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To cite this article: Peng Wan, Zhen Wu, Hui Zhang, Liyin Gao & Jingyang Wang (2016) Porous Nano-SiC as Thermal Insulator: Wisdom on Balancing Thermal Stability, High Strength and Low Thermal Conductivity, Materials Research Letters, 4:2, 104-111, DOI: [10.1080/21663831.2015.1121167](https://doi.org/10.1080/21663831.2015.1121167)

To link to this article: <https://doi.org/10.1080/21663831.2015.1121167>



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Published online: 08 Jan 2016.



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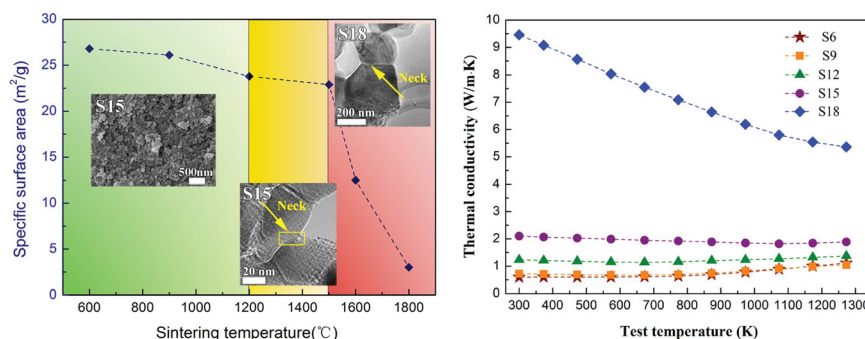
Porous Nano-SiC as Thermal Insulator: Wisdom on Balancing Thermal Stability, High Strength and Low Thermal Conductivity

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(Received 24 July 2015; final form 12 November 2015)

We herein show that by integrating nano-scale phonon-scattering mechanisms, such as interfaces and stacking faults thermal resistances in porous nano-SiC, this outstanding material (with intrinsic very high thermal conductivity) could demonstrate promising thermal insulation property. Porous nano-SiC prepared at 1,500°C exhibits a specific balanced mechanical strength (compressive and flexural strength are 26 and 13 MPa, respectively, with 57% porosity) and extremely low thermal conductivity ($2 \text{ W m}^{-1} \text{ K}^{-1}$ at 300 K); and sample sintered at 1,800°C shows excellent mechanical strength but relatively high thermal conductivity. Our work reports the novel low thermal conductivity of porous nano-SiC for the first time.



Keywords: Nano-SiC, Porous Material, Mechanical Properties, Thermal Conductivity

Advanced thermal-energy conversion systems require candidate materials possess high conversion efficiency, excellent chemical, thermal and mechanical stability in harsh high-temperature environments.[1] Conventional semiconductor materials, such as GeTe-Bi₂Te₃ and PbTe, showed attractive performances at moderate operation circumstance, but their reliabilities are typically limited below 1,000°C because of the significant decomposition, oxidation or melting at elevated temperatures. Porous silicon carbide (SiC) was suggested as potential candidate for high-temperature thermal-energy conversion because of its unique combination of excellent mechanical properties, good chemical

stability, as well as oxidation resistance at elevated temperature.[2] Unfortunately, SiC ceramic intrinsically has very high thermal conductivity which will frustrate the high efficiency of energy conversion. The promising application crucially demands new breakthrough of low thermal conductivity SiC. Ceramic foams will be one of the possible solutions because of the merits of low thermal conductivity and good thermal shock resistance.[3,4] However, increasing the porosity will greatly sacrifice mechanical strength. Therefore, new mechanisms are required to balance the thermal stability, high strength and low thermal conductivity of porous SiC.

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In this paper, we show that by integrating specific nano-scale phonon-scattering mechanisms in porous SiC, this outstanding material could demonstrate promising thermal insulation property. Porous nano-SiC sintered at 1,500°C exhibits moderate strength and very low thermal conductivity ($2 \text{ W m}^{-1} \text{ K}^{-1}$ at 300 K). The low thermal conductivity attributes to giant interfacial thermal resistance (including clean grain-boundary, amorphous SiO_2 layer at grain boundary), strong defect-phonon scattering and nano-pores within nano-scale microstructure. Sample prepared at 1,800°C shows high strength and good stiffness up to high temperature and it also illustrates relatively low thermal conductivity ($9.5 \text{ W m}^{-1} \text{ K}^{-1}$ at 300 K). The current work provides a new wisdom on balancing the excellent thermal stability, high strength and low thermal conductivity of this attractive material through controllable microstructure.

Main techniques for processing porous SiC include: partial sintering, sacrificial fugitives, replica templates and direct foaming.[5–8] Partial sintering is used to produce homogenous porous ceramics with narrow pore size distribution; sacrificial fugitives targets at porous ceramics with well-tuned shape and size of pores, which is controlled by the characteristics of the sacrificial filler; replica templates is developed to obtain highly porous ceramics with interconnected large pores/channels and open cell walls; direct foaming usually utilizes chemical or physical blowing agent to generate pores with bubble shape in ceramic slurries.[9,10] As a result of the strong Si-C covalent bonding and low self-diffusion coefficient, SiC skeleton is difficult to be densified below 2,150°C.[11] In order to lower the sintering temperature, oxides such as Al_2O_3 , MgO , SiO_2 and Y_2O_3 are usually used as sintering additives to promote densification.[12,13] However, the additives will lead to significant sintering shrinkage,[14] and also dramatically degrade high-temperature strength, stiffness and creep resistance for the formation of grain boundary glassy phase.[15] Our solution to develop porous SiC with good high-temperature strength and meanwhile low thermal conductivity is based on a partial sintering method using nano-SiC powder as the only raw material. On the one hand, the high sinterability of nano-SiC powders guarantees the formation of rigid neck at relatively low sintering temperature.[16] Fukushima et al. prepared porous SiC ceramics using nano-sized SiC powder additive, and reported enhanced mechanical strength. They attributed the well-developed sintering neck to the promotion of mass transfer by nano-powders.[17,18] On the other hand, high densities of stacking faults, interfaces and nano-pores[19,20] in nano-scale microstructures will enhance phonon scattering and reduce thermal conductivity. In this paper, thermal stability and properties of porous nano-SiC prepared from 600°C to 1,800°C are investigated. It is interesting to find that

porous nano-SiC sintered at 1,500°C exhibits a specific balanced mechanical strength and very low thermal conductivity; and sample sintered at 1,800°C shows high-mechanical strength and moderate thermal conductivity. Our work shed a light on the novel thermal insulation property of porous nano-SiC for the first time.

Nano β -SiC powder with a mean particle size of 35 nm was used as the raw material. The powder was uniaxially pressed by a steel mold to form definite shapes and further cold isostatically pressed at 260 MPa. After demolding and drying, the green bodies in a graphite crucible were placed in a graphite-heating furnace to sinter with a ramp rate of 5°C min^{-1} under pure flowing argon. The target temperatures are 600°C, 900°C, 1,200°C, 1,500°C and 1,800°C, and denoted as S6, S9, S12, S15 and S18, respectively. Dwelled for 120 min at the target temperatures, the samples were cooled down to room temperature at $10^\circ\text{C min}^{-1}$.

Phase composition was analyzed via a step-scanning X-ray diffraction (XRD) (Rigaku D/max-2400, Tokyo, Japan) with CuK_α radiation ($\lambda = 1.54178 \text{ \AA}$) at a scanning speed of 0.02° per step. The apparent density of the specimens was calculated by the ratio of measured mass to volume. Microstructural characterization of the as-prepared sample was studied by a scanning electron microscope (SEM, LEO, SUPRA35, Ammerbuch, Germany) and a transmission electron microscope (TEM, FEI, Tecnai G2 F20, Oregon, USA). The specific surface area was characterized by Brunauer–Emmett–Teller surface analysis (BET; Micromeritics Inc., ASAP 2020) with nitrogen as adsorption gas at 77 K.

Young's modulus and shear modulus were tested on rectangular bars ($3 \text{ mm} \times 15 \text{ mm} \times 40 \text{ mm}$) using the impulse-excited resonance method (resonance frequency and damping analyzer, IMCE, Diepenbeek, Belgium) from 200°C to 1,500°C. The Vickers microhardness was measured at loads of 10 N with a dwell time of 15 s. The flexural strength was determined via the three-point bending method using samples with a dimension of $3 \text{ mm} \times 4 \text{ mm} \times 36 \text{ mm}$, the span is 30 mm and the crosshead speed is 0.5 mm min^{-1} . The samples with a size of $\varnothing 8 \text{ mm} \times 8 \text{ mm}$ were used for the compression test. Strength measurements were performed in a universal testing machine (SANS, CMT4204, SANS, Shenzhen, China) using at least five samples.

Temperature dependence of thermal conductivity and thermal diffusivity of porous SiC were obtained using a disc sample with dimensions of ($\varnothing 12.7 \text{ mm} \times 1 \text{ mm}$) from 473 to 1,273 K by a FlashlineTM-5000 thermal properties analyzer (Anter Corp., Pittsburgh, PA). During the thermal conductivity test, the thermal diffusivity and heat capacity were measured by a laser pulse method. The specimens were sputter coated with a colloidal graphite layer to ensure that the laser pulse signals can be absorbed completely. Finally, we calculated the

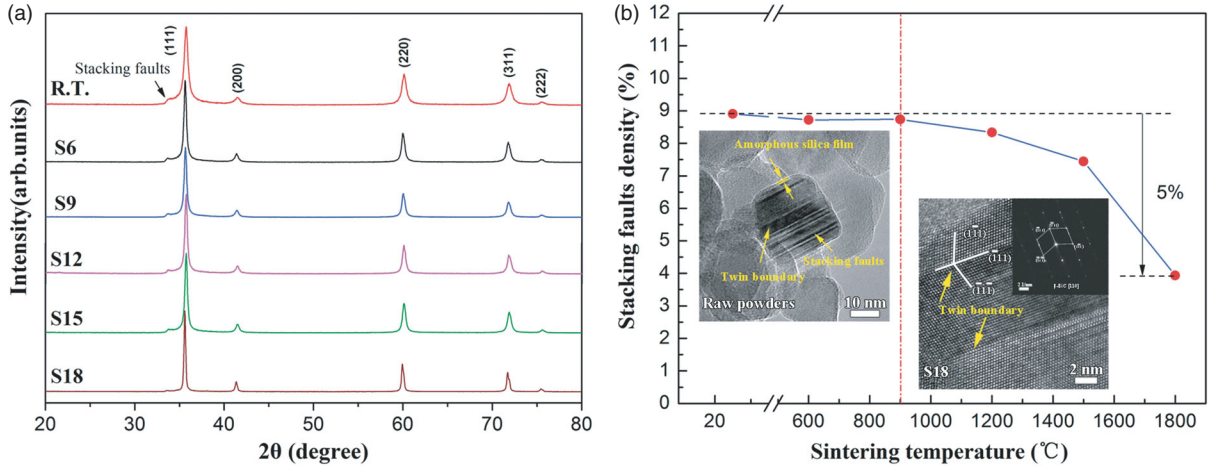


Figure 1. (a) XRD patterns and (b) stacking fault density in SiC powders as a function of sintering temperature. The inset in (b) shows bright-field TEM micrograph of SiC raw nano-powders, and bright-field TEM micrograph of S18 and the corresponding HRTEM image and the SAED pattern taken along the direction of $[110]$ zone axis [Colour online].

thermal conductivity κ ($\text{W m}^{-1} \text{K}^{-1}$) according to the following equation:

$$\kappa = \alpha \cdot C_p \cdot \rho, \quad (1)$$

in which α is the thermal diffusivity ($\text{m}^2 \text{s}^{-1}$), C_p the heat capacity ($\text{J kg}^{-1} \text{K}^{-1}$), and ρ the density (kg m^{-3}).

SiC appears in a number of polytypes in which 3C, 2H, 4H, 6H and 15R are the most common phases. The XRD patterns of porous nano-SiC specimens sintered at various temperatures are plotted in Figure 1(a). It can be seen that the porous nano-SiC mainly consists of zinc-blende structure SiC (β -SiC) and there is an additional diffraction peak at $2\theta \approx 33.7^\circ$, which may be caused by the content of polytypes or stacking faults and twins in β -SiC powders. Pujar and Cawley systematically studied the effect of stacking faults and different polytypes on the XRD profiles of β -SiC powders.[21] Based on extensive theoretical simulations, they found the high background intensity near (111) the diffraction peak was attributed to the stacking faults instead of other polytypes. Our XRD experiments show good agreement with Pujar's conclusion, and it is further confirmed by HRTEM observations, as shown in the inset of Figure 1(b). As sintering temperature increases, intensity of the additional peak caused by stacking faults decreases, indicating the reduction of the density of stacking faults. According to the XRD data, the stacking fault density ρ_{sf} can be estimated by the following equation [22]:

$$X = \frac{P_{33.6^\circ}}{P_{41.4^\circ}}, \quad (2)$$

$$\rho_{\text{sf}} = \frac{X}{6.82 \times 10^{-2}X + 2.27 \times 10^{-2}} + 1.7X^3, \quad (3)$$

where $P_{33.6^\circ}$ and $P_{41.4^\circ}$ are the intensities of the peaks at 33.6° and 41.4° (2θ), respectively. Using Equations

(2) and (3), the density of stacking faults as a function of sintering temperature is calculated and the result is shown in Figure 1(b). With the increase in temperature, the density of stacking faults maintains nearly stable below 900°C , then moderately decreases from 900°C to $1,500^\circ\text{C}$, and finally drops rapidly above $1,500^\circ\text{C}$. With reference to the previous literature,[22] decrease in the density of stacking fault is mainly attributed to the thermally activated atomic diffusions.

As a kind of stacking fault, twins are usually observed in nanostructures with a face-centered cubic (fcc) lattice.[23] The inset in Figure 1(b) also demonstrates the bright-field TEM image of S18 and corresponding high-resolution transmission microscopy (HRTEM) image and selected area electron diffraction (SAED) pattern taken along the direction of $[110]$ zone axis. The twin boundary is the $(\bar{1}\bar{1}1)$ plane and the side surface of each segment is the (111) plane. The influences of stacking faults on blocking phonon transport, as well as on lowering thermal conductivity of porous nano-SiC, will be discussed in the following sections.

Thermal stability of porous nano-SiC is explored by examining the changes of BET surface areas of porous samples with corresponding TEM images, as shown in Figure 2. This figure reveals the evolution of microstructures of porous nano-SiC sintered at different temperatures. The measured specific surface area decreases slightly with increased temperature and drops abruptly above $1,500^\circ\text{C}$. From 600°C to 900°C , the nano-SiC grains remain stable and only slightly sintering is identified. As sintering temperature rises, sintering necks form tightly between grains in S12 and S15; and grain sizes still maintain in nanoscale. Nano-pores are observed in the microstructure of S6 to S15 samples, and the type of porosity is clearly external. Typical SEM image of S15 is presented in the inset on the left part of Figure 2. When temperature reaches $1,800^\circ\text{C}$, grain growth occurs and

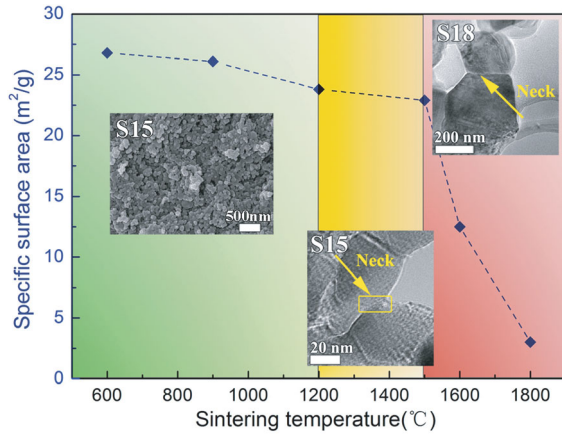


Figure 2. BET surface area of porous SiC ceramics as a function of sintering temperature. Inset images correspond to SEM micrographs of S15 and BF TEM micrographs of S15 and S18, respectively [Colour online].

Table 1. Sintering temperature T_S , porosity v_a , and measured room temperature mechanical properties of porous SiC ceramics, including bulk modulus B , shear modulus G , Young's modulus E , Vickers hardness H_v , compressive strength σ_c and flexural strength σ_f .

Samples	T_S (°C)	v_a (%)	B	G	E	H_v	σ_c	σ_f
			(GPa)				(MPa)	
S6	600	56.6	2.3	2.3	5.2	0.14	6.4	6.4
S9	900	56.6	2.9	2.7	6.1	0.20	11.9	8.7
S12	1,200	56.2	3.3	3.4	7.6	0.27	20.6	12.2
S15	1,500	57.1	3.4	3.5	7.8	0.29	25.6	12.8
S18	1,800	58.0	8.6	7.2	17.0	0.75	121.1	30.0

strong sintering necks are obtained. The results indicate that porous nano-SiC holds good thermal stability up to 1,500°C.

Table 1 summarizes the experimental room temperature mechanical properties, including bulk, shear and Young's modulus, Vickers hardness, flexural strength and compressive strength, together with corresponding sintering temperature and porosity. The porous SiC specimens remain approximately the same porosity (56.2–58%) when temperature increases from 600°C to 1,800°C. And the linear shrinkages of S6, S9, S12, S15 and S18 are 0.02%, 0.04%, 0.36%, 0.53% and 1.68%, respectively. The low shrinkages are related to the specific mass-transfer path during solid sintering of

SiC, which is dominated by surface diffusion or evaporation/condensation. The low sintering shrinkage also ensures the near-net-shape processing of porous nano-SiC. The flexural strength increases from 6.4 MPa to 12.8 MPa, as the sintering temperature enhanced from 600°C to 1,500°C due to the tighter sintering necks. When sintering temperature reaches 1,800°C, flexural strength enhances up to 30 MPa. Other mechanical properties display the similar trends to that of the flexural strength.

Flexural strengths of porous SiC fabricated by other methods are presented in Table 2 for comparison. Self-bonded porous SiC with 51% porosity had the flexural strength of 12 MPa and liquid phase bonded porous SiC with 56–58% porosities showed the flexural strengths of 15–26 MPa.[24,25] In addition, morphology of pores also affects the strength. Using technique combining freeze drying process and solid sintering (sintered at 2,150°C), Zuo K.H. et al. prepared SiC foams with porosities of 60–62% and flexural strengths of 5–21 MPa. The pore size distribution curve had multi-peaks and the main peak located at about 10.2 μm . [26] By comparison to the listed references, S18 is fabricated at the lowest temperature without sintering additives and it has superior flexural strength (30 MPa) at the comparable porosity (58%). These could be attributed to both the good sinterability of nano-powders and the rigid network structure of SiC skeleton as well as the homogenous distribution and smaller pore sizes (0.46–0.68 μm for S18). Fukushima et al. prepared porous SiC by submicro-SiC powder using 10 wt% nano-SiC as additive.[17] The reported BET surface area, compressive strength σ_c , and porosity are 9.3 m^2/g , 122 MPa, and 39%, respectively, for sample processed at 1,500°C; and 1.9 m^2/g , 513 MPa, and 39%, respectively, for sample prepared at 1,800°C. In the present work, since nano-SiC starting powders have large specific surface area, the BET values of our samples (22.9 m^2/g for sintered at 1,500°C and 3.0 m^2/g for sintered at 1,800°C) are much higher than the reported values in Ref. [17]. Therefore, the present samples have higher porosities but relatively lower strengths.

At high temperature, the critical challenge for previously reported liquid-sintered SiC foams may be the significant softening of grain boundary glass phases.[15] We studied the temperature dependence of Young's

Table 2. Comparison of sintering temperature T_S , particle size d of SiC raw powders, additives, pore former, porosity v_a , and flexural strength σ_f of porous SiC ceramics fabricated by different methods.

Methods	T_S (°C)	d (nm)	Additives	v_a (%)	σ_f (MPa)
This work	1,800	35 (β -SiC)	—	58	30
Self-bonding [24]	1,800	700–2,500 (α -SiC)	Si, C, B	51	12
Liquid phase bonding [25]	1,950	450 (α -SiC), 1,700 (β -SiC)	Al_2O_3 , Y_2O_3	56–58	15–26
Freeze-drying [26]	2,150	500 (α -SiC)	C, B_4C	60–62	5–21

modulus and internal friction of S18 sample. It is found that Young's modulus decreases slightly with the increase in temperature. The magnitude at 1,500°C (15 GPa) maintains nearly 90% of its room temperature value (17 GPa). Meanwhile the corresponding internal friction does not show obvious increment, which suggests the absence of grain boundary softening or other mechanism of energy dissipation. Therefore, the current preparation of porous nano-SiC without sintering additives endows good mechanical property from room to high temperatures.

Bulk SiC ceramic is known as an excellent thermal conductor. Hot-pressing sintered SiC with Be dopant showed the thermal conductivity as high as 270 W m⁻¹ K⁻¹ at room temperature.[27] For SiC single crystal, the value has been reported above 300 W m⁻¹ K⁻¹. [28] Although enhancing porosity will decrease the thermal conductivity of SiC, previously works reported relatively high thermal conductivities of porous SiC. Jang et al. fabricated porous SiC without sintering additives, and the sample exhibited thermal conductivities of 57–83 W m⁻¹ K⁻¹ with porosities of 30–41%. [8] Zhao and coworkers produced a series of porous SiC by addition of Al₂O₃-SiO₂-Y₂O₃ composites, and their specimens showed thermal conductivities of 30–43 W m⁻¹ K⁻¹ with porosities of 65–74%. [29] In this work, the experimental thermal conductivities of all samples are plotted in Figure 3(a) as a function of measuring temperature. For S6 and S9, the R.T. thermal conductivities exhibit ultralow magnitudes about 0.6 W m⁻¹ K⁻¹; and the values show increments at elevated temperature. These are attributed to the second sintering of nano-grains at high temperature. It is interesting to notice that, S15 and S12 own significant low thermal conductivities, about 1–2 W m⁻¹ K⁻¹ at room temperature, and remains stable at elevated temperatures. S18 has the moderate thermal conductivity of 9.5 W m⁻¹ K⁻¹ at room temperature. As test temperature enhances, the

thermal conductivity of S18 decreases with $\sim 1/T$ (T in Kelvin) due to the intrinsic Umklapp scattering among phonons and finally approaches to about 5 W m⁻¹ K⁻¹ at 1,300 K. The thermal conductivity of S18 with a porosity of 58% is much lower (1/3–1/8 in magnitude) than the previously reported data.

In general, the lattice thermal conductivity κ_L of dielectric material crucially depends on the phonon relaxation time τ . Different phonon-scattering mechanisms in this work can be incorporated by independent processes: $\tau^{-1} = \tau_U^{-1} + \tau_{gb}^{-1} + \tau_{sf}^{-1} + \tau_{others}^{-1}$, where τ_U^{-1} , τ_{gb}^{-1} , τ_{sf}^{-1} and τ_{others}^{-1} are rate related to Umklapp scattering, phonon-grain boundary scattering, phonon-stacking fault scattering and other scattering channels, respectively. For porous material, the thermal conductivity κ_p depends on both porosity and microstructural characteristics. If only porosity is considered at first, the Umklapp scattering within skeleton will affect the thermal conductivity. There exist several analytical models to describe the effective thermal conductivity κ_p of porous materials, including the Series and Parallel models, Maxwell–Eucken models (two forms) and effective medium theory (EMT) equation.[30] The pores and SiC skeleton are interconnected randomly to form a two-phase system. Therefore, the microstructure fits the assumption of EMT model:

$$\kappa_p = \frac{1}{4} \left(\frac{(3v_a - 1)\kappa_a + (2 - 3v_a)\kappa_s}{\sqrt{[(3v_a - 1)\kappa_a + (2 - 3v_a)\kappa_s]^2 + 8\kappa_s\kappa_a}} \right), \quad (4)$$

where κ_s and κ_a are the thermal conductivities of the solid matrix and air, respectively; v_a is the corresponding volume fraction of air. The effective thermal conductivity at 300 K yields the maximum theoretical value of 49.5 W m⁻¹ K⁻¹ for S15, as well as S18 possesses the theoretical value of 45 W m⁻¹ K⁻¹. However, these theoretical values obviously overestimate the experimental thermal conductivities. We need to consider other

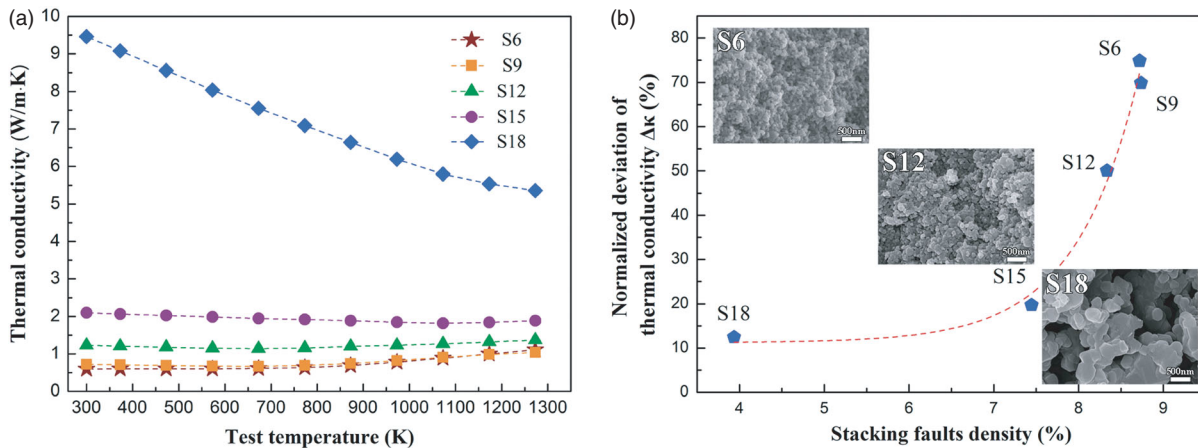


Figure 3. (a) Thermal conductivities of porous nano-SiC at elevated temperature and (b) normalized deviation between theoretical effective and experimental thermal conductivities of porous SiC ceramics with different densities of stacking faults [Colour online].

mechanism account for the extremely low thermal conductivity.

When grain size reduces down to nano-meter, the effects of phonon scattering at interfaces/grain boundaries are important. The thermal conductivity κ_{GB} of polycrystalline SiC is estimated by considering the thermal resistance of clean grain boundary. We refer to the method of Maldovan's model.[31] This model calculated the thermal conductivity for bulk polycrystalline SiC by considering intrinsic Umklapp phonon scattering, phonon-grain boundary scattering, and thermal boundary Kapitza conductance. The present grain size d inputted in Maldovan's model was determined by counting at least 100 grains in corresponding TEM images of each sample. Thereafter, the amorphous SiO₂ layer at grain boundary is further included because it has important effect on the Kapitza resistance for our samples. Clegg et al reported the reaction between SiC and surface SiO₂ started at 1,340°C, reached a maximum rate at 1,571°C and completed at 1,646°C.[32] Our experimental results are consistent with their conclusions. According to sintering temperature-dependent mass loss and TEM observations of porous SiC specimens, the thickness of amorphous SiO₂ layer is nearly unchanged below 1,200°C (~ 0.66 nm), decreases at 1,500°C (~ 0.24 nm) and completely disappears at 1,800°C. The thermal conductivity κ_{GB+SiO_2} of this hypothetic composite is calculated by [33]:

$$\begin{aligned} \kappa_{GB+SiO_2} &= \kappa_{SiO_2} \frac{2(1 - V_{SiO_2})(\kappa_{GB}/\kappa_{SiO_2} - 1) + \kappa_{GB}/\kappa_{SiO_2} + 2}{(1 - V_{SiO_2})(1 - \kappa_{GB}/\kappa_{SiO_2}) + \kappa_{GB}/\kappa_{SiO_2} + 2}, \end{aligned} \quad (5)$$

where κ_{SiO_2} and V_{SiO_2} denote the thermal conductivity and volume fraction of amorphous SiO₂ layer at grain boundary, respectively.

Integrating the interfacial thermal resistance (clean grain boundary and amorphous SiO₂ layer) with the EMT model of porous material (Equation (4)), more reasonable predictions of the effective thermal conductivities κ_{GB+SiO_2+P} of the present porous nano-SiC samples are obtained. Table 3 summarized all predicted thermal conductivities after taking into account different mechanisms. The theoretical predictions are remarkably reduced (from 49.5 W m⁻¹ K⁻¹ to 2.6 W m⁻¹ K⁻¹ for S15 and from 45 W m⁻¹ K⁻¹ to 10.8 W m⁻¹ K⁻¹ for S18) and exhibit the similar trend with experimental results. It is also noticed that when sintering temperatures are below 1,500°C, the measured thermal conductivities are noticeably lower than the predicted ones. The reason may be explained by the effects of both weak sintering neck and high concentration of stacking faults in these samples. Weak sintering neck provides less effective channel for heat conduction, and hence

Table 3. Theoretical effective thermal conductivities of porous nano-SiC ceramics after sequent consideration of clean grain boundary, amorphous SiO₂ layer at grain boundary, and porosity, together with experimental thermal conductivities for comparison.

T (°C)	κ_{GB} (W/m K)	κ_{GB+SiO_2} (W/m K)	κ_{GB+SiO_2+P} (W/m K)	$\kappa_{\text{expt.}}$ (W/m K)
600	19.2	15.1	2.4	0.6
900	19.2	15.1	2.4	0.7
1,200	19.2	15.1	2.5	1.2
1,500	19.2	17.6	2.6	2.1
1,800	81.9	81.9	10.8	9.5

significantly enhances thermal resistance. In the temperature region between 600°C and 1,200°C, the samples are not well sintered. This conclusion can be obtained from the slight change of BET values and thermal conductivities of S6 to S12. When temperature is above 1,500°C, strong sintering neck forms and the sample shows obvious enhancement of thermal conductivity as well as significant decrease in BET value. In addition, thermal conductivity is inversely proportional to the density of stacking faults ρ_{sf} due to the relationship between phonon-scattering rate τ_{sf}^{-1} and ρ_{sf} ($\tau_{sf}^{-1} = A\omega^2\rho_{sf}$). [34] In Figure 3(b), we plot the normalized deviation of thermal conductivity $\Delta\kappa$ at different density of stacking faults. $\Delta\kappa$ is calculated by:

$$\Delta\kappa = \frac{\kappa_{GB+SiO_2+P} - \kappa_{\text{expt.}}}{\kappa_{GB+SiO_2+P}} \quad (6)$$

Clearly, the data in Figure 3(b) are monotonously correlated. We noticed larger deviations for the samples sintered at lower temperatures (600°C, 900°C and 1,200°C). The difference might originate from the different morphologies of sintering necks.

Therefore it can be concluded that the nano-scale microstructural characters dominant the very low thermal conductivities of the present porous nano-SiC. However, another crucial concern may arise on the high electrical resistivity which will frustrate its promising application in high-temperature thermal-energy conversion. We measured the temperature-dependent electrical resistivity of S15 sample. Interestingly, the electric resistance decreases sharply from 27 Ω /cm at 373 K to 0.36 Ω /cm at 1,073 K. The extrapolated figure of merit (ZT) shows a reasonable value, ZT = 0.12 at 1,600 K. Therefore, the current idea for porous nano-SiC ceramic might be valuable for the development of new high-temperature thermoelectric material. We are now trying to further reduce the electrical resistivity, as well as the thermal conductivity, of porous nano-SiC by proper method, such as N-doping.

In summary, we herein investigate the thermal stability, mechanical properties and thermal conductivities of porous nano-SiC sintered at various temperatures

(from 600°C to 1,800°C). There exists high density of stacking faults in raw nano-powders and these defects significantly decrease at higher sintering temperatures. The thermal stability of nano-SiC grains could maintain up to 1,500°C. Porous nano-SiC can be fabricated with very low shrinkages (less than 0.53%) when sinter temperature is below 1,500°C and the sample keeps similar porosities (56.2–58%). The porous nano-SiC prepared at 1,500°C achieves good mechanical strength (compressive and flexural strength are 26 and 13 MPa, respectively, with 57% porosity) and simultaneously very low thermal conductivity ($2 \text{ W m}^{-1} \text{ K}^{-1}$ at 300 K). By comparison, porous SiC sample sintered at 1,800°C possesses superior R.T. mechanical strength (compressive and flexural strength are 121 and 30 MPa, respectively, with 58% porosity), excellent high-temperature stiffness (90% retention up to 1,500°C), and relatively low thermal conductivity ($9.5 \text{ W m}^{-1} \text{ K}^{-1}$ at 300 K). By analyzing the integrated effective thermal conductivity model, we disclose that the nano-scale microstructure characteristics (including high densities of grain boundaries, stacking faults and nano-pores) of porous nano-SiC are the effective factors that lead to the very low thermal conductivity. The present work sheds light on the porous nano-SiC as competitive high-temperature thermal-energy conversion material and promising thermal insulator. Our results also provide new wisdom on how to properly design performances of porous nano-SiC by solving the long-standing problem on the trade-off between high strength and low thermal conductivity through tailoring the microstructural characteristics.

Disclosure statement No potential conflict of interest was reported by the authors.

Funding This work was supported by the Natural Sciences Foundation of China under Grant Nos. [51032006] and [51372252].

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