Mechanical properties of silicon carbide—in situ zirconium carbonitride composites

1 INTRODUCTION

Silicon carbide (SiC) is an essential material for various industrial applications owing to its high thermal conductivity, high mechanical strength, and superior wear, oxidation, and corrosion resistances. In order to expand the applicability of SiC, extensive studies have been performed on the development of SiC-based composites. Such composites are designed to provide novel noninherent properties or further enhance the existing properties of monolithic SiC. With this regard, particulate-reinforced SiC composites provide the most economical solution by offering a unique combination of mechanical, electrical, thermal, and wear properties, which cannot be achieved by other classes of materials. We present below a brief review on the enhancement of mechanical, thermal, electrical, and wear properties by various reinforcements in the SiC matrix.

1. Mechanical properties: Particulate reinforcement could remarkably improve the flexural strength of monolithic SiC. For example, the strength of 647 MPa for monolithic SiC was improved to 1020 MPa for SiC–35 wt% Si₃N₄ upon sintering with a 10 wt% Y–Mg–Si–Al–ON glass. Similarly, Shaoming et al⁶ and Yeom et al⁷ reported high strength values of 810 and 720 MPa for SiC–25 vol% Si₃N₄ with 3 wt% Y₂O₃–La₂O₃ and SiC–35 vol% Si₃N₄ with 2 vol% equimolar Y₂O₃–Sc₂O₃, respectively, which were attributed to grain refinement, reduction in flaw size, and increase in fracture toughness upon Si₃N₄ addition. Among the group-13 nitrides, BN addition (≤5 vol%) leads to improved strengths of in situ SiC–BN composites, whereas for the powder-processed SiC–BN composites, a monotonic decrease in strength is observed with the BN addition owing to the intrinsic weakness of BN grains. SiC–AlN particulate composites exhibited very high strengths of ~1 GPa, attributed to grain refinement originating.
from the formation of a 2H solid solution. Titanium-based ceramic reinforcements such as TiB$_2$, TiN, and TiC could also effectively enhance the flexural strength owing to the grain refinement and induction of residual compressive stresses inside the SiC matrix.\textsuperscript{12-16} For example, the monolithic strengths increased from 620 to 750 and 616 to 921 MPa for SiC–30 wt\% TiC\textsuperscript{12} and SiC–40 vol\% TiN,\textsuperscript{13} respectively. Generally, particulate addition improves the fracture toughness owing to favorable residual thermal stresses, crack bridging, crack deflection, etc.\textsuperscript{5,7,9,12,14-16} In particular, in the case of a β-Si$_3$N$_4$-reinforced SiC composite, the acicular morphology of β-Si$_3$N$_4$ grains contributes favorably to the increase in fracture toughness by crack deflection and crack bridging. For example, an improved fracture toughness of ~7 MPa·m$^{1/2}$ was obtained for SiC–35 vol\% Si$_3$N$_4$,\textsuperscript{7} whereas the toughness value of monolithic SiC was ~4 MPa·m$^{1/2}$. In case of nonacicular particulate phases, the major factor for the increased toughness is the residual thermal stresses emerging owing to the mismatch in coefficient of thermal expansion (CTE) between SiC and particulate phases, leading to an enhanced crack deflection.\textsuperscript{9,12,14-16}

2. Electrical properties: SiC is a wide-bandgap semiconductor; its 3C, 6H, 4H, and 2H polytypes have bandgap energies of 2.4, 3.0, 3.2, and 3.3 eV, respectively.\textsuperscript{17} Its electrical properties have been successfully manipulated by doping with a particulate phase containing N-derived donors and/or Al- or B-derived acceptors. Entrapment of charge carriers by deep acceptors such as Al and B led to very high electrical resistivities of $1.03 \times 10^{12}$ and $1.1 \times 10^{10}$ Ω·cm for SiC–10 wt\% BN\textsuperscript{18} and SiC–35 vol\% AlN,\textsuperscript{19} respectively. The addition of metal nitrides could successfully enhance the electrical conductivity by acting as an N-doping source. For example, electrical conductivities of $2.9 \times 10^4$ and $2.3 \times 10^5$ (Ω·cm)$^{-1}$ were obtained for SiC–20 vol\% TiN\textsuperscript{20} and SiC–35 vol\% ZrN,\textsuperscript{21} respectively. Similarly, with the addition of conducting phases into SiC, electrical conductivities in the range of $10^2$-$10^3$ (Ω·cm)$^{-1}$ were obtained for SiC–MoSi$_2$,\textsuperscript{22} ZrC–ZrB$_2$–SiC,\textsuperscript{23} and SiC–NbC–Ti composites.\textsuperscript{24}

3. Thermal properties: In general, it has been observed that the addition of various particulate phases such as TiN, BN, AlN, and ZrN leads to reduction in the thermal conductivity of particulate-reinforced SiC composites owing to the low intrinsic thermal conductivity of secondary phases and high interfacial resistance. However, the worst degradation in properties is observed for particulate phases containing soluble elements in the SiC lattice such as Al, B, etc, as they create point defects, which adversely affect the thermal transport.\textsuperscript{9,21,25,26} The typical reported thermal conductivities for 10 vol\% additions of ZrN, TiN, BN, and AlN were 168, 131, 84, and 50 W·m$^{-1}$·K$^{-1}$, respectively.\textsuperscript{19,21,26,27} The above review shows that SiC can be reinforced with secondary particulates to obtain desirable combinations of thermal, electrical, mechanical, and wear properties, suitable for various applications. There are still various reinforcements into the SiC matrix that have not been investigated. One of the least studied reinforcement materials is ZrN, promising owing to its high melting point, high electrical conductivity, excellent chemical stability, low neutron capture capability, and resistance to corrosion against molten iron/steel.\textsuperscript{12-16} Recently, Jang et al\textsuperscript{21} reported SiC–Zr$_2$CN composites with very good thermal and electrical conductivities. The SiC–4 vol\% ZrN composite exhibited an excellent thermal conductivity of ~177 W·m$^{-1}$·K$^{-1}$; a very high electrical conductivity of $2.3 \times 10^5$ (Ω·cm)$^{-1}$ was observed for SiC–35 vol\% ZrN. Their study was the first study on ZrN-reinforced SiC, aiming to develop electrically conductive SiC by reinforcing with an electrically conductive phase together with grain-growth-assisted N-doping of the SiC matrix during liquid-phase sintering (LPS) through solution reprecipitation. Previous studies have investigated the mechanical and electrical properties of ternary systems including Si$_3$N$_4$–SiC–ZrN,\textsuperscript{37} (Zr,Hf)B$_2$–SiC–ZrN,\textsuperscript{38} ZrB$_2$–SiC–ZrN/ZrC,\textsuperscript{39,40} etc. However, no study has been reported on the mechanical behavior of the SiC–ZrN binary system. Therefore, in continuation to the previous study,\textsuperscript{21} we investigated the mechanical potential of the newly developed SiC–Zr$_2$CN composites.

The newly developed SiC–Zr$_2$CN composites have shown excellent combination of hardness and fracture toughness and are expected to show improved wear resistance properties. The present report is limited to the discussion on mechanical properties; however, the sliding wear and erosion wear studies are under investigation and would be discussed in another article.

2 | EXPERIMENTAL METHODS

Hot-pressed SiC–Zr$_2$CN composites with ZrN contents of 0, 4, 10, and 20 vol\%, denoted as SZN0, SZN4, SZN10, and SZN20, respectively, are investigated in this study. The additive content was fixed at 2 vol\% equimolar Y$_2$O$_3$–Sc$_2$O$_3$. The detailed processing procedure is reported in a previous
The sintered density of the specimens was determined using the Archimedes method. The relative density was obtained from the density measurement value and the theoretical density from the rule of mixtures. The hot-pressed samples were cut, polished, and etched with CF₄ plasma containing 10 vol% oxygen. The etched microstructures and fracture morphologies were observed using scanning electron microscopy (SEM, S4300; Hitachi Ltd., Chiyoda, Tokyo, Japan). The RT flexural strength and fracture toughness were determined according to the American Society for Testing and Materials (ASTM) C1161-13 and Japanese Industrial Standards (JIS) R1607 specifications, respectively. The high-temperature bending tests were performed in a nitrogen atmosphere at temperatures between 600 and 1500°C on four to five samples (2 × 1.5 × 25 mm) for each condition at a cross-head speed of 0.2 mm min⁻¹ using a four-point bending method with an inner and outer span of 10 and 20 mm, respectively (Instron 4505 machine; Instron Inc., Norwood, MA). Tests at high temperatures were performed by heating the specimens, avoiding thermal shock, and with a dwell time of 10 minutes to ensure a homogeneous temperature during testing. The fracture surfaces of specimens tested at high temperatures were observed using SEM after cooling the specimens to RT. The RT hardness was determined with a Vickers indenter (Model AVK-C2; Akashi Corp., Yokohama, Japan) using a load of 9.8 N and dwell time of 15 seconds.

3 | RESULTS AND DISCUSSION

3.1 | Microstructure and phase analysis

A detailed microstructure and phase analysis is presented in a previous paper. Briefly, all of the specimens including monolithic SiC and SiC–Zr₂CN composites could be densified to ≥98.5% of the theoretical density upon sintering at 2000°C for 3 hours under 40 MPa in N₂ atmosphere. An in situ Zr₂CN was synthesized by the reaction of ZrN and SiC during sintering: $4\text{ZrN} + 2\text{SiC} \rightarrow 2\text{Zr}_2\text{CN} + 2\text{Si(liquid)} + \text{N}_2$.

![Scanning electron microscopy image and qualitative compositional map of SiC–Zr₂CN composite (SZN20) determined using EPMA: A, SE image; B, carbon; C, scandium; D, BSE image; E, nitrogen; F, yttrium; G, zirconium; and H, silicon elemental analysis](image-url)
no $\beta \rightarrow \alpha$ phase transformation was observed in both monolithic and SiC–Zr$_2$CN composites. The XRD plots for monolithic SiC and SZN20 are provided as a supplementary file (Figure S1). As shown in Figure 1, the SEM image and dot mapping of Si, Zr, Y, Sc, C, and N by EPMA on the SZN20 sample revealed the locations of each element in the sample. The microstructure of SZN20 sample consists of nitrogen-doped SiC grains, Zr$_2$CN grains, and Y-Sc containing grain boundary phase. In Figure 1, the encircled region contains only Si which proves the occurrence of aforementioned reaction with the partial retention of free Si inside the composite matrix. The morphologies of SiC grains were equiaxed in both SiC and SiC–Zr$_2$CN composites, whereas the Zr$_2$CN phase exhibited two morphologies depending on the location: small and spherical grains were trapped inside SiC grains, whereas large and irregularly shaped Zr$_2$CN grains were located at grain boundaries or multigrain junctions (Figure 2). The average grain size of SiC decreased from 5.3 to 3.4 µm, whereas the Zr$_2$CN grain size increased from 2.0 to 2.9 µm with the increase in the ZrN content from 0 to 20 vol%.

3.2 RT mechanical properties

The variations in fracture toughness and flexural strength as a function of the initial ZrN content are shown in Figure 3. The fracture toughness monotonously increases with the initial ZrN content; for example, the fracture toughness of 4.2 ± 0.3 MPa-m$^{1/2}$ for the monolithic SiC increased to 6.3 ± 0.2 MPa-m$^{1/2}$ for SZN20. The crack paths became more tortuous with the increase in the ZrN addition owing to the increased crack deflection and crack bridging, as shown in Figure 4. The fracture surfaces show the imprint of the increased crack deflection manifested as a change in the fracture mode from a predominantly transgranular fracture in the monolithic SiC to a dominant intergranular fracture in SZN20 (Figure 5). The increased toughening can also be attributed to the development of thermal residual stresses owing to the thermal expansion mismatch between the SiC and Zr$_2$CN phases. Such thermal residual stresses can be quantitatively estimated by:

$$\sigma_{mr} = \frac{(\alpha_p - \alpha_m) \Delta T}{\left(\frac{1+\nu_m}{2E_m}\right) + \left(\frac{1-2\nu_p}{E_p}\right)}$$

$$\sigma_{mr} = -\frac{\sigma_{mr}}{2}$$
where \( \sigma_{mr} \) and \( \sigma_{mt} \) denote the radial and tangential matrix stresses around a Zr\(_2\)CN particle, respectively. \( \alpha_p \) and \( \alpha_m \) denote the CTE of Zr\(_2\)CN and SiC, respectively. \( \nu \) is the Poisson’s ratio. \( \Delta T \) is the temperature difference over which the stress is not relieved by the diffusive process, and the subscripts \( p \) and \( m \) refer to the particle and matrix, respectively. Using the values of 400 and 410 GPa for \( E_p \)\(^{46} \) and \( E_m \)\(^{13} \), \( 7.65 \times 10^{-6} \) and \( 4.3 \times 10^{-6} \) °C\(^{-1} \) for \( \alpha_p \)\(^{47} \) and \( \alpha_m \)\(^{13} \), 0.2 and 0.19 for \( \nu_p \)\(^{46} \) and \( \nu_m \)\(^{13} \), respectively, and 1000°C for \( \Delta T \), the radial tensile stress developed around Zr\(_2\)CN particles is 1135 MPa, while the tangential compressive stress is \(-568\) MPa. This stress field implies that an approaching crack inside the matrix is attracted by the radial tensile stress existing around the particle, whereas the compressive tangential stress diverts the crack around the particle.\(^{44} \) Such repetitive crack path deflections originating from crack-stress-field interactions lead to absorption of energy, yielding an increased fracture toughness. The crack-stress-field interactions increase with the amount of the second phase; therefore, the fracture toughness increases with the ZrN addition. Another phenomenon attributed to the improved fracture toughness is crack bridging.
The Zr$_2$CN particles act as a bridge between the opposite faces of a crack, preventing an excessive crack opening and thereby reducing the stress intensity factor, leading to the increased fracture toughness compared to that of the monolithic SiC (SZN0).

Figure 3 shows that the RT flexural strength of 546 ± 32 MPa for monolithic SiC increased to 644 ± 87 MPa for SZN20. The critical flaw sizes for monolithic SiC and SiC–Zr$_2$CN composites were calculated from the measured values of the fracture toughness ($K_{IC}$) and flexural strength ($\sigma_c$) using the equation: $K_{IC} = 1.35\sigma_c^{0.5}$, assuming semieliptical surface flaws. The calculated critical flaw sizes ($c$) for SZN0, SZN4, SZN10, and SZN20 were ~32, ~33, ~43, and ~78 µm, respectively. The $c$ values are significantly larger than the average grain sizes (3.4–5.3 µm for SiC and 2–2.9 µm for Zr$_2$CN) for all specimens, indicating that the fracture originated from external flaws. The large critical flaw size at the 20 vol% ZrN addition reflects its high susceptibility to processing defects, compared to those of the monolithic SiC or composites with smaller amounts of ZrN.

Figure 6 presents the typical fracture origins for SZN0 and SZN10, which were mostly surface pores for all specimens. The pore sizes of SZN0 and SZN10 measured using SEM micrographs were ~45 and ~60 µm, respectively, comparable to the calculated critical flaw sizes.

Despite the larger critical flaw size, SZN10 exhibited a higher flexural strength than that of SZN0, owing to residual stresses, as in the case of the improved fracture toughness discussed above. As the CTE of SiC (4.3 × 10^{-6}°C^{-1}) is smaller than that of Zr$_2$CN (7.65 × 10^{-6}°C^{-1}), a residual compressive stress is developed inside the matrix (SiC), whereas balancing tensile stresses are developed inside the particulate phase (ZrN). The previous studies on SiC–TiN$^{13}$ and SiC–Ti$_2$CN$^{16}$ described the crack-closing residual compressive stresses in the matrix as an important contribution to the increased flexural strength. Therefore, it is crucial to understand the relationship between the flexural strength and residual stresses. These stresses can be estimated by$^{44}$:

$$\sigma_p = \frac{-2 \left \{ \left ( 1 - V_f \right ) \beta \alpha \ast \right \} }{((1-V_f)(\beta+2)(1+\nu_m)+3V_f\beta(1-\nu_m))} \left \{ (1-V_f)(\beta+2)(1+\nu_m)+3V_f\beta(1-\nu_m) \right \}$$

The values of $\alpha^*$ and $\beta$ are defined as:

$$\alpha^* = \int_{T_o}^{T_p} (\alpha_p - \alpha_m) dT$$

$$\beta = \frac{(1+\nu_m)E_p}{(1-2\nu_p)E_m}$$

The values of $E_p$, $E_m$, $\alpha_p$, $\alpha_m$, $\nu_p$, and $\nu_m$ are equal to those in Equations (1) and (2), whereas $V_f$ denotes the volume fraction of the particulate phase and $T_p$ is the temperature at which residual stresses begin to develop upon cooling down to $T_o$, $T_p - T_o$ is set to 1000°C. The calculated residual matrix and particulate stress as a function of the initial ZrN content are presented in Table 1, which reveals that the residual compressive stress inside the matrix increases, whereas the tensile stress in the particulate phase decreases with the increase in the ZrN particulate addition. These compressive stresses contribute favorably to the strength by applying crack closing forces. However, the reduced strength at the 20 vol% ZrN addition is more related to coarsening of the Zr$_2$CN phase and increased inhomogeneity leading to an increased critical flaw size (~78 µm), almost twice that of SZN10 (~43 µm).

The Vickers hardness of the SiC–Zr$_2$CN composite with the 2 vol% Y$_2$O$_3$–Sc$_2$O$_3$ additive decreased from 28.5 ± 1.7 to 23.1 ± 0.8 GPa with the increase in the initial ZrN content.

### Table 1  Matrix and particulate phase thermal residual stresses for monolithic SiC and SiC–Zr$_2$CN composites

<table>
<thead>
<tr>
<th>Sample designation</th>
<th>SiC (Matrix)</th>
<th>Zr$_2$CN (Particulate)</th>
</tr>
</thead>
<tbody>
<tr>
<td>SZN0</td>
<td>0</td>
<td>–</td>
</tr>
<tr>
<td>SZN4</td>
<td>−45</td>
<td>1089</td>
</tr>
<tr>
<td>SZN10</td>
<td>−113</td>
<td>1019</td>
</tr>
<tr>
<td>SZN20</td>
<td>−226</td>
<td>903</td>
</tr>
</tbody>
</table>
from 0 to 20 vol%, as shown in Figure 7. The hardness of composites is a function of relative density and intrinsic hardness of its constituent phases. The rule of mixture is an effective tool for estimating the hardness variation with respect to second phase addition. The two governing equations are:

\[ H_c = V_f H_f + (1 - V_f) H_m \]

\[ H_e = \left( \frac{V_f}{H_f} + \frac{1 - V_f}{H_m} \right)^{-1} \]

where \( H_c \), \( H_f \), and \( H_m \) denotes the hardness of composite, particulate phase, and matrix phase, respectively. Equations (7) and (8) are based upon the isostrain and isostress conditions, respectively. Though, these equations assume stress conditions which are hardly possible in real system but they essentially provide the useful upper and lower bound for hardness variation.

The other well-known equation which describes the effect of porosity on hardness is given as:

\[ H_p = H_o e^{-bP} \]

where \( H_p \) and \( H_o \) are the hardness of porous and dense material, respectively. \( P \) denotes porosity and \( b \) is a fitting parameter. In order to include the effect of porosity and second phase simultaneously, we suggest the following equation:

\[ H_e = H_c e^{-bP} \]

where \( H_e \) denotes effective hardness of the composite and \( H_c \) is the hardness evaluated from Equations (7) and (8). The hardness was estimated by Equation (10) using the following values: 30.1 GPa (100% dense monolithic SiC corresponding to SZN0 using Equation (9)) for monolithic SiC (\( H_m \)), 19 GPa for Zr\(_2\)CN\(^{51}(H_f)\), and 5.04 for \( b \). Figure 7 shows a comparative plot for the experimentally observed hardness values vs the values obtained using Equation (10) considering both isostrain and isostress conditions. It can be observed that the experimentally determined values show a deviation of \( \sim 3\%-12\% \) from the theoretical value which increased with increasing ZrN content. This anomaly is ascribed to the limited scope of the empirical relations which does not include the effect of heterogeneous interface formed between SiC and Zr\(_2\)CN phases and the residual stresses existing in the composites as described previously (Table 1). However, both experiment and theoretical estimations show degradation in hardness with successive ZrN addition owing to the low hardness of Zr\(_2\)CN (~19 GPa). Unlike fracture toughness and flexural strength, ZrN addition proved to be deleterious to hardness. The foregoing discussion on RT mechanical properties of SiC–Zr\(_2\)CN composites suggests that a reinforcement with higher thermal expansion coefficient and hardness than SiC could increase its flexural strength, fracture toughness, and hardness simultaneously.

### 3.3 High-temperature flexural strength

During hot pressing, Y\(_2\)O\(_3\)–Sc\(_2\)O\(_3\) reacts with SiO\(_2\) and ZrO\(_2\) which are always present on SiC and ZrN particles, respectively to form a Si–Y–Zr–Sc–O–N melt and, with increasing temperature, a Si–Y–Zr–Sc–O–C–N melt is formed by dissolution of SiC. This liquid phase promotes densification via liquid-phase sintering and solidifies as a crystalline or amorphous intergranular phase (IGP) at grain boundaries during cooling. It has been reported that Y\(_2\)O\(_3\)-doped SiC promotes an amorphous IGP. So, it is probable that the IGP formed in the present monolithic SiC and SiC–Zr\(_2\)CN composites is amorphous/glassy in nature. The high temperature strength of liquid-phase-sintered ceramics is often
limited by their IGP softening temperature which in turn depends on the chemistry of the system rather than the physical nature of IGP (crystalline or amorphous). The flexural strengths of SZN0, SZN10, and SZN20 as a function of the temperature are shown in Figure 8. The monolithic SiC exhibited an increase in strength at 1200°C, whereas the composites (SZN10, SZN20) exhibited decreases in strength even at a temperature as low as 600°C, compared to their RT strengths. The increased strength for monolithic SiC may be attributed to plastic deformation-induced crack healing around the critical defect. A similar tendency has been observed in previous studies. Kim et al. reported an increase in flexural strength from 594 MPa at RT to 644 MPa at 1400°C when SiC was sintered with 10 vol% AlN–Lu2O3; SiC sintered with 10 vol% AlN–Sc2O3 exhibited an increased strength (670 MPa) at 1500°C compared to the RT strength (600 MPa). Further, as shown in the SEM micrographs in Figure 9, the fracture mode was predominantly transgranular for the SZN0 specimen fractured at 1200°C, indicating the presence of a highly refractory grain boundary (Si–Y–Sc–O–C–N IGP). In general, degradation in high-temperature strength is manifested as a predominant intergranular fracture indicating the softening of the IGP at high temperatures.

Therefore, the increased high-temperature strength of SZN0 can be attributed to a cumulative effect of crack healing and the presence of a highly refractory intergranular glass phase. However, the monolithic SiC tested at 1500°C exhibited a reduced strength of 437 ± 25 MPa, which implies softening of the intergranular film causing a predominantly intergranular fracture (Figure 9C). For the case of SiC–Zr2CN composites, with the addition of ZrN, the Si–Y–Sc–O–C–N liquid phase in the monolithic SiC was transformed to a Si–Y–Zr–Sc–O–C–N melt. The severe degradation in strength with the increase in temperature indicates that the IGP consisting of Si–Y–Zr–Sc–O–C–N has a very poor refractory property (the softening temperature is probably below 600°C), such that the RT strengths of 644 ± 87 and 528 ± 117 MPa decreased to 556 ± 51 and 408 MPa at 600°C, which further decreased to 448 ± 31 and 386 ± 19 MPa at 1200°C for SZN10 and SZN20, respectively. The fracture surfaces of SZN10 and SZN20 (Figure 10) show dominant intergranular fractures at 1200°C, which also demonstrates the softening of the IGP at high temperatures.
SiC–Zr2CN composites were fabricated from β-SiC and ZrN with 2 vol% equimolar Y2O3–Sc2O3 which could be sintered to ≥98.5% of the theoretical density upon hot pressing at 2000°C for 3 hours at 40 MPa under N2 atmosphere. Thermal residual stresses produced by the CTE mismatch between SiC and Zr2CN contributed favorably to both fracture toughness and flexural strength. The fracture toughness increased with the ZrN addition. The fracture toughness of 4.2 ± 0.3 MPa·m$^{1/2}$ for monolithic SiC increased to 6.3 ± 0.2 MPa·m$^{1/2}$ for the SiC–20 vol% ZrN composite, whereas the flexural strength had the maximum of 644 ± 87 MPa at the 10 vol% ZrN content, ~18% higher than that of the monolithic SiC (546 ± 32 MPa) owing to the increased crack-closure residual compressive forces induced inside the SiC matrix. The monolithic SiC exhibited an improved strength of 650 ± 53 MPa at 1200°C compared to the RT strength of 546 ± 32 MPa, which could be attributed to the cumulative effect of crack healing and the presence of a highly refractory intergranular glass phase. However, SiC–Zr2CN exhibited a degradation in strength even at 600°C, probably owing to the poor refactoriness of the intergranular film. The typical flexural strengths of the SiC–10 vol% ZrN composite were 644 ± 87 MPa and 448 ± 31 MPa at RT and 1200°C, respectively.

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