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Properties of Sintered SiC Whisker-Reinforced Si_3N_4 Composites

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Abstract

High-density sintered SiC whisker-reinforced Si_3N_4 materials prepared by slip casting have been characterized in terms of toughness and strength. At high temperature, a decrease in strength is observed consistent with reported observations on high additive-content silicon nitride. At room temperature bending strength increases with whisker content. This strength increase is attributed to a load-transfer mechanism and is independent of toughening which is very low. The interface bonding is therefore relatively strong. However, some whisker–crack interactions have been noticed, suggesting that some toughening mechanisms are in operation. The increase in toughness is offset by the reduction in toughness due to the whisker-induced reduction in silicon nitride grain size.

Schlickergegossenes, gesintertes und SiC-Whisker verstärktes Si_3N_4 hoher Dichte wurde auf seine Zähigkeit und Festigkeit untersucht. Bei hohen Temperaturen wird, genau wie bei Si_3N_4 mit hohem Additivgehalt, eine Festigkeitsabnahme beobachtet. Bei Raumtemperatur nimmt die Biegefestigkeit mit steigendem Whiskergehalt zu. Diese Festigkeitszunahme resultiert aus einem Lastübertragungsmechanismus und hängt nicht von einer eher geringen Zähigkeitssteigerung ab. Allerdings konnten einige Whisker–Riß Wechselwirkungen nachgewiesen werden, die doch ein Indiz für einen zähigkeitssteigernden Mechanismus sind. Dieser Zähigkeitsanstieg wird jedoch dadurch neutralisiert, daß die whiskerbedingte Reduzierung der Si_3N_4-Korngröße einen Zähigkeitsverlust mit sich bringt.

On a étudié la ténacité et la résistance mécanique de matériaux Si_3N_4 renforcés par des whiskers de SiC, préparés par coulage de barbotine et frittés jusqu'à des densités élevées. A haute température, on observe une diminution de la résistance qui est en accord avec les études concernant le nitrure de silicium à haute teneur en dopants. A température ambiante, la résistance à la rupture augmente avec une teneur croîssante en whiskers, ce que l'on peut expliquer par un mécanisme de transfert de charge, cette augmentation étant indépendante du gain de ténacité, qui reste très faible. La liaison interfaciale est par conséquent relativement forte. On a pu toutefois relever certaines interactions whisker–fissure, suggèrant la mise en jeu de mécanismes d'accroîssement de la ténacité. Ce gain de ténacité est compensé par la réduction due à la diminution de la taille de grains du nitrure de silicium provoquée par les whiskers.

1 Introduction

The reinforcement of ceramic matrices with ceramic whiskers can lead to significant improvements in mechanical properties such as strength or fracture toughness. Moreover whiskers have the advantage compared with long fibres to allow lower-cost conventional processing.

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To date most of the whisker-reinforced composites have been made by hot pressing or hot isostatic pressing in order to overcome densification difficulties that arise from the introduction of reinforcement in the powder. However, in a recent publication[1] a process to fabricate composites of silicon nitride reinforced with up to 20 wt% silicon carbide whiskers using low pressure sintering was described. Consolidation by slip casting was shown to bring the double advantage of leading to nearly dense material containing up to 15 wt% whiskers, with a highly homogeneous whisker dispersion free from agglomeration bundles.

The mechanical behaviour improvement, conferred by the whiskers, is very different for the two matrices Al_2O_3 and Si_3N_4. A very significant increase is generally reported, both in strength and toughness for the SiC–Al_2O_3 composites. Results for a Si_3N_4 matrix are less consistent and have led to some controversy. This is in part attributed to the much higher performance of the monolithic Si_3N_4 material. Another significant factor relates to the activation of the theoretical models for toughening by fibre–whisker debonding, and crack deflection, or pull out. The SiC–Si_3N_4 bond is stronger than for SiC–Al_2O_3 and often precludes debonding, and therefore the operation of these models.

The present work is concerned with the mechanical behaviour of the sintered SiC–Si_3N_4 composites, especially with respect to strength and toughness. The analysis of the toughening has been conducted by fracture surface studies, crack–whisker interaction observations, residual stress analysis and as well by comparison with monolithic material. This is concluded with some considerations toward improving material properties.

2 Experimental Procedure

The SiC whisker reinforced Si_3N_4 composites were produced by dispersing silicon nitride powder (TS10, Toyosoda, Japan) with 10 wt% of an equimolar ratio of Al_2O_3/Y_2O_3 (Al_2O_3, Janssen Chimica, Belgium; Y_2O_3, fine grade, Hermann C. Stark, Berlin) in water containing 0·3 wt% of a dispersing agent (Darvan C, Van der Bitt Co., USA) in an attrition milling device with 2h milling. The whiskers were added in the last 15 min of the milling step in order to get a good dispersion in quantities up to 20 wt%. The whiskers' dimensions were measured by direct SEM analysis giving a diameter of 0·9 μm and an average length of 15 μm. The low viscosity slip obtained was then poured into plaster of Paris moulds of cylindrical shape ($\phi = 45$ mm, $l = 65$ mm). Some preferred orientation of the whiskers in surfaces parallel to the mould–slip interface was observed. The green bodies were sintered at 1850°C for 6 h under 1 MPa (10 bar) nitrogen pressure. X-ray and TEM analysis showed no apparent degradation of the whiskers. The final densities as a function of the whisker content are represented in Table 1.

Table 1. Measured density and Young's modulus of several composites. Calculated moduli from the modulus–porosity are also listed

Whisker (wt%)	*Density (%TD)*	*Measured modulus (GPa)*	*Calculated modulus*[a] *(GPa)*	*Measured/ calculated modulus*
0	99·20		303·39	
5	99·05	302	302·00	
10	98·47	307 ± 1	296·69	1·03
15	97·69	307	289·69	1·06
20	93·36	279 ± 3	253·74	1·10

[a] Calculation from $E = E_0 e^{-3.06p}$ (eqn (1)) with whisker (wt%) as reference for determination of E_0.

Flexural strength was measured in four-point bending mode with an outer loading span of 40 mm and inner loading span of 20 mm on specimens presenting sections of 4 × 3 mm, at a crosshead speed of 0·5 mm/min. The samples were cut such that the tensile face was parallel to the whisker plane. (On this small width, the whiskers can be considered oriented in a plane with a good approximation.) Elastic moduli were determined by an audio-resonance technique from the Grindo Sonic method. The toughness values were measured by the Vickers' indentation technique calibrated with double torsion testing on similar specimens (different casting moulds).

3 Results

3.1 Modulus

The elastic moduli are listed in Table 1. In this table the error represents the maximum and minimum deviation from average values of specimens made from different green bodies. In order to appreciate the influence of whiskers on the modulus one must eliminate the density parameter. While elastic modulus porosity relationships of RBSN are well documented in the literature, the data are more sparse for dense silicon nitride. Dutta *et al.*[2] report a formula proposed by Larsen *et al.*:

$$E = E_0 e^{-3.06p} \quad (1)$$

where p = porosity and E = elastic modulus with $E_0 = 313$ GPa. Assuming that 5% whisker loading is low and replacing p and E by the respective values measured for this whisker loading, $E_0 = 311$ GPa is obtained, showing that this relation fits reasonably with the present data. In Table 1 the modulus values calculated according to relation (1) (using $E_0 =$ 311 GPa) are listed. The experimentally measured modulus values deviate significantly with increasing whisker content from calculated values, showing that the whisker addition leads to an increase of elastic modulus.

This observation can be confirmed by recalling that the elastic modulus of the whiskers is higher than the elastic modulus of the matrix. Pandey & Dayal[3] report values of the elastic tensor for SiC, namely: $C_{11} = 352{\cdot}3$ GPa, $C_{12} = 140{\cdot}4$ GPa and $C_{44} = 232{\cdot}9$ GPa, from which the tensile modulus in the direction $\langle 111 \rangle$, which is the long axis of the whiskers, may be calculated as $E_{111} = 511$ GPa.

3.2 Toughness

Toughness measurement by indentation presents several advantages, such as simple preparation and small sample size. However, the toughness analysis is a critical problem and the selection of an accurate formula to calculate the toughness is important. On this basis a formula was chosen by comparison with results obtained by double torsion on similar samples (prepared for the processing investigation[1]). Finally the following relation was used:

$$K_{Ic} = 0{\cdot}057 H \sqrt{a} (E/H)^{2/5} (C/a)^{-3/2} \qquad (2)$$

where H is the Vickers hardness, a is one-half the length of the diagonal of the Vickers impression, and C is one-half the median crack length, proposed by Singh *et al.*,[4] which has the advantage of giving results close to those from double torsion and is used in the literature to assess similar material to that of the present study. The toughness values have been calculated with elastic modulus constants listed in Table 1.

The toughness was measured on a surface parallel to the whisker plane and on a surface perpendicular to the whisker plane. In the first case, the crack planes are perpendicular to the whisker plane, while in the second case one of the cracks is parallel to the whisker plane.

The results, as a function of the whisker content, are shown in Fig. 1. There is only a modest toughening, very small compared to that which can be predicted by theoretical toughening analysis. However, since the toughness measured in the perpendicular direction seems to be lower and

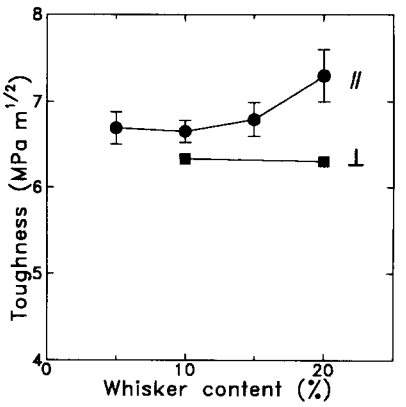

Fig. 1. Toughness versus whisker content measured by indentation. The diamond is applied on a plane. ⊥, perpendicular to whisker plane; //, parallel to whisker plane.

constant with whisker content, clearly there is a toughening effect due to whiskers. Such an orientation effect has already been reported on materials processed by hot pressing or injection moulding.[5-7]

3.3 Strength

The results at ambient temperatures in terms of strength as a function of the whisker content are represented in Fig. 2. It is noted that the strength increases with whisker content up to 15% whiskers and decreases with the 20% whisker-containing material. This decrease might be explained by the significant density decrease between 15 and 20%. In order to test this assumption, the strength that is expected from the strength–porosity relationship may be plotted. Those relations are usually described following:

$$\sigma(p) = \sigma_0 e^{-ap} \qquad (3a)$$

$$\sigma(p) = \sigma_0 (1-p)^m \qquad (3b)$$

where σ is the strength for a porosity p, σ_0 is the strength for fully dense material, and a and p are

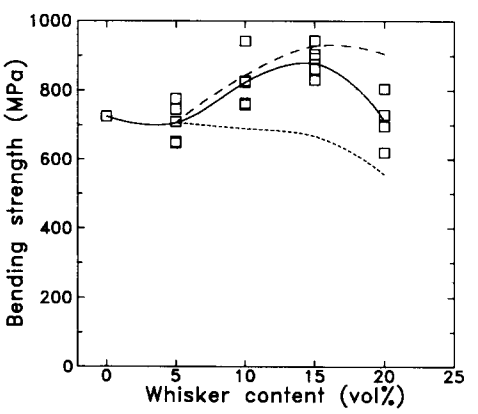

Fig. 2. Flexure stress at ambient temperature versus whisker content. ——, Measured strength values; – – –, strength calculated according to increasing porosity, and — —, extrapolated strength of a 99% TD material.

constants. Dutta *et al.*[2] report values of $a \approx 4{\cdot}2$ and $m \approx 4$ for dense silicon nitride; these values are obtained by fitting of numerous data from the literature.

Figure 2 represents the strength calculated from relations (3) where σ_0 is estimated from the strength and porosity of the 5% whisker-content material. This confirms the explanation of lost strength caused by the porosity increase. In order to estimate the whisker influence on the strength, it is also possible to extrapolate the strength of a hypothetical 99% dense material (Fig. 2). The strength increases up to 15% whisker content and then remains constant between 15 and 20%, suggesting that within the conditions explored here, the optimum material in terms of strength contains 15% whiskers.

The strength has also been measured at high temperatures in air for the 10% and 20% whisker material (Fig. 3). Strength decreases significantly with temperature as generally reported for silicon nitride sintered with high additive content. However, a more detailed observation shows that the stress variation and dispersion are not similar for both materials.

The material containing 10% whiskers reveals a glassy intergranular phase[1] and shows a regular strength decrease with temperature. Moreover, the scatter decreases, with increasing temperature up to 1200°C. This temperature corresponds to the outset of fracture originated from slow crack growth (Fig. 4). Below this temperature the fracture originates from defects under the tensile face.

The material containing 20% whiskers contains a partially crystallised intergranular phase and shows a different behaviour. The strength dispersion slowly decreases but is still significant at 1300°C. No slow crack growth zone can be observed on fracture surfaces up to the maximum tested temperature. Up to 1000°C the strength decrease is lower than in the latter case, followed by a remarkable decrease at 1200°C. At this temperature, the strength of the 20% whisker material is lower than the strength of the 10% material. However, at 1300°C the strength of the 20% composite is again higher than the strength of the other material. An analysis of the weight gain during exposure to air shows (Fig. 5) a significant minimum in the oxidation kinetics. From this it may be suggested that the fracture stress behaviour of the 20% whisker composite is related to the oxidation behaviour of the crystallised phase.

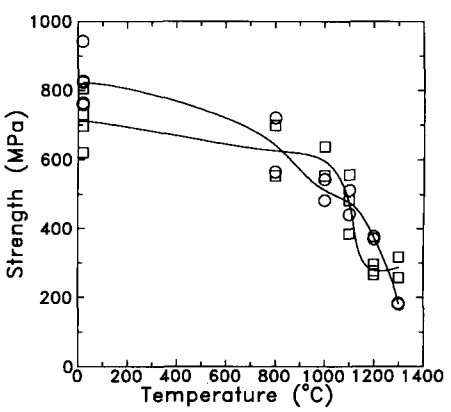

Fig. 3. Flexure strength versus temperature in air for (○) 10% and (□) 20% whisker-content composite.

The high temperature mechanical behaviour differences of both materials can not be attributed to variations in whisker content. In any case the materials are covered by an oxide layer above 1100°C which makes the observation of whiskers on the fracture surface difficult. A few samples ejected in a cooler part of the surface during fracture allowed some observations to be made, but even under these conditions the surface is smoother than the fracture surface made at ambient temperature, suggesting that the intergranular phase has, at least in part, melted, or that the sample contained some glassy phase.

Note that in the case of the 20% whisker composite, the apparent whisker length does not seem to be more important at 1300°C than at ambient temperatures, perhaps due to the fast rupture mode.

4 Discussion

4.1 Toughening

The toughness results suggest that the benefit brought by whiskers is low and even negligible. Consideration of the thermal expansion coefficients of both the silicon carbide ($3{\cdot}3$ to $4{\cdot}5 \times 10^{-6}\,K^{-1}$) and the silicon nitride ($3 \times 10^{-6}\,K^{-1}$) suggests that the whiskers are in tension and the matrix is under tangential compression and radial tensile stresses, giving tensile stresses in the interface. This has been confirmed by measurements conducted by X-ray diffraction studies on the whiskers and the matrix.[8] Here a tensile stress equal to 82 MPa in the whisker was obtained for a 20% whisker-content composite. Such a stress state is favourable for crack propagation along the interface.[9]

These two statements would suggest that the whisker–matrix interface is too strong, as has often been reported. However, an analysis of whisker–crack interactions by observation of crack paths made by indentation shows that the crack does not go systematically through the whisker, as is the

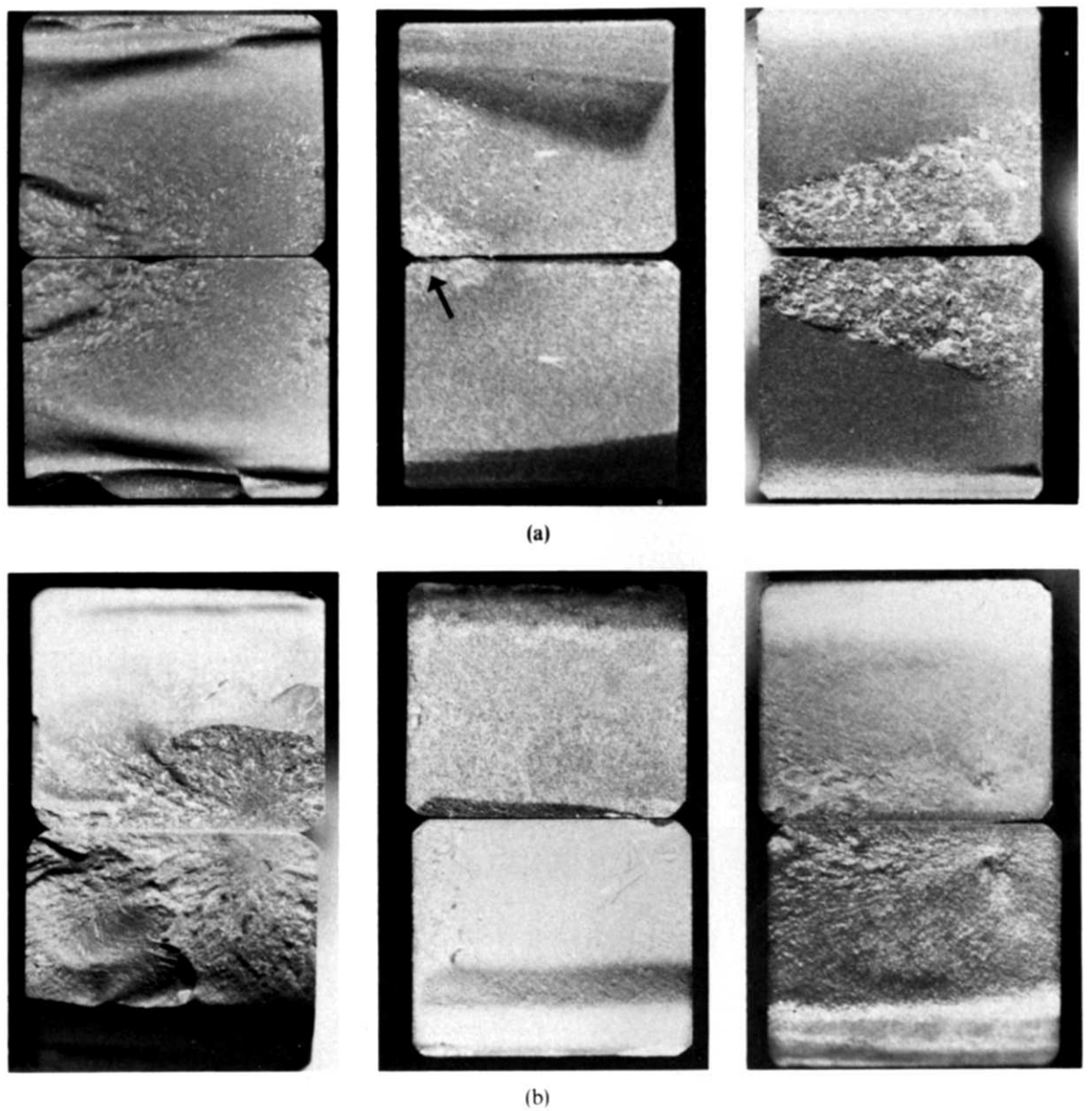

Fig. 4. Fracture surfaces of composite materials at several temperatures. The zone of slow crack growth is indicated by an arrow. (a) 10% SiC whisker and (b) 20% SiC whisker at (from left to right) 1000°C, 1200°C and 1300°C (× 10).

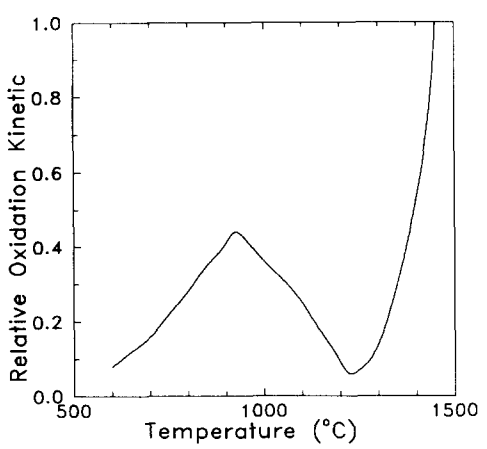

Fig. 5. Oxidation kinetics versus temperature of a 20% whisker composite material (relative scale). Calculation made from a linear heating rate oxidation thermogram.

case with a strong interface. Figure 6 shows that the crack is often deviated by whiskers. Moreover, the same figure also shows that the crack deviation increases with whisker content. At higher whisker content, the cracks which are oriented such that their plane is parallel to the whisker plane are straighter, which explains the toughness difference between perpendicular and parallel planes.

Among the several toughening mechanisms, crack bridging by the whiskers appears to be able to lead to the highest degree of toughening. Such bridging may arise when the crack extends through a whisker without breaking it, i.e. some debonding must occur along the whisker from the crack. Since such a mechanism arises for a very small crack opening, it is

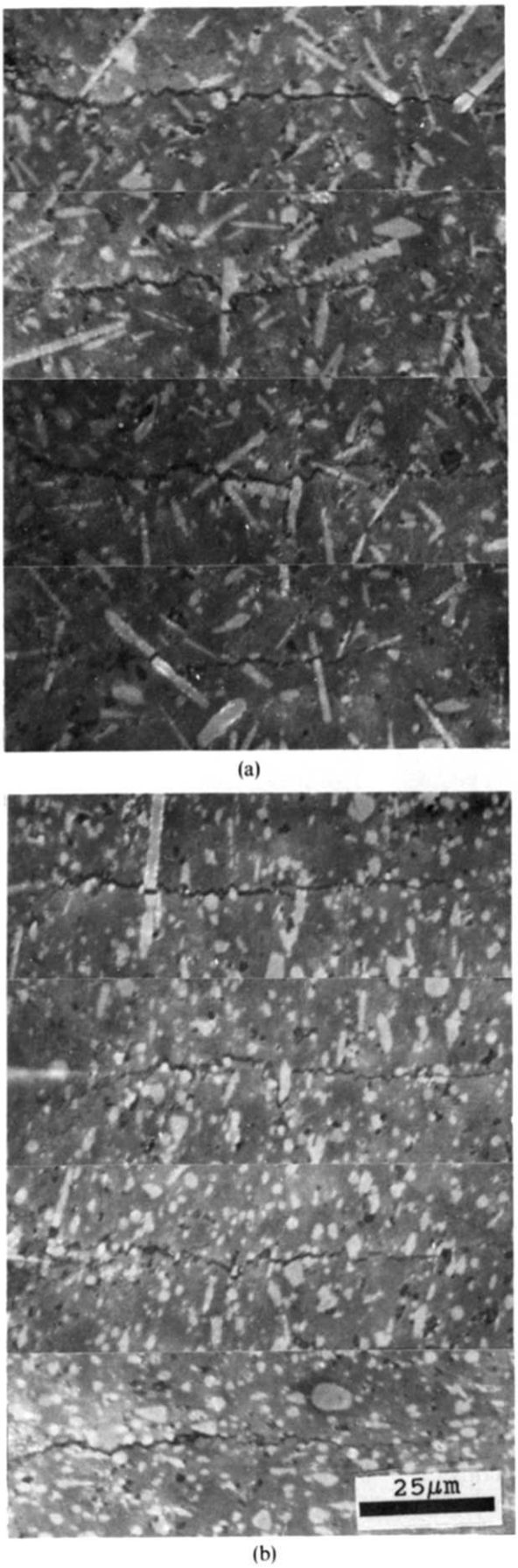

Fig. 6. Micrographs of Vickers crack profiles of a 20% whisker composite. Surface (a) parallel and (b) perpendicular to the whisker plane.

difficult to observe. Rühle *et al.*[10] proposed to observe such material under the scanning electron microscope, without conductive coating and under low tension. In the present case even with low acceleration voltage, the sample quickly became charged, making such an observation impossible. Moreover, the complex microstructure of the matrix leads to significant local variations of conductivity, which gives higher local contrast variation in the matrix than between the whiskers and the matrix.

Therefore the authors tried to make some investigations by directly observing a controlled crack propagation under the optical microscope on a polished double torsion specimen. The crack opening is too small to see whether the crack breaks the whiskers immediately or not. However, it is possible to observe the interface along the whisker after fracture on abnormally large whiskers, since on the one hand, it is easier to observe on big whiskers, and on the other hand, this mechanism is of higher probability on larger diameter whiskers. Indeed, it has been shown[11] that the stress (σ) leading to debonding is:

$$\sigma = (4E_f\tau/r)^{1/2} \tag{4}$$

where E_f is the elastic modulus of the fibre, τ interfacial shear stress and r the whisker diameter.

Figure 7 shows such an example where some debonding along the whisker is clear. However, it is

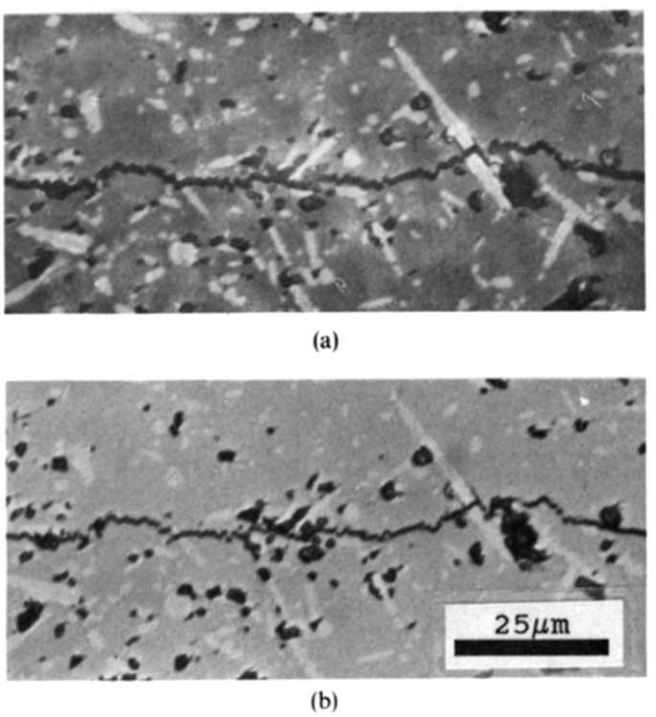

Fig. 7. Micrographs of a crack induced by double torsion. (a) Loaded specimen, (b) relaxed specimen thinly gold coated in order to increase the contrast. Note the debonding along the whisker near both sides of the crack.

not possible to assert whether the whiskers effectively exert a transient closure stress on the crack or not.

The in-situ observation of controlled crack propagation allowed the observation of the crack advancing by steps, in which the speed of propagation increases quickly and then slows down and even stops transiently before accelerating again. This behaviour is repeated for several cycles for each load increment. The crack seems to be regularly and temporarily pinned.

The in-situ observation also shows a high propensity to crack forking. When the crack slows down and becomes pinnned, some microcracks arise from the apparent crack tip. These microcracks grow and sometimes coalesce up to the next main crack advancement. After these steps, most of the microcracks close and can not be observed. A detailed observation shows as well that some (apparently) isolated microcracks arise around the main crack tip. These considerations suggest that some microcracking toughening exists. Careful observation of the video films seems to indicate that these microcracks arise from whiskers all the more that the extent of microcrackings exists but is less pronounced on monolithic ceramics.

The analyses demonstrate that the bonding is not 100% strong and that the several toughening mechanisms described in the literature occur in the material, albeit to a minor extent. In order to understand this contradiction between the poor toughening and the crack–whisker interactions, the fracture behaviour of the corresponding monolithic ceramic was analysed.

The toughness of materials containing several additive contents measured under the same conditions as the composite ceramics are listed in Table 2. A comparison of the toughness of the composite ceramics with the equivalent 10% additive monolithic ceramic shows that the toughness is similar to the 5–10% whisker-composite material. Moreover, the monolithic ceramic that contains 6% additives shows a higher toughness. That the structure of those monolithic materials consists of elongated grains is of importance for crack extension behaviour. The aspect ratio and the diameter of the Si_3N_4 grains have been measured on etched polished samples following the Wötting *et al.*[12] method, which supposes the grain aspect ratio as constant within the sample. As already reported,[13] the average grain size is lower in the composite material than in the monolithic alone, but the aspect ratio of the grains in the two materials is the same, within experimental error.

Table 2. Toughness and average diameter grain size of several monolithic materials (prepared in similar conditions) and of a 10% whisker composite

Additives (%)	*Toughness* ($MPa\ m^2$)	*Average grain size diameter* (μm)
5	$6{\cdot}2 \pm 0{\cdot}2$	$0{\cdot}99 \pm 0{\cdot}6$
6	$7{\cdot}6 \pm 0{\cdot}4$	$0{\cdot}86 \pm 0{\cdot}5$
10[a]	$6{\cdot}5 \pm 0{\cdot}2$	$1{\cdot}04 \pm 0{\cdot}6$
10% whisker	$6{\cdot}6 \pm 0{\cdot}4$	$0{\cdot}53 \pm 0{\cdot}2$

[a] 10% additives corresponds to the composite sintering aids content.

More interesting than the average value is the diameter distribution of the grains (Fig. 8). While the smallest grains of both the composites and the monolithic material are of the same order of magnitude, the largest grains are much bigger in the monolithic tough material. Calculation of the number of grains in the monolithic material shows that more than 20 vol.% of the samples consists of elongated grains having a diameter greater than $2{\cdot}5\ \mu m$, with a length of more than $15\ \mu m$.

In-situ observations of controlled crack extension and fracture surface analysis of such a monolithic material shows that the fracture is mostly intergranular. Numerous grains are partially pulled out (Fig. 9). Furthermore, the larger grains having a diameter in the range 3 to $6\ \mu m$ seem to be more debonded. In comparison, the composite materials, while also exhibiting an intergranular fracture and debonding of grains and whiskers, show more regular fracture with less debonding. This is believed to be caused in part by the smaller average grain size and also by the rougher surface of the whiskers

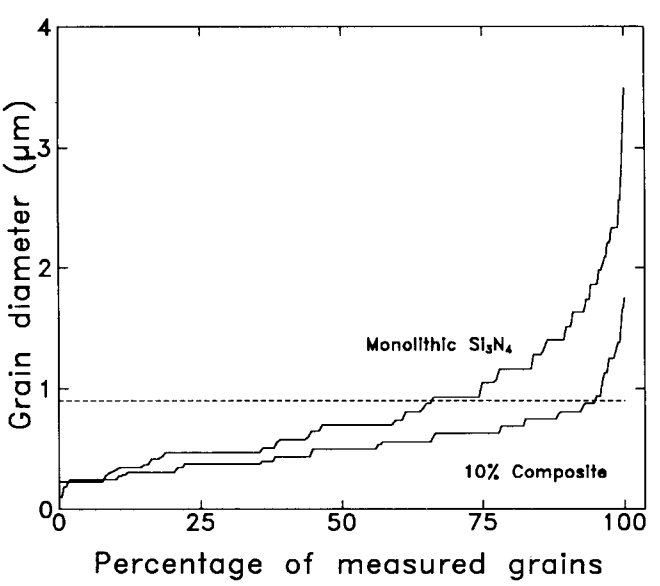

Fig. 8. β-Si_3N_4 grain diameter distribution of a 10% whisker composite and tough monolithic ceramic.

Fig. 9. SEM micrograph fracture surface of the toughest monolithic ceramic.

compared with the smooth large Si_3N_4 grains in the tough monolithic material.

Thus low acicular grains clearly reinforce the material according to the mechanisms described for whisker composites and account for the high toughness of this monolithic ceramic material. It is also important to note that the toughness of this monolithic material is all the more significant since it exhibits isotropic behaviour, while in most composites of high toughness described in the literature or prepared here, the whiskers are uniaxially oriented or are planar. While it is generally admitted that an increase of Si_3N_4 grain aspect ratio is favourable, the effect of the average grain size on toughness is not fully clarified in the literature. Some fracture energy increases with grain size have been reported[14] and associated with increasing microcracking with grain size. Buljan & Sarin[15] studied sintering silicon nitride at extended times in order to increase the grain size and recorded a simultaneous toughness and strength increase. Moreover, they proposed the following equation that relates the toughness increase to the grain size increase:

$$K_{Ic} = C \cdot K_{Ic}^{0}(d/d^{0} - 1) \tag{5}$$

where C is a constant, K_{Ic} is the toughness for a grain size d and K_{Ic}^{0} the toughness for a grain size d^{0}. This relationship, in accordance with the present authors' observations, may describe the toughness decrease concomitant with grain size decrease between the monolithic material and a pseudo monolithic material that would have the grain size of the composite.

These findings suggest that there are two conflicting effects in the composite material, namely, that the whiskers act to toughen and the smaller grains act to reduce the toughness of the pure matrix. Finally, the stress intensity in the composite may be described as:

$$K^{c} = K_{0}^{c} + \Delta K_{w} \tag{6}$$

where K_0^c is the toughness of the matrix which depends on grain size according to the relation (6), and ΔK_w is the toughening due to whiskers.

4.2 Strengthening

The flexure strength increases with whisker content between 5 and 15% whisker content, and extrapolation to dense composite suggests a strength increase up to 15% whiskers associated with whiskers. As shown previously, the strength increase cannot be attributed to a toughness increase according to the Griffith relationship:

$$\sigma_f = YK_{Ic}/\sqrt{\pi a} \tag{7}$$

where σ_f is the strength, K_{Ic} the toughness and a the defect size. Since the elastic modulus increases with whisker content, one may consider the possibility that the strengthening is due to load transfer from the matrix to the fibre through shear. This load transfer depends on the properties of the matrix and the reinforcing fibre and one must maximise the modulus ratios (E_f/E_m) and also the maximum strength of the fibre (σ_f). However, strengthening has been reported[16] in the literature for a modulus ratio as small as 2.

For a finite length fibre, a critical length below which no strengthening may be expected has been defined. Gadkaree & Chyung[17] report elastic–elastic analyses that are in agreement for good bonding (not frictional bond) which are diverging for low fibre content ($<20\%$), the more severe being the Rosen[18] which leads to the following critical aspect ratio:

$$\frac{L_c}{d_f} = \left[\left(\frac{1}{\sqrt{V_f}} - 1\right)\frac{E_f}{E_m}\right]^{1/2} \operatorname{argch}\left(\frac{1 + (1-\phi)^2}{2(1-\phi)}\right) \tag{8}$$

where L_c denotes critical length, d_f the fibre diameter, V_f the fibre fraction, E_m and E_f the matrix and fibre elasticity moduli respectively and ϕ the ratio of fibre stress to fibre stress in an infinitely long fibre (characteristic of fibre efficiency), where ϕ is in the range 0·2 to 0·9 (depending on V_f and whisker length).

For the silicon carbide whiskers–silicon nitride

matrix, the modulus ratio is close to 1·65, which is very low, but at the same time the strength ratio (fibre strength/matrix strength) is very high. Calculation of the critical length according to relation (8) gives L_c/d_f in the range 0·42 to 1·1, depending on ϕ ($V_f = 0{\cdot}1$). Thus, the whisker length is long enough to obtain some strengthening by load transfer.

However, it should be kept in mind that strengthening could also be due to some change in critical flaw size or morphology, according to relation (7). Firstly, since the shrinkage during sintering is anisotropic with increasing anisotropy with whisker content, the porosity (and flaws) is certainly anisotropic, leading to lower flaw stress intensity in the direction perpendicular to the tensile axis (note that the tensile face of the four-point bending bar is parallel to the whisker plane). Secondly, whiskers that bridge flaws or pores may exert a force opposite to crack opening and effectively reduce the stress intensity at the crack tip. Under the conditions described, this strengthening is very interesting, since it shows that the whisker distribution obtained within the matrix is highly homogeneous.

5 Conclusion

The authors have shown the possibility of fabricating sintered whisker composite materials of high ambient temperature properties. The strength increase (with whisker content) demonstrates the high distribution homogeneity of the whiskers obtained by this processing method.

Strength strongly decreases with temperature, as expected for such high additive content Si_3N_4-based material. However, it is believed that these properties could be enhanced by modification of additives, and intergranular chemistry.

The toughness of the composite material is the result of some decrease caused by grain size reduction and increase by whisker toughening. A relatively strong interface bonding has been found in these composites, which allows some strengthening by load transfer, but is sufficiently weak that some crack–whisker interactions have been found that lead to limited toughening.

Since a pull out mechanism is to be required to obtain tough and less notch-sensitive material, one must reduce the interface strength by some coating and/or use larger diameter whiskers that can simultaneously lead to load transfer and easier debonding (for a given interfacial strength).

It has been shown as well that the microstructure of the monolithic material could be engineered such that toughness could be much enhanced. A more precise grain size distribution analysis is needed.

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