Morphology modulation of SiC nano-additives for mechanical robust high thermoelectric performance $\text{Mg}_2\text{Si}_{1-x}\text{Sn}_x/\text{SiC}$ nano-composites

Kang Yin^a, Xianli Su^{a*}, Yonggao Yan^a, Hao Tang^a, Mercouri G. Kanatzidis^b, Ctirad Uher^c, and Xinfeng Tang^{a*}

a) State Key Laboratory of Advanced Technology for Materials Synthesis and Processing, Wuhan University of Technology, Wuhan 430070, P. R. China

b) Department of Chemistry, Northwestern University, Evanston, Illinois 60208, USA

c) Department of Physics, University of Michigan, Ann Arbor, Michigan 48109, USA

*Corresponding author. E-mail address: suxianli@whut.edu.cn, tangxf@whut.edu.cn

Abstract: SiC nano-powders and nano-wires with excellent toughness as well as high strength were incorporated in $\text{Mg}_{2.16}(\text{Si}_{0.3}\text{Sn}_{0.7})_{0.98}\text{Sb}_{0.02}$. The effect of the morphology and phase fraction of nano-SiC additives on the thermoelectric (TE) as well as mechanical properties of the composite was characterized in detail. It is found that, due to the pinning effect, fiber bridging and fiber pull-out mechanisms, the fracture toughness and the compressive strength of the composite with 0.8 at.% SiC nano-powders or nano-wires are improved by about 50% and 30%, respectively. And the TE properties changed little, with a maximum ZT value of ~ 1.20 at 750 K.

Keywords: $\text{Mg}_2\text{Si}_{1-x}\text{Sn}_x$; Nano-composite; Thermoelectric properties; Mechanical properties

Thermoelectric (TE) materials, which are capable of generating electrical power owing to the Seebeck effect, are promising as green and clean candidates for waste heat recovery in vehicles and industrial processes. The performance of TE materials is assessed by the dimensionless figure of merit, $ZT = S^2 \sigma T / \kappa_T$, where S , σ , T , and κ_T are the Seebeck coefficient, the electrical conductivity, the absolute temperature, and the total thermal conductivity, respectively. Therefore, high power factor ($S^2 \sigma$) and low thermal conductivity are essential for developing high performance TE materials. $\text{Mg}_2\text{Si}_{1-x}\text{Sn}_x$ ($0 \leq x \leq 1$) based solid solutions are a family of materials intended for TE power generation in the temperature range of 300-800 K, with the advantages of good TE performance, as well as inexpensive and environmentally friendly raw materials [1]. The phase composition and the TE performance of $\text{Mg}_2\text{Si}_{1-x}\text{Sn}_x$ solid solutions depend largely on the Si/Sn ratio, with the compositions near $x = 0.7$ being particularly promising. As reported recently [2, 3], band convergence occurs in $\text{Mg}_2\text{Si}_{0.3}\text{Sn}_{0.7}$, whereby the band edges of the light and heavy conduction bands coincide in energy. Such band convergence enhances the number of carrier valleys N_v participating in the transport process, leading to a significant enhancement of the density-of-states (DOS) effective mass $m^* = N_v^{2/3} m_b^*$ without increasing the band effective mass m_b^* . As a result, a sharp increase in the Seebeck coefficient is observed. The band convergence,

together with an overstoichiometric content of Mg and suitable Sb or Bi doping, enhances the ZT value of $\text{Mg}_2\text{Si}_{0.3}\text{Sn}_{0.7}$ beyond 1.30 at about 750 K [4-6]. Moreover, the most recent studies indicate that a single phase $\text{Mg}_2\text{Si}_{0.3}\text{Sn}_{0.7}$ will be obtained when heat-treated at about 1080 K [7], and a long term service life is possible when operated below 823 K in air with BN coatings [8]. Thus, solid solutions based on the $\text{Mg}_2\text{Si}_{0.3}\text{Sn}_{0.7}$ composition show excellent prospects as n-type legs for TE power generation [7].

However, for viable commercial applications, in addition to excellent TE properties, good mechanical properties are also very important. Usually, $\text{Mg}_2\text{Si}_{1-x}\text{Sn}_x$ solid solutions fail *via* the brittle fracture mode [8], i.e., no apparent plastic deformation takes place before fracture and cracks are easily observed due to the thermomechanical stress when working at high temperatures. Hence, it is challenging in the machining processes (like cutting and polishing) when fabricating modules based on $\text{Mg}_2\text{Si}_{1-x}\text{Sn}_x$ solid solutions. Previous studies found [9, 10] that $\text{Mg}_2\text{Si}_{1-x}\text{Sn}_x$ solid solutions have a reasonable value of the Vickers hardness (3.07-3.54 GPa at room temperature), but showed low fracture toughness ($\sim 0.48 \text{ MPam}^{1/2}$ at room temperature). Thus, improving the mechanical properties, especially the fracture toughness, is an important task to solve.

It has been shown that incorporating additives with high strength and high elastic modulus is a good way to improve the toughness of composites [11-13]. In this work, we make use of SiC nano-powders and nano-wires in an attempt to improve the mechanical properties of $\text{Mg}_2\text{Si}_{1-x}\text{Sn}_x$ solid solutions without any notable TE performance degradation. The choice of SiC is obvious. SiC possesses excellent mechanical properties (with a flexural strength of 640 MPa, and a fracture toughness of $3.8 \text{ MPam}^{1/2}$ at room temperature), low coefficient of thermal expansion ($2.77 \times 10^{-6} \text{ K}^{-1}$ at room temperature), low density (3.21 gcm^{-3}), and good thermal stability (melt at 3103 K) [14, 15]. The effects of the phase fraction and morphology (i.e. nano-powders *v.s.* nano-wires) of SiC on the thermoelectric and mechanical properties of $\text{Mg}_{2.16}(\text{Si}_{0.3}\text{Sn}_{0.7})_{0.98}\text{Sb}_{0.02}$ solid solution have been characterized in detail, and significant enhancements in the mechanical properties are observed while maintaining an excellent TE performance.

High purity powders of Mg (99%, 100-200 mesh), Si (99.99%, 200 mesh), Sn (99.9%, 200 mesh), Sb (99.999%, 200 mesh), SiC nano-powders (SiC_P , 3C-SiC, 99.99%, 40 nm, Aladdin), and SiC nano-wires (SiC_W , 3C-SiC, 99.9%, prepared by CVD, 0.1-0.6 μm in diameter with the length of 100-500 μm , Changsha Sinet Advanced Materials), were used as starting materials to prepare samples with the nominal composition of $\text{Mg}_{2.16}(\text{Si}_{0.3}\text{Sn}_{0.7})_{0.98}\text{Sb}_{0.02} + x \text{ SiC}_P / y \text{ SiC}_W$ ($x, y = 0, 0.002, 0.004, 0.006, 0.008$). A proven two-step solid state reaction method was applied here, with the initiated reaction at 873 K for 24 h and the second reaction at 1080 K for 24 h, samples were placed in the pyrolytic boron nitride (PBN) crucibles and sealed in the silica ampoules during the reactions [16]. The final consolidation into an ingot was done by spark plasma sintering at 930 K for 5 min as described elsewhere. An 8% excess of Mg over the stoichiometry was applied to

compensate for the evaporation loss of Mg during the preparation process [6].

The phase composition of the resulting materials was checked on a PANalytical X'Pert Empyrean X-ray diffractometer by Cu K_α radiation ($\lambda = 1.5406 \text{ \AA}$), within a range of 10-80 degrees and a step size of 0.02 degree. The morphology of the fractured surface was determined by the field-emission scanning electron microscopy (FESEM, Hitachi SU8020) equipped with EDX (XFlash 6160, Bruker). Additionally, for the sample shown in Figure 2, etching by a 2 vol.% nitric acid ethanol solution for 5-10 s was applied before the FESEM characterization. Transport properties, namely the electrical conductivity σ and the Seebeck coefficient S in the range of 300–800 K were obtained on a commercial ZEM-3 system (Ulvac Sinku-Riko) operated under helium gas. The Hall coefficient R_H at 300 K was measured by the five probe method using a physical property measurement system (PPMS, Quantum Design), using a magnetic field of $\pm 2 \text{ T}$. The carrier concentration n_H and the carrier mobility μ_H were obtained from the relations $n_H = 1/(eR_H)$ and $\mu_H = \sigma R_H$, respectively, assuming a single parabolic band and acoustic phonon scattering. The total thermal conductivity was determined from $\kappa_{\text{total}} = \lambda \rho C_p$, where λ is the thermal diffusivity acquired by the laser flash method (Netzsch LFA 457), ρ is the density measured by the Archimedes method in alcohol, and C_p is the specific heat obtained by a differential scanning calorimeter (TA DSC Q20) in argon. The overall errors in measurements of the electrical conductivity, the Seebeck coefficient and the thermal conductivity are estimated to be $\pm 5\%$, $\pm 3\%$, and $\pm 5\%$, respectively.

Compression tests and three-point flexural tests (the bending span was fixed at 12 mm) were carried out at room temperature using an MTS 810 testing system, with the sample size $3 \times 3 \times 6 \text{ mm}$ and $2 \times 2 \times 15 \text{ mm}$, respectively. 15 samples were prepared for each set of tests [17]. Hardness and fracture toughness was measured on a Vickers indenter (Wilson WH1150), with a load of 1 kg and a dwell time of 15 s. 3 samples were prepared for each set, and 10 indentions were made for each disk-shaped sample of 15 mm diameter and 3 mm thickness. Relevant equations for the calculation of the Vickers hardness H_V and the fracture toughness K_{IC} can be found elsewhere [18].

Figure 1 shows the powder X-ray diffraction patterns of the composites with different amounts of SiC nano-powders and nano-wires, together with the recalculated reference pattern of $\text{Mg}_2\text{Si}_{0.3}\text{Sn}_{0.7}$ according to JCPDS#01-089-4254. Due to the small amount of SiC ($< 1 \text{ at.}\%$), all samples seemed to be single phases, based on the XRD data. Metallographic etching followed by FESEM and EDX characterization was applied as another way to check the presence and form of SiC, namely, whether it might have reacted with $\text{Mg}_{2.16}(\text{Si}_{0.3}\text{Sn}_{0.7})_{0.98}\text{Sb}_{0.02}$, given that Mg is highly reactive. We take the structure with the highest content of SiC nano-wires (0.8 at.%) as an example, as the wires here are long enough and easy to be observed (the morphology of pure SiC nano-powders and nano-wires are presented in Figure 1, Ref. [19]). **Figure 2(a)** shows the morphology after etching, and **Figure 2(b)** depicts the elemental distribution maps corresponding to

Figure 2(a). Clearly, SiC shows no tendency to react with the matrix, and the nano-wires are dispersed randomly without any anisotropy in the composites.

Details regarding the distribution of SiC nano-powders and nano-wires in the composites are presented in **Figures 2(c) and 2(d)**. Judging from these data, the SiC nano-powders are dispersed uniformly on the grain boundaries of the matrix. The size of the nano-powder grains of 40-50 nm indicates that no growth took place during the preparation process (sharper images are shown in the inset of **Figure 2(c)** and Figure 2 in Ref. [19]). The SiC nano-wires were embedded randomly in the matrix. Some nano-wires with the diameter of 0.1-0.5 μm are observed on the fracture surface in Figure 2(d) (and more so in **Figure 4(f)** with a lower magnification). The different morphology (nano-powders or nano-wires) and the distribution pattern of SiC are expected to have a strong influence on the thermoelectric and mechanical properties of the composites.

The key transport parameters, such as the carrier concentration n_H , the carrier mobility μ_H , the reduced Fermi level η_F , the carrier effective mass m^* , and the Lorenz number L at room temperature are shown in Table 1. The data were obtained from **Eqs. (1)-(4)** assuming a single parabolic band model (SPB) under the relaxation time approximation and the dominance of acoustic phonon scattering ($r = -1/2$) [4]:

$$S = \pm \frac{k_B}{e} \left[\eta_F - \frac{(r+5/2) F_{r+3/2}(\eta_F)}{(r+3/2) F_{r+1/2}(\eta_F)} \right] \quad (1)$$

$$\eta_F = E_F / (k_B T) \quad (2)$$

$$F_i(\eta_F) = \int_0^\infty \frac{x^i dx}{1 + \exp(x - \eta_F)} \quad (3)$$

$$L = \left(\frac{k_B}{e} \right)^2 \left[\frac{3 F_0(\eta_F) F_2(\eta_F) - 4 F_1^2(\eta_F)}{F_0^2(\eta_F)} \right] \quad (4)$$

Here, η_F , $F_i(\eta_F)$, r , and k_B are the reduced Fermi level, the Fermi integral, the scattering factor, and the Boltzmann constant, respectively. As the data in Table 1 indicate, the incorporation of SiC does not alter the carrier concentration. However, due to the higher carrier mobility of SiC compared to the matrix, the carrier mobility of the composites is slightly higher than that of the matrix. This is especially the case of composites containing nano-wires [17]. [For the SiC nano-composites, the mainly difference in terms of the TE properties is ascribed to the carrier mobility, which has a strong relation with the morphology and interface structure of the nano-composites \[20-22\].](#) Surprisingly, with the increasing content of the SiC nano-phase, the carrier mobility stays essentially constant.

The temperature dependences of the electrical conductivity, the Seebeck coefficient, the power factor, the total thermal conductivity, the sum of the lattice part and the bipolar part of the thermal

conductivity, together with the ZT value of composites between 300 – 800 K are shown in **Figure 3**. As a consequence of the slightly increased carrier mobility in samples containing the SiC nano-phase, the electrical conductivity of the composites (particularly the nano-wire-based ones) is also slightly larger than the matrix, as depicted in **Figure 3(a)**. The Seebeck coefficient, shown in **Figure 3(b)** indicates an opposite trend, the neat matrix having a somewhat higher Seebeck coefficient than the SiC containing composites. This might be due to a slightly enhanced carrier mobility, according to **Eq. 5** [20]. As a consequence, the structures have comparable power factors in **Figure 3(c)**.

$$S = \frac{\pi^2}{3} \frac{k_B^2 T}{e} \left[\frac{d \ln \sigma(E)}{dE} \right]_{E=E_F} = \frac{\pi^2}{3} \frac{k_B^2 T}{e} \left[\frac{1}{n} \frac{dn(E)}{dE} + \frac{1}{\mu} \frac{d\mu(E)}{dE} \right]_{E=E_F} \quad (5)$$

The lattice and the bipolar contributions were estimated from the relation $\kappa_L + \kappa_B = \kappa_T - L\sigma T$. **Figure 3(d)** indicates that the total thermal conductivity is enhanced upon the presence of the SiC nano-phase, especially when in the form of nano-wires. However, as suggested by the data in **Figure 3(e)**, the lattice and the bipolar contributions to the total thermal conductivity change little in the presence of the nano-phase and, thus, the changes in the total thermal conductivity are due to the enhanced electrical conductivity. All in all, the composites still maintained an excellent TE performance with the figure of merit at a level of $ZT \sim 1.20$ at 750 K, see **Figure 3(f)**.

Mechanical properties, such as compressive strength, flexural strength, Vickers hardness, and fracture toughness of the composites measured at room temperature are shown in **Figures 4(a)-4(d)**. All the above mechanical parameters were improved to various degrees due to the excellent mechanical properties of SiC. Taking the composite with 0.8 at.% SiC as an example, the compressive strength of both nano-powder and nano-wire containing samples was improved by 30 % and 24 %, respectively, with respect to the matrix. The Vickers hardness was improved by about 10 %, regardless of whether the matrix was loaded with the SiC nano-powder or nano-wires. Nano-wires performed marginally better than the nano-powder regarding the fracture toughness (50 % improvement versus 42 % improvement). The nano-wires were particularly effective in improving the flexural strength, resulting in some 17 % improvement over the matrix while the nano-powder had essentially no effect. The toughening mechanism for the composites containing SiC nano-powders is mainly attributed to the pinning effect of SiC nano-powders dispersed at the grain boundaries and their ability to stop propagating cracks [23, 24]. Regarding the composites with SiC nano-wires, fiber bridging and fiber pull-out are believed to make a great contribution [25-27]. The above explanations are firmly supported by the experimental results shown in **Figures 4(e) and 4(f)**.

In conclusion, the effects of different amounts and morphology of SiC nano-additives on the

thermoelectric and mechanical properties of $\text{Mg}_2\text{Si}_{1-x}\text{Sn}_x$ were fully characterized. It is found that, because of the pinning effect, fiber bridging and fiber pull-out mechanisms, the compressive strength, the flexural strength, the Vickers hardness, and the fracture toughness of the composites were significantly improved due to the excellent mechanical properties of SiC. In particular, the fracture toughness was improved by about 50 % when 0.8 at.% of SiC nano-powders or nano-wires were added to the $\text{Mg}_2\text{Si}_{1-x}\text{Sn}_x$ matrix. Transport properties remained essentially inert to the presence of the SiC nano-phase and excellent thermoelectric performance, characterized by the maximum ZT value of 1.20 at 750 K, was maintained.

This work is financially supported by the National Basic Research Program of China under project 2013CB632502, the Natural Science Foundation of China (Grant No. 51402222, 51172174, 51521001), the 111 Project of China (Grant No. B07040), the Fundamental Research Funds for the Central Universities (WUT: 2013-YB-013), and the CERC-CVC joint U.S.-China Program supported by the U.S. Department of Energy (DE-PI0000012). We would like to thank Tingting Luo and Rong Jiang for their help in the Materials Research and Test Center of WUT.

Figure 1 PXRD patterns of $\text{Mg}_{2.16}(\text{Si}_{0.3}\text{Sn}_{0.7})_{0.98}\text{Sb}_{0.02}$ with different amounts of SiC nano-powder or nano-wires.

Figure 2 (a): FESEM images of $\text{Mg}_{2.16}(\text{Si}_{0.3}\text{Sn}_{0.7})_{0.98}\text{Sb}_{0.02}$ containing 0.8 at.% of SiC nano-wires after metallographic etching; (b): elemental distribution map of SiC nano-wires shown in Figure 2(a); (c): fracture surface of a composite containing 0.8 at.% of SiC nano-powders, inset: an enlarged view of SiC nano-grains

on the grain boundaries of the matrix; (d): fracture surface of a composite containing 0.8 at.% of SiC nano-wires.

Figure 3 Temperature dependence of (a) the electrical conductivity σ , (b) the Seebeck coefficient S , (c) the power factor PF , (d) the total thermal conductivity κ_T , (e) the sum of the lattice thermal conductivity and the bipolar contribution ($\kappa_L + \kappa_B$) as a function of temperature, and (f) the dimensionless figure of merit ZT .

Table 1 Room temperature electrical transport parameters of the composites with different amounts of SiC. Row entries represent the carrier concentration n_H , the carrier mobility μ_H , the reduced Fermi level η_F , and the effective mass m^*/m_0 .

SiC	$x, y=0$	$x=0.002$	$x=0.004$	$x=0.006$	$x=0.008$	$y=0.002$	$y=0.004$	$y=0.006$	$y=0.008$
n_H ($\times 10^{20} \text{ cm}^{-3}$)	2.33	2.32	2.31	2.35	2.34	2.31	2.33	2.32	2.35
μ_H ($\text{cm}^2 \text{ V}^{-1} \text{ s}^{-1}$)	61.35	66.91	67.61	66.54	66.69	73.28	73.19	73.42	72.84
$\eta_F = E_F / k_B T$	1.83	1.84	1.83	1.85	1.85	1.86	1.88	1.88	1.87
m^*/m_0	2.48	2.47	2.45	2.44	2.44	2.46	2.43	2.42	2.40

Figure 4 Room temperature mechanical properties, (a) the compressive strength; (b) the flexural strength; (c) the Vickers hardness; (d) the fracture toughness; (e) crack propagation in the ingot containing 0.8 at.% of SiC nano-powders; (f) fracture surface of the ingot containing 0.8 at.% of SiC nano-wires.

References

- [1] V. Zaitsev, M. Fedorov, E. Gurieva, I. Eremin, P. Konstantinov, A. Samunin, M. Vedernikov, Highly effective $\text{Mg}_2\text{Si}_{1-x}\text{Sn}_x$ thermoelectrics, *Phys. Rev. B* 74(4) (2006) 045207.
- [2] W. Liu, X. Tan, K. Yin, H. Liu, X. Tang, J. Shi, Q. Zhang, C. Uher, Convergence of conduction bands as a means of enhancing thermoelectric performance of n-type $\text{Mg}_2\text{Si}_{1-x}\text{Sn}_x$ solid solutions, *Phys. Rev. Lett.* 108(16) (2012) 166601.
- [3] X.J. Tan, W. Liu, H.J. Liu, J. Shi, X.F. Tang, C. Uher, Multiscale calculations of thermoelectric properties of n-type $\text{Mg}_2\text{Si}_{1-x}\text{Sn}_x$ solid solutions, *Phys. Rev. B* 85(20) (2012) 205212.
- [4] W. Liu, H. Chi, H. Sun, Q. Zhang, K. Yin, X. Tang, Q. Zhang, C. Uher, Advanced thermoelectrics governed by a single parabolic band: $\text{Mg}_2\text{Si}_{0.3}\text{Sn}_{0.7}$, a canonical example, *Phys. Chem. Chem. Phys.* 16(15) (2014) 6893-6897.
- [5] W. Liu, Q. Zhang, X. Tang, H. Li, J. Sharp, Thermoelectric properties of Sb-doped $\text{Mg}_2\text{Si}_{0.3}\text{Sn}_{0.7}$, *J. Electron. Mater.* 40(5) (2011) 1062-1066.
- [6] W. Liu, X. Tang, H. Li, J. Sharp, X. Zhou, C. Uher, Optimized thermoelectric properties of Sb-doped $\text{Mg}_{2(1+z)}\text{Si}_{0.5-y}\text{Sn}_{0.5}\text{Sb}_y$ through adjustment of the Mg content, *Chem. Mater.* 23(23) (2011) 5256-5263.
- [7] M.B.A. Bashir, S. Mohd Said, M.F.M. Sabri, D.A. Shnawah, M.H. Elsheikh, Recent advances on $\text{Mg}_2\text{Si}_{1-x}\text{Sn}_x$ materials for thermoelectric generation, *Renewable and Sustainable Energy Rev.* 37 (2014) 569-584.
- [8] H. Peng, C.L. Wang, J.C. Li, H.C. Wang, Y. Sun, Q. Zheng, Elastic and vibrational properties of $\text{Mg}_2\text{Si}_{1-x}\text{Sn}_x$ alloy from first principles calculations, *Solid State Commun.* 152(9) (2012) 821-824.
- [9] Y. Gelbstein, J. Tunbridge, R. Dixon, M.J. Reece, H. Ning, R. Gilchrist, R. Summers, I. Agote, M.A. Lagos, K. Simpson, Physical, mechanical, and structural properties of highly efficient nanostructured n-and p-silicides for practical thermoelectric applications, *J. Electron. Mater.* 43(6) (2014) 1703-1711.
- [10] P. Gao, I. Berkun, R. Schmidt, M. Luzenski, X. Lu, P. Bordon Sarac, E. Case, T. Hogan, Transport and mechanical properties of High-ZT $\text{Mg}_{2.08}\text{Si}_{0.4-x}\text{Sn}_{0.6}\text{Sb}_x$ thermoelectric materials, *J. Electron. Mater.* 43(6) (2014) 1790-1803.
- [11] W. Tuan, R. Brook, The toughening of alumina with nickel inclusions, *J. Eur. Ceram. Soc.* 6(1) (1990) 31-37.

- [12] T. Akao, Y. Fujiwara, Y. Tarui, T. Onda, Z.-C. Chen, Fabrication of Zn_4Sb_3 bulk thermoelectric materials reinforced with SiC whiskers, *J. Electron. Mater.* 43(6) (2014) 2047-2052.
- [13] N. Satyala, J.S. Krasinski, D. Vashaee, Simultaneous enhancement of mechanical and thermoelectric properties of polycrystalline magnesium silicide with conductive glass inclusion, *Acta Mater.* 74 (2014) 141-150.
- [14] C. Lorrette, A. Réau, L. Briottet, Mechanical properties of nanostructured silicon carbide consolidated by spark plasma sintering, *J. Eur. Ceram. Soc.* 33(1) (2013) 147-156.
- [15] O. Madelung, *Semiconductors: data handbook*, Springer Science & Business Media 2012, Chapt. 1 pp 60-69.
- [16] K. Yin, X. Su, Y. Yan, C. Uher, X. Tang, In situ nanostructure design leading to a high figure of merit in an eco-friendly stable $\text{Mg}_2\text{Si}_{0.30}\text{Sn}_{0.70}$ solid solution, *RSC Adv.* 6(20) (2016) 16824-16831.
- [17] Y. Zheng, Q. Zhang, X. Su, H. Xie, S. Shu, T. Chen, G. Tan, Y. Yan, X. Tang, C. Uher, G.J. Snyder, Mechanically robust BiSbTe alloys with superior thermoelectric performance: A case study of stable hierarchical nanostructured thermoelectric materials, *Adv. Energy Mater.* 5(5) (2014) 1401391.
- [18] R.D. Schmidt, X. Fan, E.D. Case, P.B. Sarac, Mechanical properties of Mg_2Si thermoelectric materials with the addition of 0–4 vol% silicon carbide nanoparticles (SiCNP), *J. Mater. Sci.* 50(11) (2015) 4034-4046.
- [19] K. Yin, X. Su, Y. Yan, H. Tang, M. G. Kanatzidis, C. Uher, X. Tang, Morphology modulation of SiC nano-additives for mechanical robust high thermoelectric performance $\text{Mg}_2\text{Si}_{1-x}\text{Sn}_x/\text{SiC}$ nano-composites, *Data in Brief*. submitted .
- [20] W. Zhou, X. Liu, Y. Zhang, Simple approach to β -SiC nanowires: Synthesis, optical, and electrical properties, *Appl. Phys. Lett.* 89(22) (2006) 223124.
- [21] J.Y. Fan, X.L. Wu, P.K. Chu, Low-dimensional SiC nanostructures: Fabrication, luminescence, and electrical properties, *Prog. Mater. Sci.* 51(8) (2006) 983-1031.
- [22] H.-K. Seong, H.-J. Choi, S.-K. Lee, J.-I. Lee, D.-J. Choi, Optical and electrical transport properties in silicon carbide nanowires, *Appl. Phys. Lett.* 85(7) (2004) 1256-1258.
- [23] Z.-Y. Deng, J.-L. Shi, Y.-F. Zhang, D.-Y. Jiang, J.-K. Guo, Pinning effect of SiC particles on mechanical properties of Al_2O_3 -SiC ceramic matrix composites, *J. Eur. Ceram. Soc.* 18(5) (1998) 501-508.
- [24] I. Levin, W.D. Kaplan, D.G. Brandon, A.A. Layyous, Effect of SiC submicrometer particle size and content on fracture toughness of alumina-SiC “nanocomposites”, *J. Am. Ceram. Soc.* 78(1) (1995) 254-256.
- [25] M.D. Thouless, A.G. Evans, Effects of pull-out on the mechanical properties of ceramic-matrix composites, *Acta Metall.* 36(3) (1988) 517-522.
- [26] N.-C. Lü, J. Cheng, Y.-H. Cheng, A dynamic model of bridging fiber pull-out of composite materials, *Mech. Res. Commun.* 32(1) (2005) 1-14.
- [27] S. Wan, X. Huang, P. Qiu, S. Bai, L. Chen, The effect of short carbon fibers on the thermoelectric and mechanical properties of p-type $\text{CeFe}_4\text{Sb}_{12}$ skutterudite composites, *Mater. & Des.* 67 (2015) 379-384.