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

Peng Wan\textsuperscript{a,b}, Zhen Wu\textsuperscript{a}, Hui Zhang\textsuperscript{a,b}, Liyin Gao\textsuperscript{a,b} and Jingyang Wang\textsuperscript{a}*

\textsuperscript{a}Shenyang National Laboratory for Materials Science, Institute of Metal Research, Chinese Academy of Sciences, Shenyang 110016, People’s Republic of China; \textsuperscript{b}University of Chinese Academy of Sciences, Beijing 100049, People’s Republic of China

(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.\textsuperscript{[1]} Conventional semiconductor materials, such as GeTe-Bi\textsubscript{2}Te\textsubscript{3} and PbTe, showed attractive performances at moderate operation circumstance, but their reliabilitys 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.\textsuperscript{[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.\textsuperscript{[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.

*Corresponding author. Email: jywang@imr.ac.cn

© 2016 The Author(s). Published by Taylor & Francis. This is an Open Access article distributed under the terms of the Creative Commons Attribution License (http://creativecommons.org/Licenses/by/4.0/), which permits unrestricted use, distribution, and reproduction in any medium, provided the original work is properly cited.
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 W m$^{-1}$ 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 W m$^{-1}$ 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$_2$O$_3$, MgO, SiO$_2$ and Y$_2$O$_3$ 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°C min$^{-1}$.

Phase composition was analyzed via a step-scanning X-ray diffraction (XRD) (Rigaku D/max-2400, Tokyo, Japan) with CuK$_\alpha$ radiation (λ = 1.54178 Å) 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 mm × 15 mm × 40 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 mm × 4 mm × 36 mm, the span is 30 mm and the crosshead speed is 0.5 mm min$^{-1}$. The samples with a size of Ø 8 mm × 8 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 (Ø 12.7 mm × 1 mm) from 473 to 1,273 K by a Flashline$^\text{TM}$-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
thermal conductivity \( \kappa \) (W m\(^{-1}\) K\(^{-1}\)) according to the following equation:

\[
\kappa = \alpha \cdot C_p \cdot \rho, \tag{1}
\]

in which \( \alpha \) is the thermal diffusivity (m\(^2\) s\(^{-1}\)), \( C_p \) the heat capacity (J kg\(^{-1}\) K\(^{-1}\)), and \( \rho \) 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 (\( \beta \)-SiC) and there is an additional diffraction peak at \( 2\theta \approx 33.7 \), which may be caused by the content of polytypes or stacking faults and twins in \( \beta \)-SiC powders. Pujar and Cawley systematically studied the effect of stacking faults and different polytypes on the XRD profiles of \( \beta \)-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 \( \rho_{sf} \) can be estimated by the following equation \[22\]:

\[
X = \frac{P_{33.6}}{P_{41.4}}, \tag{2}
\]

\[
\rho_{sf} = \frac{X}{6.82 \times 10^{-2}X + 2.27 \times 10^{-2} + 1.7X^3}, \tag{3}
\]

where \( P_{33.6} \) and \( P_{41.4} \) are the intensities of the peaks at 33.6° and 41.4° (2\( \theta \)), 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°C, and finally drops rapidly above 1,500°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 (111) 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°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°C, grain growth occurs and
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, compressive strength \( \sigma_c \) and flexural strength \( \sigma_f \), 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 \( \sigma_c \), and porosity are 9.3 m²/g, 122 MPa, and 39%, respectively, for sample processed at 1,500°C; and 1.9 m²/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 m²/g for sintered at 1,500°C and 3.0 m²/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
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 Al$_2$O$_3$-SiO$_2$-Y$_2$O$_3$ 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 nanograins 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 $\kappa_L$ of dielectric material crucially depends on the phonon relaxation time $\tau$. 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 $\tau_U$, $\tau_{gb}$, $\tau_{sf}$ and $\tau_{others}$ 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 $\kappa_p$ 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 $\kappa_p$ 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:

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

(4)

where $\kappa_a$ and $\kappa_s$ 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].](image)
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 $\kappa_{GB}$ 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 SiO$_2$ 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 SiO$_2$ 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 SiO$_2$ 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 $\kappa_{GB+SiO_2}$ of this hypothetic composite is calculated by [33]:

$$
\kappa_{GB+SiO_2} = \kappa_{SiO_2} \left( \frac{2}{1-V_{SiO_2}} \left( \frac{\kappa_{GB}}{\kappa_{SiO_2}} - 1 \right) + \frac{\kappa_{GB}}{\kappa_{SiO_2}} + 2 \right).
$$

(5)

where $\kappa_{SiO_2}$ and $V_{SiO_2}$ denote the thermal conductivity and volume fraction of amorphous SiO$_2$ layer at grain boundary, respectively.

Integrating the interfacial thermal resistance (clean grain boundary and amorphous SiO$_2$ layer) with the EMT model of porous material (Equation (4)), more reasonable predictions of the effective thermal conductivities $\kappa_{GB+SiO_2+p}$ 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 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 $\rho_{sf}$ due to the relationship between phonon-scattering rate $\tau_{sf}^{-1}$ and $\rho_{sf}$ ($\tau_{sf}^{-1} = A\omega^2 \rho_{sf}$).[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:

$$
\Delta \kappa = \frac{\kappa_{GB+SiO_2+p} - \kappa_{GB+SiO_2+p}}{\kappa_{GB+SiO_2+p}}.
$$

(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 $\Omega$ cm at 373 K to 0.36 $\Omega$ 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.

| $T$ (°C) | $\kappa_{GB}$ (W/m K) | $\kappa_{GB+SiO_2}$ (W/m K) | $\kappa_{GB+SiO_2+p}$ (W/m K) | $\kappa_{exp}$ (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             |

Table 3. Theoretical effective thermal conductivities of porous nano-SiC ceramics after sequent consideration of clean grain boundary, amorphous SiO$_2$ layer at grain boundary, and porosity, together with experimental thermal conductivities for comparison.
reported by the authors.

References

Funding

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

Discloser statement

No potential conflict of interest was reported by the authors.

References

[1] Pai CH. Thermoelectric properties of aluminum compo-
und-doped α-SiC ceramics. J Ceram Soc Jpn. 2014;122
(1430):870–875.

[2] Koumoto K, Shimohigoshi M, Takeda S, Yanagida H.
Thermoelectric energy conversion by porous SiC ceramics.
J Mater Sci Lett. 1987;6(12):1453–1455.

[3] Moskovskikh DO, Lin YC, Rogachev AS, McGinn PJ,
Mukasyan AS. Spark plasma sintering of SiC powders
produced by different combustion synthesis routes. J Eur
Ceram Soc. 2015;35(2):477–486.

[4] Han Y, Li CW, Bian C, Li SB, Wang CA. Porous anor-
thite ceramics with ultra-low thermal conductivity. J Eur
Ceram Soc. 2013;33(13–14):2573–2578.

[5] Ortona A, Pusterla S, Fino P, Mach FRA, Delgado A,
Biamino S. Aging of reticulated Si-SiC foams in porous
burners. Adv Appl Ceram. 2010;109(4):246–251.

[6] Mouazer R, Thijs I, Mullens S, Luyten J. SiC foams pro-
duced by gel casting: synthesis and characterization. Adv
Eng Mater. 2004;6(5):340–343.

[7] Sarikaya A, Dogan F. Effect of various pore formers
on the microstructural development of tape-cast porous
ceramics. Ceram Int. 2013;39(1):403–413.

[8] Jang BK, Sakka Y. Thermophysical properties of porous
SiC ceramics fabricated by pressureless sintering. Sci
Technol Adv Mater. 2007;8(7–8):655–659.

[9] Ohji T, Fukushima M. Macro-porous ceramics: processing
and properties. Int Mater Rev. 2012;57(2):115–131.

[10] Colombo P. Conventional and novel processing meth-
ods for cellular ceramics. Philos Trans R Soc, A.
2006;364(1838):109–124.

[11] Liang HQ, Yao XM, Zhang JX, Liu XJ, Huang ZR.
Low temperature pressureless sintering of α-SiC with
Al2O3 and CeO2 as additives. J Eur Ceram Soc.
2014;34(3):831–835.

[12] Yao XM, Tan SH, Zhang XY, Huang ZR, Jiang DL.
Low-temperature sintering of SiC reticulated porous ceramics
with MgO-Al2O3-SiO2 additives as sintering aids.
J Mater Sci. 2007;42(13):4960–4966.

[13] Fukushima M, Zhou Y, Yoshizawa YI. Fabrication and
microstructural characterization of porous SiC membrane
supports with Al2O3–Y2O3 additives. J Membr Sci.
2009;339(1–2):78–84.

[14] Nangrejo MR, Bao XJ, Edirisinghe MJ. Preparation of
silicon carbide-silicon nitride composite foams from pre-
ceramic polycarbosilane. J Eur Ceram Soc. 2000;20(11):1777–
1785.

[15] SciDi, Bellosi A. Effects of additives on densification,
microstructure and properties of liquid-phase sintered
silicon carbide. J Mater Sci. 2000;35(15):3849–3855.

[16] Chen PL, Chen JW. Sintering of fine oxide pow-
ders: 1, microstructural evolution. J Am Ceram Soc.
1996;79(12):3129–3141.

[17] Fukushima M, Zhou Y, Yoshizawa Y-I. Fabrication and
microstructural characterization of porous silicon carbide
with nano-sized powders. Mater Sci Eng, B. 2008;148(1–
3):211–214.

[18] Fukushima M, Zhou Y, Yoshizawa Y-I, Miyazaki H,
Hirao K. Preparation of mesoporous silicon carbide from
nano-sized SiC particle and polycarbosilane. J Ceram Soc
Jpn. 2006;114(1330):571–574.

[19] Sung IK, Yoon SB, Yu JS, Kim DP. Fabrication of macro-
porous SiC from templated preceramic polymers. Chem
Commun. 2002;(14):1480–1481.

[20] Wang H, Sung IK, Li XD, Kim DP. Fabrication of porous
SiC ceramics with special morphologies by sacrifi-
cing template method. J Porous Mater. 2004;11(4):
265–271.

[21] Pujar VV, Cawley JD. Effect of stacking faults on the X-
ray diffraction profiles of β-SiC powders. J Am Ceram
Soc. 1995;78(3):774–782.

[22] Seeo WS, Koumoto K. Kinetics and mechanism of stack-
ing fault annihilation and grain-growth in porous ceramics
of β-SiC. J Mater Res. 1993;8(7):1644–1650.

[23] Wang DH, Xu D, Qing W, et al. Periodically twinned SiC
nanowires. Nanotechnology. 2008;19(21):215602.

[24] Kennedy GP, Lim KY, Kim YW, Song IH, Kim
HD. Effect of SiC particle size on flexural strength
of porous self-bonded SiC ceramics. Met Mater Int.
2011;17(4):599–605.

[25] Eom JH, Kim YW, Song IH. Effects of the initial α-SiC
content on the microstructure, mechanical properties, and
permeability of macroporous silicon carbide ceramics.
J Eur Ceram Soc. 2012;32(6):1283–1290.
[26] Zuo KH, Zeng VP, Jiang DL. Mechanical properties of solid-sintered porous silicon carbide ceramics. Adv Eng Mater. 2013;15(6):491–495.

[27] Nakano H, Watari K, Kinemuchi Y, Ishizaki K, Urabe K. Microstructural characterization of high-thermal- conductivity SiC ceramics. J Eur Ceram Soc. 2004;24(14):3685–3690.

[28] Yamada I, Kume S, Nakano H, Watari K. Macro- and micro-scale thermal conductivities of SiC single crystal and ceramic. Key Eng Mater. 2009;403:179–183.

[29] Zhao HS, Liu ZG, Yang Y, Liu XX, Zhang KH, Li ZQ. Preparation and properties of porous silicon carbide ceramics through coat-mix and composite additives process. Trans Nonferrous Met Soc China. 2011;21(6):1329–1334.

[30] Gong LL, Wang YH, Cheng XD, Zhang RF, Zhang HP. Thermal conductivity of highly porous mullite materials. Int J Heat Mass Transfer. 2013;67:253–259.

[31] Maldovan M. Thermal energy transport model for macro- to-nano grain polycrystalline semiconductors. J Appl Phys. 2011;110(11):114310.

[32] Clegg WJ. Role of carbon in the sintering of boron-doped silicon carbide. J Am Ceram Soc. 2000;83(5):1039–1043.

[33] Sigl LS. Thermal conductivity of liquid phase sintered silicon carbide. J Eur Ceram Soc. 2003;23(7):1115–1122.

[34] Bilušić A, Smontara A, Dolinšek J, McGuiness P, Ott HR. Phonon scattering in quasicrystalline i-Al72Pd19.5Mn8.5: a study of the low-temperature thermal conductivity. J Alloys Compd. 2007;432(1–2):1–6.