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Interfacial reaction and strength of SiC fibres coated with aluminium alloys

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Silicon carbide fibres (Nicalon) were coated with pure aluminium and aluminium alloys containing silicon. The coated fibres were annealed to produce an interfacial reaction zone between the coated layer and the fibre. The effect of this reaction zone on the tensile strength of the fibre was investigated. During the early stages of growth the reaction zone of the fibre is thin, and the strength of the fibre is controlled by inherent defects so that the fibre retains its original strength. After the early stages, notches are formed in the reaction zone of the fibre on loading at a small strain and the fibre fractures when a notch extends into the fibre. In this stage the fibre strength is dependent on the thickness of the reaction zone. An alloying addition of 1 or 5 at% Si to the aluminium matrix was found to be effective in retarding the growth rate of the reaction zone.

1. Introduction

In fibre-reinforced metal composites, an interfacial reaction between the fibre and the matrix often arises. Such a reaction, on the one hand, is necessary to introduce a strong interfacial bonding so that the interfacial shear strength can be increased. On the other hand, the reaction zone formed on the fibre is, in general, brittle and easily fractures even at a rather small strain.

The effect of the interfacial reaction on the tensile strength of the fibre has been explained by the theory of the interfacial reaction zone [1–5]. Experimental work supporting this theory has been mainly made on such composite systems as B/Ti [1], B/Al [1], W/Ni [6], C/Ni [7], and SiC/Ti [8].

In recent years, increasing interest and attention have been paid to the composite systems reinforced by SiC fibre because of their application potential, especially the SiC/Al system which is the most promising candidate for future aviation and spacecraft materials. The SiC/Al composites are generally fabricated by a liquid infiltration or hot pressing techniques. In both cases, thermodynamic and kinetic information about the interfacial reaction is important, because during the fabrication processes, interfacial bonding could be developed by the reaction which would damage the fibre, and therefore affects the mechanical properties of the composites.

Interfacial reaction between SiC fibres and aluminium has been investigated by Towata and Yamada [9]. Observation by differential thermal analysis showed that at the interface, aluminium carbide and elemental silicon are formed by an exothermic reaction. The effect of the addition of silicon on the control of the interfacial reaction in the SiC/Al system has been reported by Kohara [10]. It has been observed that no loss in strength occurs in the fibre coated with Al–1 at % Si alloy after heating at 773 K for 25 h, but a significant loss in strength takes place for the fibre coated with pure aluminium after heating under the same conditions. Further investigation to give a quantitative explanation of the effect of interfacial reaction on the fibre strength is considered to be necessary and possible using the theory of the interfacial reaction zone, because the degree of deterioration of fibre strength must be related to the amount of interfacial reaction.

In this work, the reaction zone formed at the interface between the coated layer of aluminium alloys and the SiC fibres is produced by annealing coated fibres under several combined conditions of temperature and time. The effects of the interfacial reaction on the fibre strength are investigated. The thermodynamics of this system and the kinetics of the reaction zone growth are studied in order to control the interaction by the addition of a small amount of alloying silicon.

2. Experimental procedure

The SiC fibre used in the present experiment was commercial continuous fibres (Nicalon) supplied by Nippon Carbon Co. Nominal properties of the SiC fibre are listed in Table I.

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TABLE I Nominal properties of the as-received Nicalon SiC fibre

Diameter (μm)	15.0
Density (g cm^{-3})	2.55
Tensile strength (GPa)	3.2
Elastic modulus (GPa)	206.0
Content of oxygen (mass %)	11.2

Alloy stocks of Al–1 at % Si and Al–5 at % Si were prepared by arc melting under an argon atmosphere using aluminium and silicon of over 99.99% purity. The SiC fibres were coated with pure aluminium and Al–Si alloys using an electron-beam evaporation technique. The deposition process has been described in detail in our previous work [11–13]. Using this technique a uniformly coated layer of $\sim 0.2\,\mu$m thick could be achieved. Because the deposition temperature is relatively low, no interfacial reaction occurs during the deposition process and therefore deterioration in fibre strength cannot be anticipated.

The coated fibres were annealed at different temperatures and for various periods of time to produce the interfacial reaction zone to various degrees. Before heat treatment the fibres were encapsulated in quartz tubes under a vacuum of 10^{-5} torr to avoid the oxidation of the fibres during annealing. The tensile strength of the treated fibres was measured using a mini-Instron-type testing machine. The gauge length was taken as 20 mm. The uncoated fibre was also annealed in the same way and its tensile strength was measured to serve as a reference. An average strength of fibres was estimated based on 40 measurements.

3. Results

Fig. 1 shows the effect of annealing temperature on the tensile strength of the uncoated and coated fibres at a constant annealing time of 1.8 ksec. To make the measured tensile strength of the coated fibre comparable with that of the uncoated one, the coated layer of the fibre is removed after the heat treatment with 0.2 N NaOH aqueous solution prior to the tensile test. It is ensured that this chemical treatment has little influence on the fibre strength by comparing the measured tensile strength of the aluminium-coated fibre after the chemical treatment with that of the as-received fibre. From Fig. 1, a sharp decrease in tensile strength of the aluminium-coated fibre is observed around the melting point of aluminium, while the uncoated fibre has really its original strength. Obviously, this is due to the interfacial reaction between the layer of molten aluminium and the SiC fibre.

Fig. 2 shows the changes in tensile strength of the coated fibre with annealing time at a fixed temperature of 833 K, which is lower than the melting point of aluminium. It is first seen that the fibres coated with Al–Si alloys show less strength loss with increasing annealing time than the fibre coated with pure aluminium. Secondly, no substantial difference is found between the fibres coated with the Al–Si alloys containing different amounts of silicon. It should be noted that each curve in Fig. 2 can be divided into two stages. In the early stage the strength remains nearly constant and in the later stage decreases continuously.

The changes in tensile strength of the coated fibres with annealing temperature are shown in Fig. 3. In this case the annealing time is fixed at 16 h. The normalized strength, σ_h/σ^0 is used in Fig. 3, where σ_h is the strength of the coated fibre after the heat treatment and σ_0 that of the as-coated fibre. The same order of loss in tensile strength as that observed in Fig. 2 among the fibres coated with pure aluminium and the Al–Si alloys is found from the linear relationship of ln (σ_h/σ^0) against $1/T$ in this figure.

The heat treatment of the coated fibres results in the formation of the reaction zone. This can be easily recognized by comparison of the fractured cross-sections of the aluminium-coated fibre before annealing, as shown in Fig. 4a, and those of the aluminium-coated fibre annealed at 823 K for 16 h, as shown in Fig. 4b. In the latter case the existence of a reaction zone between the surface layer and the fibre is clearly visible.

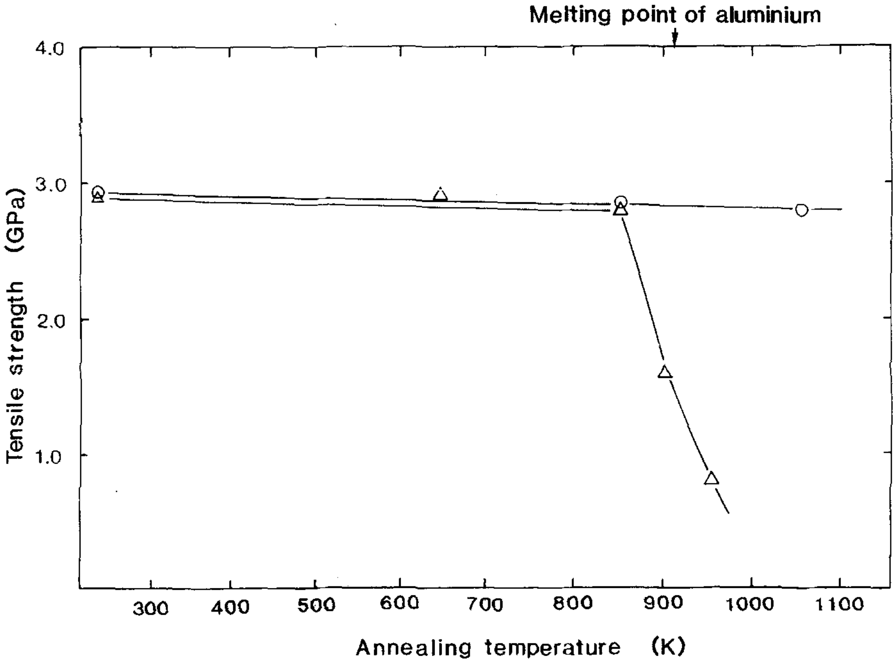

Figure 1 Annealing temperature dependence of tensile strength of SiC fibres, (○) uncoated and (△) coated with pure aluminium. Annealing time = 30 min.

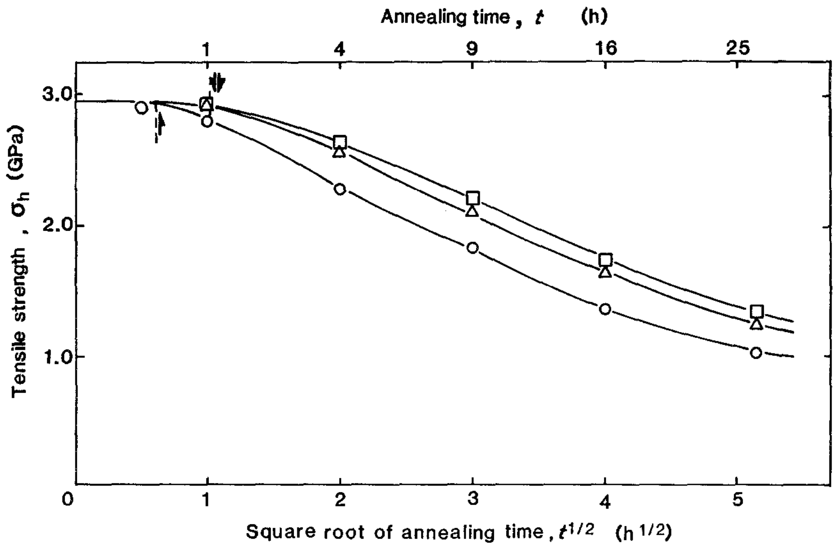

Figure 2 Relation between annealing time and the tensile strength of the SiC fibres coated with (○) pure aluminium, (△) Al-1 at % Si alloy and (□) Al-5 at % Si alloy. The arrows separate the growth stages. Annealing temperature = 823 K.

The fracture behaviour of the coated fibre is very much influenced by the growth of the interfacial reaction zone. The fracture modes of the fibres are shown in Figs 5a to h. The as-coated fibre is shown in Fig. 5a with a long pull-out length. Figs 5b to d show the fibres coated with pure aluminium, Al-1 at % Si and Al-5 at % Si alloys after heat treatment at 823 K for 1 h. The fibres with the same coatings as those in Figs 5b to d but heat-treated at 823 K for 16 h are shown in Figs 5e to g. It is seen from Figs 5b to g that the fibre coated with pure aluminium shows a shorter pull-out length than the fibres coated with Al-Si alloys. In Fig. 5h, the fibre coated with pure aluminium heat-treated at 823 K for 25 h, shows a flat fracture surface.

4. Discussion

4.1. Interfacial reaction and wettability

In order to discuss the interface reaction between aluminium and the SiC fibre, a knowledge of the Al-Si-C ternary phase diagram in the temperature range investigated is necessary. Few experimental data on the phase diagram of this system are available in spite of its significance. Possible constitutions of the Al-Si-C phase diagram have been suggested based on the known binary systems [14]. From the binary systems concerned it is known that in the Al-C system the solid solubility of carbon in aluminium is rather small, being estimated as 0.03 at %. Even in molten aluminium at 1073 K the solubility of carbon is no more than 0.22 at % [15]. However there is a great tendency to form aluminium carbide, Al_4C_3, above 773 K. In the Al-Si system, the maximum solid solubility of silicon in aluminium is ~1.59 at % at the eutectic temperature of 850 K [16]. Between aluminium and silicon no compound can be formed.

From the above information on the binary systems and some published work on the interfacial reaction in the SiC/Al composites [17-19], it is predicted that below the eutectic temperature in the Al-Si system, besides the slight dissolution of silicon into aluminium, the reaction

$$3SiC + 4Al \rightarrow Al_4C_3 + 3Si \qquad (1)$$

is expected to occur at the SiC/Al interface. The

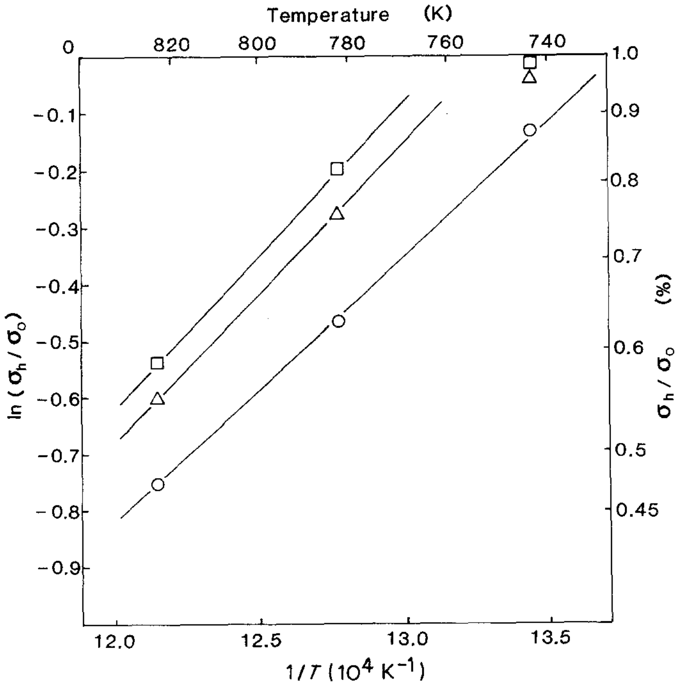

Figure 3 Annealing temperature dependence of tensile strength of the SiC fibres coated with (○) pure aluminium, (△) Al-1 at % Si alloy and (□) Al-5 at % alloy. Annealing time = 16 h.

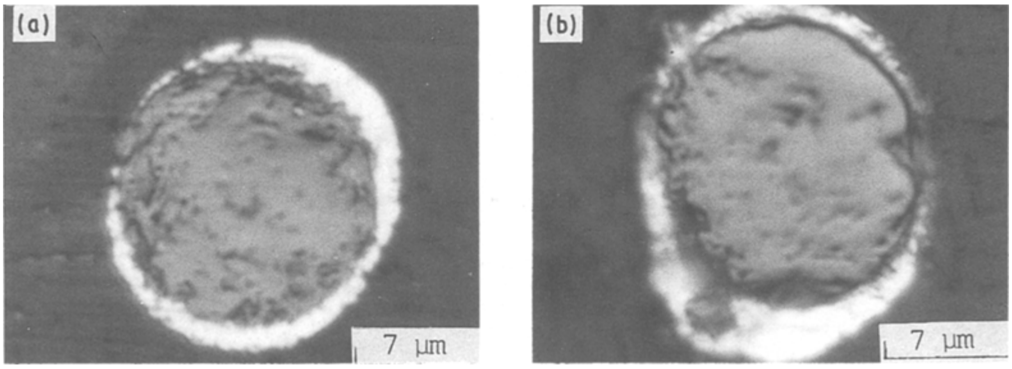

Figure 4 Fracture surface of the fibres having reaction zone between the aluminium-coated layer and SiC fibre: (a) unannealed; (b) annealed at 823 K for 16 h.

reaction is thermodynamically possible, because from the free energy of formation for Al_4C_3 (-215.8 kJ mol^{-1}) and that for SiC (-67.0 kJ mol^{-1}) [19] the standard free energy change for this reaction is estimated to be -14.8 kJ.

It is easy to understand that both the dissolution of silicon in aluminium and the reaction in Equation 1 can be suppressed by saturating aluminium with silicon [17, 20]. Figs 2 and 3 show the effects of silicon addition to aluminium where a lesser degree of strength degradation during annealing occurs for the fibres coated with Al–Si alloys. It will be seen later that the effect is actually attributed to the existence of a thinner reaction zone when the fibres are coated with Al–Si alloys. The effectiveness of silicon addition on the control of the interfacial reaction has also been confirmed by the long-term heat treatment of aluminium-coated SiC fibres by Kohara [10]. Significant strength loss has been observed for the fibre coated with pure aluminium after 25 h at 773 K, but no loss in strength has been observed for the fibre coated with an Al–1 at % Si alloy with the same treatment.

The effect of silicon on the wettability of aluminium on the SiC fibre has to be taken into account, because the interface wettability is an important factor for the development of the interfacial bonding in the fabrication process of composites. It has been generally accepted that the interface wettability between a matrix and a fibre could be improved by the formation of stable compound at the interface. This argument could be supported by observations of the different wetting behaviour between the systems of SiC/Al [20] and Al_2O_3/Al [13].

For the SiC/Al system it has been experimentally shown by Laurent *et al.* [20] that an aluminium drop spreads and wets SiC substrate at temperatures above 1015 K after a duration of 15 min. Nevertheless, in an observation of the wetting behaviour of aluminium on the Al_2O_3 fibre, it has been found that the coated layer of molten aluminium is able to cover the fibre surface smoothly, but it peels away from the fibre surface when the coated fibre is water quenched [13]. An explanation for the difference in wettability in the above examples could be given in terms of the interfacial reaction. In the SiC/Al system a stable compound, Al_2O_3, is formed at the interface [17–19], whereas in the Al_2O_3/Al system a stable compound is not formed at the interface and the reaction product between aluminium and alumina, Al_2O, is volatile [13].

Although the addition of silicon can suppress the formation of Al_4C_3, the experiment done by Laurent *et al.* [20] has shown that the wettability of aluminium on SiC substrate cannot be significantly affected by a small addition of silicon. From the view point that a stable compound formation at the interface is needed to achieve a good wettability, experiments on the Al_2O_3/Al system are currently underway, by alloying some elements such as lithium, magnesium and beryllium, which are expected to form stable compounds with Al_2O_3.

4.2. Effect of interfacial reaction on fibre strength

In most cases, such as in the fibre-reinforced metal composites of B/Ti [1], C/Ni [7], and SiC/Ti [8], the interfacial reactions are mainly controlled by diffusion through the reaction zone. For the reaction zone grown by the interfacial diffusion process, the relationship between the thickness of the reaction zone, δ, and the annealing time, t, for the reaction zone to grow is generally expressed by

$$\delta^2 = Kt \tag{2}$$

where K is the reaction zone growth rate constant [8, 21, 22]. In previous studies on the interfacial reaction in such composite systems as B/Ti [1] and SiC/Ti [8], the strength of the coated fibres was correlated with calculated thickness of the reaction zone, rather than observed ones. As is known, the strength of the fibre is usually estimated based on at least dozens of statistically distributed strength values obtained from the same number of tensile tests. But it will be rather difficult to measure the thickness of the reaction zone on each fibre tested. In the following treatment an alternative method to give an expression of tensile strength of the coated fibre is suggested, not as a function of reaction zone thickness but as that of annealing temperature and time.

According to the rule of mixtures, the tensile strength of a coated and annealed fibre, σ_h, can be described as

$$\sigma_h = \sigma_f V_f + \sigma_m V_m \tag{3}$$

where σ_f is the strength of the fibre including the

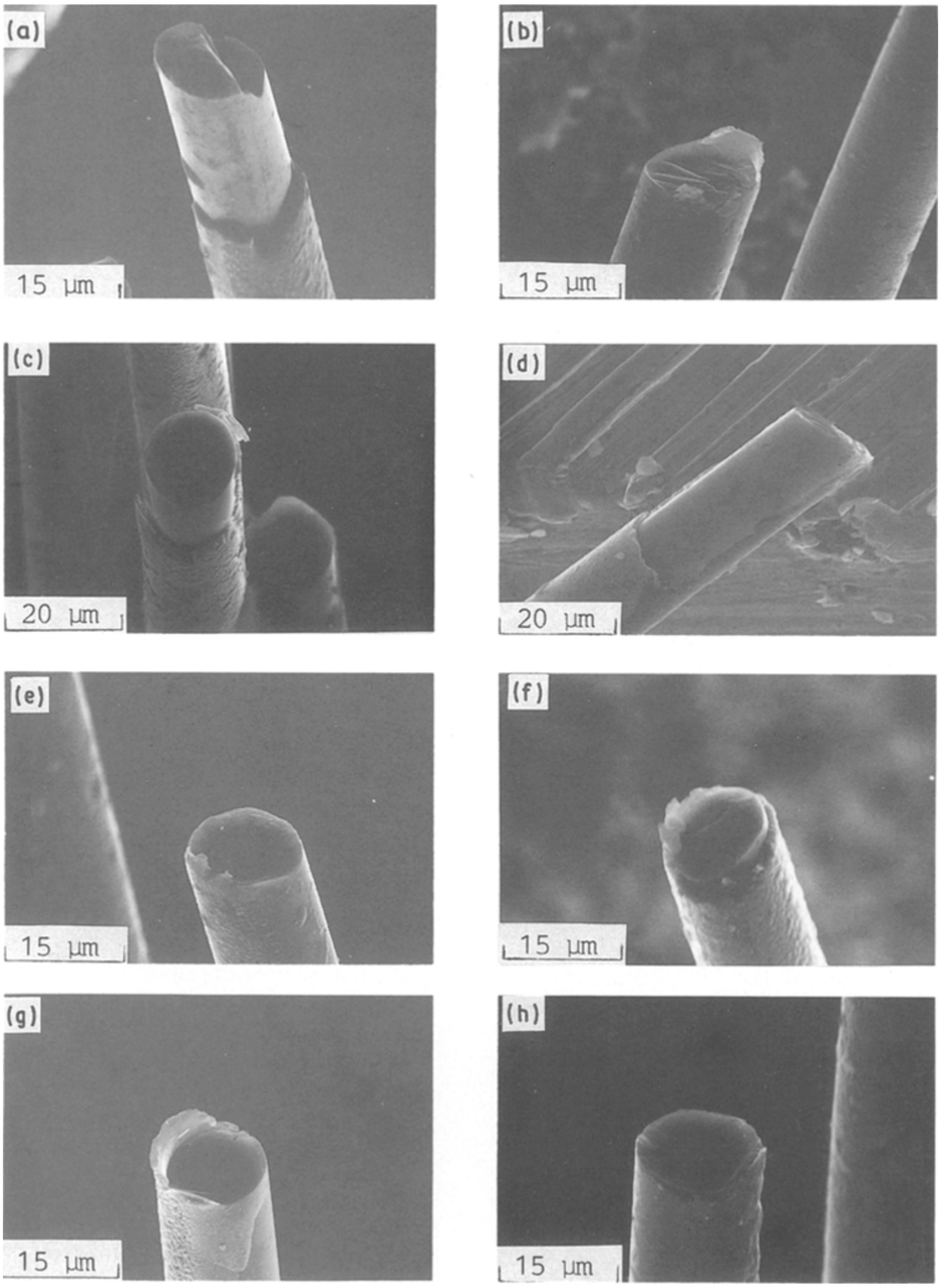

Figure 5 Fracture model of the fibres: (a) as-coated with pure aluminium; (b) coated with pure aluminium and annealed for 1 h; (c) coated with Al–1 at % Si alloy and annealed for 1 h; (d) coated with Al–5 at % Si and annealed for 1 h; (e) coated with pure aluminium and annealed for 16 h; (f) coated with Al–1 at % Si and annealed for 16 h; (g) coated with Al–5 at % Si and annealed for 16 h; (h) coated with pure aluminium and annealed for 25 h. The annealing temperature for the fibres was 823 K.

reaction zone, σ_m the stress of the matrix at the fracture strain of the fibre, V_f and V_m are the volume fractions of the fibre with the reaction zone and the matrix which remains unreacted. For the SiC fibre with a coated layer of $\sim 0.2\,\mu$m thick, the difference between the values of σ_h and σ_f is small, being estimated to be less than 4%. In Figs 2 and 3 the values of σ_h are used, because they are measurable in the tensile test, but for the convenience of theoretical treatment, in the following discussion on the results shown in Figs 2 and 3, σ_f will be used to give the changing tendency of σ_h as is usually used in the composite interface theories [1, 6, 7]. It should be assumed that the interfacial shear strength, τ_i, is strong enough to resist interfacial debonding before the fibre is broken. This will be discussed in detail later.

The value of σ_f in Equation 3, as a function of δ, has been theoretically estimated by Ochiai and Murakami [23]. According to this interface theory [1, 23] which deduces the dependence of σ_f on δ, two situations should be considered.

(A) In the early growth stage of the reaction zone, the defects caused by the interfacial reaction are less important than the inherent defects of the fibre. In this case the fibre strength is controlled by its inherent defects, and therefore the fibre remains in its original strength, i.e.

$$\sigma_f = \sigma^* \tag{4}$$

where σ^* is the tensile strength of the as-received fibre. This is what happens in this stage in Fig. 2, where the fibre strength remains constant with annealing time.

(B) With increasing annealing time the reaction zone grows, and the effect of the notch formed in the

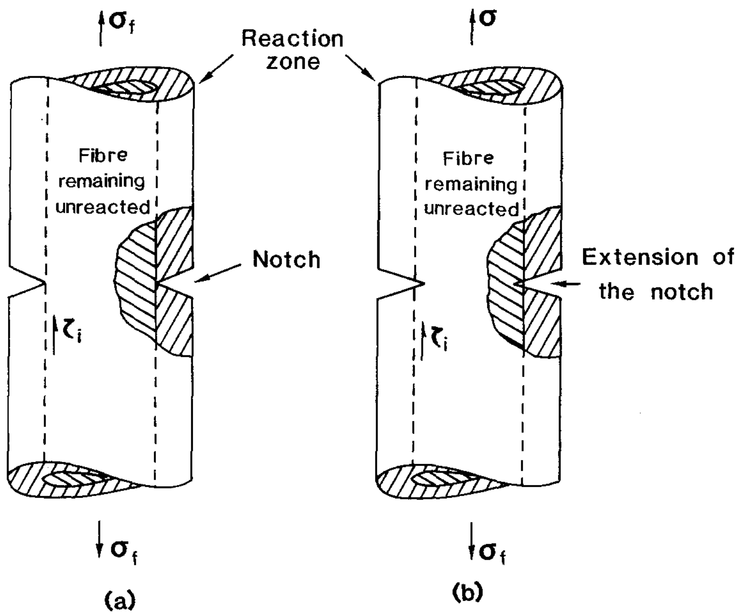

Figure 6 Models for analysis of the effects of notch formation on the tensile strength of the fibre: (a) a notch forms but does not extend; (b) a notch forms and extends into the fibre.

reaction zone becomes more important to limit the fibre strength. Then, the fibre strength decreases as the thickness of the reaction zone increases.

In Case B, it is reasonable to assume that the reaction zone is brittle and fractures on loading at a small strain. Under such an assumption, from the geometric consideration of the fibre with the reaction zone, it seems that the value of σ_f could be estimated from the relation

$$\sigma_f = \sigma^*[r_u^2/(r_u + \delta)^2] \qquad (5)$$

where r_u is the radius of the fibre core remaining unreacted. When a notch forms through the reaction zone but does not extend into the fibre, as in Fig. 6a, the value of σ_f could be predicted by Equation 5. But if the notch extends into the fibre, as in Fig. 6b, the value of σ_f will be lower than that predicted by Equation 5.

Suppose that a notch resulting in fracture forms at a stress level σ_{hf} and extends into the fibre at a stress level of σ_{he}. If $\sigma_{hf} < \sigma_{he}$, the notch forms first and then extends into the fibre when the stress reaches σ_{he}. In this case the value of σ_f is given by σ_{he}. If $\sigma_{hf} > \sigma_{he}$, as soon as the notch forms, it propagates catastrophically through the fibre. In the latter case the value of σ_f is given by σ_{nf}. In a theoretical treatment by Ochiai *et al.* [23] the expressions for σ_{hf} and σ_{he} are given as

$$\sigma_{hf} = C(1/2\pi r_u L)^{1/m}(E_f/E_z)\Gamma(1 + 1/m)\delta^{-1/m} \qquad (6)$$

and

$$\sigma_{he} = \frac{1}{1.122} K_0 \pi^{-1/2} \delta^{-1/2} \qquad (7)$$

where C is a constant, L the gauge length of the tensile specimen, m the Weibull distribution shape parameter, Γ the gamma function, K_0 the fracture toughness of the fibre, and E_f and E_z are the modulus for the fibre and that for the reaction zone. The values of m for the fibre treated in the present experiment have been evaluated as $m > 2$ by the Weibull distribution analysis by the present authors [24].

According to Equations 4, 6 and 7, the variations of σ^*, σ_{hf} and σ_{he} with increasing reaction thickness δ, can be schematically shown in Fig. 7a. Using the conditions that in the early stage of reaction zone growth, $\sigma_f = \sigma^*$; in the later stage, as discussed in Case B, if $\sigma_{hf} < \sigma_{he}$, $\sigma_f = \sigma_{he}$, if $\sigma_{hf} > \sigma_{he}$, $\sigma_f = \sigma_{hf}$, then the tensile strength of the fibre with the reaction zone, σ_f, will vary with increasing δ along the route of A → B → C → D, as shown in Fig. 7b. Comparing Fig. 7b with Fig. 2 we can find that σ_f changes with δ in the same way as σ_h does with $t^{1/2}$, because σ_f contributes the main part of σ_h in Equation 3 and δ is proportional to $t^{1/2}$ as shown in Equation 2.

The reasons for fibre failure in Cases A and B can also be identified by analysis of the tensile test results by the Weibull distribution which has been carried out in a previous work by the present authors [24]. It has been found that for the fibre in Case A, the strength values satisfy the multi-modal Weibull distribution, where the inherent defects on the surface and those inside the fibre competitively control the fibre strength. In Case B, the strength of the fibre can be described by a composite Weibull distribution model, where the strength at higher level is mainly controlled by the inherent surface defects and the strength at lower level by the defects caused by the interface reaction.

4.3. Growth kinetics

From Equation 2 we know that a constant K is an important parameter to determine the growth rate of the reaction zone. The expression for K is given by the Arrhenius law as

$$K = A \exp(-Q/RT) \qquad (8)$$

where A is a material constant, Q the apparent activation energy for the interface reaction, T the absolute temperature and R the gas constant. Because K determines the growth rate of the reaction zone, it also affects the tensile strength as affected by the thickness of the reaction zone which varies with annealing conditions.

The relationship between the tensile strength, σ_f, and the annealing temperature, T, has been found experimentally, as shown in Fig. 3. From this figure

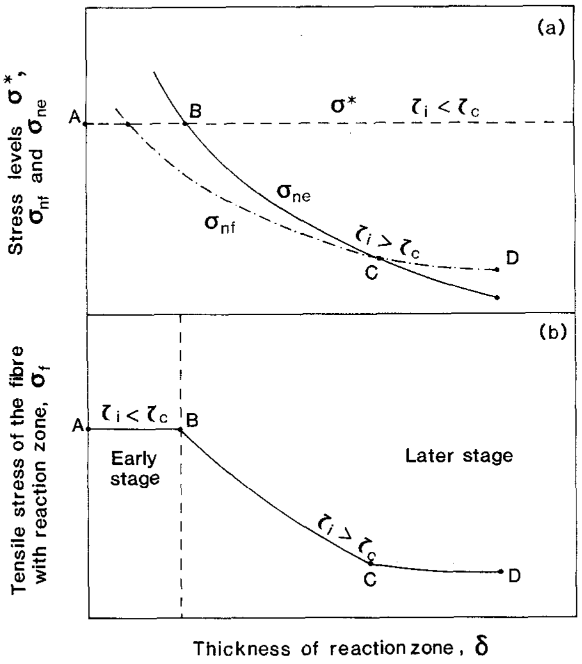

Figure 7 Schematic illustrations of the variations in: (a) the tensile strength levels, σ^*, σ_{hf}, σ_{he}; and (b) strength of the fibre, σ_{f}, with increasing thickness of the reaction zone, δ.

we could estimate the apparent activation energy for the interfacial reaction. For this purpose an expression of σ_{f} as a function of the annealing temperature, T, should be introduced. In Case B it is reasonable to assume that if the reaction zone is not too thick, the stress, σ_{hf}, required for notch formation is lower than the stress, σ_{he}, required for notch propagation. Therefore, the fracture process should be initiated with a notch formation, and then with the increase in stress the notch extends into the fibre leading to fracture. Under such an assumption, σ_{f} should mainly be limited by the strength level, σ_{he}, as expressed in Equation 7. Introducing Equations 2 and 8 into Equation 7, and dividing both sides of the derived equation by σ^0, we have

$$\sigma_{\text{f}}/\sigma^0 = K_0/(1.122\sigma^0\pi^{1/2}A^{1/4}t^{1/4})\exp(Q/4RT) \tag{9}$$

Taking the logarithm of both sides, we obtain

$$\ln(\sigma_{\text{f}}/\sigma^0) = \ln[K_0/(1.122\sigma^0\pi^{1/2}A^{1/4}t^{1/4})] + Q/4RT \tag{10}$$

It is seen by such a mathematical treatment that the tensile strength of the fibre with a reaction zone, σ_{f}, could be correlated with the annealing temperature, T, by the linear relation between the quantity $\ln(\sigma_{\text{f}}/\sigma^0)$ and the reciprocal of absolute temperature, $1/T$, with $Q/4R$ being the slope of the straight line. From the measured slopes of the plots in Fig. 3 the values of the activation energy, Q, for the interfacial reaction of the coated fibres is estimated to be about 180 kJ mol^{-1}.

This value is a little lower than the value of 195 kJ mol^{-1} evaluated for the interfacial reaction between SiC fibres and aluminium or Al–5 at % alloy by Towata and Yamada [9]. The evaluated activation energy value is thought to account for the diffusion of aluminium through the reaction zone which determines the rate of reaction shown in Equation 1. The activation energy for carbon to diffuse through the reaction zone of Al_4C_3 has been evaluated as 147 kJ mol^{-1} [25] and that for silicon through liquid aluminium as 137 kJ mol^{-1} [19]. Both these values are fairly lower than the activation energy derived here. In an observation by Iseki *et al.* [26], sintered SiC was joined with aluminium by heating at 1073 and 1273 K. Interface analysis by EPMA has shown the existence of Al_4C_3 and silicon in the reaction zone. Silicon in the reaction zone comes from the decomposition of SiC as shown in Equation 1. We believe that for the SiC/Al system the reaction takes place most possibly at the interface between the fibre and the reaction zone.

4.4. Fracture mode of the coated fibre

Because the change in σ_{f} in the early stages of growth of the reaction zone is found to be different from that in the later stages, discussion of the fracture behaviour of the coated and annealed fibre should be made at the respective reaction zone growth stage. It is easy to understand that with increasing thickness of the reaction zone, the interfacial shear strength, τ_{i}, increases. The value of τ_{i} with respect to a critical value of the interfacial shear strength, τ_{c}, determines the fracture mode of the coated fibre. τ_{c} is such a shear strength level: when $\tau_{\text{i}} < \tau_{\text{c}}$, interfacial debonding occurs prior to the failure of the fibre; when $\tau_{\text{i}} > \tau_{\text{c}}$, the fibre fractures before interfacial debonding occurs. In the early growth stages of the reaction zone, as shown in Figs 2 and 7, the reaction zone is thin and the interfacial shear strength is weak, so that $\tau_{\text{i}} < \tau_{\text{c}}$. In the case of a uniaxial tensile test on a coated fibre, interfacial debonding occurs prior to notch extension and the fibre fractures due to its inherent defects, resulting in a long pull-out length of the fibre, as shown for the as-coated fibre in Fig. 5a and the coated fibres annealed at 823 K for 1 h in Figs 5b to d.

In the later growth stages of the reaction zone, with increasing thickness, the interfacial shear strength, τ_{i}, becomes strong so that $\tau_{\text{i}} > \tau_{\text{c}}$. In this stage the strength of the coated fibre is reduced, depending on the thickness of the reaction zone. Fracture of the coated fibre occurs due to the extension of a notch, resulting in a short pull-out length as shown in Figs 5e, f and g, for the coated fibres annealed at 823 K for 16 h, or in a flat fracture surface as shown in Fig. 5h, for the coated fibre annealed at the same temperature for 25 h. It should be noted that for the fibres coated with Al–Si alloys annealed at 823 K for 1 h the pullout length is longer than that of the aluminium-coated fibre under the same treatment. Such a fact shows the effectiveness of the addition of silicon on the suppression of the interfacial reaction zone growth.

5. Conclusion

The influence of the interfacial reaction zone growth on the tensile strength of the SiC fibre coated with aluminium and the effect of an alloying addition of silicon on the control of the zone growth are investigated. The following results were obtained

1. The growth kinetics of the reaction zone between the SiC fibre and pure aluminium or Al–Si alloys is divided into two stages. In the early stage the reaction is small, the fibre strength is not affected by the

interfacial reaction, and fibre fracture is caused by its inherent defects. In the later stage, the interfacial reaction is greater and the fibre strength is reduced with increasing thickness of the reaction zone. Fibre fracture is caused by the propagation of a notch formed in the reaction zone into the fibre.

2. Interfacial shear strength determines the fracture mode of the fibre. In the early growth stages of the reaction zone, the shear strength is weak, and debonding occurs before the notch propagates into the fibre. Fracture of the fibre leads to a long pull-out length of the fibres. In the later growth stages, the interfacial shear strength becomes stronger, under longitudinal loading a notch forms and then extends into the fibre before interfacial debonding occurs. Fracture of the fibre results in a short pull-out length or in a flat fracture surface.

3. Alloying addition of silicon to aluminium could retard the growth rate of the reaction zone. The strength loss of the fibre coated with Al–Si alloys after heat treatment is much less than that of the fibre coated with pure aluminium.

4. The results obtained here are expected to be applicable in the fabrication process of the SiC/Al composites by the hot-pressing method, because the temperatures employed here are well below the melting point of aluminium. For the fabrication process by the infiltration method, the effect of silicon addition on the interfacial reaction control is expected to be more extensive, because molten aluminium reacts with SiC at a faster rate.

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