Effects of Filament-Matrix Interfaces on the Mechanical Properties of SiC-Reinforced Si₃N₄: A Review

C. H. Schilling

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EFFECTS OF FILAMENT-MATRIX INTERFACES
ON THE MECHANICAL PROPERTIES OF SiC-
REINFORCED Si_3N_4 : A REVIEW

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SUMMARY

This report summarizes the results of a literature review of the effects of filament-matrix interfaces on the mechanical properties of ceramic composites composed of SiC-filament-reinforced Si_3N_4 . A general review of the processing and mechanical properties of SiC-filament-reinforced Si_3N_4 is presented with special emphasis on research pertaining to processing-related effects on filament-matrix interfaces and the resulting effects of these interfaces on fracture behavior. A review of coating techniques for ceramic filaments is also presented, and recommendations are made for future directions in processing SiC-filament-reinforced Si_3N_4 with mechanical properties that are enhanced by the microstructure of the filament-matrix interface.

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INTRODUCTION

Ceramic composites of SiC-filament-reinforced Si_3N_4 are prime candidates for advanced heat engine applications, as these materials have the ability to resist creep and corrosion at elevated temperatures. The fracture toughness of these and other filament-reinforced ceramics is enhanced by engineering the shear strength of filament-matrix interfaces by materials processing. Brittle fracture ensues if the degree of filament-matrix bonding is too strong; whereas in tough materials, weaker interfaces permit crack deflection, crack bridging, and filament pullout to occur. Theoretical models describing these phenomena are presented by Evans and co-workers.^(a)

Experimental research has been performed on a variety of ceramic-based composites to enhance fracture toughness by processing-induced modifications to filament-matrix interfaces. Examples include ZrO_2 -matrix composites reinforced with BN-coated SiC fibers (Bender et al. 1986); glass-matrix composites reinforced with SiC fibers having C- and NbC-based layers at fiber-matrix interfaces (Brennan 1986 and 1987; Chaim and Heuer 1987); amorphous matrices reinforced with SiC-based fibers;^(b) alumina-matrix composites reinforced with SiC filaments (Tiegs, Becher, and Harris 1987; Vaughn, Homeny, and Ferber 1987a and 1987b); and several Si_3N_4 -SiC composites that are reviewed in the following sections.

The primary purpose of this literature review is to evaluate processing-related effects on filament-matrix interfaces in SiC-filament-reinforced Si_3N_4 and the resulting effects of these interfaces on mechanical properties. The review is divided into four sections: 1) a general review of the processing and mechanical properties of SiC-filament-reinforced Si_3N_4 , 2) a review of research pertaining to fiber-matrix interfaces in SiC-reinforced Si_3N_4 , 3) a review of research pertaining to whisker-matrix interfaces in

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- (a) Marshall, Cox and Evans 1985; Marshall and Evans 1985; Faber and Evans 1983a and 1983b; Evans and McMeeking 1986; Budiansky, Hutchinson, and Evans 1986; Evans 1988.
 - (b) Godard 1988; Chaim, Brandon, and Baum 1988; Singh 1988b; Goettler and Faber 1988; Mah and Subramanian 1988; Cooper and Chyung 1986; Coyle et al. 1987.

SiC-reinforced Si_3N_4 , and 4) a review of filament coating processes which may be used for tailoring filament-matrix interfaces in ceramic-ceramic composites. This review is followed by recommendations for processing SiC-filament-reinforced Si_3N_4 with mechanical properties that are enhanced by filament-matrix interfaces.

REVIEW OF PROCESSING AND MECHANICAL PROPERTIES

A detailed tabulation of reported processing methods and mechanical properties of SiC-filament-reinforced Si_3N_4 appears in the Appendix. Several processing variations are mentioned in the literature, which primarily entail differences in the amounts and types of raw materials and differences in powder consolidation and sintering procedures, including slip casting, hot pressing, hot and cold isostatic pressing, extrusion, reaction bonding, pressureless sintering, injection molding, and several combinations thereof. Variations in raw materials include the quantity and type of sintering additive (for example, MgO , Y_2O_3 , $\text{Y}_2\text{O}_3\text{-Al}_2\text{O}_3$, $\text{Y}_2\text{O}_3\text{-La}_2\text{O}_3$); the diameter, length, quantity, orientation, and purity of the filaments; and the types of surface coatings which may have been applied to the filaments. All of the processing variations mentioned in the Appendix are expected to result in uniquely defined properties of filament-matrix interfaces by virtue of the specific chemical equilibria and transport processes existing in each case.

Differences in mechanical properties data mentioned in the Appendix are expected to result from variations in processing approaches as well as the several different types of equipment and procedures used to generate fracture data. Thus, it is difficult to generalize processing/properties relationships by correlating one reference to the next. Nevertheless, there are some noteworthy trends in mechanical properties which may be observed. In particular, improvements in room temperature fracture toughness, work-of-fracture, and rupture modulus have been observed in several studies with reinforced (as opposed to unreinforced) Si_3N_4 . Specifically, room temperature fracture toughnesses of SiC-filament-reinforced Si_3N_4 range between approximately 0.5 to 12.5 $\text{MPa}\cdot\text{m}^{1/2}$, with one exception having a maximum of approximately 56 $\text{MPa}\cdot\text{m}^{1/2}$ (Rice et al. 1980). In contrast, unreinforced Si_3N_4 -based materials are reported having a maximum fracture toughness of approximately 8.3 $\text{MPa}\cdot\text{m}^{1/2}$ at room temperature (Kandori et al. 1987). Several authors do not report specific values of fracture toughness, although these authors do indicate "tough" or "brittle" behavior. A significant example of fiber-enhanced toughening is provided by Corbin, Rossetti, and Hartline (1986), who report that fiber reinforcement can

increase the room temperature work of fracture above monolithic Si_3N_4 by at least one order of magnitude. At 1200°C, Buljian and co-workers^(a) also reported an increased toughness from approximately $6 \text{ MPa}\cdot\text{m}^{1/2}$ in unreinforced Si_3N_4 to approximately $7.9 \text{ MPa}\cdot\text{m}^{1/2}$ as a result of SiC whisker reinforcement. Two reports also indicate no change in fracture toughnesses between reinforced and unreinforced Si_3N_4 at 1200°C ($\approx 6 \text{ MPa}\cdot\text{m}^{1/2}$ in Willkens and Corbin 1987; $\approx 15 \text{ MPa}\cdot\text{m}^{1/2}$ in Gibbs, Petrovic, and Honnell 1987).

Although improvements in rupture moduli have been shown in reinforced Si_3N_4 ,^(a) evaluation of individual references in the Appendix indicates that flexure strengths may or may not increase as a result of SiC filament reinforcement. Some of the decreases in flexure strength are attributed to processing-related flaws, such as whisker impurities, porosity, exaggerated growth of Si_3N_4 grains, agglomerated whiskers, or closely spaced SiC-based fibers that have sintered together.^(b) In addition, tensile strengths at room temperature and 1300°C are reported to be greater for unreinforced Si_3N_4 , as opposed to reinforced Si_3N_4 . Specifically, the maximum tensile strength for unreinforced material at room temperature is reported to be 1050 MPa, whereas filament-reinforced material of similar matrix composition is only 650 MPa (Ueno and Toibana 1983; Hayami et al. 1986). At 1300°C, the same unreinforced and reinforced materials exhibit tensile strengths of approximately 750 MPa and 600 MPa, respectively (Ueno and Toibana 1983; Hayami et al. 1986).

In general, most of the references mentioned in the Appendix report that toughening behavior relates to processing history, the degree of filament-matrix debonding, and the microstructure of the filament-matrix interface. However, most of these discussions are brief, and exhaustive correlations among macroscopic physical properties, chemical equilibria at the interface, and thermo-mechanical stresses at the interface have only been performed in a

(a) Buljian 1987; Buljian et al. 1987 and 1988; Buljian, Baldoni, and Huckabee 1987.

(b) Rice et al. 1980; Bender et al. 1978; Lundberg et al. 1987; Shetty et al. 1985; Shalek et al. 1986; Hayami et al. 1986; Ueno and Toibana 1983; Porter, Lange, and Chokshi 1987.

few of these references. The most pertinent references are discussed in further detail in the following sections.

SiC-Fiber-Reinforced Si_3N_4

Correlations between the mechanical properties and fiber-matrix phases in Nicalon-fiber-reinforced Si_3N_4 ^(a) have been reported by Jones et al. (1988) and Schoenlein et al. (1988). Based on analyses by transmission electron microscopy (TEM), two layers of C were observed at fiber-matrix interfaces, which allowed some pullout to occur during fracture. The layers comprised an amorphous C layer, which was adjacent to the fiber, and a graphitic C layer, which was adjacent to the matrix. Fiber degradation occurred during hot pressing, and consequently the room temperature fracture properties of these composites were inferior to unreinforced Si_3N_4 prepared under similar conditions. Fiber degradation was attributed to SiC grain growth and the development of an amorphous intergranular phase in the fiber. This intergranular phase was enriched in Mg, which apparently diffused from the matrix and into the fiber as a result of the MgO sintering aid. In some cases, Mg enrichment was sufficient to crystallize the amorphous phase in the fibers to Mg_2SiO_4 .

Microstructure/property relationships for SiC-fiber-reinforced reaction-bonded Si_3N_4 have been reported by Corbin, Rossetti, and Hartline (1986), and significant changes in the room temperature strength, strain-to-failure, and work-of-fracture of AVCO-fiber-reinforced Si_3N_4 ^(b) were attributed to the thickness of a C layer which existed on the fibers prior to nitridation (see Table 1). Thick coatings (3 μm on AVCO type SCS-6 fibers) resulted in enhanced toughening by way of weak interfacial bonding, fiber pullout, and crack deflection along the middle of the C layer at the fiber-matrix interface. This behavior was less pronounced in samples with thin C coatings (1 μm on AVCO type SCS-2 fibers), as the thin coatings degraded during nitriding to form a strong fiber-matrix bond. In the case of the thin coatings,

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- (a) Nicalon fibers are manufactured by Nippon Carbon Company, Ltd. and distributed by Dow Corning Corporation in the U.S.
 - (b) AVCO fibers are manufactured by AVCO Specialty Materials/Textron Inc., Lowell, Massachusetts.

TABLE 1. Room Temperature Mechanical Properties of AVCO SiC-Fiber-Reinforced Si_3N_4 (Corbin, Rossetti, and Hartline 1986)

Sample	Maximum Flexure Strength (MPa)	Strain-to-Failure (%)	Work-of-Fracture ($\times 10^6$ erg/cm ²)
RBSN ^(a)	325	0.18	4.62
RBSN ^(a) + SiC fiber with 1 μm C-coat	479	2.2	42.9
RBSN ^(a) + SiC fiber with 3 μm C-coat	755	over 3.0	94.5

(a) Reaction-bonded Si_3N_4 .

cracks propagated through the fibers, resulting in lower flexural strength and lower strain-to-failure. Fracture properties were not measured above room temperature, and it is expected that elevated temperatures may result in C-layer oxidation accompanied by degradation of mechanical properties.

Corbin, Rossetti, and Hartline (1986) also reported a "two-layer structure" existing at fiber-matrix interfaces in nitrided samples containing the thick-coated fibers. This observation agrees with the surface structures observed on AVCO type SCS-2 fibers by Marzik (1984); Wawner, Teng, and Nutt (1983a, 1983b); and Nutt and Wawner (1982). The inner layer consisted of amorphous C and was approximately 0.5 μm thick, and the outer layer contained beta SiC crystallites dispersed in C. The thick coatings on the SCS-6 fibers are reported to have a similar morphology (Marzik 1984). Scanning Auger microanalysis of fracture surfaces showed that the thick-coated samples exhibited C layers bonded to the fibers and matrix troughs (Corbin, Rossetti, and Hartline 1986). In contrast, the thin-coated samples did not exhibit a C layer on fiber surfaces or in matrix troughs, and it was believed that the C was consumed by the $\text{Si} + \text{C} \rightarrow \text{SiC}$ reaction during nitriding at 1380°C for 48 hours. Matrix troughs in the thin-coated samples also exhibited "small black deposits" that were believed to be SiC crystallites, based on knowledge of the SiC/amorphous C surfaces of AVCO fibers (Marzik 1984; Wawner, Teng, and Nutt 1983a, 1983b; Nutt and Wawner 1982). However, the composition of these deposits was not reported.

It may be possible to alleviate degradation of the C layer during nitridation of these samples by reducing the maximum time and temperature during nitridation. One possible option is to use the methods described by Lundberg et al. (1987); Pompe, Hermansson, and Carlsson (1982); and Falk, Pompe, and Dunlap (1985), which entail consolidating powders of Si, Si_3N_4 , Al_2O_3 , and Y_2O_3 , followed by nitriding at 1350°C for 1 hour before hot isostatic pressing. Another option entails use of high-purity, laser-synthesized, ultra-fine Si powder, which has been reported to nitride to 99% completion (with no Fe additive) within 1 hour and at a range of temperatures between 1150°C and 1350°C (Sheldon and Haggerty 1988; Haggerty et al. 1988). A significant reduction of nitridation time and temperature was cited, entailing greater than 97% nitridation within 10 minutes at 1250°C (Sheldon and Haggerty 1988).

Shetty et al. (1985) evaluated room temperature mechanical properties of Si_3N_4 reinforced with AVCO type SCS-2 fibers. The samples were prepared by 1) slurry coating the fibers with a mixture of Si_3N_4 , Y_2O_3 , and Al_2O_3 powder; 2) cold pressing preforms of laminated sheets of unidirectionally aligned fibers; and 3) binder removal followed by hot pressing at 1750°C . Nonlinear stress-strain behavior was exhibited in the fiber-reinforced Si_3N_4 , as opposed to unreinforced Si_3N_4 . The authors reported that "The SiC filaments were compatible with the Si_3N_4 matrix, and there was no evidence of excessive fiber-matrix interaction" based on scanning electron microscope (SEM) fractography and optical ceramography. It was also reported that the average crack growth resistance for the fiber-reinforced composites (7300 to 8150 J/m^2) was nearly two orders of magnitude greater than the critical strain energy release rate for unreinforced Si_3N_4 (82.5 J/m^2), suggesting that large contributions from inelastic energy dissipation processes were operative, such as fiber pullout and debonding. It is possible that toughening behavior was affected by the C layer, which is known to be present on the fiber surfaces in the as-received condition. Based on the reported data, it is uncertain to what extent the C layers underwent microchemical changes during the processing of the composites.

The fiber-reinforced composites prepared by Shetty et al. (1985) exhibited poor fracture strengths as a result of cracks in the fibers that

may have been introduced during sample preparation. Several reasons for the cracks were suggested, including thermal expansion stresses from hot pressing, oxidation during hot pressing, and thermal degradation of fiber tensile strength. In the latter case, thermally induced degradation of mechanical properties was reported in AVCO fibers above 900°C (McHenry and Tressler 1975) and at room temperature after high temperature exposure (Ahmad, Hill, and Heffernan 1976).

Bhatt (1985 and 1986) also reported enhanced toughening in reaction-bonded Si_3N_4 that was reinforced with AVCO type SCS-2 fibers. It was reported that the C coatings on the fibers resulted in weak bonding between the fibers and matrices, causing nonlinear stress-strain behavior and crack deflection along fiber-matrix interfaces. Microstructure examination indicated little evidence of chemical reaction between the fiber and matrix. Bhatt mentions that use of these fibers is attractive from the standpoint of being able to specify various coatings by chemical vapor deposition. Detailed analyses of interface microstructures and phase equilibria were not reported.

Effects of thermal expansion mismatch on interfacial shear stresses at fiber-matrix interfaces were evaluated by Brun and Singh (1987) in samples comprising various matrices that were reinforced with AVCO type SCS-6 fibers. Matrices including silica glass, mullite, and hot-pressed Si_3N_4 were selected having different thermal expansion coefficients relative to the fibers, and fiber push-out tests were performed to determine interfacial shear stresses. The interfacial shear stress was reported to increase for samples containing matrices with larger expansion coefficients relative to the fiber. In the case of the Si_3N_4 , which has a slightly higher thermal expansion coefficient than the AVCO fiber, a measurable interfacial shear stress of 12 to 16 MPa was observed. A "two-step" push-out test was reported, which enabled separation of shear stress components attributed to both chemical and frictional bonding. Si_3N_4 -matrix composites exhibited 16 MPa shear stresses on the first push, which was a measure of chemical plus frictional bonding. The second push, which was exerted on the same fiber but in the opposite direction, indicated a shear stress of 12 MPa, which was attributed to frictional bonding alone. Mullite matrices, which have a higher degree of

thermal expansion mismatch relative to the fiber, exhibited somewhat higher shear stresses (30 to 85 MPa) than the Si_3N_4 -matrix samples. Fracture surface morphologies for all samples were evaluated, and higher degrees of pullout were observed for samples with the low interfacial shear stresses.

Similar push-out tests were performed to evaluate interfacial shear stresses in mullite-matrix composites containing BN-coated AVCO fibers by Brun and Singh (1987) and Singh and Brun (1987 and 1988). The uncoated fibers were tightly bound to the mullite matrix, as they could not be displaced at stresses above 200 MPa. In contrast, the BN-coated fibers could be pushed with an interfacial shear stress of approximately 85 MPa. Significant effects of coating thickness on interfacial stress were observed by Brun and Singh (1987), as BN coatings of increasing thickness resulted in a reduction of shear stress from 85 to 50 MPa during the first pushout. In addition, the difference between the interface stresses for the two thicknesses of the BN coating in the mullite-matrix composites was small (≈ 5 MPa) on the second push. The authors mention that it may be possible that slippage occurs between the BN layer and the matrix, so that coating thickness has a small effect on the frictional stress. Experiments with BN-coated fibers in Si_3N_4 matrices were not reported by Brun and Singh (1987) and Singh and Brun (1987 and 1988), although similar effects of coating thickness on interfacial shear stress may be possible. Laughner et al. (1986) also reported push-out tests which were performed on several varieties of SiC-fiber-reinforced Si_3N_4 . Unfortunately, Laughner et al. (1986) presented only a short abstract that did not include mechanical properties data.

Interface-controlled toughening in reaction-bonded Si_3N_4 that was reinforced with four different types of SiC-based fibers, Nicalon, MPDZ, HPZ, and TYRANNO,^(a) was reported by Corbin, Willkens, and Hartline (1987a). The authors reported that "Nitridation parameters can be controlled such that a useful interface will form in situ during matrix processing. Composites which do not fail in a brittle manner and can carry loads after matrix

(a) MPDZ and HPZ are designations for experimental fibers produced by Dow Corning. TYRANNO fibers are manufactured by Ube Industries, Ltd., Japan.

microcracking have been demonstrated. Due to the different composition and thermal stability of the fibers evaluated, each requires a different nitridation scheme to produce composites with desirable properties." Experiments were primarily focused on evaluating how room temperature fracture behavior was affected as a result of the time-dependent formation of a reaction zone at fiber-matrix interfaces during nitriding. Reaction zones were produced in samples using 8 or 48 hours of nitridation at 1350°C.

The Nicalon-reinforced samples that were nitrided for 48 hours exhibited nonlinear stress-strain behavior and noncatastrophic failure. In comparison, the same material that was nitrided for 8 hours exhibited linear stress-strain behavior and catastrophic, brittle failure, with cracks propagating through the fibers. The 48-hour nitridation caused the Nicalon fibers to become noncircular in cross section with the interface reaction zones growing to approximately 2 μm in size. The interface reaction zone that developed in the Nicalon fibers over the 48-hour nitridation provided a crack deflection/debonding zone, causing matrix cracks to propagate along the fiber length rather than penetrating the fibers. The 48-hour nitridation of the Nicalon-containing samples also resulted in a trace amount of recrystallized beta SiC, based on x-ray diffraction analysis. In contrast, SiC was not detected with the same material nitrided for only 8 hours. In addition, scanning Auger microanalysis of fractured samples indicated that the 48-hour nitridation resulted in the fiber surfaces exhibiting less O and a 1:1 atomic ratio between Si and C, suggesting stoichiometric SiC. Also, the surfaces of the matrix troughs in the same samples indicated C and low O concentrations. Apparently, two grades of Nicalon fiber were used: ceramic grade and nonceramic grade. The authors indicated a smaller interface reaction zone was formed during nitridation with the ceramic-grade Nicalon fiber.

Based on the data presented in this report, it is uncertain which chemical equilibria and chemical transport processes affected the formation of interfacial reaction zones during nitridation of Nicalon-reinforced Si_3N_4 . Possible reactions are likely to involve elemental Si in the matrix, N gas in the nitriding furnace, and the Si-C-O-N glass, which is known to co-exist

with beta-SiC crystallites, in the as-received Nicalon fibers.^(a) The authors reference the earlier work of Lucek, Rossetti, and Hartline (1985), in which the stability of Nicalon fibers had been evaluated under various RBSN processing conditions. Additionally, several references exist pertaining to thermal degradation of the Nicalon fibers, resulting in evolution of SiO and CO gases.^(b) It is plausible that these gases may react with elemental Si or N gas during the nitridation process.

Corbin, Willkens, and Hartline (1987a) reported similar behavior for MPZD-fiber-reinforced Si_3N_4 . The 48-hour nitridation resulted in an interfacial reaction zone which enhanced tensile strength and toughness by way of crack deflection and fiber-matrix debonding. Fracture surfaces of this sample were analyzed by scanning Auger microanalysis, and a C layer was detected at the surfaces of pulled out fibers and matrix troughs. In contrast, no reaction zone was found in a similar sample which underwent the 8-hour nitridation. Analyses were not reported pertaining to chemical equilibria at the interface.

Enlarged reaction zones were formed by the longer nitriding times, as opposed to shorter nitriding times, in samples that were reinforced with HPZ and TYRANNO fibers (Corbin, Willkens, and Hartline 1987a). The HPZ-reinforced sample exhibited an interfacial gap of 0.5 and 3 μm after 8 and 48 hours of nitriding, respectively. In both samples, the HPZ fibers were reduced in diameter, indicating fiber degradation. The TYRANNO-reinforced material, although exhibiting an interface reaction layer after 48 hours of nitridation, exhibited brittle and catastrophic fracture behavior. Analyses were not reported pertaining to transport processes and chemical equilibria at the interface.

In a companion paper, Corbin, Willkens, and Hartline (1987b) evaluated room temperature mechanical properties of reaction-bonded Si_3N_4 -matrix composites that were reinforced with Nicalon fibers having various chemically

(a) Andersson and Warren 1984; Chaim, Heuer, and Chen 1988; Jones et al. 1988; Schoenlein et al. 1988.

(b) Luthra 1986; Simon and Bunsell 1984; Clark et al. 1985; Mah et al. 1984; Johnson et al. 1988.

vapor-deposited coatings. Coatings of SiC, Si₃N₄, and Al₂O₃ did not enhance toughening or strength, as these coatings were strongly bonded to the Si₃N₄ matrix and brittle behavior was observed. In contrast, toughening behavior was exhibited in samples containing C-coated fibers, as cracks propagated along the coating-matrix interface, and the coatings adhered to the fibers after fracture. With exception to the C and Al₂O₃ coatings, the remaining coatings resulted in composite tensile strengths that were lower than unreinforced Si₃N₄ prepared under similar conditions. In addition, all of the above coatings resulted in slightly decreased tensile strengths of the fibers; tensile strengths were approximately 2.3 and 2.0 GPa for uncoated and coated fibers, respectively. Detailed analyses were not reported pertaining to chemical equilibria at the interface.

Bender et al. (1978) evaluated room temperature mechanical properties of Nicalon-reinforced Si₃N₄, which was prepared by hot pressing with MgO sintering aid. It was reported that all samples exhibited strong bonding at fiber-matrix interfaces, although no evidence of fiber degradation by interaction with the matrix or densification aid was observed. The Nicalon fibers exhibited grain growth after hot pressing, and the fracture origins appeared to be at regions where fiber clusters were sintered together. Detailed analyses were not reported pertaining to chemical equilibria at the interface. Bender et al. (1978) also presented a footnote estimating tensile-compressive stresses existing in SiC-fiber-reinforced Si₃N₄ as a result of thermal expansion differences between SiC and Si₃N₄: "Depending on the degree of 'plastic' relaxation in the Si₃N₄ matrix during cooling, the fibers would be in 100,000 to 200,000 psi tension giving approximately 30,000 to 60,000 psi compression (for approximately 25 vol% fibers). If fiber stresses are too high, lower temperature processing, e.g. by CVD techniques, may be more desirable."

SiC-Whisker-Reinforced Si₃N₄

Room temperature mechanical properties and microstructures of whisker-matrix interfaces were correlated by Jones et al. (1988) and Schoenlein et

al. (1988) for reaction-bonded Si_3N_4 reinforced with VLS whiskers.^(a) The samples were prepared with various coatings applied to the whiskers, including SiO_2 , BN, and C. Alterations to processing parameters resulted in significant changes to the microstructures of whisker-matrix interfaces, based on analyses by TEM with x-ray energy dispersive spectroscopy and Auger electron spectroscopy. Composites which were reaction bonded with an Fe_2O_3 additive exhibited faceted whisker surfaces as a result of a reactive Fe silicide phase, despite the C, BN, or SiO_2 coatings applied to the whiskers. When no sintering aid was used, the uncoated whiskers were not degraded and appeared to be strongly bonded to the matrix. Unfortunately, the samples containing no sintering aid possessed large areas of unreacted Si, which most likely acted as strength- and toughness-limiting defects. All whisker-reinforced samples exhibited reductions in flexure strength and toughness as compared to unreinforced Si_3N_4 , which had been prepared under similar conditions. Reduced flexure strengths and toughnesses of the whisker-reinforced composites containing the Fe_2O_3 additive were also attributed to whisker degradation and lower toughness matrices as a result of a lesser alpha-to-beta- Si_3N_4 ratio. The composites reinforced with SiO_2 -coated whiskers exhibited the highest fracture strength and toughness, whereas the composites with BN-coated whiskers exhibited the lowest strength and toughness, respectively.

Room temperature mechanical properties of VLS-whisker-reinforced Si_3N_4 , which was prepared by hot pressing with MgO sintering aid, were also evaluated by Shalek et al. (1987). Slightly nonlinear stress-strain behavior was observed in whisker-reinforced samples with little evidence of pullout. Nonetheless, a maximum toughness of $12.5 \text{ MPa}\cdot\text{m}^{1/2}$ was observed in one sample containing 10 vol% whiskers. The authors reported that whisker-matrix interfaces were strongly bonded and contained a Mg-based glass phase. Large processing-related flaws served as fracture origins and may have overshadowed effects of interface-controlled toughening. Atomistic analyses of

(a) VLS whiskers are produced by Los Alamos National Laboratory using the vapor-liquid-solid process (Milewski et al. 1985; Shalek et al. 1987).

whisker-matrix interfaces in these samples were reported by Nutt and Phillips (1986) and Davis et al. (1987) using high-resolution TEM.

Nutt and Phillips (1986) and Davis et al. (1987) reported three significant features of whisker-matrix interfaces in VLS-whisker-reinforced Si_3N_4 : 1) the VLS whiskers exhibited rounded triangular cross sections with {112} faces adjacent to the Si_3N_4 matrix; 2) whisker-interface planes were composed of {111}-type microfacets arranged in step-ledge structures, which tended to be configured in spirals around the whisker circumference; and 3) whisker-matrix interfaces contained an amorphous phase which sometimes devitrified to form a Mg-sialon phase that was epitaxial on the SiC whisker. Another significant observation was that whisker degradation occurred after hot pressing at 1850°C. Chemical reactions responsible for whisker degradation were not analyzed. Whisker degradation was not observed in samples that were hot pressed at a lower temperature. Nutt and Phillips (1986) reported that the effects of interface microstructure on mechanical properties were not clear and that higher toughnesses by way of a weaker whisker-matrix bond may be possible by increasing the thickness of the glass layer.

Gibbs, Petrovic, and Honnell (1987) analyzed the mechanical properties of hot-pressed VLS-whisker-reinforced Si_3N_4 containing MgO as sintering aid. Flexure strengths were reported for reinforced and unreinforced samples as a function of temperature up to 1500°C. Unreinforced samples retained their strength of approximately 650 MPa from room temperature to about 1200°C, whereas a strength of only 50 MPa resulted at temperatures above 1300°C. A whisker-reinforced sample exhibited a room temperature strength of 520 MPa, which decreased to 200 and 100 MPa at 1300 and 1500°C, respectively. Room temperature toughnesses of 7 and 10 $\text{MPa}\cdot\text{m}^{1/2}$ were exhibited by unreinforced and reinforced samples, respectively, whereas, at 1200°C, both reinforced and unreinforced samples exhibited a toughness of 15 $\text{MPa}\cdot\text{m}^{1/2}$. Additionally, a toughness of 7 $\text{MPa}\cdot\text{m}^{1/2}$ was exhibited by reinforced and unreinforced samples at 1400°C. Whisker pullout was occurring at elevated temperatures, based on fractography analysis. Fracture surfaces generated at room temperature showed evidence of pullout, whisker-matrix debonding, and crack bridging. The whiskers also did not exhibit degradation after processing.

Willkens and Corbin (1987) prepared Tateho-whisker-reinforced Si_3N_4 ^(a) by reaction bonding followed by glass encapsulation and hot isostatic pressing. Various coatings were applied to the whiskers by chemical vapor deposition, and effects on fracture toughness were compared at room temperature using indentation experiments. The coatings included a BN-based coating and 4 types of "dual layer" coatings, which were designed to have a debonding/parting layer (BN- or C-based coatings) protected by an overcoating layer (SiC- or Si_3N_4 -based coatings). The dual layer coatings included SiC/BN, Si_3N_4 /BN, Si_3N_4 /C, and SiC/C; these coatings were nonstoichiometric. The highest room temperature toughness of $4.4 \text{ MPa}\cdot\text{m}^{1/2}$ was exhibited by the sample containing whiskers coated with a single layer of BN. Analysis of this sample by TEM indicated that the BN dissolved into the matrix. The SiC/BN samples and the Si_3N_4 /BN samples exhibited somewhat lower toughnesses of approximately 3.9 and $3.2 \text{ MPa}\cdot\text{m}^{1/2}$, respectively. Samples with uncoated whiskers and no whiskers exhibited slightly lower toughnesses of approximately 3.0 and $3.1 \text{ MPa}\cdot\text{m}^{1/2}$, respectively. The lowest toughnesses of approximately 2.9 and $2.7 \text{ MPa}\cdot\text{m}^{1/2}$ were exhibited by samples with Si_3N_4 /C and SiC/C coatings, respectively. Aside from the TEM analysis mentioned above, detailed analyses of interfaces in the remaining samples were not reported.

Effects of different whisker types on fracture toughnesses of reaction-bonded Si_3N_4 -matrix composites were also evaluated by Willkens and Corbin (1987). Based on fracture toughness measurements, the most promising fabrication approaches entailed use of Huber whiskers^(b) and two types of Tateho whiskers (SCW-1 and T-44). The room temperature fracture toughnesses of samples containing these whiskers were approximately equal (4.9 to $5.0 \text{ MPa}\cdot\text{m}^{1/2}$). Flexure strength was highest for the Huber-reinforced material (823 MPa), and this was attributed to a sedimentation process used for removing SiC particulates from the whisker batch during sample preparation. Detailed analyses of whisker-matrix interfaces were not reported for any of these samples. In the same report, Willkens and Corbin

(a) Tateho whiskers are manufactured by Tateho Chemical Industries, Ltd., Japan.

(b) Huber whiskers are manufactured by J. M. Huber Corp., Langley, South Carolina.

(1987) evaluated several types of as-received whiskers for physical dimensions, shape uniformity, and concentrations of O, N, and Ca.

Willkens and Corbin (1987) also measured fracture toughnesses in extruded samples and observed fracture toughness anisotropy as a result of preferred orientation of whiskers. The highest toughnesses were observed where cracks propagated perpendicular to the whisker axes. The effect was not observed in isopressed samples where whiskers were randomly oriented. Again, detailed analyses of whisker-matrix interfaces were not reported.

Mechanical properties of ACM-whisker-reinforced Si_3N_4 ,^(a) which was hot pressed with Y_2O_3 and Al_2O_3 sintering aids, were evaluated by Buljian and co-workers.^(b) Major results were that increases in fracture toughness resulted from 1) increasing the amount of SiC whisker, 2) increasing the temperature at fracture from 25°C to 1200°C, and 3) prolonging the time-at-maximum-temperature during hot pressing to enlarge Si_3N_4 grain sizes and thereby increase crack deflection in the matrix. Fractography indicated limited amounts of whisker pullout, which suggests that the whiskers were strongly bonded to the matrices. Analyses by TEM revealed no evidence of whisker degradation or reaction phases at whisker-matrix interfaces (Buljian and Zilberstein 1987).

Fracture toughness anisotropy of SiC-whisker-reinforced Si_3N_4 was also reported as a result of preferred orientation of whiskers during hot pressing (Buljian, Baldoni, and Huckabee 1987; Buljian et al. 1987). Preferred orientation of SiC whiskers was also reported in injection-molded samples of whisker-reinforced Si_3N_4 (Buljian and Baldoni 1988). Changes in the geometry of the injection mold apparatus resulted in changes in preferred orientation of whiskers, which ultimately affected shrinkage anisotropy during densification by hot isostatic pressing. Analyses of whisker-matrix interfaces were not reported in these injection-molded samples.

(a) ACM whiskers are manufactured by Advanced Composite Materials Corp., Greer, South Carolina.

(b) Buljian 1987; Buljian et al. 1987 and 1988; Buljian, Baldoni, and Huckabee 1987; Buljian and Baldoni 1988; Buljian and Zilberstein 1987.

The oxidation behavior of ACM whiskers, ACM-whisker-reinforced Si_3N_4 , and unreinforced Si_3N_4 was evaluated by Buljian et al. (1988). Although microstructure analyses of whisker-matrix interfaces were not reported, several results of these experiments are noteworthy. X-ray diffraction analyses were performed on a batch of as-received whiskers which had been heated to 1200°C in air, and results indicated formation of silica on whisker surfaces accompanied by an increase in the amount of beta SiC, as opposed to alpha SiC, with increased oxidation time. Scanning Auger microanalyses were performed on oxidized surfaces of reinforced and unreinforced Si_3N_4 , and results indicated the presence of a silica layer in both samples after 1200°C heating in air. This layer contained C and Ca only in the whisker-reinforced samples; the C and Ca originated as impurities in the as-received whiskers. A significant result was that the oxidation behavior was nearly identical for unreinforced and reinforced samples, based on analyses of weight gain as a function of oxidation time. The rupture moduli of oxidized composites were also measured at room temperature and 1000°C, and results were nearly identical to that of an unoxidized composite evaluated at room temperature. In addition, the rupture modulus of an oxidized composite measured at 1200°C was approximately 88% of that of an unoxidized composite evaluated at room temperature. Buljian et al. (1988) also reported measurements of several physical properties of whisker-reinforced and unreinforced Si_3N_4 , including thermal conductivity, thermal expansion, Young's modulus, Poisson's ratio, bulk modulus, shear modulus, and Knoop hardness.

High temperature deformation of SiC-whisker-reinforced Si_3N_4 was evaluated by Nixon et al. (1987). Samples were prepared with $\text{Y}_2\text{O}_3\text{-Al}_2\text{O}_3$ sintering aids according to procedures reported by Buljian et al. (a). Samples were evaluated by steady-state creep tests, and results were correlated with TEM studies to evaluate mechanisms responsible for deformation. Nixon et al. (1987) reported that the temperature as well as the crystallization of the amorphous grain boundary phase in the Si_3N_4 matrix were principal factors controlling high temperature deformation behavior. Creep rates in the

(a) Buljian et al. 1987 and 1988.

whisker-containing samples exceeded those of the unreinforced samples. These results were attributed to increased amounts of devitrified intergranular phase in the whisker-reinforced composites, based on analyses by TEM. The authors suggested that diffusion of glass modifiers (Ca, Mn, Fe, and Mg) from the whiskers and into the matrix may have reduced the tendency of the intergranular glass phase to crystallize, thus promoting creep. The TEM analyses of creep-tested samples indicated that the glass continued to devitrify with time. Results of the same experiments are also mentioned by Buljian et al. (1988).

Processing and creep behavior of SiC-whisker-reinforced Si_3N_4 were also analyzed by Porter, Lange, and Chokshi (1987). Processing entailed mixing slurries of SiC whiskers and Si_3N_4 with solutions containing Al- and Y-based dopants as dissolved nitrates and subsequently increasing the pH to precipitate the additive hydroxides. Sedimentation procedures were used to eliminate agglomerated Si_3N_4 powders and SiC whiskers; the processing method resulted in hot-pressed samples exhibiting one of the highest room temperature flexure strengths reported to date for SiC-filament-reinforced Si_3N_4 (1250 MPa). Whisker reinforcement was shown to increase room temperature flexural strength and fracture toughness. Scanning electron microscopy revealed little evidence of whisker pullout on fracture surfaces, suggesting that the whiskers were strongly bonded to the Si_3N_4 matrix. Additional microstructure analyses at whisker-matrix interfaces were not reported. High temperature creep performance was unaffected by whisker reinforcement; several reasons for this behavior were suggested, including impurities in the whiskers, reducing the eutectic melting point of the grain boundary phase, and evidence of more rapid grain growth in the unreinforced composites, with the creep rate being primarily affected by Si_3N_4 grain size.

Haynes, Martin, and Yeh (1987) reported processing and room temperature mechanical properties of SiC-whisker-reinforced Si_3N_4 . Extensive degradation of Tateho whiskers, as opposed to Arco whiskers,^(a) was observed in

(a) Arco whiskers were manufactured by Atlantic Richfield Corp. The manufacturing facility is currently operated by Advanced Composite Materials Corp., Greer, South Carolina.

composites as a result of reactions with N gas during a presintering process. This was verified in separate experiments where Tateho whiskers were heated in N to 1500°C for 2 hours, after which x-ray diffraction indicated the formation of extensive amounts of Si_3N_4 . In contrast, the same experiment was performed with Arco whiskers, and negligible amounts of Si_3N_4 were found. Analysis by scanning TEM was performed on the composite containing Arco whiskers, and an amorphous film was detected at whisker-matrix interfaces. This film was likely to have resulted in strong whisker-matrix bonds, as pullout was not observed on fracture surfaces of the Arco-reinforced Si_3N_4 . Haynes, Martin, and Yeh (1987) and Yeh et al. (1988) also presented scanning TEM analyses of several types of commercially available whiskers, and the authors discussed morphological features of each whisker type. Tateho whiskers were shown to be hollow in one micrograph.

Processing and mechanical properties characterization of SiC-whisker-reinforced Si_3N_4 were reported by Haynes and Yeh (1985). Samples were prepared by slip casting and hot isostatic pressing methods, and fracture toughnesses as high as $10.98 \text{ MPa}\cdot\text{m}^{1/2}$ were reported along with evidence of whisker pullout on fracture surfaces. Analyses of whisker-matrix interfaces were not reported. The authors reported results of slip-casting experiments using suspensions of Si_3N_4 powder mixed with Al_2O_3 , Y_2O_3 , and several types of SiC whiskers including coated Arco type SC-9 whiskers. The coating method and material were not specified, although it was mentioned that the source of the coatings was the Naval Research Center. Unfortunately, difficulties in slip preparation resulted from use of these coated whiskers, and test samples were not prepared for microstructure analyses or mechanical properties evaluation.

Computer-assisted analyses of phase equilibria were reported by Nickel et al. (1987) and Yeh et al. (1988) to determine a thermodynamic stability regime for SiC-whisker-reinforced Si_3N_4 as functions of temperature and N partial pressure. Above the stability regime, excess temperatures and N partial pressures were predicted to degrade whiskers according to the nitridation reaction $3\text{SiC} + 2\text{N}_2 \rightarrow \text{Si}_3\text{N}_4 + 3\text{C}$. Below the stability regime, lower N partial pressures were predicted to degrade the Si_3N_4 matrix according to the reaction $\text{Si}_3\text{N}_4 \rightarrow 3\text{Si} + 2\text{N}_2$. Experiments were presented by

Nickel et al. (1987) showing whisker degradation as a result of hot isostatic pressing at conditions above the thermodynamic stability regime. Similar experimental results were presented by Yeh et al. (1988) showing that nitridation slightly above the stability regime may be catalyzed by small concentrations of Fe on surfaces of as-received Tateho whiskers. In comparison, minimal evidence of nitridation was shown in samples which were processed under similar conditions containing Fe-free Arco whiskers. Yeh et al. (1988) also reported extensive analyses of the processing and mechanical properties characterization of SiC-whisker-reinforced Si_3N_4 prepared by slip cast and hot isostatic pressing.

Extensive evaluation of the mechanical properties of Tokai-whisker-reinforced Si_3N_4 ^(a) at temperatures between room temperature and 1300°C was reported by Hayami et al. (1986) and Ueno and Toibana (1983). Samples were prepared by hot pressing using two types of sintering aids: $\text{Y}_2\text{O}_3\text{-Al}_2\text{O}_3$, and $\text{Y}_2\text{O}_3\text{-La}_2\text{O}_3$. The Al_2O_3 -based sintering aid was found to degrade the whiskers, whereas, the La_2O_3 -based sintering aid prevented whisker attack based on analysis of whisker-matrix interfaces by SEM. No further evaluation of whisker-matrix interfaces was reported. Fracture energies of whisker-reinforced samples were reported to exceed those of unreinforced Si_3N_4 by approximately 30%. Whisker-reinforced samples exhibited impressive tensile strengths of 600 and 650 MPa at room temperature and 1300°C, respectively. The authors also presented experiments indicating that both room temperature tensile strength and Weibull modulus increase when large SiC particles and agglomerates were removed from the as-received whiskers by a filtration process. Based on the superior fracture properties reported by these authors, it is possible that similar fabrication methods may result in interface-enhanced toughening by use of coated filaments.

Lundberg et al. (1987) evaluated Tokai-whisker-reinforced Si_3N_4 , which was prepared by a nitridation of bodies containing a mixture of Si, Si_3N_4 , and whiskers. Pullout was not observed in samples which were fractured at room temperature. The authors mentioned that whiskers were strongly bonded to the matrix. Fracture origins were determined to be Co and Fe inclusions

(a) Tokai whiskers are manufactured by Tokai Carbon Co., Japan.

which originated as impurities in the Tokai whiskers. Detailed analyses of interface microstructures and phase equilibria were not reported in this study. In an earlier study, Lundberg et al. (1986) also reported SEM of SiC-whisker-reinforced Si_3N_4 which was prepared by nitridation of bodies containing a mixture of Si, Si_3N_4 , and whiskers, followed by hot isostatic pressing. Based on this study, the authors reported that there was no evidence of chemical reactions at fiber-matrix interfaces after the nitridation step or after the hot isostatic pressing.

Room temperature mechanical properties were correlated with microstructures of whisker-matrix interfaces by Kandori et al. (1987). Samples of Tateho-whisker-reinforced Si_3N_4 containing Y_2O_3 and MgAl_2O_4 sintering aids were prepared by injection molding followed by hot isostatic pressing; fracture behavior of reinforced and unreinforced samples was correlated with analyses of whisker-matrix interfaces by fractography and TEM. Fracture surfaces did not show evidence of whisker pullout, indicating that the whiskers and matrices were strongly bonded. Before the hot isostatically pressed treatment, voids in samples were preferentially segregated at whisker-matrix interfaces, based on TEM analyses reported earlier by the same authors (Kobayashi, Kandori, and Wada 1986). After the hot isostatically pressed treatment, TEM analyses indicated that the voids were eliminated and that a 1-nm-thick amorphous layer existed at whisker-matrix interfaces with no evidence of a reaction layer.

The authors reported that the good bonding provided by the amorphous layer at whisker-matrix interfaces enhances load transfer but retards pullout upon fracture (Kandori et al. 1987). Whisker-reinforced samples exhibited relatively high toughnesses of 9 to 10 $\text{MPa}\cdot\text{m}^{1/2}$, despite the strong whisker-matrix bonding and the lack of evidence for pullout. The fracture surfaces of whisker-reinforced composites were rougher than those of unreinforced samples, suggesting that crack deflection and/or crack branching existed due to the presence of whiskers. The reinforced samples also possessed relatively high flexure strengths of 800 to 900 MPa. In addition, improvements in fracture toughness, fracture energy, and flexural strength were exhibited by reinforced samples, as opposed to unreinforced samples.

REVIEW OF FILAMENT COATING TECHNIQUES

Techniques for coating continuous ceramic fibers were recently reviewed by Cranmer (1988). These techniques include chemical vapor deposition, organometallic precursor deposition, polymer precursor deposition, electrodeposition, and electroless plating. Cranmer (1988) evaluated the advantages and disadvantages of each of these coating technologies and stated that major disadvantages to many of the techniques entail deficiencies in uniformly and reproducibly coating the fibers. Brief comments on these techniques are made in the following paragraphs.

Chemical vapor deposition has been most commonly used for coating continuous ceramic fibers and is considered advantageous from the standpoint of producing a large number of ceramic and metal compositions.^(a) The chemical vapor deposition process is slow, although mass-production methods may be set up which may provide deposition of multiple coatings with similar or dissimilar compositions. For example, AVCO-SiC fibers are prepared by a chemical vapor deposition process which is capable of producing multiple coatings of several ceramic-based materials (Marzik 1984). Another concern is that coatings by chemical vapor deposition may reduce the strength of ceramic fibers (Langley and Li 1987; Langley, Li, and Rabe 1987; Corbin, Willkens, and Hartline 1987b).

Cranmer (1988) cites three references pertaining to organometallic precursor deposition (sol-gel) techniques: Cranmer and Speece (1986), Katzman (1983), and Hurwitz et al. (1987). The technique primarily entails dipping a filament into an alkoxide-based solution, followed by hydrolysis or ammonialysis of the solution to form oxide or nitride phases. This process is followed by higher temperature treatments to pyrolyze any remaining organics. Most of the coatings formed to date are simple oxides. Non-uniform thicknesses or inhomogeneities in composition may be formed in multicomponent systems, as different constituents may react or precipitate at different times (Cranmer 1988). Multiple dippings are required for coatings

(a) Cranmer 1988; Blocher and Withers 1970; Langley and Li 1987; Langley, Li, and Rabe 1987; Stinton, Besmann, and Lowden 1988.

in excess of 200 nm in thickness, and this process may be used to develop coatings with graded compositions. In addition, fiber coating techniques using organometallic precursors may be envisioned by use of several reported methods for fabricating thin films or monoliths.^(a)

Polymer precursor coatings are formed in a similar manner as the organometallic precursor coatings (Cranmer 1988), in that fibers are dipped into a polymer-based solution, followed by heat treatment to form ceramic coatings by pyrolysis. The technique has been used to form C coatings from pitch precursors (Katzman 1983) and SiC coatings from polycarbosilane (Katzman 1986). Similarly, several of the reported polymer-pyrolysis routes to forming monolithic ceramics could be modified to produce ceramic coatings on SiC-based fibers.^(b)

Electroless plating and electrodeposition are techniques which are used to produce metallic coatings. Both techniques require electrically conductive substrates for deposition to take place (Cranmer 1988). To the author's knowledge, it has not been reported whether these methods may be successfully applied to SiC-based filaments.

Continuous fibers may also be coated by slurry-dipping techniques, such as reported by Brennan (1986 and 1987) and Shetty et al. (1985) to produce glass-matrix composites and SiC-fiber-reinforced Si_3N_4 , respectively. These fabrication techniques primarily entail 1) dipping fibers into a colloidal suspension containing appropriate binders and plasticizers, 2) producing preforms of aligned or woven fibers, 3) heat treating the preforms to remove organics, and 4) densification by sintering. Using multiple dipping procedures, the slurry-dipping process may be envisioned to produce multiple layer coatings of similar or dissimilar compositions. In addition, several coating materials are possible with this technique, as colloidal suspensions for the dipping process may be prepared with a wide variety of ceramic and metal particulates using appropriate solution chemistry approaches. Several methods may then be envisioned for fabricating Si_3N_4 -matrix composites with

(a) Mazdidasni and Cooke 1973; Hensch and Ulrich 1984 and 1986.

(b) Rice 1983; Walker et al. 1983; Bender, Rice, and Spann 1985; Jamet et al. 1984; Hensch and Ulrich 1984 and 1986.

coated SiC fibers, including chemical vapor infiltration (Stinton, Besmann, and Lowden 1988), injection molding (Batha and Corbett 1988), slurry impregnation by slip casting (Singh 1988a), impregnation by organometallic precursor solutions (Baumgartner 1978; Jamet et al. 1984; Mazdiyasni, West, and David 1978), and melt infiltration (Singh 1988a; Hillig 1987, 1988a, and 1988b). In the latter case, it may be feasible to infiltrate molten Si followed by solidification and nitridation.

Reports pertaining to whisker-coating technology are not so prevalent in the literature. Chemical vapor deposition has been used to produce ceramic coatings on SiC whiskers in reports by Jones et al. (1988), Schoenlein et al. (1988), and Willkens and Corbin (1987). However, the latter reference cites difficulties with the chemical vapor deposition process based on a personal communication with D. Zacharich of Norton Co: "The coating of fine materials by chemical vapor deposition is very difficult due to the large surface area associated with these materials and mean free path considerations which can limit infiltration of gas species into whisker bundles." Based on the limited availability of information in the published literature, it is uncertain whether this process may be used to economically produce whisker coatings with uniform composition and thickness.

RECOMMENDATIONS

Several reports indicate that the room temperature fracture behavior of SiC-filament-reinforced Si_3N_4 may be improved by introducing coatings on the filaments prior to sintering. However, ideal coatings have not been reported that are capable of mechanical properties enhancement at both room and elevated temperatures. Coatings which are reported to enhance room temperature fracture behavior primarily consist of C- or BN-based phases. Based on reaction thermodynamic arguments, these phases are not expected to resist oxidation at elevated temperatures, and this is likely to result in degradation of mechanical properties at room and elevated temperatures.

There are several critical issues associated with selecting an interface modification technique which will enhance mechanical properties: 1) identifying chemical compatibility of the filament, matrix, and phases that are at the filament-matrix interface, 2) optimizing thermal and mechanical stresses at the interface after high-temperature processing, 3) identifying an interfacial phase that raises fracture toughness by assisting debonding and pullout, 4) identifying an interfacial phase that also possesses enough strength to provide load transfer and thereby increasing the fracture strength and creep resistance, 5) identifying an interface modification technique that does not degrade mechanical properties of the filament or matrix, 6) optimizing the processing of the matrix such that it exhibits resistance to creep and fracture, and 7) identifying a filament coating that is stable during processing and during usage in high temperature corrosive environments.

To establish chemically compatible filament/matrix/coating materials it is recommended to start with analyses of the reaction thermodynamics and kinetics of the entire composite system in the processing and use environments (Cranmer 1988). This rationale has been considered in the processing of filament-matrix interfaces in high toughness composites containing glass and glass-ceramic matrices.^(a) As indicated by Cranmer (1988), an evaluation

(a) Mendelson 1985; Benson, Spear, and Pantano 1987 and 1988; Spear, Benson, and Pantano 1988; Mah and Subramanian 1988.

of the reaction thermodynamics/kinetics of the entire composite system serves to answer three significant questions: 1) what are the chemical reactions in the filament/coating/matrix system which will give the desired interface, 2) what are the competing reactions which will impede or prevent formation of a desired interface, and 3) under which processing or service environments will the filament/matrix/coating maintain acceptable physical properties (Bhatt and Phillips 1988; Kriven et al. 1987)? A rigorous evaluation of these questions is possible using computer-assisted analyses of chemical thermodynamic databases. Such analyses have been reported by Nickel et al. (1987) and Yeh et al. (1988) to evaluate a thermodynamic stability regime for SiC-whisker-reinforced Si_3N_4 as functions of temperature and N partial pressure. Similar analyses that examine the sensitivity of additional parameters on phase equilibria in SiC-filament-reinforced Si_3N_4 are recommended. These parameters include impurities in raw materials, use of coated filaments, the amount and type of sintering aid, and the various combinations of gas composition, pressure, and temperature which are encountered during processing and usage of these composites.

Analyses of thermal and mechanical stresses at filament-matrix interfaces is possible using theoretical approaches^(a) and interfacial stress measurements by neutron diffraction (Kupperman, Majumdar, and Singh 1988), single fiber tensile tests (Goettler and Faber 1988; Desmukh, Freiman, and Cranmer 1988; Desmukh and Coyle 1988), and single fiber push-out tests using indentation techniques.^(b) For example, fiber push-out tests have been used to evaluate interfacial shear stresses as a function of fiber coating and thermal expansion differences between SiC-based fibers and several ceramic and glass matrices by Brun and Singh (1987) and Singh and Brun (1987 and 1988). Similar analyses are needed to evaluate SiC-filament-reinforced Si_3N_4

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- (a) Marshall, Cox, and Evans 1985; Faber and Evans 1983a, 1983b; Evans and McMeeking 1986; Budiansky, Hutchinson, and Evans 1986; Mandell and Jao 1988; Gupta et al. 1988; Chiang and Chou 1988; Evans 1988; Mikata and Taya 1985 and 1986.
 - (b) Marshall 1984 and 1987; Marshall and Evans 1985; Laughner et al. 1986; Brun and Singh 1987; Singh and Brun 1987 and 1988; Lowden and Stinton 1988 and 1987a, 1987b; Laughner et al. 1986; Singh 1988b; Weihs and Nix 1988.

as a function of additional parameters, such as coating composition and thickness. Studies are also needed to evaluate strength degradation to the filaments as a result of the coating formation process (Langley and Li 1987; Langley, Li, and Rabe, 1987; Corbin, Willkens, and Hartline 1987b). In addition, parametric studies are needed to analyze stress fields in and around coated filaments in Si_3N_4 -matrix composites as a function of filament geometry and the thermomechanical properties of the entire filament/coating/matrix system. Such analyses have been reported by Mikata and Taya (1985 and 1986) for aluminum-matrix composites reinforced with coated graphite filaments. Results indicate that stress fields within filament coatings are sensitive to the thickness, stiffness, and thermal expansion coefficient of the coating, and that high tensile stresses in the coating may lead to cracking of the coating layer.

After meeting all of the above criteria and identifying candidate coatings and composite processing conditions, it then remains to evaluate whether the interfacial properties, especially the strength of the filament-matrix bond, are suitable to enhance fracture toughness by way of crack deflection, crack bridging, and fiber pullout. A possible solution, as recommended by Caputo, Lackey, and Stinton (1985), may be to utilize a filament coating such as BN, which could tightly adhere to both the filament and matrix, but the BN itself would slip along basal planes.

Additionally, the composite processing scenario must ensure that the matrix has a minimal population of flaws, that the largest flaws are small, and that the matrix possesses maximum strength and creep resistance. In sintered Si_3N_4 materials, maximum strength and creep resistance are ensured by several methods including maximizing the beta-to-alpha Si_3N_4 ratio,^(a) optimizing the Si_3N_4 grain size (Buljian, Baldoni, and Huckabee 1987), selecting the appropriate sintering aid,^(b) and eliminating processing-related fracture

(a) Buljian 1987; Buljian, Baldoni, and Huckabee 1987; Corbin, Willkens, and Yeckley 1988; Lange 1979 and 1983.

(b) Lange 1983; Porter, Lange, and Chokshi 1987; Sheppard 1986; Edington, Rowcliffe, and Hensel 1975; Buljian et al. 1987 and 1988; Nixon et al. 1987; Kandori et al. 1987; Tanaka, Greil, and Petzow 1985; Iwasa and Watanabe 1987; Govila, Ezis, and Beckwith 1984.

origins by particulate size fractionation^(a) and hot isostatic pressing (Kandori et al. 1987; Kobayashi, Kandori, and Wada 1986; Tanaka, Greil, and Petzow 1985). For reaction-bonded Si_3N_4 materials, maximum strength and creep resistance are ensured by reducing the amount of porosity and unreacted Si, reducing the maximum nitridation temperature and the time-at-maximum-temperature, and by hot isostatic pressing after nitridation.^(b) Additionally, precautions should be made to appropriately select the sizes and amounts of the filaments relative to the sizes and amounts of the matrix powders.^(c)

Another concern is the elimination of defective SiC filaments used during processing. Several authors have characterized the physical and chemical properties of commercial SiC filaments.^(d) Yeh et al. (1988) report that control over the surface chemistry of SiC whiskers is important for minimizing whisker degradation during processing of Si_3N_4 -matrix composites; in one example, Fe contamination on Tateho whisker surfaces was shown to result in increased nitridation of the whiskers under the processing environment used in that study.

Anisotropy of mechanical properties is another concern resulting from preferred orientation of whiskers during processing. This phenomenon has been observed in SiC-filament-reinforced Si_3N_4 prepared by hot pressing (Buljian, Baldoni, and Huckabee 1987; Buljian et al. 1987) and injection molding (Buljian and Baldoni 1988; Kandori et al. 1987; Kobayashi, Kandori, and Wada 1986). Preferred orientation of SiC whiskers in slip cast Si_3N_4 -matrix composites has also been reported by Nagel et al. (1987), whereas Yeh et al. (1988) reported random orientation of SiC whiskers in similar slip

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- (a) Hayami et al. 1986; Ueno and Toibana 1983; Porter, Lange, and Chokshi 1987; Willkens and Corbin 1987.
 - (b) Mangels 1977; Mazdiasni, West, and David 1978; Willkens and Corbin 1987; Lundberg et al. 1986 and 1987; Baumgartner 1978; Birch, Wilshire, and Godfrey 1978; Sheldon and Haggerty 1988; Haggerty et al. 1988; Takao, Cannon, and Danforth 1986; Koczak and Barsoum 1988.
 - (c) Milewski 1986; Shalek et al. 1986; Starr 1986; Takao, Cannon, and Danforth 1986.
 - (d) Nutt 1984; Willkens and Corbin 1987; Haynes, Martin, and Yeh 1987; Yeh et al. 1988.

cast composites. Shaw and Faber (1987) reported a stress analysis predicting conditions for whisker pullout and whisker fracture based on evaluation of bending moments applied to whiskers as a function of whisker orientation, whisker strength, and interfacial shear strength.

Several authors report that fracture origins in SiC-filament-reinforced Si_3N_4 may be eliminated by processing modifications. These flaws primarily consist of impurities in raw materials and non-uniform packing of powders and whiskers. For example, Ueno and Toibana (1983) and Hayami et al. (1986) presented an extensive reduction of flaws which contributed to fracture using a simple filtration technique for removing large agglomerates of SiC whiskers; the process resulted in significant increases in both the Weibull modulus and tensile strength of SiC-whisker-reinforced Si_3N_4 . Hot isostatic pressing has also been utilized to reduce porosity and thus improve mechanical properties of SiC-filament-reinforced Si_3N_4 .^(a) These and all reasonable precautions should be made in order to ensure that processing-related flaws do not overshadow possible effects of properties enhancement which are attributed to the filament-matrix interface.

In terms of producing creep-resistant matrices, several reports indicate methods of processing Si_3N_4 -based matrices with a minimum amount of phases that are susceptible to thermally induced softening. These methods include use of reaction bonding (Lundberg et al. 1986 and 1987; Willkens and Corbin 1987), yttria-based sintering aids,^(b) scandia-based sintering aids (Mecartney 1987), sintering aids based on other rare earth oxides (Hirosaki, Okada, and Matoba 1988; Hayami et al. 1986; Ueno and Toibana 1983), and development of sialon-based matrices.^(c)

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- (a) Lundberg et al. 1986 and 1987; Kandori et al. 1987; Willkens and Corbin 1987; Greil, Petzow, and Tanaka 1987.
(b) Lewis, Leng-Ward, and Jasper 1988; Lange 1983; Porter, Lange, and Chokshi 1987; Buljian et al. 1987 and 1988; Nixon et al. 1987; Corbin, Willkens, and Yeckley 1988; Govila, Ezis, and Beckwith 1984.
(c) Lewis, Leng-Ward, and Jasper 1988; McMurtry, Boecker, and Lau 1988; Lumby 1982; Sheppard 1986; Loehman 1982; Lange 1983.

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APPENDIX

MECHANICAL PROPERTIES OF SiC-FILAMENT-REINFORCED Si_3N_4

APPENDIX: MECHANICAL PROPERTIES OF SiC-FILAMENT-REINFORCED Si_3N_4

This appendix contains two tables listing mechanical properties of SiC-filament-reinforced Si_3N_4 reported in the open literature. Table 1 presents a detailed list of mechanical properties from selected references, and Table 2 presents an abbreviated summary of the range of properties which are listed in Table 1. The nomenclature used in Table 1 is explained below:

Nomenclature	Material Description
HP	Hot pressed
HIP	Hot isostatically pressed
HIPRBSN	Hot isostatically pressed reaction bonded Si_3N_4
HPSN	Hot pressed Si_3N_4
RBSN	Reaction bonded Si_3N_4
SSN	Sintered Si_3N_4
HIPSN	Hot isostatically pressed Si_3N_4
SCHIPSN	Slip cast and hot isostatically pressed Si_3N_4
CVISN	Chemical vapor infiltrated Si_3N_4
ARCO	SiC whiskers supplied by Arco Corp.
ACM	SiC whiskers by Advanced Composite Materials Corp.
AVCO	SiC fibers supplied by AVCO Corp.
Nicalon	SiC-based fiber supplied by Nippon Carbon Co., Japan
Huber	SiC whiskers supplied by Huber Corp.
Tateho	SiC whiskers supplied by Tateho Chemical Industries
Norton	Si_3N_4 supplied by Norton Co.
MPDZ	SiC-based fiber supplied by Dow Corning
TOKAI	SiC whisker supplied by Tokai Carbon Co.
VLS	SiC whiskers from vapor-liquid-solid process (Milewski et al. 1985; Shalek et al. 1987)
CVD	Chemical vapor deposition
UTS	Ultimate tensile strength
STF	Strain to failure
WOF	Work of fracture ($\times 10^6$ ergs/cm ²)

Table 1: Selected Mechanical Properties of SiC-Filament-Reinforced Si₃N₄

Material	Flexure Strength (MPa)	Fracture Toughness (MPa•m ^{1/2})	Miscellaneous Mechanical Properties	Reference
HPSN-MgO	379-545 @25°C			Bender et al. (1978)
HPSN-MgO 6 vol% Nicalon	448-586 @25°C			"
HPSN-MgO 25 vol% Nicalon	310-324 @25°C			"
RBSN	107 @25°C			Bhatt (1985 and 1986a)
RBSN 32 vol% AVCO	539-717 @25°C		325 MPa UTS @25°C	"
RBSN 40 vol% AVCO	616-958 @25°C		536 MPa UTS @25°C	"
HPSN-Y ₂ O ₃ -Al ₂ O ₃ + AVCO			Interfacial shear stress 12-16MPa @25°C	Brun and Singh (1987)
HIPSN or SSN		6 @1200°C	500 MPa MOR @1200°C	Buljian (1987)
Tateho + (HIPSN or SSN)		4.7-5.4 @25°C		"
30 vol% Tateho + (HIPSN or SSN)		7.9 @1200°C	600 MPa MOR@1200°C	"
Oxidized (HIPSN or SSN)			773 MPa MOR @25°C	"
Oxidized (HIPSN or SSN) + 30 vol% ARCO			859 MPa MOR @25°C	"

Table 1 (continued): Selected Mechanical Properties of SiC-Filament-Reinforced Si₃N₄

Material	Flexure Strength (MPa)	Fracture Toughness (MPa•m ^{1/2})	Miscellaneous Mechanical Properties	Reference
HPSN-Y ₂ O ₃ -Al ₂ O ₃		3.4-4.6 @25°C 4.9 @1000°C 6.1 @1200°C	773 MPa MOR @25°C 585 MPa MOR @1000°C 486 MPa MOR @1200°C	Buljian (1987); Buljian, Baldoni, and Huckabee (1987); Buljian et al. (1987)
HPSN-Y ₂ O ₃ -Al ₂ O ₃ + 30 vol% ARCO		3.0-6.4 @25°C 7.7 @1000°C 7.8 @1200°C	975 MPa MOR @25°C 819 MPa MOR @1000°C 597 MPa MOR @1200°C	" " "
CVISN-SiC fiber		3.4-4.1 @25°C		Caputo, Lackey, and Stinton (1985)
RBSN	325 @25°C		0.18% STF @25°C 4.62 WOF @25°C	Corbin, Rossetti, and Hartline (1986)
RBSN + AVCO with 1.0 μm C-coating	479 @25°C		2.2% STF @25°C 42.9 WOF @25°C	"
RBSN + AVCO with 3.0 μm C-coating	755 @25°C		above 3.0% STF @25°C 94.5 WOF @25°C	"
RBSN + Nicalon	125-200 @25°C			Corbin, Willkens, and Hartline (1987a)
RBSN + MPDZ	125-200 @25°C			"
RBSN		"brittle" @25°C	315 MPa max. stress @25°C	Corbin, Willkens, and Hartline (1987b)
RBSN + 20 vol% Nicalon		"brittle" @25°C	252 MPa max. stress @25°C	"

Table 1 (continued): Selected Mechanical Properties of SiC-Filament-Reinforced Si₃N₄

Material	Flexure Strength (MPa)	Fracture Toughness (MPa•m ^{1/2})	Miscellaneous Mechanical Properties	Reference
RBSN + 20 vol% Nicalon with Al ₂ O ₃ coating		"brittle" @25°C	376 MPa max. stress @25°C	Corbin, Willkens, and Hartline (1987b)
RBSN + 7 vol% Nicalon with SiC coating		"brittle" @25°C	194 MPa max. stress @25°C	"
RBSN + 21 vol% Nicalon with C coating		"tough" @25°C	390 MPa max. stress @25°C	"
RBSN + 18 vol % Nicalon with Si ₃ N ₄ coating		"brittle" @25°C	180 MPa max. stress @25°C	"
HPSN-MgO	650 @25°C 50 @1300-1500°C	7 @25 & 1400°C 15 @1200°C		Gibbs, Petrovic, and Honnell (1987)
HPSN-MgO + VLS whiskers	520 @25°C 200 @1300°C 100 @1500°C	10 @25°C 15 @1200°C 7 @1400°C		"
HPSN + AVCO	no mechanical properties reported			Laughner et al. (1986)
SSN + AVCO	" "	" "	" "	"
RBSN + AVCO	" "	" "	" "	"

Table 1 (continued): Selected Mechanical Properties of SiC-Filament-Reinforced Si₃N₄

Material	Flexure Strength (MPa)	Fracture Toughness (MPa•m ^{1/2})	Miscellaneous Mechanical Properties	Reference
HPSN-Y ₂ O ₃ -La ₂ O ₃		"brittle" @25°C	Weibull 5-10 @25°C UTS 1050 MPa @25°C UTS 750 MPa @1300°C	Hayami et al. (1986); Ueno and Toibana (1983)
HPSN-Y ₂ O ₃ -La ₂ O ₃ + 30 vol% Tokai		"tough" @25-1300°C	Weibull 10 @25°C UTS 650 MPa @25°C UTS 600 MPa @1300°C	"
HPSN-Y ₂ O ₃ -La ₂ O ₃ + 30 vol% filtered Tokai		"tough" @25°C	Weibull 29 @25°C UTS 710 MPa @25°C	"
HIPSN-Y ₂ O ₃ -Al ₂ O ₃		5.37 @25°C		Haynes et al. (1986)
HIPSN-Y ₂ O ₃ -Al ₂ O ₃ + (Tateho or ARCO)		5.08-6.45 @25°C	UTS 195 MPa @25°C	"
RBSN			UTS 55 MPa @25°C	Lindley and Godfrey (1971)
RBSN + CVD SiC on Tungsten core			UTS 55 MPa @25°C WOF 0.9 @25°C	"
RBSN + HIP or PS	900 @25°C	6.0 @25°C		Lundberg et al. (1987)
RBSN + HIP or PS + 20% Tokai	600 @25°C	5.0 @25°C		"
HIPSN + Tateho		8.0 @25°C		Nagel et al. (1987)
SSN + Tateho		4.0 @25°C		"

Table 1 (continued): Selected Mechanical Properties of SiC-Filament-Reinforced Si₃N₄

Material	Flexure Strength (MPa)	Fracture Toughness (MPa•m ^{1/2})	Miscellaneous Mechanical Properties	Reference
Si ₃ N ₄ /SiC whisker extruded + HIP	650 @25°C	3.6-6.3 @25°C		Willkens, Corbin, and Pujari (1987)
Si ₃ N ₄ + whisker isostatic press	650 @25°C	4.0 @25°C		"
HPSN + whisker	700 @25°C	5.0 @25°C		"
HIPSN + whisker	700 @25°C	6.3 @25°C		"
HPSN-MgO	413-565 @25°C			Rice et al. (1980)
HPSN-MgO + 6.4 vol% Nippon fiber	592 @25°C			"
HPSN-MgO + 23.6 vol% Nippon Fiber	303 @25°C	56 @25°C		"
HPSN-MgO	400-700 @25°C	5.0-7.8 @25°C		Shalek et al. (1986)
HPSN-MgO + 10 vol% VLS	370-600 @25°C	12.5 @25°C		"
HPSN-MgO + 30 vol% VLS	270-510 @25°C	7.3-10.8 @25°C		"
RBSN + 25 vol% chopped Nicalon	150-190 @25°C	1.8-2.7 @25°C		Starr, Harris, and Mohr (1987)

Table 1 (continued): Selected Mechanical Properties of SiC-Filament-Reinforced Si₃N₄

Material	Flexure Strength (MPa)	Fracture Toughness (MPa·m ^{1/2})	Miscellaneous Mechanical Properties	Reference
HIPRBSN-Y ₂ O ₃		3.1 @25°C		Willkens and Corbin (1987)
HIPRBSN-Y ₂ O ₃ + Tateho		3.0 @25°C		"
HIPRBSN-Y ₂ O ₃ + BN-coated Tateho		4.4 @25°C		"
HIPRBSN-Y ₂ O ₃ + SiC/BN-coat Tateho		3.9 @25°C		"
HIPRBSN-Y ₂ O ₃ + Si ₃ N ₄ /BN-coat Tateho		3.2 @25°C		"
HIPRBSN-Y ₂ O ₃ + Si ₃ N ₄ /C-coat Tateho		2.9 @25°C		"
HIPRBSN-Y ₂ O ₃ + SiC/C-coat Tateho		2.7 @25°C		"
HIPRBSN-Y ₂ O ₃ + 30 vol% Tateho T-44	700 @25°C 427-483 @1370°C	5.0 @25°C		"
HIPRBSN-Y ₂ O ₃ + 20 vol% Tateho SCW-1	593 @25°C 427-483 @1370°C	5.0 @25°C		"
HIPRBSN-Y ₂ O ₃ + 30 vol% Huber	823 @25°C 427-483 @1370°C	5.0 @25°C		"
RBSN Norton type NC132		6.0 @25°C 6.0 @25-1200°C		"
HIPSN + unknown whisker		7.3 @25°C 6.0 @1200°C		"

Table 1 (continued): Selected Mechanical Properties of SiC-Filament-Reinforced Si₃N₄

Material	Flexure Strength (MPa)	Fracture Toughness (MPa•m ^{1/2})	Miscellaneous Mechanical Properties	Reference
HPSN-Y ₂ O ₃ -Al ₂ O ₃	930 @25°C	7.0 @25°C		Shetty et al. (1985)
HPSN-Y ₂ O ₃ -Al ₂ O ₃ + 5-44 vol% AVCO	400 @25°C	7.0 @25°C		"
HPSN-Y ₂ O ₃ -Al ₂ O ₃ + 30 vol% AVCO		8.2 @25°C		"
HPSN + MgO	700 @25°C			Jones et al. (1988)
HPSN + MgO + 10 vol% Nicalon	250 @25°C			"
HPSN + MgO	700 @25°C	4.0 @25°C		Schoenlein et al. (1988)
HPSN + MgO + 10 vol% Nicalon	250 @25°C	2.0 @25°C		"
RBSN	221 @25°C	2.0 @25°C		"
RBSN + 5 vol% coated VLS	83-700 @25°C	0.5-2.0 @25°C		"
HIPSN-Y ₂ O ₃ -MgAl ₂ O ₄	650-790 @25°C	6.5-8.3 @25°C	WOF 0.07-0.12 @25°C	Kandori et al. (1987)
HIPSN-Y ₂ O ₃ -MgAl ₂ O ₄ + 10 vol% Tateho	800-925 @25°C	9.0-10.0 @25°C	WOF 0.125-0.16 @25°C	"

Table 1 (continued): Selected Mechanical Properties of SiC-Filament-Reinforced Si₃N₄

Material	Flexure Strength (MPa)	Fracture Toughness (MPa·m ^{1/2})	Miscellaneous Mechanical Properties	Reference
HPSN	375 @25°C	4 @25°C		Singh et al. (1987)
HPSN + 20 vol% Arco whisker	550 @25°C	7 @25°C		"
HIPSN		4.8-7.9 @25°C	715.6 MPa MOR @25°C	Haynes and Yeh (1985)
HIPSN + 10 vol% Arco whisker		6.4 @25°C		"
HIPSN + 20 vol% Arco whisker		10.98 @25°C	625 MPa MOR @25°C	"
SCHIPSN + 10 wt% ACM whisker	725 @21°C 691 @1000°C 430 @1200°C			Yeh et al. (1988)
SCHIPSN + 20 wt% ACM whisker	604 @21°C 803 @1000°C 270 @1200°C			"
SCHIPSN + 30 wt% ACM whisker	674 @21°C 674 @1000°C 266 @1200°C			"
SCHIPSN + 40 wt% ACM whisker	439 @21°C 643 @1000°C 230 @ 1200°C			"
HPSN-Y ₂ O ₃ -Al ₂ O ₃	850 @25°C	4.2 @25°C		Porter, Lange, and Chokshi (1987)
HPSN-Y ₂ O ₃ -Al ₂ O ₃ + 15 vol% ARCO	1250 @25°C	4.8-6.0 @25°C		"

Table 1 (continued): Selected Mechanical Properties of SiC-Filament-Reinforced Si₃N₄

Material	Flexure Strength (MPa)	Fracture Toughness (MPa·m ^{1/2})	Miscellaneous Mechanical Properties	Reference
HPSN	830 @25°C	6 @25°C		Tamari et al. (1982)
HPSN + 10 vol% whisker	730 @25°C	5.6 @25°C		"
HPSN + 20 vol% whisker	660 @25°C	5.3 @25°C		
HPSN + 30 vol% whisker	730 @25°C	4.8 @25°C		

Table 2: Abbreviated Summary of Reported Mechanical Properties for SiC-Filament-Reinforced Si_3N_4

Material	Flexure Strength (MPa)	Tensile Strength (MPa)	Rupture Modulus (MPa)	Fracture Toughness ($\text{MPa}\cdot\text{m}^{1/2}$)	Work of Fracture ($\times 10^6 \text{ erg/cm}^2$)
.....					
Si_3N_4 mono-lith	107-930 @25°C 50 @1300-1500°C	55-1050 @25°C 750 @1300°C	716-773 @25°C 585 @1000°C 486-500 @1200°C	2.0-8.3 @25°C 6-15 @1200°C	0.07-4.6 @25°C
Si_3N_4 + SiC filament	83-1250 @25°C 483 @ 1370°C 200-680 @1300°C	55-710 @25°C 600 @1300°C	625-975 @25°C 819 @1000°C 600 @1200°C	0.5-12.5 @25°C 56 @25°C 4.9-7.7 @1000°C 6-15 @1200°C 7 @1400°C	0.125-94.5 @25°C
.....					

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