Near-seamless joining of C\textsubscript{f}/SiC composites using Y\textsubscript{3}Si\textsubscript{2}C\textsubscript{2} via electric field-assisted sintering technique

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Abstract: A novel Y\textsubscript{3}Si\textsubscript{2}C\textsubscript{2} material was synthesized at a relatively low temperature (900 °C) using a molten salt method for the first time, and subsequently used as the joining material for carbon fiber reinforced SiC (C\textsubscript{f}/SiC) composites. The sound near-seamless joints with no obvious remaining interlayer were obtained at 1600 °C using an electric field-assisted sintering technique (FAST). During joining, a liquid phase was formed by the eutectic reaction among Y\textsubscript{3}Si\textsubscript{2}C\textsubscript{2}, γ(Y–C) phase, and SiC, followed by the precipitation of SiC particles. The presence of the liquid promoted the sintering of newly formed SiC particles, leading to their complete consolidation with the C\textsubscript{f}/SiC matrix. On the other hand, the excess of the liquid was pushed away from the joining area under the effect of a uniaxial pressure of 30 MPa, leading to the formation of the near-seamless joints. The highest shear strength (τ) of 17.2±2.9 MPa was obtained after being joined at 1600 °C for 10 min. The failure of the joints occurred in the C\textsubscript{f}/SiC matrix, indicating that the interface was stronger than that of the C\textsubscript{f}/SiC matrix. The formation of a near-seamless joint minimizes the mismatch of thermal expansion coefficients and also irradiation-induced swelling, suggesting that the proposed joining strategy can be potentially applied to SiC-based ceramic matrix composites (CMCs) for extreme environmental applications.

Keywords: C\textsubscript{f}/SiC; joining; Y\textsubscript{3}Si\textsubscript{2}C\textsubscript{2}; electric field-assisted sintering technique (FAST)
1 Introduction

Advanced SiC-based ceramic matrix composites (CMCs), such as carbon fiber reinforced SiC (Cf/SiC) and SiC fiber reinforced SiC (SiCf/SiC), have recently attracted increasing attention for aerospace and nuclear applications [1–4]. These materials possess superior high-temperature physical properties, oxidation resistance, low density, and excellent mechanical properties [5–7]. However, their manufacturing as large components with complex shapes is extremely difficult. Therefore, the application of CMCs will depend on a reliable joining method to assemble different simple-shaped pieces into large complex components [8–10].

Various techniques for the joining of CMCs have been developed, such as active metal brazing [11–13], solid-state diffusion bonding [14–16], ceramic precursor joining [17], and MAX phase (Mn+, A, X) joining method to assemble different simple-shaped pieces into large complex components [8–10].

Among these, MAX phases exhibit typical properties of both metals and ceramics, such as excellent high-temperature mechanical properties and oxidation resistance [26–28]. Therefore, Ti3SiC2 (one of the typical MAX phase) has been utilized as the joining filler for SiC-based materials [29,30]. However, a mismatch between coefficients of thermal expansion (CTE) of SiC substrates (~4.4×10⁻⁶ K⁻¹) [5] and Ti3SiC2 joining layer (~9.1×10⁻⁶ K⁻¹) [21] leads to the formation of cracks in the joining area. The presence of cracks would cause a leakage of transmutation gas if such crack-containing joints were used in nuclear reactor applications.

RE3Si2C2 (where RE represents a rare earth element) is a relatively new group of ternary layered materials, which has recently attracted increasing attention due to the ability of these materials to form a liquid phase by the eutectic reaction with SiC [31–33]. In Refs. [34,35], a pair of monolithic SiC ceramics were near-seamlessly joined with a RE film as the initial joining layer. During joining, Ti3SiC2 ceramic was used as a joining material for Cf/SiC. For the first time, the near-seamless Cf/SiC joints were obtained by the arc-melting of the Y, Si, and C powder mixture, followed by its annealing at 900 °C in an evacuated silica tube furnace for 30 d. This approach was too demanding on time, thereby not cost-effective. In our recent work, Y3Si2C2 ceramic (with 5.5 wt% Y2O3) was successfully fabricated by the spark plasma sintering (SPS) at 1500 °C for 30 min [37]. It was an effective approach to synthesize Y3Si2C2 ceramics, but a relatively large grain size (~8.8 μm) would inhibit the interfacial reactions if such a material was used as the joining layer between a pair of Cf/SiC. Therefore, some other techniques must be considered to synthesize Y3Si2C2 ceramics, which would be appropriate as a joining filler. High-temperature molten salt approach can provide a high solvation power for metal particles, which facilitates mass transport and nucleation processes [38]. This approach has been successfully used to synthesize a series of new MAX phases [39].

In the present work, such a molten salt approach was employed to synthesize Y3Si2C2 powder, which was then used as a joining material for Cf/SiC. For the first time, the near-seamless Cf/SiC joints were obtained using Y3Si2C2 slurry as the joining layer via an electric field-assisted sintering technique (FAST). The interface reactions, phase evolution, and shear strength (τ) of the joints were investigated. The joining mechanism was proposed based on the interfacial reactions occurring during joining, according to the calculated Y–Si–C phase diagram.
2 Experimental

2.1 Starting materials

The two-dimensional (2D) C\textsubscript{f}/SiC composites were fabricated using a plain carbon cloth of 1K polyacrylonitrile (PAN)-based carbon fibers (Toray T300). Polycarbosilane (PCS) with a molar weight of ~1300 g/mol was used as the precursor of SiC matrix. The PCS/xylene solution with a mass ratio of 1:1 was used to prepare the C\textsubscript{f}/SiC composites. The carbon fiber cloth was cut into 200 mm × 200 mm pieces. Twenty pieces of the cloth were stacked and stitched in Z direction with 2 needles/cm. The preform was then infiltrated in a vacuum and pyrolyzed under inert gas. The details about the processing were described in Ref. [40]. The density and open porosity of the C\textsubscript{f}/SiC composites were ~1.93 g/cm\textsuperscript{3} and ~13.5%, respectively. Except C and SiC, no other phases were detected in the C\textsubscript{f}/SiC composites by the X-ray diffraction (XRD).

The τ of the C\textsubscript{f}/SiC composites measured perpendicularly to Z direction was ~28 MPa. The cuboid-shaped C\textsubscript{f}/SiC blocks with the dimensions of 6 mm × 15 mm × 15 mm were used as the substrate materials for joining.

2.2 Joining process

The as-obtained Y\textsubscript{3}Si\textsubscript{2}C\textsubscript{2} powder was synthesized using a molten salt technique. The powder of YH\textsubscript{2} (Sinopharm Chemical Reagent Co., Ltd., Shanghai, China; purity = 99.5%, mean particle size = 75 μm), β-SiC (Eno Material, Qinhuangdao, China; purity = 99.5%, mean particle size = 0.5 μm), and NaCl and KCl (Sinopharm Chemical Reagent Co., Ltd., Shanghai, China; purity = 99%, mean particle size = 1 μm) were used as raw materials. The powders with the composition of YH\textsubscript{2}:SiC = 3:2 and NaCl:KCl = 1:1 (molar ratio) were mechanically mixed using an agate mortar under Ar atmosphere in a glove box. The powder mixtures were heated at 900 °C for 5 h in a molten salt furnace at a heating rate of 8 °C/min under Ar atmosphere. The Y\textsubscript{3}Si\textsubscript{2}C\textsubscript{2} powder was washed and filtered using deionized water, and further dried at 50 °C in a vacuum furnace.

2.3 Joint characterization

The microstructure and phase evolution of the samples were characterized by a scanning electron microscope (SEM; Quanta 250 FEG, FEI, USA) and a transmission electron microscope (TEM; Talos F200X, Thermo Fisher Scientific, USA), equipped with an energy dispersive spectroscopy (EDS) detector. The samples for TEM observations were machined using the focused ion beam (FIB) technique (Auriga, Carl Zeiss, Germany). The crystalline phases in the samples were identified by the X-ray diffractometer (D8 Advance, Bruker AXS, Germany) using Cu Kα radiation under 40 kV and 30 mA.

Fig. 1 (a, b) Schematic illustrations of the joining assembly and the joining set-up using FAST, respectively; (c) optical image of the joined sample showing the solidified liquid on the side of the joint; and (d) schematic diagram of the shear strength measurement set-up.

C\textsubscript{f}/SiC was placed on the Y\textsubscript{3}Si\textsubscript{2}C\textsubscript{2}-coated C\textsubscript{f}/SiC counterpart, and the assembly was placed between two graphite punches with a diameter of 20 mm, while no graphite die was used (Figs. 1(a) and 1(b)). The joining process was performed using a SPS machine (HP D25, FCT Systeme GmbH, Germany). \(T_j\) was measured at the interior bottom surface of the top graphite punch (~2 mm away from the sample surface) using an axial arranged infrared pyrometer, as shown in Fig. 1(b). The solidified liquid, which was squeezed out of the joining interface, is shown on the side surface of the joint in Fig. 1(c). The joining was performed at the temperatures of 1500, 1550, 1600, 1650, and 1700 °C for 10 min under Ar atmosphere. The applied uniaxial pressure was 30 MPa. Both the heating and cooling rates were 100 °C/min.
The $\tau$ of the joints was tested using a laboratory made assembly (Fig. 1(d)) by a universal electromechanical testing system (CMT5105, MTS, USA). A cross-head speed of 0.5 mm/min was used. At least five samples were measured per each joining parameter, and the mean value was calculated. The $\tau$ of each joint was calculated using Eq. (1):

$$\tau = \frac{F}{S}$$  (1)

where $F$ is the maximum load (N) at failure and $S$ is the actual joining area (mm$^2$).

3 Results and discussion

3.1 Phase composition and microstructure of Y$_3$Si$_2$C$_2$

The XRD pattern of the Y$_3$Si$_2$C$_2$ powder synthesized by the molten salt route at 900 °C is shown in Fig. 2(a). Except the main Y$_3$Si$_2$C$_2$ phase (JCPDS Card No. 51-0835), a small amount of Y$_2$O$_3$ (JCPDS Card No. 43-1036) was also detected. The SEM image of the Y$_3$Si$_2$C$_2$ powder having the plate-like morphology is shown in Fig. 2(b). Figure 2(c) shows the corresponding elemental distributions of Y, Si, C, and O. Both the XRD pattern and the uniform distribution of Y, Si, and C in their respective atomic layers indicated that the Y$_3$Si$_2$C$_2$ phase was successfully synthesized. The mechanism of the formation of Y$_3$Si$_2$C$_2$ phase in the molten salts at a relatively low temperature (900 °C) can be described as follows: At the beginning, Y from YH$_2$ dissolved into the molten salt, and diffused to the surface of SiC particles. Then, Y$_3$Si$_2$C$_2$ was formed by Reaction (2):

$$3Y + 2SiC \rightarrow Y_3Si_2C_2$$  (2)

The similar mechanism has also been reported for the formation of MAX phases via the molten salt method [38,41]. The surface energy of the reactants can be significantly improved by the polarization effect of the molten salt, which can decline the potential barrier of the formation of Y$_3$Si$_2$C$_2$. Moreover, the liquid molten salt, as a reaction medium, can remarkably increase the diffusivity of Y, Si, and C atoms, which may facilitate the growth of Y$_3$Si$_2$C$_2$ phase. A small amount of Y$_2$O$_3$ (observed on the XRD pattern in Fig. 2(a)) could have been formed by the oxidation of YH$_2$ with oxygen in Ar gas and/or oxygen on the surface of SiC powder.

3.2 Microstructure analysis of the joining layer

Figure 3 shows both the low- and high-magnification back-scattered electron (BSE) SEM images of the cross sections of the samples joined at different temperatures, and their corresponding EDS elemental mappings. The specimen joined at 1500 °C failed during machining, it was not investigated further. No obvious joining interface between the two pieces of Cf/SiC composites was observed (Figs. 3(a)–3(h)). This clearly indicates that the near-seamless Cf/SiC joints were obtained. It was observed that the Y-containing phase diffused into the pores of Cf/SiC near the interface (Figs. 3(a) and 3(e)), but this became gradually less significant with the increasing $T_j$ up to 1700 °C (Figs. 3(d) and 3(h)).

Furthermore, the EDS mapping clearly showed that some C-rich zones were present near the joining area, as highlighted by the yellow frames in Fig. 3(i). As a possible explanation, we propose that it might have been formed by squeezing of the liquid phase out of the joining area, resulting in the formation of C-rich precipitates near the interface. According to the calculated Y–Si–C ternary phase diagram, a liquid phase is formed by a ternary eutectic reaction among Y$_3$Si$_2$C$_2$, $\gamma$(Y–C) phase, and SiC at the temperature of ~1560 °C [31]. Some free carbon, which is usually present in Cf/SiC composites after the pyrolysis of PCS, could diffuse to the interface during joining. Since the liquid phase was formed at the interface, the C-rich zone could also be formed near the interface after the
liquid had been squeezed out of the joining area. A similar C-rich zone was also observed in the SiC joints with a SiC/Al₄SiC₄ joining layer [42].

We suppose that the near-seamless joining can be explained by the densification of the newly-formed SiC particles, which were precipitated from the liquid, with the Cf/SiC matrix. The liquid phase, which was in-situ formed by the reaction between the Y₃Si₂C₂ joining filler and the SiC matrix, first assisted densification, but its remaining volume was later squeezed out of the interface under the effect of external uniaxial pressure (~30 MPa). Indeed, the grain growth and densification of the precipitated SiC can be observed on the fracture surfaces of the joining layer after the shear strength measurements (Fig. 4). When $T_j = 1600 \, ^\circ C$, the joints failed in the matrix, so the morphology of the joining layer could not be observed directly by SEM. Therefore, the TEM analysis of the interface was performed for the sample joined at 1600 \, ^\circ C$. 

![Fig. 3](image_url)  
Fig. 3  Low- and high-magnification BSE images of the samples joined at different temperatures: (a, e) 1550 \, ^\circ C; (b, f) 1600 \, ^\circ C; (c, g) 1650 \, ^\circ C; (d, h) 1700 \, ^\circ C; and (i) their corresponding EDS elemental mappings.
It can be clearly observed that the amount of Y-containing phase (bright phase) in the joining layer gradually decreased with the increase of temperature, as shown in Figs. 4(a)–4(c). When $T_j = 1550 \degree C$, some Y-containing phases remained in the joining layer (Fig. 4(a)). It showed a porous and ribbon-like morphology (Figs. 4(d) and 4(e)) with a thickness of ~500 nm and a length of ~3 $\mu$m. The EDS analysis (point 1 in Fig. 4(d)) suggested that it could be a Y–C–O phase (Table 1). The precipitated spherical SiC with the grain size of ~1.5 $\mu$m (point 2 in Fig. 4(e) and Table 1) could be observed.

When $T_j$ increased to 1650 $\degree C$ (Fig. 4(f)), the residual Y-containing phase (point 4 in Fig. 4(f)) was mostly located at the grain boundaries of SiC (point 3 in Fig. 4(f)). Since the viscosity of a liquid phase decreased with the increasing temperature, a greater amount of liquid could be more easily squeezed out of the joining area under the effect of a uniaxial pressure. At the same time, some liquid infiltrated into the pores in the C/SiC matrix, and solidified upon cooling.

When $T_j$ further increased to 1700 $\degree C$, a very small amount of Y-containing phase remained at the grain boundaries of SiC (Fig. 4(g)). The grain size of the precipitated SiC further increased to 3–6 $\mu$m. Near fully dense microstructure was observed on the fracture surface (Fig. 4(c)). This clearly confirms that the density of the joining layer significantly increased with the increasing $T_j$.

To investigate the microstructures and phase compositions of the joining layers, the TEM analysis was performed for the sample joined at 1600 $\degree C$. Figure 5 shows a high angle annular dark field (HAADF) image of the joining layer of the sample joined at 1600 $\degree C$ and the corresponding elemental distributions of C, Si, Y, and O. No initial Y$_3$Si$_2$C$_2$ joining filler was found in the joining layer, confirming that the near-seamless joint was obtained. A “nail like” interface was observed because of the reactions between the initial Y$_3$Si$_2$C$_2$ and C$_f$. The “nail like” morphology was reported to be beneficial to improve the $\tau$ of the joints [43]. The bright Y-containing phase

![Fig. 4](image-url) Fracture surfaces of the samples joined at different temperatures: (a, d, e) 1550 $\degree C$; (b, f) 1650 $\degree C$; and (c, g) 1700 $\degree C$.

| No. | Composition (at%) | Possible phase    |
|-----|-------------------|-------------------|
| 1   | 3.4 12.6 55.5 28.5 | Y–C–O            |
| 2   | — 18.5 79.8 1.7   | SiC + C          |
| 3   | — 29.2 70.8 —     | SiC + C          |
| 4   | 39.4 13.5 34.5 12.6 | Y$_3$Si$_2$C$_2$ |
| 5   | 1.2 8.1 90.7 —     | C                |

Table 1 EDS results of points 1–5 in Fig. 4

A plate-like C was detected by the EDS analysis, as shown by point 5 in Fig. 4(f) (Table 1). The presence of the precipitated SiC grains with a size of 2–3 $\mu$m suggested that the grain growth occurred during the joining at a higher temperature (when compared to ~1.5 $\mu$m SiC grains after being joined at 1550 $\degree C$). At the same time, the analysis of fracture surfaces revealed that the microstructure of the joining area was relatively dense (Fig. 4(b)).
was mainly observed at the grain boundaries of SiC, but also at the interface between SiC and Cf and/or precipitated C. In order to further investigate the microstructures and the phase compositions of Cf, precipitated SiC, and Y-containing phase, high-resolution transmission electron microscopy (HR-TEM) with the EDS analysis was employed, as shown in Figs. 6 and 7, and Table 2.

The semi-quantitative EDS analysis of points 1–8 in Figs. 6(a) and 6(d) is shown in Table 2. The presence of carbon fibers in the formation of graphite single crystals was confirmed by the HR-TEM (black square in Fig. 6(b)) analysis, the corresponding EDS results of point 1 and SAED pattern shown in the inset of Fig. 6(b). The lattice fringe spacing of 0.333 nm was assigned to the (002) planes of graphite. The TEM analysis also clearly confirmed the presence of precipitated C-rich area (Figs. 6(d)–6(f)), as the SAED pattern revealed a polycrystalline structure in the inset of Fig. 6(e).

In addition, two different types of Y-containing phase, Y–Si–C (points 3 and 6 in Fig. 6) and Y–C–O (points 2, 5, and 7 in Fig. 6), were detected by the EDS analysis. Figure 7 shows the HR-TEM image of the typical Y-containing phases in a triple junction (yellow square in Fig. 6(b)). The lattice fringe spacing of 0.248 nm (Fig. 7(b)) and 0.247 nm (Fig. 7(c)) were assigned to the (231) plane of $Y_3Si_2C_2$ and (200) plane of $Y_2OC$, respectively. This was consistent with the EDS results (points 2 and 3 in Table 2). These results confirmed that a major part of the initial $Y_3Si_2C_2$ interlayer decomposed during joining, but a very small amount was still present in the joining area.

3.3 $\tau$ of the joints

Figure 8 shows the $\tau$ of the Cf/SiC samples joined with the $Y_3Si_2C_2$ layer at different temperatures. The $\tau$ of the samples joined at 1550 °C was 16.4±3.4 MPa. As described above, the precipitated SiC grains were not yet fully densified with the Cf/SiC matrix at this temperature, resulting in the porous joining layer, as shown in Figs. 4(a) and 4(d). Since the $\tau$ of the porous interface was lower than that of the Cf/SiC matrix, the failure mainly occurred at the interface, as shown in Fig. 9(a).

The $\tau$ of the samples joined at 1600 °C was 17.2±2.9 MPa. At this temperature, the precipitated SiC grains were fully densified and well bonded with the Cf/SiC matrix, resulting in the formation of a robust interface. Although all of the tested joints failed in the Cf/SiC matrix (Fig. 9(b)), the $\tau$ of the joints was still lower than that of the as-received unjoined Cf/SiC (~28 MPa). This can be attributed to the fact that carbon fibers were slightly damaged due to their reaction with the $Y_3Si_2C_2$ filler during joining (Fig. 10(a)). According to the Y–Si–C phase diagram (Fig. 11), $Y_3Si_2C_2$ reacts with C to generate the Y–C phase [31]. At the same time, the applied uniaxial pressure of ~30 MPa could also cause some damage to the Cf/SiC matrix [44], resulting in the deterioration of the $\tau$ of the Cf/SiC joints.

When $T_j$ further increased to 1650 and 1700 °C, the $\tau$ of the joints sharply decreased to 9.5±2.4 and 8.6±3.7 MPa, respectively. The samples mainly failed at the interface, but partially also in the matrix near the
Fig. 6  (a, b) HAADF and bright field TEM images of carbon fiber and (c–f) precipitated carbon zones are the HR-TEM images of the area highlighted by the black squares in (b) and (e). The insets in (b) and (e) are the selected area electron diffraction (SAED) patterns of the area shown by a red circle.

Fig. 7  HR-TEM images of the two different Y-containing phases located in a triple junction: (a) bright field image; (b) HR-TEM images of the highlighted yellow squares in (a).

Table 2  EDS results of points 1–8 in Fig. 6

| No. | Composition (at%) | Possible phase |
|-----|------------------|----------------|
| 1   | —                | Cr             |
| 2   | 29.6 1.8        | Y–C–O          |
| 3   | 10.2 37.6       | Y₃Si₂C₂        |
| 4   | — 65.1          | SiC            |
| 5   | 15.2 4.9        | Y–C–O          |
| 6   | 20.4 37.1       | Y₃Si₂C₂        |
| 7   | 16.9 2.6        | Y–C–O          |
| 8   | — 0.8           | Cr             |

Fig. 8  τ of the C₇/SiC samples joined with the Y₃Si₂C₂ filler at different temperatures.

Joining interface (Fig. 9(c)). Such a significant decrease in τ can be well explained by a significant degradation of carbon fibers (Fig. 10(b)). This was caused by a sharp increase of the reaction degree between Y₃Si₂C₂ and C as the temperature increased to 1650 °C.

In addition, the τ of the joints measured in the present work was higher than that of the C₇/SiC (with M40J 3K carbon fibers) joined with Ti₃SiC₂ (5.5±1.8 MPa), and comparable with that of the C₇/SiC (with T300 1K carbon fibers) joined with Ti₃SiC₂ (19.1±2.0 MPa) [45]. On the other hand, the τ
of the present joints was lower than that of the chemical vapor deposition (CVD)-SiC coated C_{f}/SiC joined with Ti_{3}Si_{2}C_{2} (31.1±4.0 MPa) [45]. This indicates that a CVD-SiC coating on the C_{f}/SiC surface apparently acted as a protective layer to protect C_{f} from damage, which led to the improvement of the τ of joints.

3.4 Joining process discussion

Considering the calculated Y–Si–C ternary phase diagram, the interface reactions and the joining process can be described as follows. At a low T_{j} (1500–1550 °C), Y_{3}Si_{2}C_{2} reacts with C_{f} by Reaction (3) [31]:

\[ 6C + Y_{3}Si_{2}C_{2} \rightarrow 3YC_{2} + 2SiC \]  \( (\Delta G = -36,326 \text{ J/mol}, T_{j} = 1500 \text{ °C}) \)  (3)

At around 1560 °C, Y_{3}Si_{2}C_{2} transforms to a liquid phase by the ternary eutectic reaction with the in-situ formed γ(Y–C) phase and SiC (Fig. 11):

\[ L \leftrightarrow SiC + \gamma + Y_{3}Si_{2}C_{2} \]  (4)

Furthermore, Y_{3}Si_{2}C_{2} melts at around 1590 °C (Fig. 11), while thermodynamically more stable γ(Y–C) phase and Si–C phases are generated. The Y–C–O phase can be formed by the oxidation of Y–C with the oxygen impurity in C_{f}/SiC and/or Ar gas. At the same time, new SiC particles precipitate from the liquid upon cooling. Thus, both the Y–C–O and SiC phases were detected on the fracture surfaces of the samples joined at 1550 and 1600 °C. The precipitated SiC grains and the ribbon-like Y–C–O were not fully consolidated with the C_{f}/SiC. The τ of the joints was relatively low, as a consequence of the porous joining layer and a weak bonding interface.
However, in the case of the C_f/SiC joined at 1600 °C, a large amount of liquid was formed, which significantly promoted the densification of the precipitated SiC grains with the C_f/SiC matrix. At the same time, the excess of the liquid was squeezed out of the joining interface under the effect of uniaxial pressure. Thus, a sound near-seamless C_f/SiC joint was obtained.

As the $T_j$ increased to 1650–1700 °C, carbon fibers were significantly damaged due to their reactions with the Y_3Si_2C_2 joining filler, which led to a significant decrease of the $\tau$ of the joints.

### 3.5 Perspectives

In summary, the near-seamless C_f/SiC joints were successfully obtained via FAST at 1600 °C using Y_3Si_2C_2 as the initial joining layer. The formation of the near-seamless C_f/SiC joints may definitely solve the issues with thermal mismatch and irradiation-induced swelling mismatch that inherently exist between the C_f/SiC matrix and the traditional joining filler, such as MAX phases, Ti foil, and glass–ceramic. On the other hand, harsh interface reactions between C_f and Y_3Si_2C_2 at high temperatures caused significant damage to C_f, resulting in a deterioration of the mechanical properties of the C_f/SiC matrix. As a result, although the interface was stronger than the matrix, the overall $\tau$ of the joints was lower than that of the unjoined C_f/SiC composites. This issue should be considered while designing C_f/SiC composites, so it may be necessary to protect the surface of C_f by a protective coating, such as BN, pyrolytic carbon (PyC), SiC, and/or PyC/SiC multilayer coatings. Another approach would be to further decrease $T_j$ to obtain the near-seamless joints by developing a new RE_3Si_2C_2 joining material with potentially lower temperature of liquid phase formation via the eutectic reaction between RE_3Si_2C_2 and SiC.

### 4 Conclusions

The near-seamless C_f/SiC joints were successfully obtained via the electric FAST using the as-synthesized ternary layered Y_3Si_2C_2 material as the initial joining filler. The Y_3Si_2C_2 material was synthesized by a novel molten salt approach at the temperature of 900 °C, and then used in the form of ethanol-based slurry for the joining of C_f/SiC. During joining, a desired liquid phase was in-situ formed at the joining interface via the eutectic reaction among Y_3Si_2C_2, γ(Y–C) phase, and SiC. In the next stage, new SiC grains precipitated from the liquid. The sound near-seamless C_f/SiC joints were obtained by near fully densification of the precipitated SiC grains with the C_f/SiC matrix. The in-situ formed liquid phase first improved the densification, while its excess volume was squeezed out of the joining interface under the effect of uniaxial pressure (30 MPa). The $\tau$ of the near-seamless joints made at 1600 °C was 17.2±2.9 MPa, and the fracture always occurred in the C_f/SiC matrix, away from the joining interface. On the other hand, the further increase of $T_j$ to 1650 and 1700 °C resulted in a deterioration of $\tau$, due to the carbon fiber degradation. The proposed concept for the formation of the near-seamless C_f/SiC joints using RE_3Si_2C_2 could also be applied for the joining of SiC-based CMCs.

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