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Fabrication of SiCw/Ti3SiC2 composites with improved thermal conductivity and mechanical properties using spark plasma sintering

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Abstract: High strength SiC whisker-reinforced Ti3SiC2 composites (SiCw/Ti3SiC2) with an improved thermal conductivity and mechanical properties were fabricated by spark plasma sintering. The bending strength of 10 wt% SiCw/Ti3SiC2 was 635 MPa, which was approximately 50% higher than that of the monolithic Ti3SiC2 (428 MPa). The Vickers hardness and thermal conductivity (k) also increased by 36% and 25%, respectively, from the monolithic Ti3SiC2 by the incorporation of 10 wt% SiCw. This remarkable improvement both in mechanical and thermal properties was attributed to the fine-grained uniform composite microstructure along with the effects of incorporated SiCw. The SiCw/Ti3SiC2 can be a feasible candidate for the in-core structural application in nuclear reactors due to the excellent mechanical and thermal properties.

Keywords: SiC whisker; Ti3SiC2; mechanical property; thermal conductivity; spark plasma sintering

1 Introduction

Ti3SiC2 is a typical MAX phase material (M: early transition metal, A: group A element, and X: C or N), which was developed in 1967 [1]. Because Ti3SiC2 unit cell consists of alternating covalently-bonded Ti6C octahedra and planar metallic-bonded Si layers along the c-axis direction [2], it shows combined properties of ceramics and metals, such as high elastic modulus, fracture toughness, thermal shock, and corrosion resistance [2–7]. However, the flexural strength (260–450 MPa based on four-point bending test), Vickers hardness (4 GPa), and thermal conductivity (k) of Ti3SiC2 are known to be insufficient for nuclear reactor applications, which are inferior to those of SiC [4].

One feasible way to compensate these weaknesses is the incorporation of strengthening filler, which possesses good mechanical and thermal properties [8]. Among many possible fillers, because SiC is thermodynamically stable along with excellent mechanical properties and high k [9–13], the Ti3SiC2 reinforced with SiC particles has attracted much attention [14–22]. It was reported that the Vickers hardness, fracture toughness, k, and oxidation resistance of Ti3SiC2 could be improved by the addition of SiC particles [14–16]. However, the
room-temperature strength of Ti$_3$SiC$_2$ decreased by the addition of SiC due to the retained tensile stress in the Ti$_3$SiC$_2$, originating from the increased thermal expansion mismatch between SiC and Ti$_3$SiC$_2$ [14,16].

On the other hand, SiC whisker (SiC$_w$) can be a promising one-dimensional filler, which can strengthen the composite by crack deflection, bridging, and whisker pull-out, due to its excellent strength combined with high aspect ratio [23]. Indeed, the bending strength of Al$_2$O$_3$/Ti$_3$SiC$_2$ was improved to 688 MPa when 20 wt% SiC$_w$ was added [24]. Furthermore, the presence of SiC$_w$ in the SiC$_w$/TiC/Ti$_3$SiC$_2$ was reported to increase the room-temperature strength. Hashimoto et al. [24] reported a very high bending strength of 1 GPa for the Ti$_3$SiC$_2$ containing 15 vol% SiC$_w$, which was explained by the preferential grain orientation, TiC precipitates, and the effects of SiC$_w$. However, the presence of TiC is not desirable because it deteriorates the oxidation resistance of Ti$_3$SiC$_2$ [26–28]. Moreover, the Ti–Si intermetallic compounds, which are formed simultaneously upon the decomposition of Ti$_3$SiC$_2$ during sintering, are not desirable for nuclear applications due to their poor corrosion resistance [29].

Therefore, compared to SiC particles, one-dimensional SiC whiskers have remarkable effects on the improving flexural strength of ceramics. This study aimed to investigate the effects of SiC$_w$ on the reinforcement of Ti$_3$SiC$_2$. The SiC$_w$/Ti$_3$SiC$_2$ composites were fabricated using a heteropolar coagulation method followed by spark plasma sintering, which can minimize the decomposition of Ti$_3$SiC$_2$. The mechanical and thermal properties of SiC$_w$/Ti$_3$SiC$_2$, as a function of SiC$_w$ content, were analyzed in comparison with the monolithic Ti$_3$SiC$_2$.

2 Experimental

2.1 Starting materials

Commercial Ti$_3$SiC$_2$ powder with 98.5% purity was used as a starting material (Beijing Jinhezhi Materials Co., China), which contains Al$_2$O$_3$ and TiC impurities. SiC$_w$ with a diameter of 0.4–0.9 μm and a length of 6–120 μm was used as a reinforcement (Union Materials Co., Republic of Korea), while polyethyleneimine (PEI, molecular weight ≈ 10,000, Aladdin Co., China) was used as a cationic dispersant to improve the uniform distribution of SiC$_w$ in the Ti$_3$SiC$_2$ matrix.

2.2 Composite fabrication

SiC$_w$ and Ti$_3$SiC$_2$ particles were added into water containing 1.5 wt% PEI, while the pH of the solution was adjusted to 3 using HCl to achieve a uniform dispersion. The pH of the suspension was then increased to 9 by adding NH$_4$OH followed by ultrasonication for 45 min, where a uniform dispersion by heterocoagulation between the Ti$_3$SiC$_2$ particles and SiC$_w$ could be achieved due to the opposite surface charge. After drying the suspension at 80°C in a vacuum, a uniform SiC$_w$ and Ti$_3$SiC$_2$ mixture could be obtained.

The mixture was consolidated at 1350°C for 20 min using a spark plasma sintering (SPS, HP D25/1, FCT System GmbH, Germany) in an Ar atmosphere after loading the powder into a graphite die with an inner diameter of 60 mm. The heating and cooling rate was 50°C/min, and 50 MPa uniaxial pressure was applied during sintering. The amount of SiC$_w$ was controlled at 1 wt%, 5 wt%, and 10 wt%, which corresponded to 1.24 vol%, 6.2 vol%, and 12.4 vol%, respectively, in the SiC$_w$/Ti$_3$SiC$_2$, while a pure monolithic Ti$_3$SiC$_2$ was also fabricated at the same condition for comparison.

2.3 Characterization

The phases in the composite were analyzed with a X-ray diffractometer (XRD, D8 Advance, Bruker AXS, Germany) using Cu Kα radiation under an operating voltage 40 kV and current 30 mA. The microstructure of SiC$_w$, powder mixture, and sintered specimen was observed with a scanning electron microscope (SEM, Hitachi S-4800, Japan) equipped with an energy dispersive spectroscope (EDS). A transmission electron microscope (TEM, JEM-2010) equipped with an EDS was also utilized to analyze the interface between SiC$_w$ and Ti$_3$SiC$_2$ matrix.

A zeta potential analyzer (Zetasizer Nano ZS, Malvern, UK) was used to measure the surface charge of the SiC$_w$ and Ti$_3$SiC$_2$ in deionized water. The pH was controlled at 3–12 with or without PEI addition. The composite density was measured using an Archimedes’ method. Three-point bending strength was measured for rectangular specimen (20 mm × 4 mm × 2 mm) after polishing the sample surface using a diamond suspension. A universal electromechanical testing system (CMT5105, MTS, USA) was used for the strength test at room temperature with a cross-head speed of 0.5 mm/min. At least 5 samples were tested for each type of specimen. The indentation test was
performed to measure the Vickers hardness at room temperature using a Vicker’s diamond indenter (HV-1000; Shanghai Lianer Testing Equipment Co., China). An indentation load \(P\) of 9.8 N and a dwell time of 15 s were used. At least 20 indents were tested for each specimen composition, and the mean value and standard deviation were calculated.

A laser flash method (LFA 457, Netzsch, Germany) was used to measure the thermal diffusivity and specific heat capacity \(C_p\). The measurement was performed for the compacts having a diameter of 12.7 mm and a thickness of 2 mm in a nitrogen atmosphere at room temperature. The \(k\) \(\text{W/(m·K)}\) was calculated from Eq. (1):

\[
k = \frac{\lambda}{\rho C_p}
\]

where \(\lambda\) is the thermal diffusivity coefficient and \(\rho\) is the density of the material.

3 Results and discussion

3.1 SiC whisker dispersion in Ti₃SiC₂

The uniform distribution of SiCₕ in Ti₃SiC₂ is believed to enhance the mechanical and thermal properties of the final composite. Because PEI is a cationic surfactant with a high molecular weight, it can facilitate the dispersion of SiCₕ by both electrostatic and steric mechanism [30]. Figure 1 shows the zeta potential behavior of SiCₕ in an aqueous system with different amounts of PEI as a function of pH. The zeta potential of SiCₕ was increased from 22 to 58 mV at pH = 3 by adding 1.5 wt% PEI. Because the zeta potential of Ti₃SiC₂ (25 mV) was similar to that of SiCₕ without PEI at pH = 3, firstly SiCₕ was dispersed in water after adding 1.5 wt% PEI, and then the Ti₃SiC₂ powder was added followed by the pH adjustment to 3 to obtain the uniform dispersion of SiCₕ and Ti₃SiC₂.

The pH of the suspension was then increased to 9 by adding NH₄OH, followed by ultrasonication for 45 min. A uniform mixing between the Ti₃SiC₂ (zeta potential = –30 mV) and SiCₕ (zeta potential = +25 mV) was obtained by heteropolar coagulation originated from opposite surface charge. The insets in Fig. 1 compare the image of SiCₕ/Ti₃SiC₂ dispersion at pH = 3 and 9.

3.2 Microstructure of SiCₕ/Ti₃SiC₂ composites

The relative density for SiCₕ/Ti₃SiC₂ with different SiCₕ contents after SPS at 1350 ℃ is shown in Fig. 3, where the density \(\geq 98.4\%\) was obtained for all samples, regardless of the amount of SiCₕ. The theoretical composite density was calculated by considering the relative amount of SiCₕ and Ti₃SiC₂.

Figure 4 shows the XRD patterns of SiCₕ/Ti₃SiC₂ and pure Ti₃SiC₂ after SPS at 1350 ℃, which indicates the presence of Ti₃SiC₂ (JCPDS No. 65-3559) and SiC.
Fig. 4 XRD patterns of the SiC_w/Ti_3SiC_2 and pure Ti_3SiC_2 for 2θ = (a) 20°–80° and (b) 32°–45° after SPS at 1350 °C.

(JCPDS No. 29-1129). No trace for the second phase, such as TiC and intermetallic Ti-Si compounds, is detected. Moreover, the Ti_3SiC_2 in both monolith and composites seems to have a random orientation, based on a magnified XRD pattern shown in Fig. 4(b), showing characteristic peaks for various planes. This is opposite to the result for the SiC_w/TiC/Ti_3SiC_2 system from Hashimoto et al. [25], where the Ti_3SiC_2 grains were preferentially aligned. Due to the grain alignment along with the TiC precipitates, the SiC_w/TiC/Ti_3SiC_2 showed very high bending strength, especially when the composite contained 15 vol% SiC_w. However, the SiC_w/TiC/Ti_3SiC_2 revealed the decomposition of components during heat treatment, resulting in the formation of intermetallic Ti-Si compounds, such as TiSi_2 and Ti_5Si_3, which are deleterious for nuclear applications.

The reason for the decomposition of Ti_3SiC_2 in Ref. [25] was attributed to the flake-shaped Ti_3SiC_2 particles formed by ball milling. Because Ti_3SiC_2 contains TiC octahedrons, which are aligned to (111) plane, the flake-shaped Ti_3SiC_2 particles promote the formation of TiC by the decomposition of Ti_3SiC_2. The enhanced decomposition of Ti_3SiC_2 can also be achieved by the mechano-chemical activation using a high energy ball milling [31]. However, the SiC_w/Ti_3SiC_2 composites were fabricated by a soft hetero-coagulation process in this study, without exposing them to a severe milling. Therefore, no formation of TiC was observed in this study. Resultantly, the decomposition of Ti_3SiC_2 and the formation of TiC could be minimized because the Ti_3SiC_2 particles remained as randomly oriented polycrystalline.

The SiC_w phase distribution in the Ti_3SiC_2 matrix was observed from the polished-surface of the samples with different content of SiC_w. Figures 5(a)–5(c) show the SEM images for the SiC_w/Ti_3SiC_2 containing different amounts of SiC_w. The gray phase corresponds to Ti_3SiC_2 matrix, while the dark phase is SiC_w, according to the EDS analysis (not shown here). The SiC_w is distributed uniformly in the Ti_3SiC_2 matrix for the 1 wt% SiC_w/Ti_3SiC_2 sample. Some agglomeration of whiskers in the 5 wt% and 10 wt% SiC_w/Ti_3SiC_2 composites was observed (Figs. 5(b) and 5(c)). Figure 5(d) shows a high-angle annular dark-field (HAADF) image for a SiC_w embedded in the Ti_3SiC_2 matrix. The Ti_3SiC_2 and SiC_w can be distinguished by the intensities in the HAADF image, which are proportional to the square of atomic number [32]. No crack is found at the interface between SiC_w and Ti_3SiC_2 matrix.

Figures 5(e)–5(g) show the semi-quantitative EDS analysis results for points 1, 2, and 3 shown in Fig. 5(d). Only Si and C are detected for point 1, as shown in Fig. 5(e), which is the SiC_w. Ti_3SiC_2 is the phase present for point 2, while Al and O are detected at point 3, as shown in Fig. 5(g). Because Al was intentionally added to facilitate the synthesis of Ti_3SiC_2, it seems that Al is segregated at point 3 after oxidation into Al_2O_3. The presence of Al_2O_3 is reported to improve the oxidation resistance, strength, and fracture toughness of Ti_3SiC_2, while decreasing the possibility for the formation of TiC [33,34].

Both TEM and EDS results indicate that there is no obvious interfacial reaction between the SiC_w and Ti_3SiC_2 matrix. Spencer et al. [35] could not observe the reaction between SiC fibers and Ti_3SiC_2 up to 1550 °C either. This thermochemical stability between SiC_w and Ti_3SiC_2 would be beneficial for improving the mechanical and thermal properties of the composites.

Randomly oriented Ti_3SiC_2 grains can be observed with the fractured surface images for both monolithic Ti_3SiC_2 and SiC_w/Ti_3SiC_2, as shown in Fig. 6. Ti_3SiC_2 grains typically show plate shape because of the
layered TiC and Si structure. Abnormally grown Ti$_3$SiC$_2$ grains were observed for the pure Ti$_3$SiC$_2$ and composites containing up to 5 wt% SiC$_w$, as shown in Figs. 6(a)–6(c). The abnormal grain size of Ti$_3$SiC$_2$ decreased with increasing SiC$_w$ content. The very large abnormal grains seemed to be mostly impinging each other and dominated the microstructure of the monolithic Ti$_3$SiC$_2$, but as SiC$_w$ was added, they were reduced in terms of the occurrence and size so that they became isolated in the fine Ti$_3$SiC$_2$ matrix grains.

The largest abnormal grain size for monolithic Ti$_3$SiC$_2$ is 150 μm approximately, while those for SiC$_w$/Ti$_3$SiC$_2$ are 52, 56, and 19 μm for the SiC$_w$ content of 1 wt%, 5 wt%, and 10 wt%, respectively, when four low magnification SEM photomicrographs of random 0.25 mm × 0.22 mm area fracture surfaces (not shown here) were examined for each specimen. Notably, only 1 wt% SiC$_w$ addition suppresses the abnormal grain growth significantly similar to that of 5 wt% SiC$_w$ addition, which might be attributed to the good dispersion of SiC$_w$. The abnormal grain growth persisted until the SiC whisker content increased to 10 wt%, where a fine-grained uniform microstructure could be obtained (Fig. 6(d)). Moreover, the addition of SiC$_w$ seemed to affect not only the abnormal grain growth but the overall microstructure, which resulted in finer Ti$_3$SiC$_2$ matrix as more SiC$_w$ was added. Besides, the toughening mechanisms, such as the whisker bending and whisker pull-out as well as interface debonding could be observed in the high magnification SEM image of the fracture surface of the 10 wt% SiC$_w$/Ti$_3$SiC$_2$ sample (Fig. 7). It implies that the interface bonding strength between SiC$_w$ and Ti$_3$SiC$_2$ is not that strong.
3.3 Mechanical properties of SiC<sub>w</sub>/Ti<sub>3</sub>SiC<sub>2</sub> composites

3.3.1 Bending strength

Figure 8 presents the bending strength of the SiC<sub>w</sub>/Ti<sub>3</sub>SiC<sub>2</sub> composites at room temperature as a function of SiC<sub>w</sub> content. Considering the bending strength of the monolithic Ti<sub>3</sub>SiC<sub>2</sub> was 428 MPa, the strength of composite did not change much up to 5 wt% SiC<sub>w</sub>. On the other hand, the bending strength of composite containing 10 wt% SiC<sub>w</sub> reached to 636 MPa, which is significantly improved from the monolithic Ti<sub>3</sub>SiC<sub>2</sub> by 48%.

This strength behavior as a function of SiC<sub>w</sub> might be affected by the microstructure shown in Fig. 6, where the largest abnormal grain size decreased from 150 to 19 μm by increasing the SiC<sub>w</sub> content from 0 to 10 wt%. Because the critical flaw size generally decreases with the decrease in the largest grain size, it is reasonable that the SiC<sub>w</sub>/Ti<sub>3</sub>SiC<sub>2</sub> with 10 wt% SiC<sub>w</sub> showed the highest bending strength because the strength of ceramics is limited by the critical flaw size. Besides, the higher strength also resulted from the finer grain size according to the Hall–Petch relationship and Ref. [36]. Therefore, the uniform and fine grain sizes with 10 wt% SiC<sub>w</sub>, as shown in Fig. 6, seemed to contribute to the significantly improved strength. Hashimoto et al. [25] also reported a similar trend for the SiC<sub>w</sub>/TiC/Ti<sub>3</sub>SiC<sub>2</sub> by varying the content of SiC<sub>w</sub>. However, it is notable that the composites containing up to 5 wt% SiC<sub>w</sub> showed similar strength compared to monolithic Ti<sub>3</sub>SiC<sub>2</sub> despite of microstructural difference shown in Fig. 6. Therefore, the bending strength behavior of the composites cannot be explained with the grain size in the microstructural alone.

A few more factors other than the microstructure must be considered to better understand the effects of the SiC<sub>w</sub> on the composite strength. The crack bridging by the SiC<sub>w</sub> in the wakes behind the tips of cracks can be another reason (Fig. 9(d)), in addition to the fine microstructure, to increase the SiC<sub>w</sub>/Ti<sub>3</sub>SiC<sub>2</sub> composite strength [37]. On the other hand, there are two possible opposing effects of whiskers to decrease the strength of the composite. Ti<sub>3</sub>SiC<sub>2</sub> is non-cubic and has a layered structure, resulting in extensive grain bridging. That is why Ti<sub>3</sub>SiC<sub>2</sub> intrinsically has relatively high fracture toughness. However, the addition of SiC<sub>w</sub> decreased this matrix grain bridging effect significantly, as most of the large abnormal grains disappeared, and bridging grain sizes became much smaller, as shown in Fig. 6. The reduced grain bridging would harm the composite strength [38]. Besides, as Wan et al. [16] observed in the SiC-particulate composites, the strength of the composite could be decreased as a result of the residual tensile stresses in the matrix due to the thermal expansion mismatch with the SiC<sub>w</sub>.

Thus, the effects of SiC<sub>w</sub> on the strength of the SiC<sub>w</sub>/Ti<sub>3</sub>SiC<sub>2</sub> composites may be complicated by these various phenomena occurring at the same time. Therefore, for the composites containing 5 wt% SiC<sub>w</sub> or less, it is speculated that the strengthening effects of the smaller grain sizes and the crack bridging by the SiC<sub>w</sub> might have been balanced by the negative effects caused by the reduced matrix grain bridging and residual tensile stresses in the matrix. As a result, the strength of the composites containing up to 5 wt% SiC<sub>w</sub> was at the same level as that of the monolithic Ti<sub>3</sub>SiC<sub>2</sub>. When the SiC<sub>w</sub> content increased to 10 wt%, the fine-grained uniform microstructure and crack bridging by the SiC<sub>w</sub> (Fig. 9(d)) dominated over the negative effects, and hence the significant strengthening was observed.

The addition of 10 wt% SiC<sub>w</sub> in Ti<sub>3</sub>SiC<sub>2</sub> increased
the bending strength by 50% compared to the pure Ti$_3$SiC$_2$. However, the addition of particulate SiC to Ti$_3$SiC$_2$ or Ti$_3$Si(Al)C$_2$ resulted in the decrease of flexural strength [14–16], where the degree of strength reduction increased with the increase in SiC particle content. This difference in reinforcement between SiC particle and whisker comes from the toughening mechanisms by whiskers, such as crack bridging, whisker bending and whisker pull-out, and interfacial debonding.

### 3.3.2 Vickers hardness

Figure 9 shows SEM images of the surface after the indentation test for monolithic Ti$_3$SiC$_2$ and 10 wt% SiC$_w$/Ti$_3$SiC$_2$. The indent shape was irregular with exfoliated surfaces and deformed particles for monolithic Ti$_3$SiC$_2$, as shown in Figs. 9(a) and 9(b). Even though the extended cracks are usually observed at the corners of the indent for brittle ceramics, such cracks were difficult to found in this study, owing to the ductile nature of Ti$_3$SiC$_2$.

The deformation around the indent was asymmetric, while the damaged region extended approximately 50% from the indent. Some specimens showed exfoliated directional Ti$_3$SiC$_2$ grains around the indent. When the basal Ti$_3$SiC$_2$ plane was parallel to the load, laminated fracture due to the kink, delamination, and slipping was observed, as shown in Fig. 9(b). Besides, exfoliation occurred when the basal plane of Ti$_3$SiC$_2$ grain was perpendicular to the shear stress, as shown in Fig. 9(a). The deformation during the indentation of Ti$_3$SiC$_2$ is generally caused by the basal plane slipping and debonding at the Ti$_3$C$_2$ octahedron and Si layer interface [4,15].

The incorporation of SiC$_w$ decreased the indent size, as shown in Fig. 9. Although the basic deformation behavior was similar to that of the monolithic Ti$_3$SiC$_2$, the degree of grain buckling and fracturing of the SiC$_w$/Ti$_3$SiC$_2$ is less severe, probably due to finer grain size and the presence of whisker. Notably, the presence of SiC$_w$ did not decrease significantly the ductile nature of Ti$_3$SiC$_2$. The interfacial debonding between SiC$_w$ and Ti$_3$SiC$_2$ matrix seemed to have similar effects on the mechanical properties of SiC$_w$/Ti$_3$SiC$_2$ to the basal plane slipping and debonding between the Ti$_3$C$_2$ octahedron and Si layer occurring in the Ti$_3$SiC$_2$ unit cell.

Figure 10 shows the Vickers hardness of SiC$_w$/Ti$_3$SiC$_2$ as a function of SiC$_w$ content. The hardness of 5.8 GPa for pure Ti$_3$SiC$_2$ increased to 7.9 GPa for the 10 wt% SiC$_w$/Ti$_3$SiC$_2$ composite. The relatively low hardness of pure Ti$_3$SiC$_2$ is attributed to the weak bonding between the Ti–C–Ti–C–Ti chain and Si layer [39,40]. Continuous increase in hardness was found with the increase in SiC$_w$ content, as shown in Fig. 10. The significant microstructural change by adding only 1 wt% SiC$_w$, as shown in Fig. 5, would be the reason for a drastic increase in hardness from pure Ti$_3$SiC$_2$ to 1 wt% SiC$_w$/Ti$_3$SiC$_2$. Because hardness is known to follow the Hall–Petch relationship for microcrystalline ceramics [41], the finer microstructure of 1 wt% SiC$_w$/Ti$_3$SiC$_2$ compared to the pure Ti$_3$SiC$_2$ caused the increase in hardness. By increasing the amount of SiC$_w$ further, however, the overall grain size gradually decreased, showing a modest linear increase in hardness, as shown in Fig. 10.

### 3.4 $k$ of SiC$_w$/Ti$_3$SiC$_2$ composites

$k$ is very important for nuclear application of SiC$_w$/Ti$_3$SiC$_2$, which directly influences the energy exchange efficiency of the reactor. Figure 11 compares the thermal conductivities of SiC$_w$/Ti$_3$SiC$_2$ composites to the monolithic Ti$_3$SiC$_2$. The $k$ increased with the increase

Fig. 10 Variation of Vickers hardness in the SiC$_w$/Ti$_3$SiC$_2$ as a function of SiC$_w$ content.

Fig. 11 $k$ of SiC$_w$/Ti$_3$SiC$_2$ as a function of SiC$_w$ content.
in SiCw content, where the theoretical k of single-crystal SiC is approximately 490 W/(m·K) [42]. The k of 10 wt% SiCw/Ti3SiC2 was 40.3 W/(m·K), which is almost 25% higher than that of the monolithic Ti3SiC2 (32.4 W/(m·K)). Zhang et al. [14] measured the k of Ti3SiC2 composites reinforced with SiC nanoparticles. When the amount of SiC nanoparticles was 30 vol%, the highest k of 39 W/(m·K) was obtained, which was slightly lower than that of the composite containing 10 wt% SiCw in this study. Therefore, it seems that the effect of SiCw is significantly higher than that of nanoparticles on the enhancement of thermal conductivity, which may be attributed to the large aspect ratio of SiCw.

4 Conclusions

SiCw/Ti3SiC2 with high strength and k were fabricated using SPS. The effects of SiCw on the strength of the composites were complicated by multiple phenomena occurring at the same time. As a result, the strengths of the composites containing up to 5 wt% SiCw remained at a similar level as that of the monolithic Ti3SiC2. However, when the SiCw content increased to 10 wt%, the bending strength of the composite (635 MPa) was approximately 50% higher than that of the monolithic Ti3SiC2 (428 MPa). The remarkable strengthening effect with 10 wt% SiCw was attributed to the fine-grained uniform microstructure and crack bridging by SiCw in the composite. The Vickers hardness also increased to 7.9 GPa with 10 wt% SiCw from 5.8 GPa for pure Ti3SiC2. Furthermore, the addition of 10 wt% SiCw also increased the k by 25%, which is attributed to the high k of the SiCw and the one-dimensional nature of the long whiskers.

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