Fabrication and Characterization of SiC-Hybridized Carbon Fibre Reinforced Aluminium Matrix Composites

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Carbon fibre reinforced aluminium matrix(CF/Al)composites were fabricated with a hybridization technology through pressure casting and characterization of the hybrid composites obtained was carried out. Hybridization with SiC additive could improve the infiltration performance of fibre preform, control the fibre volume fraction, and increase the fibre-strength transfer efficiency of the hybrid composites. The longitudinal tensile strength of the hybrid composites was greatly improved compared with that of the conventional composite. It was found that the strength deterioration of high-modulus cabon fibres(HMCFs) during fabrication was not heavy and depended upon the type of aluminium matrix, while the strength of high-strength carbon fibres(HSCFs)was greatly reduced by aluminium matrix, but the degradation did not depend on the variety of the matrix. As a result, the longitudinal tensile strength of hybrid HMCF/Al composites was higher than 800 MPa, while that of hybrid HSCF/Al composites was about 400 MPa. Contrariwise, the transverse tensile strength of hybrid HSCF/Al composites was much greater than that of hybrid HMCF/Al composites. The results of thermal exposure of hybrid HMCF/Al-Si composites at 773 K showed that their longitudinal tensile strength increased, then decreased, while their transverse tensile strength increased montonically, as the exposure was proceeded. The thermal exposure slightly reduced fibre strength, caused chemical interactions, and resulted in changes of fracture morphology and fibre pull-out length of the composites. It was also concluded that fibre/matrix interfacial bonding has an important effect on the mechanical properties of CF/Al composites and intermediate interfacial bonding can be expected to result in good longitudinal and transverse tensile strengths of CF/Al composites.

1. Introduction

Metal matrix composite materials(MMCs)have been developed to offer attractive properties including high tensile strength and stiffness, improved fatigue life, and good mechanical properties at elevated temperatures. Many fabrication technologies for MMCs have been developed. One of the most successful fabrication technologies is of intruding liquid metal into a fabric on prearranged fibre configuration called a preform. This intrusion can be carried out under pressure, vacuum, or combination of pressure and vacuum. The techniques able to be utilized in this area include die casting, pressure infiltration, and vacuum infiltration, etc. The potential advantage for these processes is near-shape part fabrication in a simple and cost-effective manner. Among these techniques, pressure infiltration, often called pressure casting or squeezing casting, is the most commonly used technique. Some developments[1, 2] show that, as one of the most important MMCs, carbon fibre reinforced aluminium(CF/Al)

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composites can also permit the use of this technique, thereby eliminating the need for precursor wires as an intermediate step.

Nevertheless, this pressure casting process also suffers from critical problems. (i) Because aluminium does not spontaneously wet carbon fibres, the direct infiltration of molten aluminium into the preform of carbon multifilaments needs high pressure and high melt temperatures. As a result, not only do fibre contacts in the CF/Al composites obtained occur and composite failures initiate at these contact points[3], but also chemical interaction on fibre/matrix interface is accelerated by the high temperatures. (ii) The fibre volume fraction of CF/Al composites is difficult to control in a sense of satisfying various requirements of different customers with the minimum fibre consumption, or saving expensive carbon fibres.

Recently, one entirely different approach has been developed for the manufacture of CF/Al composites directly by a casting process. It is based on the pre-distribution of SiC whiskers or particles among carbon fibres[4]. It has been reported that the longitudinal flexural strength of the hybrid composites was improved in comparison with that of the conventional composites.

Although CF/Al composites can be developed using the above fabrication methods to offer excellent mechanical properties, all these properties strongly depend on fabrication methods and conditions, varieties of carbon fibres, and aluminium matrixes. Moreover, it is known that, though the properties of CF/Al composites may be excellent at ambient temperatures, in most cases, they are disappointing at elevated temperatures. This is because chemical interactions occur at the interface of carbon fibres and aluminium matrix, causing fibre degradation and matrix embrittlementt[5,6].

A number of research works on the fibre degradation of CF/Al composites and on the properties of CF/Al precursor wires after thermal exposure have been presented[2,7-9]. Nevertheless, most of these investigations were mainly focused on the strength degradation of carbon fibres and on the properties of CF/Al composite precursor wires after heat treatment; they did not deal with the behaviour of CF/Al composite bodies. Because CF/Al composites are generally considered for applications at intermediate or elevated temperatures, it is important to understand the thermal exposure behaviour of CF/Al composite bodies and the effect of thermal exposure on the fibre/matrix interface of CF/Al composites.

In the present work, unidirectional SiC particle-and SiC whisker-hybridized carbon fibre reinforced aluminium matrix composites (abbreviated as SiC_p−CF/Al and SiC_w−CF/Al composites, respectively) were prepared through pressure casting and were characterized. The functions of SiC particles or whiskers in the infiltration process and their effect on the strength of the composites obtained were discussed. Finally, the effect of thermal exposure on the longitudinal and transverse tensile strengths and on the fibre degradation behaviour of hybrid CF/Al composites were investigated.

2 . Experimental

The carbon fibres used in this investigation were polyacrylonitrile (PAN) based high-modulus and high-strength fibres (HMCFs and HSCFs, respectively). Some properties of the fibres, along with those of SiC particles and whiskers and aluminium alloys, are listed in Table 1. The tensile strengths of carbon fibres and aluminium alloys were measured in the present work, whereas the other data in Table 1 were quoted from the manufacturers' technical specifications. Tensile strength test of carbon fibres was performed on single fibres and the number of testing was 50 having a gauge length of 25 mm at a crosshead speed of 8. 3 μm/s. The results were evaluated via the Weibull distribution theory.

The sequences adopted to prepare hybrid carbon fibre preforms included the following two main steps.

(i) Distribution of SiC particles or whiskers among carbon fibres. The carbon fibres were impregnated into an aqueous suspension of 0−15 wt% SiC particles or whiskers, using a polymer as a binding agent and an organometallic compound as a dispersing agent. After this treatment, the particles and whiskers were rather uniformly distributed among the carbon fibres with the binding of the binder. However, when additive concentra-

Table 1 Typical properties of the raw materials used

Material	Diameter (μm)	Tensile Strength*(MPa)	Tensile Modulus(GPa)
Reinforcement:			
HMCF	6.7	2331	359
HSCF	7.0	3160	236
Additive:			
α−SiC particle	0.6	− − −	− − −
β−SiC whisker	0.1−0.6	− − −	− − −
Matrix:			
99.99%Aluminium	− − −	38	− − −
Al−12 wt% Si alloy	− − −	172	− − −
6061 Al Alloy	− − −	158	− − −
6061 Al Alloy−T6	− − −	294	− − −

*the values measured in the present work

tion of the suspension was high, the additives agglomerated between carbon fibres. This behaviour was especially obvious in the case of SiC whiskers because of their longer length.

(ii) Carbon fibre preform preparation. The impregnated fibres were dried to a certain extent, then cut to a length of 80 mm and laminated in one direction into a Shirasu Balloon (expanded volcanic glass) preform mould to obtain a fibre preform 80 mm long, 9 mm wide and 3−5 mm thick.

Then, the infiltration of hybrid carbon fibre preforms with molten aluminium was carried out through pressure casting. The preform described above was preheated to a temperature of 648 K along with cast iron mould, and aluminium matrix was heated to a temperature of 1053 K. The pressure casting was performed with a pressure of 49 MPa for 60 s. Both preheating and casting were practised in air.

The fibre volume fraction (V_f) of all the composites prepared was measured according to the systematic point-count method. The points counted were 360 in a measuring area of 96x60 mm^2. For every specimen, 12 to 20 pieces of optical cross-section photographs (x1000) were counted and an average value was calcuated as the V_f of the composite. The additive volume fraction (V_{add}) of hybrid composites was computed according to the relative volume of additives stuck on carbon fibres in the pre-treatment and the V_f of the related composites.

Carbon fibre reinforced pure aluminium (CF/P-Al), aluminium-silicon alloy (CF/Al-Si), and 6061 alloy (CF/6061) composites prepared with the hybridization method were examined in the as-cast condition (AC). CF/6061 composites were also investigated after subjected to a T6 heat treatment (T6). The T6 heat treatment was conducted on the as-cast composite billets so that the composites 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 hour followed by immediate water quenching for solution treatment, and 413 K for 18 hours for artificial ageing. These conditions were determined through measuring the Vicker's hardness of 6061-alloy samples T6-treated in various conditions in this study.

Some of SiC_p-HMCF/Al-Si composites were subjected to thermal exposure at a temperature of 773 K for 72, 144, 216, 288, and 360 hours, respectively. CF/Al composites are expected to be used at intermediate temperatures of 673 K to 773 K. Therefore the temperature of 773 K was selected as the experimental temperature. Moreover, Sawada and Bader[2] have reported that, in their study on the interface between carbon fibres and aluminium alloys, only after 240 hours at 773 K, some cracking, cavity formation, and fibre degradation were observed. Thus, thermal exposure time was chosen to cover several hundred hours to understand fully the strength change of CF/Al composites with thermal exposure. The thermal exposure was conducted in air with the as-cast billets so that the

composites in the centre of the billets could be protected from oxidation during the treatment.

From all these composites, "dumb-bell" shaped longitudinal tensile specimens having a gauge length of 25 mm, a width of 4 mm, and a thickness of $1-1.5$ mm were prepared. Transverse tensile samples were 30 mm in length, 8 mm in width, and $1-1.5$ mm in thickness. All the longitudinal and transverse tensile specimens were polished with SiC paper to 1000 grit before tensile testing. Tensile testing was performed at room temperature at a crosshead speed of 8.3 μm/s using an Autograph testing machine. The tensile fracture morphologies of these specimens were studied with a scanning electron microscope (SEM).

Some of CF/Al composites were also investigated with a transmission electron microscope (TEM) to clarify their interface characteristics. The specimens for TEM observation were mechanically polished to $30-50$ μm thick, and then thinned with an ion-thinner in the following sequence: 18 degrees and 5 KV for 3 hours; 15 degrees and 4.5 KV for 10 hours; and 10 degrees and 4 KV for 4 to 5 hours.

To understand the strength change of carbon fibres after composite fabrication and thermal exposure, single fibre tensile test was performed on the fibres after preheating and extracted 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 having a gauge length of 25 mm. Tensile testing was carried out at a crosshead speed of 8.3 μm/s. The results obtained were evaluated via the Weibull distribution theory.

3. Results and Discussion

3.1 Effect of SiC hybridization on preparation and properties of HMCF/Al-Si composites

3.1.1　Microstructure of hybrid HMCF/Al-Si composites

The microstructure of transverse cross sections of the HMCF/Al-Si composites obtained is shown in Fig. 1.
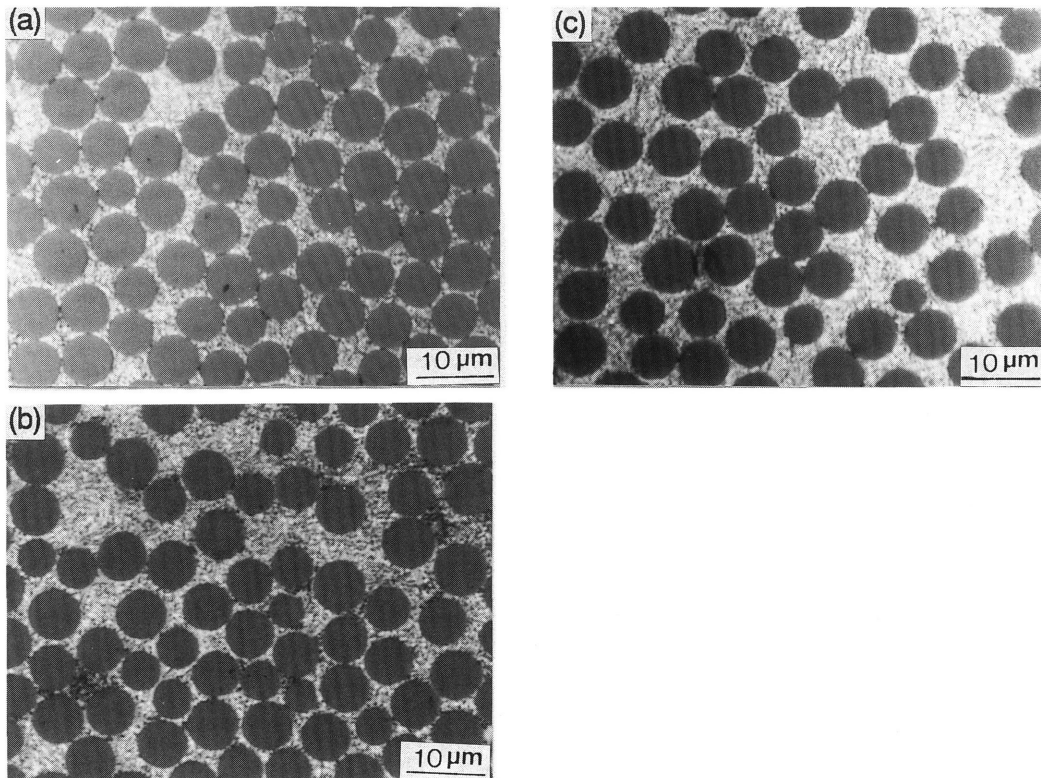

Fig. 1　Optical micrographs of transverse cross sections of (a) HMCF/Al－Si, (b) SiC_p-HMCF/Al-Si(V_{add}: 1.0 vol%) and (c) SiC_w-HMCF/Al-Si(V_{add}: 1.1 vol%) composites

Apparently, the structure of the hybrid composites is quite different from that of the conventional composite. It was found that the carbon fibres in the hybrid composites were infiltrated with aluminium matrix to yield a sound billet. Very few carbon fibres made contacts with one another and little voids existed between the fibres and the matrix. On the other hand, in the conventional composite shown in Fig. 1 a, the carbon fibres were closely packed one by one. Not only did most of them make contact with the neighbouring fibres, in some areas almost perfect hexagonal packing being observed, but also at the sites close to the contacts, impregnation of aluminium was incomplete, and voids existed. It can also be noted that the concentration of carbon fibres in the conventional composite was much higher than that in the hybrids. Comparing Fig.1b with c, there seems to be no big difference in microstructure in addition to their fibre volume fraction between SiC_p-HMCF/Al-Si and SiC_w-HMCF/Al-Si composites. Although the volume fraction of SiC particles was almost the same as that of whiskers, the distribution of carbon fibres in SiC_p-HMCF/Al-Si composites was denser than that of SiC_w-HMCF/Al-Si composites. This behaviour indicates that SiC whiskers can make larger spaces between carbon fibres than SiC particles.

3.1.2 Longitudinal tensile strength and fibre volume fraction of hybrid HMCF/Al-Si composites
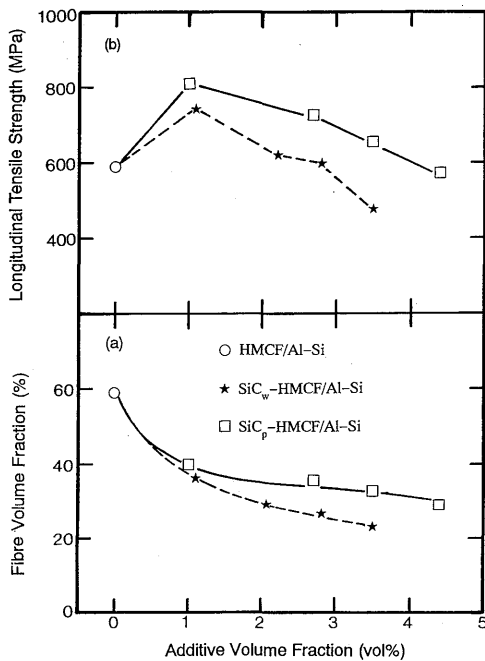

Fig. 2 Effects of additive volume fraction on (a) fibre volume fraction and (b) longitudinal tensile strength of hybrid HMCF/Al-Si composites·

The longitudinal tensile strength (LTS) and V_f of the hybrid HMCF/Al-Si composites obtained are shown in Fig. 2 as a function of the V_{add} of the composites. Evidently, as the V_{add} increased, the V_f of the hybrid composites decreased monotonically, because SiC particles and whiskers have widened the space among carbon fibres. However, the LTS of the hybrid composites did not have the same tendency as that of the V_f as illustrated in Fig. 2 b. The LTS increased sharply as the V_{add} increased to around 1 vol%, even though the V_f of the hybrid composites was lower than that of the conventional composite. After the V_{add} exceeded 1 vol%, the LTS went down mainly because of the obvious decrease of their V_f. Also, when much more than the suitable amount of SiC additives was added, because agglomerated grains were formed with extra SiC additives, the strength of the hybrid composites was reduced. Although the same amount of SiC particles or whiskers was distributed in the composites, the V_f of SiC_p-HMCF/Al-Si composites was a little higher than that of SiC_w-HMCF/Al-Si composites. As a result, the LTS of the former was also greater than that of the latter.
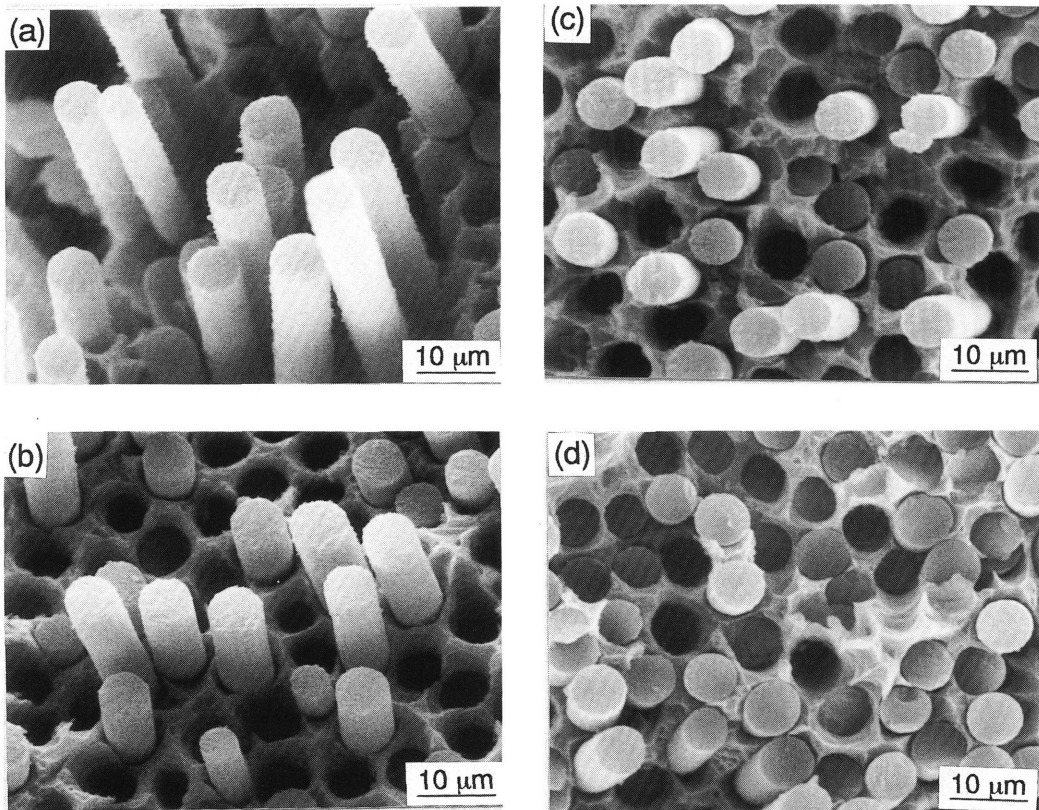
3.1.3 Fracture morphology of hybrid HMCF/Al-Si composites

Figure 3 gives the SEM fractographs of the hybrid HMCF/Al-Si composites. For the conventional composite shown in Fig. 3 a, it can be seen that most carbon fibres were totally pulled out from the aluminium matrix or stripped off the interface due to interlamilar failure. However, in the hybrid cases as shown in Fig. 3 b—d, although many carbon fibres were pulled out from the matrix as well, the pull-out length was much shorter than that in the conventional case, and interlamilar failure rarely existed. Further, when the amount of SiC additives increased, the pull-out length decreased, as observed in Fig. 3 b and c. Comparing Fig. 3 c with Fig. 3 d, it appears that the fracture behaviour of SiC_p-HMCF/Al-Si and SiC_w-HMCF/Al-Si composites was almost the same.

Fig. 3   SEM micrographs of longitudinal tensile fracture surfaces of (a) HMCF/Al-Si, (b) and (c) SiC$_w$-HMCF/Al-Si (V$_{add}$: 1.1 and 2.2 vol%, respectively), and (d) SiC$_p$-HMCF/Al-Si (V$_{add}$: 2.7 vol%) composites

The above observations indicate that the interfacial behaviour of the conventional and hybrid composites is different. It is known that there is generally a large difference in the thermal expansion coefficient of aluminium matrix and carbon fibres. The fabrication route involved cooling from elevated temperature to room temperature, and thus CF/Al composites contained thermally induced residual stress before any external loading. Moreover, the thermal residual stress would provide an initial compressive stress clamping the matrix to carbon fibres, resulting in improved interface bonding. Wilkinson[10] pointed out that, in continuous fibre reinforced MMCs, the average radial compression at the interface became smaller, as the intervening thickness of the matrix between fibres was reduced by increasing fibre content. This means that the clamping effect of the matrix around the fibres is decreased by the V$_f$ increase, which would weaken the interfacial bonding. Therefore, it can be understood that the interface bonding of the hybrid composites has been improved by their V$_f$ decrease, which caused shortened fibre pull-out length, and consequently promoted the tensile strength of the hybrid composites.

3.1.4   Functions of SiC additives in hybrid HMCF/Al-Si composites

a) Improvement of infiltration performance of CF preform

Carbon fibres without pretreatment cannot be completely impregnated by molten aluminium through pressure casting, and most carbon fibres make contact with the neighbouring fibres as shown in Fig. 1 a. One reason for this is the poor wettability of carbon fibres against aluminium; another is that carbon fibres themselves are subjected to the applied stress during pressing, causing densification of carbon fibre preforms and forming a closely-packed microstructure.
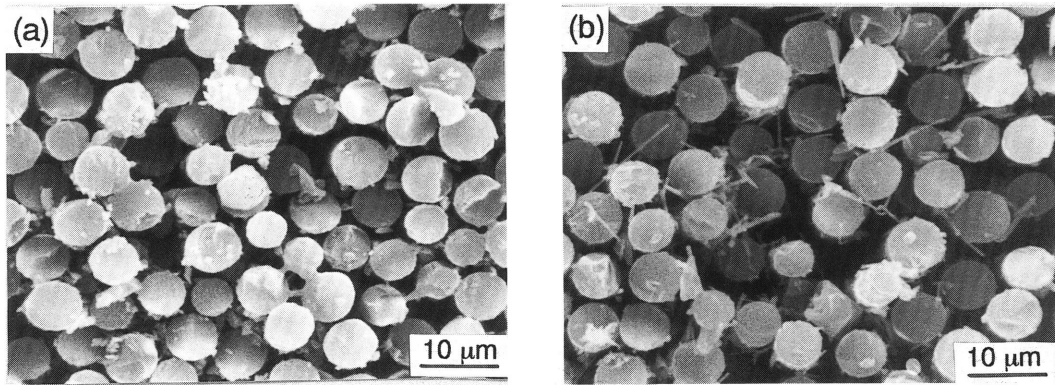
Fig. 4   SEM micrographs of transverse cross sections of (a) SiC particle-distributed (volume ratio of particle to fibre: 7.3%) and (b) SiC whisker-distributed (volume ratio of whisker to fibre: 7.7%) carbon fibre preforms

In the case where SiC additives are uniformly distributed among carbon fibres, although the hybridization cannot lower the wet angle of carbon fibres and aluminium, it can enlarge the spacing distance between carbon fibres to produce many bigger capillaries in the fibre preforms, as illustrated in Fig. 4, which will reduce infiltration pressure. When pressure casting is applied, with their high compressive strength, the distributed additives are capable of enduring the applied pressure and maintaining the formed fibre separation to a certain degree, to allow molten aluminium and the following plastic solidifying aluminium to fill these capillaries up. The greater the amount of SiC particles or whiskers is added, the bigger the capillaries form; as a result, the easier the liquid infiltration and the lower the V_f of hybrid composites. Because SiC whiskers have a longer length and lower packing density, their capacity to separate carbon fibres is greater than that of SiC particles. This is consistent with the varying tendency of V_f given in Fig. 2 a.

b) Tailoring fibre volume fraction

In the direct-casting process (unlike other processes in which one can design the V_f of composites by arranging the volume of precursor wires in advance), fibre preform prepared from multifilament bundles is densified in a closely packed manner under the applied stress during pressure infiltration. The CF/Al composites obtained usually have a V_f of as high as 60 vol% [11]. Therefore it is almost impossible to manufacture composites with a certain V_f as required by practical applications. When considering the variations of applications and reduction of cost, nevertheless, it is essential to be capable of controlling V_f to satisfy property requirements with the minimum fibre consumption.

From Fig. 2 a, it can be seen that the V_f of the hybrid composites had a definite relation to the V_{add}, and the V_f variation was in a range of 59 vol% to 25 vol% with LTS not lower than that of the conventional composite. This indicates that V_f can be easily controlled to such an extent as to meet the requirements of various applications with the minimum consumption of costly carbon fibres. Moreover, in comparison with the conventional composite, the decrease in V_f of the hybrid composites does not imply a sacrifice of composite properties, in particular, when LTS is the main concern.

c) Promotion of strength transfer efficiency of carbon fibres

Figure 5 shows the relation of V_{add} and transfer efficiency of fibre strength of the hybrid composites (η), which is defined as follows

$$\sigma_c = \sigma_f V_f \eta + \sigma_m (1 - V_f)$$

Where σ represents tensile strength; V is the volume fraction; subscripts c, f, and m refer to composite, fibre, and matrix, respectively; and σ_m is the matrix stress at the strain corresponding to the composite ultimate tensile strength. In the calculation, the contribution of SiC additives was ignored because of its small volume fraction and ineffectiveness as a reinforcement compared with carbon fibres. It can be seen that the fibre strength transfer efficiency of the hybrid composites was much higher than that of the conventional one.
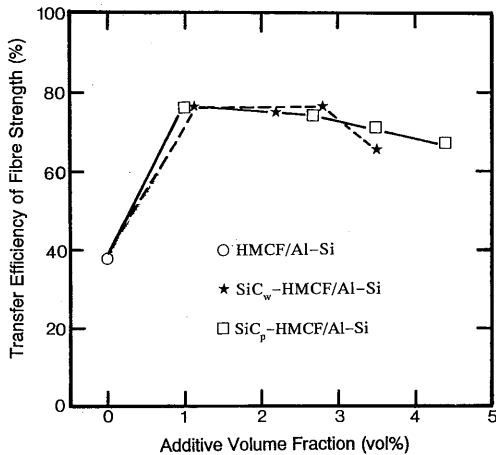
Fig. 5   Effect of additive volume fraction on strength transfer efficiency of carbon fibres for hybrid HMCF/Al-Si composites

In the conventional composite, the interfacial bonding of carbon fibres and aluminium matrix is too weak to transfer efficiently the applied stress to carbon fibres. Moreover, because of the incomplete impregnation, direct fibre contacts and voids exist, namely, not enough aluminium matrix presents around the circumferences of carbon fibres. Therefore, concentrated stress caused by failure at the weak points of carbon fibres cannot be released by the plastic deformation of aluminium matrix. The stress concentration grows very quickly even at low stress levels, and causes interlamilar failure in which the carbon fibres are broken at the weakest points and then stripped off interface. It is considered that they result in a lower strength of the composite and a lower strength transfer efficiency of carbon fibres.

The hybrid composites obtained had much higher strength than the conventional composite, but it is not reasonable to consider that this promotion originates from the direct contribution of the additives as strength reinforcement, because of their very low volume fractions. As mentioned above, additive distribution can effectively prohibit fibre contacts and enhance the complete infiltration of aluminium, which leads to production of good billets. The intermediate length of fibre pull-out, observed on the fracture surfaces of the hybrid composites, evidently indicates that carbon fibres have played a more important role in the composites. Therefore, it can be assumed that the strength increase of the hybrid composites originates from the great improvement of strengthening efficiency of the carbon fibres themselves.

From the above discussion on additive functions, it can be understood that this hybridization method not only can be used in the fabrication of CF/Al composites to improve fibre strengthening efficiency and to tailor fibre volume fraction, but also can be applied to other fibre reinforced composite systems.

3.2   Characterization of Several SiC$_p$-CF/Al Composite systems

3.2.1   Fibre strength degradation of SiC$_p$-CF/Al composites

Using the above hybridization method, HMCF/P-Al, HSCF/Al-Si, HMCF/6061, and HSCF/6061 composites were prepared with the addition of 1.0 Vol% SiC particles. In order to investigate the srtength degradation of carbon fibres, single-fibre tensile test was performed on the carbon fibres after preheating and extracted from the composites. The results obtained are illustrated in Fig. 6, along with the computed degradation rate, which is a ratio of strength decrease of the treated carbon fibres to the srtength of the as-received carbon fibres.

In the case of HMCFs, as shown in Fig. 6a, the fibre strength was marginally reduced after preheating. This decrease was considered to be caused by mechanical damage during preparation of hybrid fibre preforms by hand. After incorporated with Al-Si matrix and 6061 matrix, HMCFs maintained 85% of the strength of the as-received counterpart. However, HMCFs from the pure aluminium and T6-treated 6061 matrix composites were deteriorated. Because interfacial chemical interactions affect the strength of carbon fibres, the strength change of
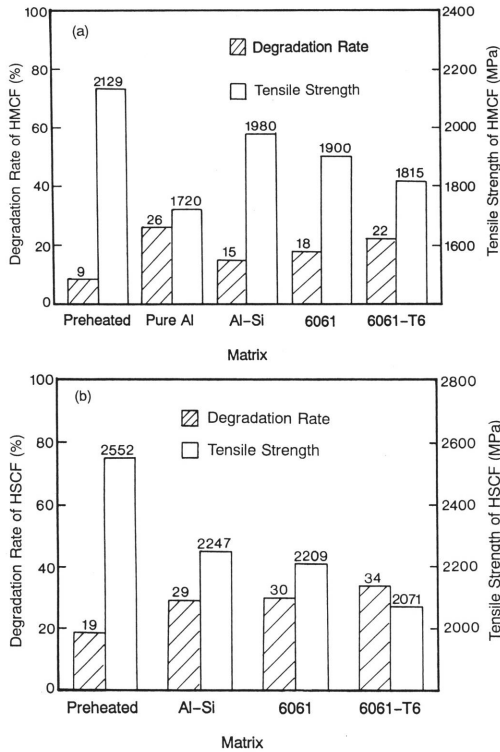
Fig. 6 Tensile strength and degradation rate of (a) high-modulus and (b) high-strength carbon fibres in SiC$_p$-CF/Al composites
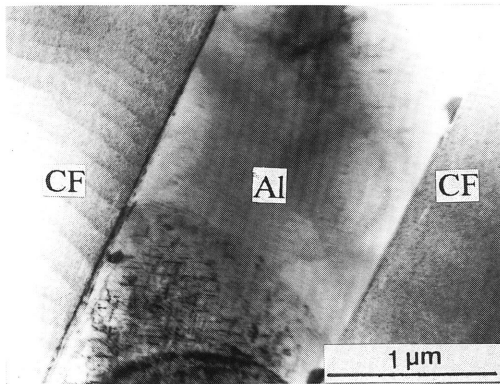

Fig. 7 TEM micrograph of interface of SiC$_p$-HMCF/6061-T6 composite

carbon fibres in those composites implies that aluminium matrix has important influences on the interfacial characteristics. There appears a sequence of matrix activity, from the highest, pure aluminium, to 6061 alloy, and to the lowest, Al-Si alloy. The strength of HMCFs in the T6-treated 6061 matrix composite was reduced, which supposed that interfacial reactions might occur during the solution treatment at 793 K. However, on the fibre/matrix interface, interface zone or foreign substance cannot be observed, as shown in Fig. 7 . This fact indicates that the chemical interactions between HMCFs and 6061 matrix were not greatly accelerated by T6 treatment, although the fibre strength was decreased.

On the other hand, it can be seen from Fig. 6 b that, in the case of HSCFs, the fibre strength was decreased by preheating treatment. Obviously, this decrease cannot be explained only by mechanical damage during processing. However, other reasons for the unexpectedly rapid decrease were not definitely identified. The most probable factor considered may be the oxidation of HSCFs during preheating in air. It is known that HSCFs begin to oxidize at much lower temperatures than HMCFs. Thus, oxidation of HSCFs might take place during preheating, and consequently caused the degradation of fibre strength. Because of the high reactivity of HSCFs with aluminium, the fibre strength degradation after casting and after the T6 treatment was very distinguished. Moreover, the degradation rate seemed not depend upon the type of aluminium matrix.

The result of the light degradation of HMCFs and the heavy degradation of HSCFs in Al-Si and 6061 alloy matrix composites is consistent with those reported by other researchers [5,7,12]. This difference in degradation behaviour by the fibre type can be generally attributed to the difference in crystal structure and surface properties, which determine the reactivity, of carbon fibres. Figure 8 gives etched surfaces of SiC$_p$-HMCF/6061-T6 and SiC$_p$-HSCF/6061-T6 composites. The difference in their interfacial morphology between the two types of carbon fibres is clearly shown. Obviously, the interfacial interactions between HSCFs and aluminium were much more active. As a result, the fibre strength was greatly deteriorated.

3.2.2 Tensile strength of SiC$_p$-CF/Al composites

Table 2 gives the properties of all the SiC$_p$-CF/Al composites obtained. The η_{app} and η_{real} in the table stand for the
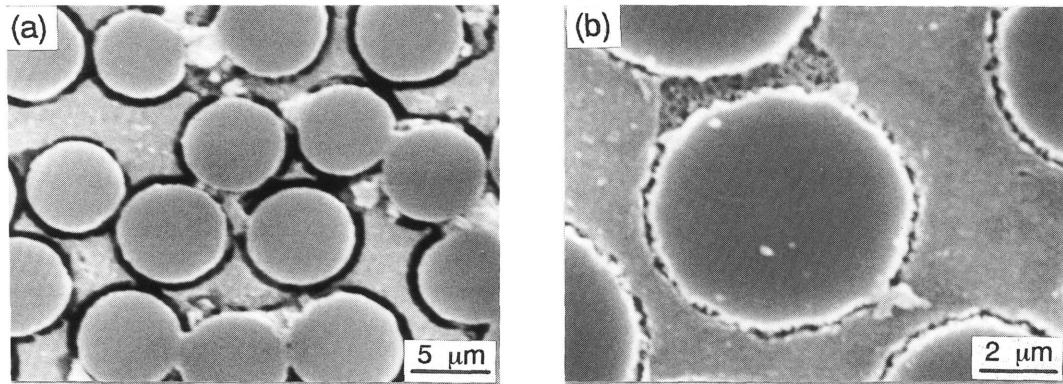
Fig. 8   SEM micrographs of (a) SiC_p-HMCF/6061-T6 and (b) SiC_p-HSCF/6061-T6 composites

apparent and real transfer efficiency of fibre strength, respectively, as defined above. For the calculation of η_{app}, the strength of the as-received carbon fibres should be used for the σ_f. For the calculation of η_{real}, the strength of those fibres extracted from the composites should be used for the σ_f. All the composites had a similar V_f of about 41 vol%. The LTS of these composites, however, was considerably different, from the greatest value of 1020 MPa to the lowest of 407 MPa. In particular, the LTSs of SiC_p-HMCF/Al composites were much greater than those of SiC_p-HSCF/Al composites, although the tensile strength of the HSCFs was much higher than that of the HMCFs. On the basis of transfer efficiency of fibre strength, the difference between SiC_p-HMCF/Al and SiC_p-HSCF/Al composites was even more outstanding. These results suggest that the HMCFs have effectively strengthened the aluminium matrixes, but the HSCFs have hardly transferred their excellent strength to the composites. In contrary to the LTSs, the transverse tensile strengths (TTSs) of SiC_p-HSCF/Al composites, above 60 MPa, were much greater than those of SiC_p-HMCF/Al composites, about 16 MPa.

It is worth noting that the η_{app} of SiC_p-HMCF/6061-T6 composite was higher than that of SiC_p-HMCF/6061-AC composite. Contrarily, the η_{app} of SiC_p-HSCF/6061-T6 composite was lower than that of SiC_p-HSCF/6061-AC composite. In HMCF/Al system, the interactions on fibre/matrix interface could hardly happen during fabrication. This indicates that almost no chemical bonding exists in the interface. Because the surfaces of the fibres are quite smooth, mechanical bonding is also weak. The weak TTS, 16 MPa given in Table 2 for the as-cast SiC_p-HMCF/Al composites, is consistent with this implication, because the transverse behaviour of a composite is closely and directly

Table 2   The properties of SiC_p-CF/Al composites obtained

Composite	V_f(Vol%)	LTS(MPa)	TTS(MPa)	η_{app}	η_{real}
SiC_p-HMCF/P-Al	40	575	– – –	0.60	0.81
SiC_p-HMCF/Al-Si	40	807	16	0.76	0.90
SiC_p-HMCF/6061-AC	40	811	16	0.77	0.94
SiC_p-HMCF/6061-T6	41	1020	20	0.89	1.14
SiC_p-HSCF/Al-Si	41	412	61	0.24	0.34
SiC_p-HSCF/6061-AC	42	407	63	0.24	0.34
SiC_p-HSCF/6061-T6	41	425	83	0.19	1.30

*LTS: longitudinal tensile strength η_{app}: apparent transfer efficiency of fibre strength

TTS: transverse tensile strength η_{real}: real transfer efficiency of fibre strength

related to the interfacial bonding properties of the composite[13]. Thus, the interface bonding of HMCF/Al composites is very weak, which results in a lower fibre strength transfer. However, as chemical interactions might occur between HMCFs and aluminium matrix above 773 K, somewhat chemical bonding in SiC_p-HMCF/6061-T6 composite might be formed during the solution treatment at 793 K for an hour. The TTS increase of SiC_p-HMCF/6061-T6 composite compared with the as-cast one agrees with this suggestion. Therefore, it is important to be able to control the degree of bonding between aluminium matrix and HMCFs so that applied load can be effectively transferred from the matrix to the fibres through fibre/matrix interface. In the case of HSCF/Al system, because heavier interfacial interactions between the fibres and aluminium matrix occurred during casting, the TTSs of SiC_p-HSCF/Al composites were much higher than those of the HMCF/Al composites. This implies that there was a very strong interface in this system, which caused poor LTS of the composites[11].

3.3 Behaviour of SiC_p-HMCF/Al-Si Composites after Thermal Exposure

3.3.1 Tensile strength of SiC_p-HMCF/Al-Si composites

It is known that the LTS of CF/Al composites is reduced by thermal exposure at high temperatures even for HMCF/Al composites[14]. However, it was found that, as shown in Fig. 9 , which is a plot of the LTS and TTS of SiC_p-HMCF/Al-Si composites as a function of thermal exposure time at 773 K, the LTS gradually increased with increasing exposure time up to 216 hours. At this point, the strength reached its maximum value of 899 MPa. Extending the thermal exposure time caused the LTS of the composites to decrease. However, the minimum strength of the composite was 652 MPa, even after exposure for 360 hours. These results suggest that SiC_p-HMCF/Al-Si composites can have a relatively long lifetime when they are to be used below 773 K.
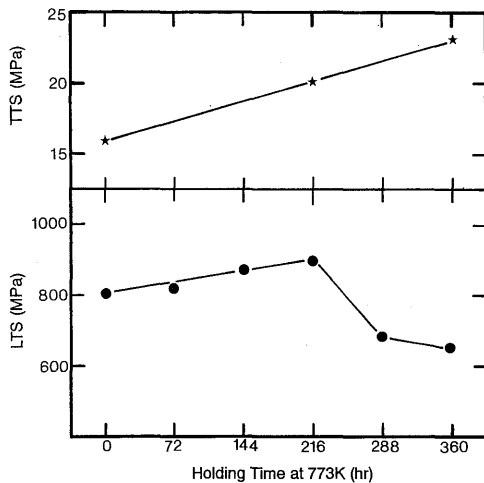
Among the important behaviour of a composite resulting from thermal exposure, the first to be considered is the changes that occur at the matrix/fibre interface of the composite. This is because the variation of interfacial bonding will directly influence its mechanical properties. The direct measurement of interfacial bonding strength of MMCs is extremely difficult. However, some studies have indicated that the transverse behaviour can be colsely and directly related to the interfacial properties[13, 15]. Therefore, the TTS of SiC_p-HMCF/Al-Si composites was measured to represent indirectly the interfacial bonding srtength. In strong conrtast to the variation of the LTS, the TTS of the composites increased monotonically with thermal exposure time, i.e. the interfacial bonding strength of the composites was improved as a result of the thermal exposure. The changes of the LTS and TTS of the composites imply that the greatest LTS of SiC_p-HMCF/Al-Si composites can be achieved when the composites have moderate interfacial strength, which is consistent with other observations on CF/Al composite wires[16].

Fig. 9   Effect of holding time of thermal exposure on
         longitudinal and transverse tensile strengths
         of SiC_p-HMCF/Al-Si composites

It is worthy to note that the LTS of the composites sharply decreased when the exposure time was over 216 hours. Two reasons for this behaviour can be suggested. One is that the strength of carbon fibres might be reduced by excessive interfacial interactions during the long-term thermal exposure. The other is that the mechanism of crack propagation and the fracture mode of the composites might be changed by srtonger interfacial bonding after the thermal exposure. These points are discussed in detail below.
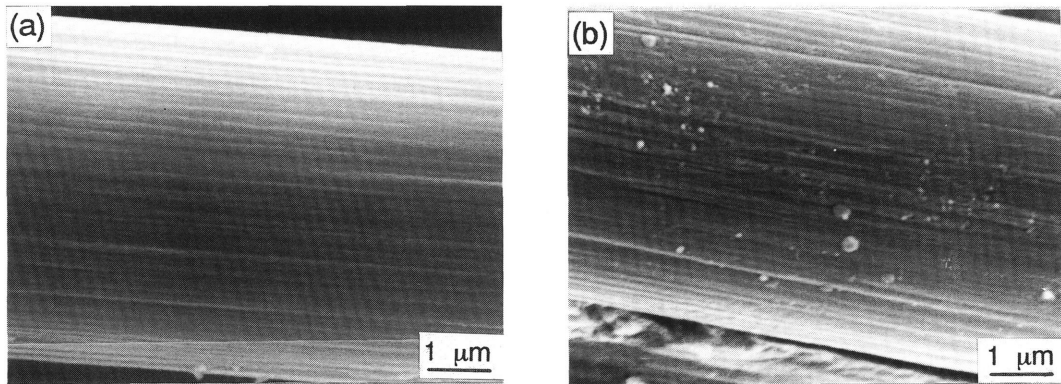
3.3.2   Fibre strength of SiC$_p$-HMCF/Al-Si composites

Fig.10   SEM photographs of carbon fibres in SiC$_p$-HMCF/Al-Si composites (a) unexposed and (b) exposed for 360 hours at 773 K
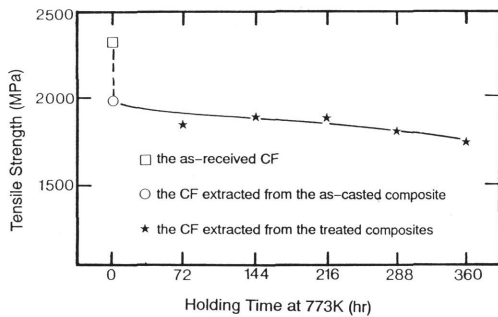

Fig.11   Effect of holding time of thermal exposure on tensile strength of carbon fibres extracted from SiC$_p$-HMCF/Al-Si composites

Figure 10 shows carbon fibre surfaces taken from the as-cast and thermally exposed composites. The fibre surface of the as-cast composite was very smooth and was free from foreign particles, like an as-received carbon fibre. However, it can be seen that, on the fibre surface of the composite thermally exposed at 773 K for 360 hours, there are many foreign particles and erosion pits. It is considered that the fine particles were formed from, and the pits were caused by, the fibre/matirx interactions during the exposure.

A plot of the tensile strength of the fibres extracted from the thermally exposed composites as a function of thermal exposure time at 773 K is shown in Fig. 11. The fibre strength gradually decreased with increasing thermal exposure time, which is not consistent with the steep decrease of the LTS of the composites for exposure time exceeding 216 hours. These results indicate that the LTS decrease of the composites with long-term thermal exposure at 773 K can not be totally attributed to the fibre degradation caused by thermal exposure.

3.3.3   Fracture morphology of SiC$_p$-HMCF/Al-Si composites

As mentioned above, the TTS results suggest that the interface of the composites became progressively stronger with increasing thermal exposure time. This was confirmed by tensile fracture morphologies of the composites shown in Fig.12. It can be found that the behaviour of fibre pull-out greatly changes with the thermal exposure time. In the composite treated for 216 hours moderate fibre pull-out occurs, which corresponds to the higher LTS of the composite. However, the composites untreated and treated for 72 hours reveal considerable fibre pull-out and that treated for 360 hours has little fibre pull-out, both cases referring to lower values of LTS. This observation indicates that fibre pull-out behaviour has an important effect on the longitudinal mechanical properties of the composites.

Generally, fibre pull-out in a composite is associated with the interfacial bonding of the composite[13]. If the interface is strong, propagation of cracks initiated in the composite under an applied stress cannot be blunted by debon-
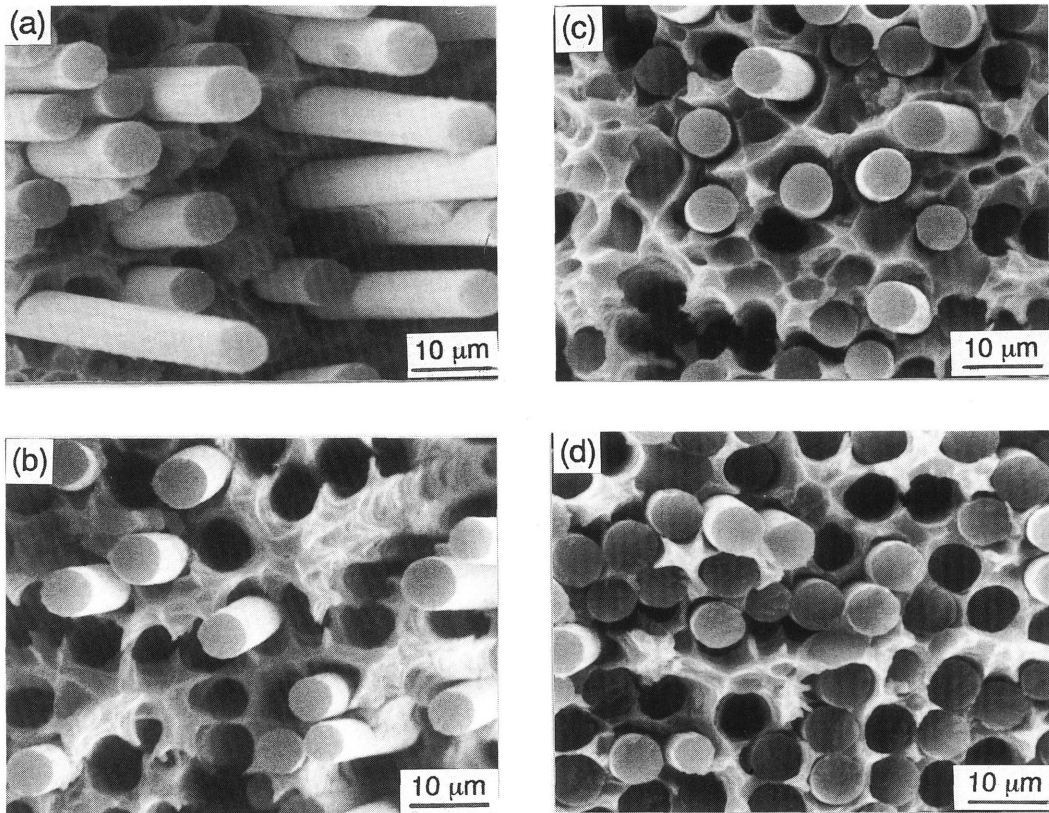
Fig.12　SEM photographs of longitudinal tensile fracture surfaces of SiC_p-HMCF/Al-Si composites (a) unexposed, (b) exposed for 72 hours at 773 K, (c) exposed for 216 hours at 773 K, and (d) exposed for 360 hours at 773 K

ding of the interface. Therefore, concentration of stress occurs rapidly and this leads to progressive fibre overload which eventually results in catastrophic failure with a planar fracture surface without fibre pull-out and a somewhat weaker strength of the composite. Where the interfacial bonding is suitably weak, if the stress exerted on the interface exceeds the interfacial bonding strength, debonding occurs and the cracks are blunted. As a result, the fibre can be protected from notch damage by the cracks. On the other hand, the interfacial strength should be sufficient to transfer efficiently the applied stress from the matrix to the stress-bearing fibres so that the fibre srtength can be effectively used, thereby strengthening the composite. The fracture mode of this type of composite is cumulative and the fracture surface appears to have a moderate fibre pull-out. Too weak an interface, however, cannot suffi-ciently transfer the applied stress to the strong fibres, and consequently the strength of the composite is relatively low. Although Fig.12 is not a typical illustration for the three interfacial states, they can satisfactorily explain, from the differences in length of the fibre pull-out, the effects of interfacial bonding on the composites.

From the above results and discussion, it can be drawn the conclusion that some controlled amount of interac-tions at matrix/reinforcement interface may even be desirable for achieving good transverse properties and obtain-ing a bonding between the reinforcement and the matrix strong enough to translate effectively the applied stress from the matrix to the reinforcement, and that a strong interfacial bonding, however, will adversely degrade com-posite longitudinal properties.

4　Conclusions

1) The hybridization with SiC additive could improve the infiltration performance of fibre preforms, control the

fibre volume fraction, and increase the fibre strength transfer efficiency of hybrid composites. The longitudinal tensile strength of the hybrid HMCF/Al composites was greatly improved, although their fibre volume fraction was very low compared to that of the conventional composite.

2) The strength degradation of high-modulus carbon fibres during fabrication was not heavy and depended upon the type of aluminium matrix. The strength of high-strength carbon fibres was greatly reduced after incorporated with aluminium matrix, but the degradation did not depend on the variety of aluminum matrix. As a result, the SiC$_p$-HMCF/Al composites having a longitudinal tensile strength of more than 800 MPa were successfully achieved, while SiC$_p$-HSCF/Al composites had only a longitudinal tensile strength of about 400 MPa. Contrariwise, the transverse tensile strength of SiC$_p$-HSCF/Al composites was much greater than that of SiC$_p$-HMCF/Al composites.

3) The longitudinal tensile strength of SiC$_p$-HMCF/Al-Si composites at first increased, then decreased, while the transverse tensile strength of the composites monotonically increased, as the thermal exposure at 773 K was prolonged. The thermal exposure slightly degraded the fibre strength and caused chemical interactions on the interface of the composites.

4) Thermal exposure at 773 K resulted in changes of fracture morphology of SiC$_p$-HMCF/Al-Si composites. It was found that an intermediate fibre pull-out and a planar fracture pattern corresponded to a good and poor longitudinal tensile strength of the composites, respectively.

5) The results and discussion suggest that fibre/matrix interfacial bonding has an important effect on the mechanical properties of CF/Al composites and intermediate interfacial bonding can be expected to result in good longitudinal and transverse tensile strengths of CF/Al composites.

References

1 . Towata, S., Yamada, S. and Ohwaki, t.; Trans. Jpn. Inst. Metals, 26 (1986), 563.

2 . Sawada, Y. and Bader, M. G.; in Proceedings of the 5th International Conference on Composite Materials, San Diego, 1985, edited by W. C. Harrigan, Jr. et al. (1985), 785, The Metall. Soc., Inc..

3 . Towata, S. and Yamada, S.; Trans. Jpn. Inst. Metals, 27 (1986), 709.

4 . Towata, S., Ikuno, H. and Yamada, S.; Trans. Jpn. Inst. Metals, 29 (1988). 314.

5 . Khan, I. H.; Met. Trans., 7 A(1976), 1281.

6 . Amateau, M. F.; J. Comp. Mater., 10 (1976), 279.

7 . Kohara, S. and Muto, N.; J. Jpn. Inst. Metals, 52 (1988), 1063.

8 . Pepper, R. T., Upp, J. W., Rossi, R. C. and Kendall, E. G,; Metall. Trans., 2 (1971), 117.

9 . Chen, R., Zhang, G. D. and Wu, R. J.; in Proceedings of the 8th International Conference on Composite Materials, Honolulu, edited by S. W. Tsai and G. S. Springer(1991), 19F, SAMPE.

10. Wilkinson, A. J.; Scripta Metall. et Mater., 26 (1992), 387.

11. Cheng, H. M., Kitahara, A., Akiyama, S., Kobayashi, K. and Zhou, B. L.; Scripta Metall. et Mater., 26 (1992), 1475.

12. Goddard, D. M.; J. Mater. Sci., 13 (1978), 1841.

13. Fishman, S. G.; in Proceedings of the 8th International Conference on Composite Materials, Honolulu, edited by S. W. Tsai and G. S. Springer (1991), 19A, SAMPE.

14. Asanuma, H. and Okura, A.; J. Jpn. Inst. Metals, 48 (1984), 1198.

15. Steckel, G. L., Fowers, R. H. and Amateau, M. F.; "Transverse strength properties of graphite-aluminum composites", Report TOR 0078 (3726−03)− 4 , Naval Surface Weapons Center,1978.

16. Shorshorov, M. Kh., Chernyshova, T.A. and Kobeleva, L. T.; in Proceedings of the 4th International Conference on Composite Materials, Tokyo, 1982, p. 1273, Japan Society for Comp. Mater..