, (b) HMCF/6061-AC, (c) HMCF/6061-T6, (d) HMCF/P–Al.

transfer from carbon fibres to the composites, and that a very good longitudinal tensile strength corresponds to an intermediate bonding that causes an intermediate fibre pull-out.

The pure aluminium matrix composite had a relatively flat fracture surface with some short fibre pull-out on it (Fig. 7d). This suggests that interfacial interactions might have happened between the fibres and the matrix, in comparison with those of the other two matrices discussed previously. This pattern with little fibre pull-out relates to a longitudinal strength intermediate between those of the other HMCF/Al and HSCF/Al composites.

The high-strength carbon fibre–aluminium system has a high reactivity between the fibres and the aluminium matrix which can result in formation of a carbide phase at the fibre–aluminium interface. It has been reported that the reaction-formed carbide phase grows as acicular inclusions that embed themselves both in the matrix and in the fibres to make interfacial bonding very strong and cause surface defects on the carbon fibres [14]. It is therefore considered that the carbide phase and the strong interfacial bonding not only result in premature fibre damage by a notch effect but also cause the unexpected planar fracture of CF/Al composites without fibre pull-out on the fracture surface. Fig. 8 is the longitudinal tensile fracture morphology of HSCF/Al composites showing an approximately planar fracture with no fibre pull-out, although partial debonding of the interface between the fibres and aluminium matrices took place. From Fig. 8a and b it can be seen that cracks originated at the brittle interfacial zone and propagated along the fibre–matrix interface near the fracture surface, but the cracks could not completely debond the interface to let the fibres pull out. Moreover, the notch effect of brittle products caused a high stress concentration and consequently a premature fibre failure. This planar fracture mode made the carbon fibres unable to fully use their strength before final failure of the composite. Furthermore, it is worth noting the direction of crack propagation in the 6061 matrix composite after T6 treatment as shown in Fig. 8c. Because the solution treatment caused further chemical interactions between the fibres and the matrix, cracks originating in the interface proceeded through the fibres and split them into two parts instead of debonding the interface. This implies that the interfacial strength between the fibres and the matrix was even stronger than the shear strength of the carbon fibres. These cracks undoubtedly deteriorated the composite properties.

The above observations indicate that cracks in HSCF/Al composites cannot debond the interface between carbon fibres and aluminium matrices to propagate along the fibre axis, because of the strong



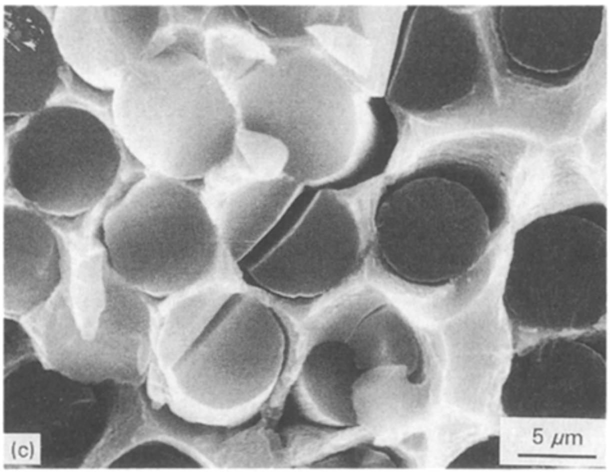


Figure 8 Longitudinal tensile fracture morphologies of composites: (a) HSCF/Al–Si, (b) HSCF/6061-AC, (c) HSCF/6061-T6.

interfacial bond produced by the probable formation of aluminium carbide, but instead propagate through the plane of fracture by breaking carbon fibres perpendicular to the fibre axis. Fig. 9 is an SEM micrograph of HSCF/Al–Si composite taken in the direction parallel to the fibre axis in a specimen prepared from a tensile-tested sample. The existence of cracks perpendicular to the fibre axis around the interface of the carbon fibre and aluminium matrix can be clearly observed at the locations denoted by arrows. It can also be suggested from this micrograph that considerable attack on the carbon fibre has happened from the evidence of many deep grooves on its surface.

In general, the above results and discussion imply that the interface characteristics of carbon fibres and aluminium matrices have an important bearing on the transfer efficiency of fibre strength and the mechanical properties of CF/Al composites. An intermediate fibre pull-out on the fracture surface, i.e. an intermediate interfacial bonding and no heavy interfacial reactions, corresponds to a good transfer efficiency of fibre strength.

4. Conclusions

The properties and microstructures of high-strength and high-modulus carbon fibre-reinforced composites with various aluminium matrices have been characterized. The composites evaluated were fabricated by a pressure casting technique based on the hybridization method. The following conclusions can be drawn from the present investigation:

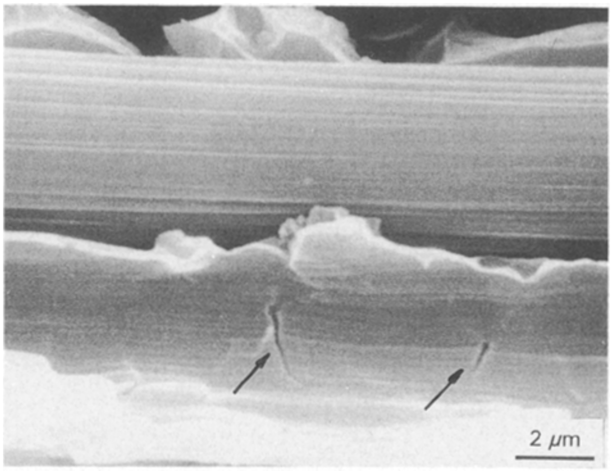


Figure 9 SEM micrograph of HSCF/Al–Si composite. Note the perpendicular cracks denoted by arrows.

1. The strength degradation of high-modulus carbon fibres in aluminium matrix composites is not heavy and depends upon the composition of the matrix. The sequence of the compatibility of aluminium matrices with the fibres is Al–12 wt % Si alloy to Al–Mg–Si (6061) alloy to 99.99% aluminium. On the other hand the strength of high-strength carbon fibres in aluminium matrix composites is greatly degraded, with degradation rates of more than 30%. Moreover, the degradation seems to be independent of the variety of aluminium matrices.

2. The longitudinal tensile strength of CF/Al composites is very different between HMCF/Al and HSCF/Al composites. The HMCF/Al composites achieve a longitudinal tensile strength of more than 800 MPa, but the HSCF/Al composites obtained have only about 400 MPa, although the fibre volume

fraction of both types is similar. In contrast, the transverse tensile strength of the HSCF/Al composites, above 60 MPa, is much higher than that of the HMCF/Al composites, about 16 MPa.

3. The microscopic observations show some evidence of chemical interactions at the interfaces of high-strength carbon fibres and aluminium matrices. There is no evidence for chemical products between high-modulus carbon fibres and Al–Si alloy and 6061 alloy matrices. However, it is considered that some interfacial reactions take place in pure aluminium matrix composites.

4. Observation of the fracture morphology of CF/Al composites indicates that an intermediate fibre pull-out and a planar fracture pattern correspond respectively to good and very poor LTS and fibre strength transfer.

5. The experimental results and discussion suggest that the interfacial bonding between carbon fibres and aluminium matrices has an important bearing on the mechanical properties of CF/Al composites, and an intermediate interfacial bonding is expected to achieve good longitudinal and transverse tensile strengths of CF/Al composites.

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