Processing and Mechanical Properties of Fiber-Reinforced Oxide Matrix Composites with High Damage Tolerance

Yoshihiro HIRATA and Rulin DONG

Department of Advanced Nanostructured Materials Science and Technology, Graduate School of Science and Engineering, Kagoshima University, 1-21-40, Korimoto, Kagoshima 890-0065, Japan

Processing and mechanical properties of dense and porous composites were studied with the nonoxide fiber/oxide matrix system and the oxide fiber/oxide matrix system. Preparation of dense composites needs a high processing temperature (1500°C–1700°C), resulting in degradation of fibers, formation of large residual stresses during cooling process and low pseudoductility. On the other hand, porous composites processed at a low temperature (500°C–1100°C) provide a high strength and great damage tolerance, which are associated with significant delamination of layered fabric or fiber pullout. The chemical affinity and difference in the thermal expansion coefficients and Young’s moduli between fiber and oxide matrix in the composite play important role on the mechanical properties.

Key-words: Fiber-reinforced oxide matrix composite, Interface, Residual stress, Pseudoductility, Fracture energy, Fiber pullout, Delamination

1. Introduction

Continuous fiber-reinforced ceramic matrix composites (CFCCs) display high damage tolerance with pseudoductility and can be applied to high temperature structural material applications such as aircraft engine components or spacecraft engine parts. The development of mechanically reliable composites requires a weak chemical bond between the fibers and matrix to enhance fiber bridging and fiber pullout during crack propagation. Amorphous C and BN are used as interlayer because of their lubricant property, which enhances the pullout of fibers in CFCCs. Recently, oxide interphases such as LaPO₄, hexaluminate or ZrO₂ have been studied because of their high stability at high temperatures in an oxidizing environment. The debonding length of fibers in pullout depends on the roughness of the fiber surface, the strength of the fibers and the residual stresses around the interfaces resulting from the difference in thermal expansion coefficients and Young’s moduli between the fibers and matrix.  The introduction of the interphase affects the residual stresses, the chemical reaction at the interfaces, the roughness of the fibers and the resultant pseudoductility.

The matrix phase of CFCCs is usually fabricated by net-shape or near-net-shape processing such as chemical vapor infiltration (CVI), polymer impregnation and pyrolysis (PIP) or reaction sintering (RS). The CVI and RS processes offer a dense matrix but the PIP process leaves residual pores. Although PIP takes a longer processing time than CVI or RS, it has advantages in low temperature processing and use of simple processing equipments (low cost). The structure of composites processed by the PIP method is analogous to that of a concrete material. The fibers, the ceramic powder filler and the precursor polymer in the PIP process correspond to the steel bars, rock and sand, and cement in the concrete material, respectively. Such a low temperature PIP process is favorable for fiber to avoid degradation during processing. In this paper, we review the processing-microstructure-mechanical properties relationship of the nonoxide fiber/oxide matrix and oxide fiber/oxide matrix composites.

2. Dense composites

2.1 Si-Ti-C-O fabric/mullite matrix composites with graphite interphase

Table I summarizes the general feature of processing and resultant mechanical properties of dense and porous CFCCs. In the processing of dense CFCCs, a relatively high processing temperature (1500°C–1700°C) is necessary to achieve a dense matrix. As a result, nonoxide fiber is often selected rather than oxide fiber because of the high temperature stability of nonoxide fiber in a low oxygen partial pressure atmosphere. In oxide fibers, degradation due to the fast grain growth is more crucial at a high temperature. The fracture behavior of the composite is greatly influenced by the fiber/matrix interfacial properties. The interphase layer, inserted between the fiber and matrix, enhances the crack deflection along the interfaces. Laminated Si-Ti-C-O woven fabrics (6–10 vol%)/mullite matrix composites with nearly full density (>99%) were produced by hot-pressing under a pressure of 39 MPa in a N₂ atmosphere at 1500°C–1750°C. The fracture toughness of monolithic mullite hot-pressed at 1650°C (1.6 MPa·m¹⁄₂) was improved to 3.9 MPa·m¹⁄₂ for the composite with 6 vol% Si-Ti-C-O fiber and 4.7 MPa·m¹⁄₂ for the composite with 10 vol% Si-Ti-C-O fiber. However, the increase of the fabric content to 20–30 vol% resulted in delaminated porous composites after hot-pressing due to the exfoliation of filament yarn in woven fabrics. This result was explained by the combination of compressive residual stress induced to the waved yarn in the fabric and the tensile residual stress induced to the mullite matrix at room temperature. A linear relation of flexural stress-displacement in fracture behavior of monolithic mullite changed to a nonlinear relation in the composites with 10–30 vol% of Si-Ti-C-O fiber. Addition of Si-Ti-C-O fabric decreased the four point flexural strength of monolithic mullite (328 MPa) hot-pressed at 1650°C to 292 MPa at 6 vol %, 271 MPa at 10 vol% and 59–78 MPa at 20–30 vol% of fabrics. A main crack in the composite with 10 vol% fabric progressed straightly in the mullite layers perpendicular to the direction of lamination and was deflected at the Si-Ti-C-O fabric layers with compressive residual stress. The crack resistance became higher when the crack tip approached the fabric layer, followed by the fracture of the fibers by shear stress (Fig. 1).

The improvement of mechanical properties in the
Table 1. General Feature of Processing and Resultant Mechanical Properties of Dense and Porous Continuous Fiber-reinforced Oxide Matrix Composites.

<table>
<thead>
<tr>
<th>Processing &amp; properties</th>
<th>Dense composites</th>
<th>Porous composites</th>
</tr>
</thead>
<tbody>
<tr>
<td></td>
<td>Nonoxide (f) / oxide</td>
<td>Oxide (f) / oxide</td>
</tr>
<tr>
<td>Fiber content</td>
<td>Low (&lt; 20 vol%)</td>
<td>High</td>
</tr>
<tr>
<td>Interphase</td>
<td>Not necessary</td>
<td>Necessary</td>
</tr>
<tr>
<td>Processing temperature</td>
<td>High</td>
<td>Intermediate</td>
</tr>
<tr>
<td>Fiber damage</td>
<td>Pyrolysis &amp; crystallization</td>
<td>Grain growth</td>
</tr>
<tr>
<td>Bulk density</td>
<td>High</td>
<td>High</td>
</tr>
<tr>
<td>Residual stress</td>
<td>High (concentrating on fabric node)</td>
<td>Low</td>
</tr>
<tr>
<td>Strength</td>
<td>High (comparable to monolithic ceramics)</td>
<td>High</td>
</tr>
<tr>
<td>Pseudoductility</td>
<td>Small</td>
<td>Small</td>
</tr>
<tr>
<td>Fracture energy</td>
<td>Low</td>
<td>Low</td>
</tr>
<tr>
<td>Deformation mechanism</td>
<td>Bridging &amp; pullout of fiber</td>
<td>Bridging &amp; pullout of fiber</td>
</tr>
</tbody>
</table>

\[
\text{SiTi}_{0.325}\text{C}_{1.38}\text{O}_{0.232} (s) \\
\rightarrow 0.689\text{SiC}(s) + 0.023\text{TiC}(s) + 0.354\text{C}(s) + 0.312\text{SiO}(g) + 0.312\text{CO}(g) \tag{1}
\]

When the amorphous Si–Ti–C–O fabric (10 vol%)/mullite composite was hot-pressed at 1500°C in a N₂ atmosphere, partial decomposition of Si–Ti–C–O fibers caused the diffusion of C and Ti from the fibers to the interfaces between Si–Ti–C–O fibers and mullite matrix of granular-shaped grain.⁴⁶ The diffusion of Al and O from mullite matrix to fiber was also observed. The diffusion of these elements produced graphite layers 30–60 nm thick on the surface of the fibers (Fig. 2) and crystalline Ti–C–N particles of about 80 nm in size at the triple points among the fibers.⁴⁵,⁴⁶ Diffusion of Al and O from mullite matrix into Si–Ti–C–O fibers was interpreted as being due to the reaction between mullite and C or SiC segregated on the surface of fibers (Eqs. (2) and (3)).

\[
\text{Al}_2\text{O}_3(s) + 2\text{C}(s) \rightarrow \text{Al}_2\text{O}(g) + 2\text{CO}(g) \tag{2}
\]

\[
\text{Al}_2\text{O}_3(s) + \text{SiC}(s) \rightarrow \text{SiO}(g) + \text{Al}_2\text{O}(g) + \text{CO}(g) \tag{3}
\]

Partial pyrolysis of the fibers in the mullite matrix during hot-pressing at 1700°C was accompanied by the crystallization into fine nanometer-sized β-SiC, producing pores of 1.5–2.0 μm diameter inside of the fibers.⁴⁶ The molar fraction of pyrolysis of the fibers was estimated to be about 10% at 1700°C. The strength of composites with 10 vol% fabric increased from 171 MPa at 1500°C of hot-pressing to 480 MPa at 1700°C of hot-pressing. The increased strength of the composite was due to strengthening of fibers with crystallization into β-SiC.⁴⁶ As described above, the fracture toughness and strength of the dense nonoxide fiber/oxide matrix composites are dominated by (1) the difference of Young’s moduli and thermal expansion coefficients between fiber and matrix, producing the residual stresses at a low temperature, (2) fiber content, (3) thermal stability of fiber and (4) chemical reactivity between fiber and matrix.

Formation of an interphase with low Young’s modulus at

Fig. 1. A main crack extension (a) and corresponding stress intensity factor (b) of Si–Ti–C–O fabric (10 vol%)/mullite composite with an initial V-notch size of a/W = 0.213 (a: notch length, W: width of beam).

Si–Ti–C–O fabric/mullite composites is also influenced by the thermal decomposition of Si–Ti–C–O fibers. Compared with monolithic fibers, the pyrolysis of amorphous Si–Ti–C–O fibers in dense mullite matrix, as expressed by Eq. (1), was suppressed because of the difficulty of diffusion of gases evolved.⁴⁵
the interface between fiber and matrix is effective to decrease the residual stresses, and enhances the fiber bridging and pull-out. Figure 3 shows the effect of the graphite layer thickness on the residual stresses at the interface between Si-Ti-C-O fiber and mullite matrix, cooled with temperature difference \( \Delta T = 1000^\circ \text{C} \). The compressive residual stress induced to mullite matrix in radial direction and the tensile residual stress induced in the circumference of Si-Ti-C-O fiber, decrease sharply with increasing the thickness of graphite layer. That is, the relatively thick graphite layer is effective to reduce the residual stresses induced at the interfaces of Si-Ti-C-O fibers/mullite matrix. Residual stress distribution around the woven fiber bundles was also analyzed by a Finite Element Method to understand the fracture behavior of the laminated Si-Ti-C-O fabric (10 vol\%) mullite matrix composite under anticlastic bending test where tension-compression biaxial stresses were applied. The analysis revealed that the residual stress was concentrated on the fabric node, a cross point of the longitudinal and horizontal fiber bundles. The result could explain the formation of surface cracks in the mullite layer during anticlastic bending test, which originated from the fabric node. We have conducted the anticlastic bending test on \( 0^\circ/90^\circ \) crossply laminate composites of the Si-Ti-C-O fibers (10 vol\%) mullite system to evaluate the influence of fabric node on the fracture behavior. No surface crack was formed in the crossply laminate composites with no fabric node. The fracture strength of the crossply laminate composite was twice that of the layered composite with woven fabric. In other words, the fabric node of the fiber bundles greatly affects the mechanical reliability of layered composites.

2.2 SiC fiber/alumina-mullite matrix composites with oxide interphase

Oxide interphase with high thermal stability and high phase compatibility with nonoxide fiber and oxide matrix, is more desirable than nonoxide interphase in the use of composite in an oxidizing atmosphere. In our previous study, no reaction occurred in the mixture of the alumina-ceria system at 1300–1500°C in air. In the alumina-silica-ceria system at 1300–1500°C, alumina, mullite and ceria coexisted as the stable phases. These results suggest a possibility that ceria may be used as an interphase with high phase compatibility with alumina or silica-alumina matrix. In addition, doping of a rare earth element into ceria is reported to suppress the grain growth of ceria. Based on the above results, Yb–doped ceria with a composition of \( \text{Ce}_{0.8}\text{Yb}_{0.2}\text{O}_{1.9} \) was used as interphase of SiC fiber/alumina-mullite matrix composite. Yb–doped ceria was coated on the SiC fibers by a solution immersion route using Ce(NO\(_3\)) \(_3\) and Yb(NO\(_3\)) \(_3\). The heat-treatment of the mixture of Yb–doped ceria/SiC fibers with an Al sintering additive (<1 mass\%) in an Ar atmosphere at 1300°C produced CeAlO\(_3\) and 7CeO\(_2\)-9SiO\(_2\). Subsequent heating of the fibers in air at 1100–1300°C decomposed the CeAlO\(_3\) and 7CeO\(_2\)-9SiO\(_2\) into CeO\(_2\), Al\(_2\)O\(_3\) and SiO\(_2\). The SiC fibers coated with Yb–doped ceria and 80 mass% mullite–20 mass% alumina sheets of 75 \( \mu \text{m} \) thickness, were laminated and hot-pressed at 1200°C in an Ar atmosphere. As hot-pressed composite with 15 vol\% fibers and the composite subsequently heated in air at 1100°C showed ductile and brittle properties in the stress-strain relations with the fracture strength of 90 and 187 MPa, respectively. The possible reasons for the pseudoductility are (1) the formation of mechanically weak interfacial compound, and (2) the formation of porous interface layer. The high porosity (~60%) of the Yb-doped ceria layer may be an important factor in the pseudoductility. Figure 4 shows the fractured surface of the composite hot-pressed in an Ar atmosphere. The pullout of fibers was observed in the fiber yarn, reflecting the weak interfaces. These results are strongly associated with the phase stability at the interfaces between the fibers and matrix. The change to brittle composite after the heat-treatment in air was accompanied by the disappearance of CeAlO\(_3\) and 7CeO\(_2\)-9SiO\(_2\) and crystallization of SiO\(_2\) and CeO\(_2\). The oxidation of interfacial compounds and SiC causes a volume expansion near the interfaces. Furthermore, the high chemical interaction between SiO\(_2\) formed and SiC fibers oxidized or mullite-alumina matrix results in the strong contact, explaining the increase of the interfacial bonding. The above factors suppressed the sliding effect of SiC fibers.
2.3 Control of interface of oxide fiber/oxide matrix composites

A sol-gel-processed mullite powder and alumina woven fabrics were formed into laminated composites (5 vol% alumina fabric) by filtration of the aqueous mullite suspension containing polyacrylic ammonium at pH = 8.8 which were poured into the spaces separated with alumina fabrics. The mullite suspensions with 30–55 vol% solids showed a Newtonian flow and formed high density green compacts of 60–63% of theoretical density (TD). The green composites were densified to near full density by hot-pressing at 1650°C for 1 h at 39 MPa in a N₂ atmosphere. The mechanical properties (strength, hardness and fracture toughness) of the dense composites (>99% TD) were measured at room temperature. However, no significant improvements of strength and fracture toughness were obtained due to the formation of large cracks (>100 μm) across the alumina fabric and disappearance of long fiber characteristics after sintering. That is, control of the difference of thermal and mechanical properties between the fibers and matrix is crucial to avoid the crack formation due to the residual stresses induced at a low temperature.

The surface of alumina and silica-alumina fibers was coated with ceria to avoid the sintering between the fibers and alumina matrix and to promote the pullout of fibers in the oxide fibers/alumina matrix composites. A cerium nitrate solution of 0.25–0.50 M at pH 2.0–6.0 was infiltrated into the spaces of woven fabric of the oxide fibers. The infiltrated cerium nitrate was thermally decomposed at 400°C in air to form a cerium oxide layer. In the alumina-silica-ceria system, α-alumina, mullite (3Al₂O₃·2SiO₂) and ceria coexisted at 1300–1500°C in air. The ceria-coated alumina fabric was immersed in a well-dispersed alumina suspension of 15–25 vol% solid to form alumina matrix. The treated fabric of 16 sheets was hot-pressed at 1200°C to produce a dense composite of ceria-coated alumina fiber (47.5–56.6 vol%)/alumina matrix (43.4–52.5 vol%) with 98.4% of theoretical density. However, these composites fractured in the absence of plastic deformation. The grain growth of ceria and the degradation of the Al₂O₃ fiber during the hot-pressing increased the roughness of the interface, resulting in a brittle fracture behavior. On the other hand, Chawla et al. produced a mullite fiber/mullite matrix composite with pseudoductility by hot-pressing using BN or BN/SiC interphase. The composite showed bending strength of 258 MPa, fracture energy of 1.6 kJ/m² and fracture toughness of 8.5 MPa·m²/². Parthasarathy et al. evaluated the interface of porous ZrO₂–SiO₂ layer and monazite (LaPO₄) coatings in the aluminosilicate fiber/blackglass™ composite. The composite with porous oxide coating showed pseudoductility but the monazite coating resulted in no appreciable fiber pullout. This result was explained by the increased frictional resistance for monazite coating than for BN or porous oxide coating. The key factors required for the interphase to enhance the fiber pullout are (1) the suppression of grain growth during the processing, (2) phase compatibility with fibers and matrix and (3) low Young’s modulus and low mechanical strength.

3. Porous composites

3.1 Influence of porosity on mechanical properties

Si-C-O fibers (20 vol%)/alumina matrix composites of 60–99% relative density were fabricated by vacuum sintering or hot-pressing. A high-purity alumina powder with the chemical composition of 99.95 mass% Al₂O₃ and 0.05 mass% MgO (average grain size 0.5 μm), was ball-milled in distilled water for 3 h at pH 9.1 to prepare aqueous alumina suspensions of 49 vol% solids. Polyacrylammonium (dispersant) of 0.6 mass% and acryl type emulsion (binder) of 1.6 mass% against the weight of alumina powder were added to the suspensions before ball-milling. The supplied Si–C–O yarn of 500 filament consisted of amorphous fibers of 15 μm diameter (chemical composition (mass%): 56.6 Si, 30.1 C, 11.8 O and 0.3 H). The Si–C–O fibers of 85 mm length were attached to polypropylene frame to make unidirectional Si–C–O fiber-sheets. The aqueous alumina suspension was poured into the Si–C–O fiber-sheets under a reduced pressure (1 kPa). The Si–C–O sheets of 9 layers (0.3 mm thick/sheet) were laminated and dried in air at 25°C. The green composites were heated under vacuum (0.01 Pa) at 5 K/min to a given temperature in the range from 1000°C to 1400°C, and kept for 1 h. The green composites were also hot-pressed at 1100°C–1500°C for 1 h under a pressure of 39 MPa in vacuum. The flexural strengths of test specimens with sizes of 35 mm long, 4 mm wide and 3 mm high, polished with No. 600 SiC abrasive paper, were measured along hot-pressing direction at room temperature by the three-point flexural method over the span of 30 mm at a cross-head speed of 0.05 mm/min.

The nonlinear stress-strain relations were analyzed by measuring the stress-strain curves and Young's moduli in addition to the observation of the fractured surfaces of the Si–C–O fibers/alumina matrix composites. The composite of 70% relative density, processed by vacuum sintering at 1300°C, showed a linear increase of stress up to 1% of strain (Young modulus ~32 GPa) and fractured with a nonlinear decrease of stress above 1% of strain. The Young modulus of composites in the early stage of deformation was almost equal to the product of tensile modulus (176 GPa) and volume fraction (20 vol%) of Si–C–O fibers, suggesting little contribution of porous alumina matrix to the elastic deformation and strength of composite. However, this type of weak interface is effective to give the bridging and pullout of Si–C–O fibers over 1% of strain. Figure 5 shows the stress-strain curves (a) and apparent Young's modulus (b) for the composites hot-pressed at 1100°C–1500°C. In the range of 70–90% of relative density, the composite showed an elastic deformation up to 0.1% of strain (Young modulus ~135 GPa) and a maximum strength (~300 MPa) at 1% of strain. This fracture behavior was interpreted by (1) relatively strong interface between matrix and fibers in the early stage of deformation, (2) bridging and pullout effects of fibers leading to the increase of stress and the fracture of alumina matrix, and (3) pullout of fibers in the
fractured matrix in final stage of deformation. Densification of composite above 90% of relative density formed many microcracks in the fibers and matrix owing to the large internal stress resulting from the large difference of Young’s moduli and thermal expansion coefficients between matrix and fibers. The formation of microcracks decreased the flexural strength to 100–200 MPa. In these dense composites, no elastic deformation was measured in the early stage of deformation. The above results apparently indicate a higher damage tolerance for porous composites than for dense composites.

3.2 Polymer Impregnation and pyrolysis (PIP) process 3.2.1 Si-Ti-C-O fabric/mullite filler/polytitanocarbo­silane composites

A polytitanocarbo­silane (a precursor of Si-Ti-C-O fiber) xylene solution was infiltrated into a laminated porous mullite composite with 35–37 vol% Si-Ti-C-O fabric and thermally decomposed to an amorphous solid at 1000°C in an argon atmosphere to decrease the porosity and residual stresses induced by the difference in thermal and mechanical properties between the Si-Ti-C-O fabric and mullite. The volume of open pores of the composite after n PIP sequences is expressed by Eq. (4),

\[ P = P_0 \left( 1 - \frac{D_p}{D_s} \right)^n \]  

where \( P_0 \) is the initial open porosity of the composite, \( C_p \) is the polymer concentration, \( D_p \) and \( D_s \) are the density of the polymer and polymer-derived amorphous solid, and \( Y \) is the ceramic yield of solid-to-polymer (\( Y = 0.88 \)). Equation (4) indicates that (1) the pore-elimination efficiency, \( dP/dn = P_0 \left( 1 - YC_pD_p/D_s \right)^n \ln \left( 1 - YC_pD_p/D_s \right) \), decreases as the number of PIP sequence increases, (2) decreased \( P_0 \) and increased \( C_p \) effectively reduce porosity for a given polymer, and (3) using a polymer with a high \( Y \) value (low loss of mass during pyrolysis of polymer) is desirable for decreasing porosity.

Figure 3 shows the flexural stress-displacement curve of the laminated composite with 31 vol% Si-Ti-C-O fabric/26 vol% mullite filler/27 vol% polymer-derived solid/15 vol% pores, processed by PIP method.

Fig. 6. Stress-displacement curve for the laminated composite with 31 vol% Si-Ti-C-O fabric/26 vol% mullite filler/27 vol% polytitanocarbo­silane-derived solid/15 vol% pores, processed by PIP method.

Fig. 7. Appearance of the laminated composite of the Si-Ti-C-O fabric (31 vol%) /mullite filler/polytitanocarbo­silane system after the measurement of flexural strength.
and the fracture energy were enhanced by controlling following factors: (1) incorporation of mullite filler in the filament yarn (formation of a narrow pore size distribution), (2) densification with a low viscosity polytitanocarbosilane (PTC) solution, (3) increase of the fraction of the PTC-derived solid, and (4) decrease in the porosity. The Si–Ti–C–O fabric (22–36 vol%)/mullite powder/polytitanocarbosilane composite with 80–85% theoretical density showed a nonlinear increase of stress up to 0.7–1.0% of tensile strain and fractured in the tensile strength range from 102 to 265 MPa. The analysis of the apparent Young’s modulus, tensile strength and fracture energy57 concludes that (1) the strength is dominated by the product of the effective fabric content, the Young’s modulus of the filament and elongation of the composite and (2) the fracture energy (19–51 kJ/m²) of the composite is greatly affected by the tensile strength because of the similar elongation at fracture. Shearing-test parallel to the fabric layers of the composite showed the nonlinear stress-strain relationship.59 Both of the shear strength and shear fracture energy have a tendency to increase with increasing relative density of the composite. The shear strength is about one-order-of-magnitude smaller than the maximum bending stress irrespectively of the relative density. Accordingly, the shear fracture energy is also several times smaller than the work of deformation obtained by bend testing. These results clearly indicate that the composites are very weak in shearing parallel to the fabric layer, and that the low shear strength and fracture energy are attributed to the delamination observed in the bent composites.

3.2.2 SiC or aluminosilicate fiber/mullite filler/mullite precursor composites

Based on the successful PIP process of Si–Ti–C–O fabric/mullite filler/PTC composite, we studied the processing, microstructure, and mechanical properties of layered composites formed from ceramic fabric (SiC or aluminosilicate)/mullite powder/mullite precursor solution by the PIP process at 500°C in air.60 An ethylenolacohol solution containing Si(OC₂H₅)₄ and Al(NO₃)₃ with a composition of mullite, 3Al₂O₃–2SiO₂, was infiltrated into a porous laminated woven fabric of 22–33 vol% SiC or Al₂O₃–SiO₂ fibers and 8–42 vol% mullite powder. The solution was then decomposed at 500°C in air. This PIP method was repeated 20–23 times to produce a laminated mullite matrix composite of 73–79% density. The pseudocductility of the densified composite when tested in four-point bend, was caused by delamination cracks. This mechanical property was enhanced by (1) good dispersion of the mullite filler in the filament yarn, (2) large amounts of dispersed filler, (3) formation of weak bonds in the filament yarn from the pyrolysis of the mullite precursor, and (4) lower amount of mullite precursor-derived amorphous solid. The infiltration of the mullite filler dispersed electrostatically in an aqueous solution into the woven fabric offered significant pseudocductility, but infiltration of the filler dispersed in the ethylalcohol solution of the mullite precursor resulted in a brittle composite.

Summary

(1) The fracture toughness and strength of dense nonoxide fiber/oxide matrix composites are dominated by (1) the difference of Young’s moduli and thermal expansion coefficients between fiber and matrix, producing the residual stresses at a low temperature, (2) fiber content, (3) thermal stability of fiber and (4) chemical reactivity between fiber and matrix.

(2) The key factors required for the interphase to enhance the bridging and pullout of fibers are (1) a low grain growth rate during the processing, (2) phase compatibility with fibers and matrix and (3) low Young’s modulus and low strength. Oxide interphase with high thermal stability and high phase compatibility with nonoxide fiber and oxide matrix, is desirable than nonoxide interphase in the use of composite in an oxidizing atmosphere.

(3) Porous composites offer a higher damage tolerance based on the large pseudocductility than dense composites. PIP process is effective to produce layered porous composites with high strength and great fracture energy. The pseudocductility of the PIP-processed composites is caused by buckling after the elastic deformation, and followed by the delamination along the direction of fabric layers.

References

27) Parthasarathy, T. A., Boakye, E., Keller, K. A. and Hay, R. S.,