Fabrication of $\text{SiC}_w/\text{Ti}_3\text{SiC}_2$ composites with improved thermal conductivity and mechanical properties using spark plasma sintering

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Abstract: High strength SiC whisker-reinforced Ti_3SiC_2 composites ($\text{SiC}_w/\text{Ti}_3\text{SiC}_2$) with an improved thermal conductivity and mechanical properties were fabricated by spark plasma sintering. The bending strength of 10 wt% $\text{SiC}_w/\text{Ti}_3\text{SiC}_2$ was 635 MPa, which was approximately 50% higher than that of the monolithic Ti_3SiC_2 (428 MPa). The Vickers hardness and thermal conductivity (k) also increased by 36% and 25%, respectively, from the monolithic Ti_3SiC_2 by the incorporation of 10 wt% SiC_w . This remarkable improvement both in mechanical and thermal properties was attributed to the fine-grained uniform composite microstructure along with the effects of incorporated SiC_w . The $\text{SiC}_w/\text{Ti}_3\text{SiC}_2$ 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; Ti_3SiC_2 ; mechanical property; thermal conductivity; spark plasma sintering

1 Introduction

Ti_3SiC_2 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 Ti_3SiC_2 unit cell consists of alternating covalently-bonded Ti_6C 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 Ti_3SiC_2 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 Ti_3SiC_2 reinforced with SiC particles has attracted much attention [14–22]. It was reported that the Vickers hardness, fracture toughness, k , and oxidation resistance of Ti_3SiC_2 could be improved by the addition of SiC particles [14–16]. However, the

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room-temperature strength of Ti_3SiC_2 decreased by the addition of SiC due to the retained tensile stress in the Ti_3SiC_2 , originating from the increased thermal expansion mismatch between SiC and Ti_3SiC_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 $\text{Al}_2\text{O}_3/\text{Ti}_3\text{SiC}_2$ was improved to 688 MPa when 20 wt% SiC_w was added [24]. Furthermore, the presence of SiC_w in the $\text{SiC}_w/\text{TiC}/\text{Ti}_3\text{SiC}_2$ was reported to increase the room-temperature strength. Hashimoto *et al.* [25] reported a very high bending strength of 1 GPa for the Ti_3SiC_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_3SiC_2 [26–28]. Moreover, the Ti–Si intermetallic compounds, which are formed simultaneously upon the decomposition of Ti_3SiC_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_3SiC_2 . The $\text{SiC}_w/\text{Ti}_3\text{SiC}_2$ composites were fabricated using a heteropolar coagulation method followed by spark plasma sintering, which can minimize the decomposition of Ti_3SiC_2 . The mechanical and thermal properties of $\text{SiC}_w/\text{Ti}_3\text{SiC}_2$, as a function of SiC_w content, were analyzed in comparison with the monolithic Ti_3SiC_2 .

2 Experimental

2.1 Starting materials

Commercial Ti_3SiC_2 powder with 98.5% purity was used as a starting material (Beijing Jinhezhi Materials Co., China), which contains Al_2O_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 $\approx 10,000$, Aladdin Co., China) was used as a cationic dispersant to improve the uniform distribution of SiC_w in the Ti_3SiC_2 matrix.

2.2 Composite fabrication

SiC_w and Ti_3SiC_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_4OH followed by ultrasonication for 45 min, where a uniform dispersion by hetero-coagulation between the Ti_3SiC_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_3SiC_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 $\text{SiC}_w/\text{Ti}_3\text{SiC}_2$, while a pure monolithic Ti_3SiC_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 K α 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_3SiC_2 matrix.

A zeta potential analyzer (Zetasizer Nano ZS, Malvern, UK) was used to measure the surface charge of the SiC_w and Ti_3SiC_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 $\times$ 4 mm $\times$ 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 (W/(m·K)) was calculated from Eq. (1):

$$k = \rho \lambda C_p \quad (1)$$

where λ is the thermal diffusivity coefficient and ρ is the density of the material.

3 Results and discussion

3.1 SiC whisker dispersion in Ti_3SiC_2

The uniform distribution of SiC_w in Ti_3SiC_2 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 SiC_w by both electrostatic and steric mechanism [30]. Figure 1 shows the zeta potential behavior of SiC_w in an aqueous system with different amounts of PEI as a function of pH. The zeta potential of SiC_w was increased from 22 to 58 mV at pH = 3 by adding 1.5 wt% PEI. Because the zeta potential of Ti_3SiC_2 (25 mV) was similar to that of SiC_w without PEI at pH = 3, firstly SiC_w was dispersed in water after adding 1.5 wt% PEI, and then the Ti_3SiC_2 powder was added followed by the pH adjustment to 3 to obtain the uniform dispersion of SiC_w and Ti_3SiC_2 .

The pH of the suspension was then increased to 9 by adding NH_4OH , followed by ultrasonication for 45 min. A uniform mixing between the Ti_3SiC_2 (zeta potential = -30 mV) and SiC_w (zeta potential = +25 mV) was obtained by heteropolar coagulation originated from opposite surface charge. The insets in Fig. 1 compare the image of $\text{SiC}_w/\text{Ti}_3\text{SiC}_2$ dispersion at pH = 3 and 9. Figure 2 shows SEM images of the SiC_w and $\text{SiC}_w/\text{Ti}_3\text{SiC}_2$ mixture, where the uniform distribution of SiC_w in Ti_3SiC_2 is found due to the dispersion by electrosteric mechanism combined with heteropolar coagulation.

Fig. 1 Zeta potential of Ti_3SiC_2 and SiC_w with/without PEI dispersant in an aqueous system. The insets compare the image of $\text{SiC}_w/\text{Ti}_3\text{SiC}_2$ dispersion at pH = 3 and 9.

Fig. 2 SEM images of (a) well-dispersed SiC_w with PEI at pH = 3 and (b) 10 wt% $\text{SiC}_w/\text{Ti}_3\text{SiC}_2$ mixture intentionally agglomerated at pH = 9.

3.2 Microstructure of $\text{SiC}_w/\text{Ti}_3\text{SiC}_2$ composites

The relative density for $\text{SiC}_w/\text{Ti}_3\text{SiC}_2$ with different SiC_w contents after SPS at 1350 °C is shown in Fig. 3, where the density $\geq 98.4\%$ was obtained for all samples, regardless of the amount of SiC_w . The theoretical composite density was calculated by considering the relative amount of SiC_w and Ti_3SiC_2 .

Figure 4 shows the XRD patterns of $\text{SiC}_w/\text{Ti}_3\text{SiC}_2$ and pure Ti_3SiC_2 after SPS at 1350 °C, which indicates the presence of Ti_3SiC_2 (JCPDS No. 65-3559) and SiC

Fig. 3 Relative density of $\text{SiC}_w/\text{Ti}_3\text{SiC}_2$ as a function of SiC_w after SPS at 1350 °C.

Fig. 4 XRD patterns of the SiC_w/Ti₃SiC₂ and pure Ti₃SiC₂ 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 Ti₃SiC₂ 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/Ti₃SiC₂ system from Hashimoto *et al.* [25], where the Ti₃SiC₂ grains were preferentially aligned. Due to the grain alignment along with the TiC precipitates, the SiC_w/TiC/Ti₃SiC₂ showed very high bending strength, especially when the composite contained 15 vol% SiC_w. However, the SiC_w/TiC/Ti₃SiC₂ revealed the decomposition of components during heat treatment, resulting in the formation of intermetallic Ti–Si compounds, such as TiSi₂ and Ti₅Si₃, which are deleterious for nuclear applications.

The reason for the decomposition of Ti₃SiC₂ in Ref. [25] was attributed to the flake-shaped Ti₃SiC₂ particles formed by ball milling. Because Ti₃SiC₂ contains TiC octahedrons, which are aligned to (111) plane, the flake-shaped Ti₃SiC₂ particles promote the formation of TiC by the decomposition of Ti₃SiC₂. The enhanced decomposition of Ti₃SiC₂ can also be achieved by the mechano-chemical activation using a high energy ball milling [31]. However, the SiC_w/Ti₃SiC₂ 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 Ti₃SiC₂ and the formation of TiC could be minimized because the Ti₃SiC₂ particles remained as randomly oriented polycrystalline.

The SiC_w phase distribution in the Ti₃SiC₂ 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/Ti₃SiC₂ containing different

amounts of SiC_w. The gray phase corresponds to Ti₃SiC₂ matrix, while the dark phase is SiC_w, according to the EDS analysis (not shown here). The SiC_w is distributed uniformly in the Ti₃SiC₂ matrix for the 1 wt% SiC_w/Ti₃SiC₂ sample. Some agglomeration of whiskers in the 5 wt% and 10 wt% SiC_w/Ti₃SiC₂ 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 Ti₃SiC₂ matrix. The Ti₃SiC₂ 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 Ti₃SiC₂ 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. Ti₃SiC₂ 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 Ti₃SiC₂, it seems that Al is segregated at point 3 after oxidation into Al₂O₃. The presence of Al₂O₃ is reported to improve the oxidation resistance, strength, and fracture toughness of Ti₃SiC₂, 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 Ti₃SiC₂ matrix. Spencer *et al.* [35] could not observe the reaction between SiC fibers and Ti₃SiC₂ up to 1550 °C either. This thermochemical stability between SiC_w and Ti₃SiC₂ would be beneficial for improving the mechanical and thermal properties of the composites.

Randomly oriented Ti₃SiC₂ grains can be observed with the fractured surface images for both monolithic Ti₃SiC₂ and SiC_w/Ti₃SiC₂, as shown in Fig. 6. Ti₃SiC₂ grains typically show plate shape because of the

Fig. 5 SEM images for the polished surface of (a) 1 wt%, (b) 5 wt%, and (c) 10 wt% SiC_w/Ti₃SiC₂. (d) HAADF image for the 10 wt% SiC_w/Ti₃SiC₂ and (e–g) EDS results for the points marked in (d).

Fig. 6 SEM images for the fractured surfaces of SiC_w/Ti₃SiC₂ containing different amount of SiC_w after SPS at 1350 °C: (a) 0 wt%, (b) 1 wt%, (c) 5 wt%, and (d) 10 wt%.

layered TiC and Si structure. Abnormally grown Ti₃SiC₂ grains were observed for the pure Ti₃SiC₂ and composites containing up to 5 wt% SiC_w, as shown in Figs. 6(a)–6(c). The abnormal grain size of Ti₃SiC₂ 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 Ti₃SiC₂, but as SiC_w was added, they were reduced in terms of the occurrence and size so that they became isolated in the fine Ti₃SiC₂ matrix grains.

The largest abnormal grain size for monolithic Ti₃SiC₂ is 150 μm approximately, while those for SiC_w/Ti₃SiC₂ 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 Ti₃SiC₂ 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/Ti₃SiC₂ sample (Fig. 7). It implies that the interface bonding strength between SiC_w and Ti₃SiC₂ is not that strong.

Fig. 7 SEM image of the fractured surface for the SiC_w/Ti₃SiC₂ containing 10 wt% SiC_w.

3.3 Mechanical properties of SiC_w/Ti₃SiC₂ composites

3.3.1 Bending strength

Figure 8 presents the bending strength of the SiC_w/Ti₃SiC₂ composites at room temperature as a function of SiC_w content. Considering the bending strength of the monolithic Ti₃SiC₂ was 428 MPa, the strength of composite did not change much up to 5 wt% SiC_w. On the other hand, the bending strength of composite containing 10 wt% SiC_w reached to 636 MPa, which is significantly improved from the monolithic Ti₃SiC₂ by 48%.

This strength behavior as a function of SiC_w 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_w 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_w/Ti₃SiC₂ with 10 wt% SiC_w 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_w, 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_w/TiC/Ti₃SiC₂ by varying the content of SiC_w. However, it is notable that the composites containing up to 5 wt% SiC_w showed similar strength compared to monolithic Ti₃SiC₂ 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_w on the composite strength. The crack bridging by the SiC_w in the wakes behind the tips of cracks can

Fig. 8 Bending strength of the SiC_w/Ti₃SiC₂ containing different amount of SiC_w.

be another reason (Fig. 9(d)), in addition to the fine microstructure, to increase the SiC_w/Ti₃SiC₂ composite strength [37]. On the other hand, there are two possible opposing effects of whiskers to decrease the strength of the composite. Ti₃SiC₂ is non-cubic and has a layered structure, resulting in extensive grain bridging. That is why Ti₃SiC₂ intrinsically has relatively high fracture toughness. However, the addition of SiC_w 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_w.

Thus, the effects of SiC_w on the strength of the SiC_w/Ti₃SiC₂ composites may be complicated by these various phenomena occurring at the same time. Therefore, for the composites containing 5 wt% SiC_w or less, it is speculated that the strengthening effects of the smaller grain sizes and the crack bridging by the SiC_w 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_w was at the same level as that of the monolithic Ti₃SiC₂. When the SiC_w content increased to 10 wt%, the fine-grained uniform microstructure and crack bridging by the SiC_w (Fig. 9(d)) dominated over the negative effects, and hence the significant strengthening was observed.

The addition of 10 wt% SiC_w in Ti₃SiC₂ increased

Fig. 9 SEM images of the surface after indentation test for (a, b) monolithic Ti₃SiC₂ and (c, d) 10 wt% SiC_w/Ti₃SiC₂.

the bending strength by 50% compared to the pure Ti_3SiC_2 . However, the addition of particulate SiC to Ti_3SiC_2 or $\text{Ti}_3\text{Si}(\text{Al})\text{C}_2$ 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_3SiC_2 and 10 wt% $\text{SiC}_w/\text{Ti}_3\text{SiC}_2$. The indent shape was irregular with exfoliated surfaces and deformed particles for monolithic Ti_3SiC_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_3SiC_2 .

The deformation around the indent was asymmetric, while the damaged region extended approximately 50% from the indent. Some specimens showed exfoliated directional Ti_3SiC_2 grains around the indent. When the basal Ti_3SiC_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_3SiC_2 grain was perpendicular to the shear stress, as shown in Fig. 9(a). The deformation during the indentation of Ti_3SiC_2 is generally caused by the basal plane slipping and debonding at the Ti_3C_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_3SiC_2 , the degree of grain buckling and fracturing of the $\text{SiC}_w/\text{Ti}_3\text{SiC}_2$ 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_3SiC_2 . The interfacial debonding between SiC_w and Ti_3SiC_2 matrix seemed to have similar effects on the mechanical properties of $\text{SiC}_w/\text{Ti}_3\text{SiC}_2$ to the basal plane slipping and debonding between the Ti_3C_2 octahedron and Si layer occurring in the Ti_3SiC_2 unit cell.

Figure 10 shows the Vickers hardness of $\text{SiC}_w/\text{Ti}_3\text{SiC}_2$ as a function of SiC_w content. The hardness of 5.8 GPa for pure Ti_3SiC_2 increased to 7.9 GPa for the 10 wt%

$\text{SiC}_w/\text{Ti}_3\text{SiC}_2$ composite. The relatively low hardness of pure Ti_3SiC_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_3SiC_2 to 1 wt% $\text{SiC}_w/\text{Ti}_3\text{SiC}_2$. Because hardness is known to follow the Hall–Petch relationship for microcrystalline ceramics [41], the finer microstructure of 1 wt% $\text{SiC}_w/\text{Ti}_3\text{SiC}_2$ compared to the pure Ti_3SiC_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 $\text{SiC}_w/\text{Ti}_3\text{SiC}_2$ composites

k is very important for nuclear application of $\text{SiC}_w/\text{Ti}_3\text{SiC}_2$, which directly influences the energy exchange efficiency of the reactor. Figure 11 compares the thermal conductivities of $\text{SiC}_w/\text{Ti}_3\text{SiC}_2$ composites to the monolithic Ti_3SiC_2 . The k increased with the increase

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

Fig. 11 k of $\text{SiC}_w/\text{Ti}_3\text{SiC}_2$ as a function of SiC_w content.

in SiC_w content, where the theoretical k of single-crystal SiC is approximately 490 W/(m·K) [42]. The k of 10 wt% SiC_w/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% SiC_w in this study. Therefore, it seems that the effect of SiC_w is significantly higher than that of nanoparticles on the enhancement of thermal conductivity, which may be attributed to the large aspect ratio of SiC_w.

4 Conclusions

SiC_w/Ti₃SiC₂ with high strength and k were fabricated using SPS. The effects of SiC_w 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% SiC_w remained at a similar level as that of the monolithic Ti₃SiC₂. However, when the SiC_w 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% SiC_w was attributed to the fine-grained uniform microstructure and crack bridging by SiC_w in the composite. The Vickers hardness also increased to 7.9 GPa with 10 wt% SiC_w from 5.8 GPa for pure Ti₃SiC₂. Furthermore, the addition of 10 wt% SiC_w also increased the k by 25%, which is attributed to the high k of the SiC_w and the one-dimensional nature of the long whiskers.

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