Enhanced thermal conductivity and tensile strength of Al–17Si–3.5Cu with SiC-nanoparticle addition

D. P. Jiang* and J. K. Yu

The morphology and size of primary Si has a significant influence on the thermal conductivity (TC) and strength of Al–17Si–3.5Cu. In this study, the effect of a 1–3 wt% SiC nanoparticle (SiC\textsubscript{nps}) addition on TC and tensile strength of Al–17Si–3.5Cu was investigated. Nanoparticles distributed at the interface between primary Si and Al led to a significant refinement of primary Si; for example, a primary Si size of 2 \( \mu \)m with 3 wt% SiC\textsubscript{nps} addition was achieved. TC of SiC\textsubscript{nps}/Al–17Si–3.5Cu improved with an increase in nanoparticle content. Nanoparticles distributed at the interface between Si and Al reduced the interfacial thermal resistance. Thus, the effective TC of eutectic Si increased. Owing to the refinement of the primary Si and the increased interfacial thermal resistance, originating from the high content of SiC\textsubscript{nps} at the interface, the effective TC of primary Si decreased. Compared with Al–17Si–3.5Cu, contribution to the improvement of the TC of SiC\textsubscript{nps}/Al–17Si–3.5Cu resulted mainly from eutectic Si. Due to the refinement of primary Si, the tensile strength of SiC\textsubscript{nps}/Al–17Si–3.5Cu improved with an increase in SiC\textsubscript{nps} content. When the SiC\textsubscript{nps} content was 3 wt%, the yield strength, ultimate tensile strength and elongation of SiC\textsubscript{nps}/Al–17Si–3.5Cu were ~176 MPa, 418 MPa and 7%, respectively, which were improved by 37.5%, 53.7% and 218%, respectively, when compared with Al–17Si–3.5Cu.

1. Introduction

Hypereutectic Al–Si alloys are used extensively in many fields, such as automobiles, electronic packaging and military applications, due to their low mass, high corrosion resistance, high thermal conductivity (TC) and outstanding mechanical performance.\(^1\)–\(^4\) In the vehicle field, as piston materials, hypereutectic Al–Si–Cu alloys are required to have a high TC and excellent mechanical properties. Primary Si, an important phase that consists of hypereutectic Al–Si–Cu alloys, exerts a significant influence on the properties of the alloy. Coarse and irregular primary Si deteriorates the mechanical properties and machinability of Al alloys. As a result, numerous efforts have been made to refine primary Si. The refinement of primary Si in hypereutectic Al–Si alloys can be achieved by various methods, such as Al–P, Al–Fe–P and the Al–Ti–P–C master alloy, rare element and ultrasonic treatment, and electromagnetic stirring.\(^5\)–\(^14\) Although primary Si can be refined by traditional processes, many methods are either complex and cannot be used on a large scale, or they reduce the TC of the Al–Si alloy considerably.

Recent investigations have shown that Al\(_2\)O\(_3\) nanoparticles can refine the primary Si and eutectic Si in the local hypereutectic Al–Si–Cu alloy area simultaneously.\(^15\) Compared with these traditional methods, nanoparticle addition can refine primary Si and activate the Orowan strengthening effect in hypereutectic Al–Si alloys. Al\(_2\)O\(_3\) nanoparticles hinder the primary Si phase from growing by distributing on the Al/Si interface, thereby leading to the refinement of primary Si. It has also been illustrated that nanoparticles in hypoeutectic Al–Si alloys lead to the refinement of the eutectic Si phase, such as TiCN and AlN.\(^16\)–\(^17\) However, some investigations have focused on the effect of nanoparticles on the microstructures of hypereutectic Al–Si, inadequate studies exist on the effect of nanoparticles on the TC of hypereutectic Al–Si. Although nanoparticles have refined the microstructures of Al–Si, the effect of nanoparticles on the TC of the alloy remains unclear. Therefore, additional research focused on the TC of Al–Si is required.

Hasselman reported that the TC of composites is related to the thermal boundary resistance (TBR).\(^18\) The effective medium theory that was derived by Nan considered the interfacial thermal resistance and the size and shape of the reinforcement to predict the TC of composites that were reinforced with a low fraction of microsized particles.\(^19\) However, effective medium theory is not applicable for particles at the nanoscale. Due to increased interface scattering and because the nanoparticle size approaches the phonon mean-free path, TC of the nanoparticle and matrix are not equal to their bulk values. To address this problem, a modified effective medium approximation (EMA)
that considers the increased interfacial scattering and interfacial thermal resistance based on Nan’s model was proposed by Minnich and Chen. According to the modified EMA, the interfacial density and TBR affect the TC of the nanocomposite. In addition, Huang et al. investigated the effect of the standard deviation (SD) of the nanoparticle size as well as the chemical bonding of the TC of nanocomposites. SD can significantly affect the TC of nanocomposites. In addition, thermal boundary resistance can be greatly reduced with an increase in the chemical bonding ratio between two crystal phases.

These models are based on a uniform distribution of nanoparticles in the matrix. At present, several investigations are available on the TC of nanocomposites for graphene and CNT. The distribution of reinforcement is uniform in graphene- and CNT-reinforced metal matrix nanocomposites fabricated by solid-state or other processes. Unlike graphene- and CNT-reinforced nanocomposites, during solidification, nanoparticles will redistribute and segregate in the matrix via a liquid process. Furthermore, the redistribution of nanoparticles leads to the refinement of microstructures and the modification of microstructures’ morphology. Thus, additional research on the effect of nanoparticles on TC is necessary.

To address these issues, we studied Al–17Si–3.5Cu with the addition of 1.0–3.0 wt% SiC nanoparticles (SiC nanoparticles) via a solidification process combined with ultrasonic treatment. Unlike graphene- and CNT-reinforced nanocomposites, during solidification, nanoparticles will redistribute and segregate in the matrix via a liquid process. Furthermore, the redistribution of nanoparticles leads to the refinement of microstructures and the modification of microstructures’ morphology. Thus, additional research on the effect of nanoparticles on TC is necessary.

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3.5Cu alloy, and in addition to CuAl2, no other intermetallic crystalline materials were detected. Therefore, the SiCnps did not react visibly with the matrix alloy.

The morphology and size of primary Si in the Al–17Si–3.5Cu with different SiCnps additions are presented in Fig. 2. The microstructures of Al–17Si–3.5Cu were characterized by coarse

Fig. 2. Morphology of primary Si in Al–17Si–3.5Cu with different SiCnps additions: (a) Al–17Si–3.5Cu, (b) 1 wt% SiCnps/Al–17Si–3.5Cu, (c) 2 wt%/Al–17Si–3.5Cu, (d) 3 wt%/Al–17Si–3.5Cu, and (e) average size of primary Si vs. content of SiCnps.
primary Si with an average grain size of 30 μm and eutectic microstructures. Fig. 2(b)–(d) show that with an increase in SiCnps content from 1 wt% to 3 wt%, the primary Si is refined significantly and the coarse primary Si content per unit volume is reduced clearly. When the addition of SiCnps reaches 3 wt%, the coarse primary Si is refined completely. Hence, the addition of SiCnps leads to a refinement of primary Si in the hypereutectic Al–Si alloy. The average size variation of primary Si with the SiCnps content is shown in Fig. 2(e). When the SiCnps content is 1 wt%, the average size of the primary Si is 16 μm, which is reduced by 47% compared with Al–17Si–3.5Cu. The average size of the primary Si with 3 wt% SiCnps addition was reduced to 2 μm, which is a 97% decrease compared with Al–17Si–3.5Cu.

Fig. 3 shows SEM images of the 3.0 wt% SiCnps/Al–17Si–3.5Cu. Fig. 3(a) shows that compared with Al–17Si–3.5Cu, the primary Si and eutectic Si were refined extensively in 3.0 wt% SiCnps/Al–17Si–3.5Cu. The cross-sectional morphology of many Si phases became spheroidized. Because of the nanoparticles, the morphology of some Si phases transformed from platelets to more curved and rounder. Fig. 3(b) shows that the nanoparticles distribute uniformly and are dispersed completely in the eutectic Al. To obtain the nanoparticle composition, EDS spot analysis of the area marked with “+” in Fig. 3(b) was conducted. According to the EDS results, four elements were detected: C, Si, Al and Cu. According to the XRD results, CuAl2 is present in 3.0 wt% SiCnps/Al–17Si–3.5Cu. Hence, the Cu signal
comes from CuAl$_2$. Therefore, in addition to CuAl$_2$, SiC$_{\text{nps}}$ are present in eutectic Al.

To illustrate the distribution of nanoparticles, microstructures and elemental mapping micrographs for 3.0 wt\% SiC$_{\text{nps}}$/Al–17Si–3.5Cu are shown in Fig. 4. Fig. 4(a) shows the SiC$_{\text{nps}}$/Al–17Si–3.5Cu microstructures. Fig. 4(b)–(e) reveal the spatial distributions of C, Al, Si and Cu, respectively. Elemental C that is derived mainly from SiC$_{\text{nps}}$ is distributed mainly around the Si phase and on the Si phase surface. The remaining elemental C distributes in the eutectic Al. The elemental Cu is not enriched around the Si phase. This result shows that the refinement of Si phase results because of nanoparticle distribution on the surface of the Si phase.

SEM images of samples that were etched for 2 h are presented in Fig. 5. The three-dimensional morphology of the Si phase in Al–17Si–3.5Cu is characterized by a large thin plate-like eutectic Si and blocky primary Si. No particles adhere to the Si phase surface. Fig. 5(b) shows that some nanoparticles are distributed on the surface of primary Si in 1 wt\% SiC$_{\text{nps}}$/Al–17Si–3.5Cu. Fig. 5(b)–(d) show that with an increasing SiC$_{\text{nps}}$ content in the Al–17Si–3.5Cu, nanoparticle distribution on the primary Si surface increased. Fig. 5(d) shows that the primary Si is encapsulated by a dense layer of nanoparticles. The inset in Fig. 5(d) shows that nanoparticles on the surface of primary Si are clear, and except for some nanoclusters, the nanoparticles are well dispersed. Fig. 5(e) shows that some nanoparticles

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**Fig. 4** Microstructures and elemental mapping micrographs of 3.0 wt\% SiC$_{\text{nps}}$/Al–17Si–3.5Cu. (a) SEM images, and (b–e) mapping micrographs for elemental C, Al, Si and Cu, respectively.
Fig. 5  SEM images of primary Si in Al–17Si–3.5Cu with different SiC\textsubscript{np} contents that were etched using a 0.5% HF–water solution for 2 h: (a) Al–17Si–3.5Cu, (b) 1.0 wt% SiC\textsubscript{np}/Al–17Si–3.5Cu, (c) 2.0 wt% SiC\textsubscript{np}/Al–17Si–3.5Cu, (d) 3.0 wt% SiC\textsubscript{np}/Al–17Si–3.5Cu. Inset in (d) is the magnification of the local area on the surface of primary Si, (e) SEM images of eutectic Si in 3.0 wt% SiC\textsubscript{np}/Al–17Si–3.5Cu, and (f) EDS analysis of the area marked by “+” in (d).
Fig. 6  TEM bright field images of 3.0 wt%SiCnp/Al–17Si–3.5Cu (a) and the interface between Al and primary Si (b). (c) TEM bright field image of the eutectic Al. (d) HRTEM image of the interface between Al and primary Si. (e) and (f) EDS analysis of the area marked by “+” and “*” in (b) respectively, and (g)–(i) are the fast Fourier transformations (FFT) of primary Si, SiCnp and Al, respectively.
distribute on the surface of eutectic Si in 3 wt% SiC_{nps}/Al–17Si–3.5Cu. Because of the increase of nanoparticles in alloys, some nanoparticles distribute at the interface between eutectic Si and Al. To obtain the nanoparticle composition, EDS analysis was conducted on the marked area in Fig. 5(d). Fig. 5(f) shows that only elemental Si and C were detected. Therefore, nanoparticles on the primary Si surface are SiC. This result is consistent with the EDS element mapping analysis in Fig. 4.

TEM analysis was used to investigate the effect of SiC_{nps} on the primary Si microstructure. Fig. 6 shows a series of HRTEM images from 3.0 wt% SiC_{nps}/Al–17Si–3.5Cu. Fig. 6(a) shows that many nanoparticles, including CuAl_2 and SiC_{nps}, are distributed in the SiC_{nps}/Al–17Si–3.5Cu microstructures. Fig. 6(b) shows that some nanoparticles are distributed at the interface between the primary Si and Al. Nanoparticles at the interface between Si and Al restrict primary Si growth by inhibiting solute atom diffusion. Fig. 6(c) shows that some acicular and granular second phases, namely CuAl_2 and SiC_{nps}, respectively, are distributed in the eutectic Al. The nanoparticles are dispersed uniformly. Fig. 6(d) shows a HRTEM image of the interface between the primary Si and Al. SiC_{nps} with a 30 nm diameter and an interplanar distance of 0.24 nm distributes at the interface between primary Si and Al with an interplanar distance of 0.32 nm and CuAl_2 with an interplanar distance of 0.20 nm. Fig. 6 (e) and (f) show EDS analysis of the area marked by “+” and “*” in (b), respectively. From Fig. 6(e), four elements were detected: C, Si, Al and Cu. It indicates that two phases, SiC and CuAl_3, exist in eutectic Al. This result is consistent with Fig. 6(b). According to EDS results shown in Fig. 6(f), it can be seen that this phase is the Si phase. Fig. 6 (g)–(i) show the fast Fourier transformation of primary Si, SiC_{nps} and Al, respectively. The electron diffraction spot of the phases in Fig. 6 (g)–(i) was consistent with that of Si, SiC_{nps} and Al, respectively. No interfacial product was observed at the interface between the SiC_{nps} and Si and Al.

To evaluate the interface between the primary Si and Al, HADDF images of the interface between primary Si and Al are shown in Fig. 7(a). Many nanoparticles are distributed at the interface and acicular phase, and some granular phases exist in the eutectic Al. Fig. 7(b) and (c) show EDS spot analysis of the area marked with “*” and “+” in Fig. 7(a), respectively. EDS results of

![Fig. 7](image-url)

**Fig. 7** (a) High-angle annular dark-field (HADDF) images of the interface between primary Si and Al, (b) EDS spot analysis of area marked with “*” in (a); (c) EDS spot analysis of the area marked “+” in (a), and (d) EDS line-scan results of elemental Al, C, Si and Cu along white line in (a).
The granular phase shows that elemental Al, Si, and C are present. In addition to Al, Si, and C, the signal derives mainly from SiC nanoparticles. According to the EDS result in Fig. 7(c), elemental Cu and Al originate from the acicular phase. Combined with XRD analysis, the acicular phase is CuAl2. EDS line-scanning results of the interface between the primary Si and Al are shown in Fig. 7(d). The line-scanning results show that the elements Si and C are enriched at the interface between the primary Si and Al, which indicates that the SiC nanoparticles that are distributed at the interface lead to the refinement of primary Si. These results are consistent with the EDS mapping analysis in Fig. 4.

### 3.2 Effect of SiC nanoparticles on thermal conductivity and tensile strength of SiC nanoparticles/Al–17Si–3.5Cu

Fig. 8 shows that the thermal conductivity (TC) of 1 wt% SiC nanoparticles/Al–17Si–3.5Cu, 2 wt% SiC nanoparticles/Al–17Si–3.5Cu, 3 wt% SiC nanoparticles/Al–17Si–3.5Cu, and Al–17Si–3.5Cu was 98, 108, 114, and 89 W m\(^{-1}\)K\(^{-1}\), respectively. As the content of SiC nanoparticles increases, the TC of the SiC nanoparticles/Al–17Si–3.5Cu increases. Compared with Al–17Si–3.5Cu, the improvement in TC of the 3 wt% SiC nanoparticles/Al–17Si–3.5Cu was 28%, which indicates that the TC of SiC nanoparticles/Al–17Si–3.5Cu can be improved by SiC nanoparticles addition.

Engineering stress–strain curves and tensile properties of the SiC nanoparticles/Al–17Si–3.5Cu and Al–17Si–3.5Cu are shown in Fig. 9. The tensile properties of the SiC nanoparticles/Al–17Si–3.5Cu are significantly improved compared with those of Al–17Si–3.5Cu. The yield strength (YS), ultimate tensile strength (UTS), and elongation (EL) of SiC nanoparticles/Al–17Si–3.5Cu increased with the increase...
in SiC<sub>nps</sub> content. The YS, UTS and EL of 3 wt% SiC<sub>nps</sub>/Al–17Si–3.5Cu were ~176 MPa, 418 MPa and 7% respectively, which represents an increase of 37.5%, 53.7% and 218% compared with the Al–17Si–3.5Cu, respectively. The primary Si is refined significantly because of SiC<sub>nps</sub> distribution at the interface between the primary Si and Al, which results in an improvement in the UTS, YS and EL of the SiC<sub>nps</sub>/Al–17Si–3.5Cu. Furthermore, nanoparticles can inhibit dislocation motion, which leads to the Orowan strengthening effect. Compared with previous work,<sup>27–30</sup> the average size of the primary Si in this study is smaller, and the UTS and TC of our samples are higher than those previously reported, as shown in Fig. 9(c).

4. Discussion

4.1 Analyses of thermal properties

The previous analysis indicates that an increase in TC of SiC<sub>nps</sub>/Al–17Si–3.5Cu occurs because the addition of SiC<sub>nps</sub> and the distribution of SiC<sub>nps</sub> at the interface between the primary Si and Al leads to refinement of the primary Si. Therefore, the SiC<sub>nps</sub> distribution is closely related to TC of the sample. To investigate the effect of SiC<sub>nps</sub> distribution at the interface between Si and Al on SiC<sub>nps</sub>/Al–17Si–3.5Cu TC, SiC<sub>nps</sub> distributed at the interface between Si and Al were assumed to be a SiC<sub>nps</sub>/Al nanocomposite layer. Fig. 10 shows that compared with the Al/Si interface in Al–17Si–3.5Cu, hypothetical interfacial nanocomposite layers were added to the Al/Si interface in SiC<sub>nps</sub>/Al–17Si–3.5Cu.

To simplify the analysis, Si phases, including primary Si, eutectic Si and SiC<sub>nps</sub>, are regarded as spheres. CuAl<sub>2</sub> and Al are regarded as a whole, that is, a matrix phase.

Interfacial scattering needs to be considered in nanocomposites in which the characteristic nanoparticle length is of the order of or smaller than the phonon mean free path (MFP). Because of boundary scattering of the energy carriers in the nanoparticles with the nanoparticle surface, the TC \( k_p \) of the nanoparticles is given by:<sup>31</sup>

\[
k_p = \frac{k_p}{1 + l_p/r}
\]

where \( k_p \) is the bulk TC of the nanoparticles, \( l_p \) is the MFP of the energy carriers in the nanoparticles and \( r \) is the nanoparticle radius.

We consider the scattering of energy carriers in the matrix with the nanoparticle surface, and the TC \( k_m \) of the matrix is as follows:<sup>31</sup>

\[
k_m = \frac{k_m}{1 + l_m/\sigma_f}
\]

where \( k_m \) is the bulk TC of the matrix, \( l_m \) is the MFP of the energy carriers in the matrix, \( f \) is the content of nanoparticles in the nanocomposites and \( \sigma_f \) is the collision cross-section per unit volume of the nanoparticles. For spherical particles:

\[
\sigma_f = \frac{3}{4r}
\]

By replacing \( k_p \) and \( k_m \) in Nan’s model with their corresponding modified values, given by eqn (1) and (2), the TC \( K \) of the matrix, considering of interfacial scattering and interfacial thermal resistance, is determined by:

\[
K = \frac{k_p(1 + 2\alpha) + 2k_m + 2f[k_p(1 - \alpha) - k_m]}{k_p(1 + 2\alpha) + 2k_m - f[k_p(1 - \alpha) - k_m]}
\]

Compared with Al–17Si–3.5Cu, after SiC<sub>nps</sub> addition, the TC increment \( \Delta k^* \) of the matrix phase is given by:

\[
\Delta k^* = \frac{f_m[k_p(1 - \alpha) - k_m]}{k_p(1 + 2\alpha) + 2k_m - f_m[k_p(1 - \alpha) - k_m]}
\]

where \( f_m \) is the SiC<sub>nps</sub> content in the matrix and \( \alpha \) is a dimensionless parameter that is determined from:

\[
\alpha^{-1} = k_mk_p + a_k/r
\]

where \( a_k = R_{K}k_m \) called the Kapitza radius, and \( R_{K} \) is the interfacial thermal resistance \( (i.e., 6.85 \times 10^{-9} \text{ m}^2 \text{ K}^{-1} \text{ W}^{-1} \text{ for SiC and Al})^{32} \)

However, due to the concentration of nanoparticles distributed at the interface being higher than that of nanoparticles in the matrix, the size distribution of the nanoparticles has a great influence on the interfacial nanocomposite layer. Nanoparticle size distribution obeys a log-normal distribution, and ln \( \mu \) and \( \sigma \) are the mean and standard deviation of the variable’s natural logarithm, respectively. Considering the size distribution of the

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**Fig. 10** Schematic diagram of the interface between Si and Al in Al–17Si–3.5Cu and SiC<sub>nps</sub>/Al–17Si–3.5Cu: (a) the interface between Si and Al in Al–17Si–3.5Cu, (b) the interface between Si and Al in Al–17Si–3.5Cu with adding a small amount of SiC<sub>nps</sub>, and (c) the interface between Si and Al in Al–17Si–3.5Cu with adding a large amount of SiC<sub>nps</sub>.
nanoparticles, the average TC of the nanoparticles and the matrix are expressed as follows, respectively:

$$k_p^* = \frac{k_p}{1 + \mu k_p e^{-\frac{1}{2}}}$$  \hspace{1cm} (7)

$$k_m^* = \frac{k_m}{1 + \frac{3k_m}{4(1-\mu)} \exp \left(-\frac{5}{2}\frac{1}{\sigma^2}\right)}$$  \hspace{1cm} (8)

and the TC $k_{in}^{eff}$ of the interfacial nanocomposites layer is determined as:

$$\frac{k_{in}^{eff}}{k_m} = \frac{k_p^*(1 + 2\alpha) + 2k_m^* + 2f\left[k_p^*(1 - \alpha) - k_m^*\right]}{k_p^*(1 + 2\alpha) + 2k_m^* - f\left[k_p^*(1 - \alpha) - k_m^*\right]}$$  \hspace{1cm} (9)

Furthermore, to simplify the calculation, the SD of a log-normal distribution is as follows:

$$SD = \mu e^{\sigma^2/2} \sqrt{e^{\sigma^2} - 1}$$  \hspace{1cm} (10)

The total interfacial thermal resistance $R$ between Al and Si in SiCnps/Al–17Si–3.5Cu is expressed as:

$$R = R_{Si-in} + R_{Al-in} + R_{in}$$  \hspace{1cm} (11)

where $R_{Si-in}$ is the interfacial thermal resistance between Si and the interfacial nanocomposite layer, $R_{Al-in}$ is the interfacial thermal resistance between Al and the interfacial nanocomposite layer, and $R_{in}$ is the thermal resistance of the interfacial nanocomposite layer:

$$R_m = \frac{2\tau}{k_{in}^{eff}}$$  \hspace{1cm} (12)

We assume that the relationship for diffuse scattering of the interfacial thermal resistance $R$ from Chen[13] can be used:

$$R = 4\left(\frac{C_{1v_1} + C_{2v_2}}{C_{1v_1}C_{2v_2}}\right)$$  \hspace{1cm} (13)

where $\tau$ and $C$ are the phonon group velocity and the volumetric specific heat respectively.

The specific heat $C_{in}$ of the interfacial nanocomposites layers is defined as:

$$C_{in} = (1 - f)C_{Al} + fC_{SiC}$$  \hspace{1cm} (14)

The combined phonon group velocity $v_{in}$ of the interfacial nanocomposites layers is determined as:

$$\frac{1}{v_{in}} = \frac{1}{3\tau_{Al}} \left[\frac{\rho_{Al}}{\rho_{Al}} \left(1 - \frac{E_{Al} - E_{SiC}}{E_{SiC}} f + 2\sqrt{1 - \frac{G_{Al} - G_{SiC}}{G_{SiC}}} f\right)\right]$$  \hspace{1cm} (15)

where $v_{Al}$ is the phonon group velocity of Al; $E_{SiC}$, $G_{SiC}$ and $E_{Al}$, $G_{Al}$ are the Young’s modulus and shear modulus of Al and SiC, respectively; $\rho_{Al}$ and $\rho_{SiC}$ are the density of Al and SiC (i.e., 2.7 and 3.2 g cm$^{-3}$) and $\rho_{in} = (1 - f)\rho_{Al} + f\rho_{SiC}$.

The effective TC $k_{Si}^{eff}$ of Si is expressed as:

$$k_{Si}^{eff} = \frac{k_{Si}}{1 + 2R_{Si}k_{Si}}$$  \hspace{1cm} (16)

where $k_{Si}$ is the TC of Si, $R$ is the thermal resistance and $d$ is the Si size. The validity of eqn (16) for calculating the effective TC of irregular inclusions, such as graphite, has been confirmed previously.15

The thermal resistance $R$ between Al and Si in Al–17Si–3.5Cu is given by:

$$R_{Al-Si} = 4\left(\frac{\frac{C_{Al}V_{Al} + C_{Si}V_{Si}}{C_{Al}V_{Al}C_{Si}V_{Si}}}{C_{Al}V_{Al}C_{Si}V_{Si}}\right)$$  \hspace{1cm} (17)

Fig. 11 shows the effective TC of Si based on the parameters in Table 1 for SiCnps/Al–17Si–3.5Cu. Fig. 11(a) shows that compared with the effective TC of Si in Al–17Si–3.5Cu, the effective TC of Si is improved with a SiCnps addition, especially when the content of SiCnps is within a certain range (i.e., 0–0.5 for SD = 10); the Si size and SD have a significant effect on the effective TC of the Si. Fig. 11(b) shows that the interfacial thermal resistance between Si and Al increases with an increase in nanoparticle content at the interface between Si and Al, and decreases with an increase in SD in SiCnps/Al–17Si–3.5Cu. The rate of increase for the interfacial thermal resistance with SiCnps content in a certain range (i.e., 0–0.4 for SD = 30) is slow. Due to the small number of nanoparticles added, the number of nanoparticles at the interface is not very high. Hence, the number of nanoparticles at the interface has little effect on the thermal resistance between Si and Al. Moreover, many studies illustrate that Si can form a strong chemical bond with SiC.18,19

According to thermal boundary resistance, which decreases with increased bonding,22 the interfacial thermal resistance between SiC and Si is lower than that between Si and Al. The interfacial bonding between SiC and Al is similar to that between SiC and Si. Hence, the interfacial thermal resistance between Si and Al in SiCnps/Al–17Si–3.5Cu is consistently lower than in Al–17Si–3.5Cu, which indicates that nanoparticles can reduce the interfacial thermal resistance between Si and Al.

The key to analyze the influence of nanoparticles on the TC of SiCnps/Al–17Si–3.5Cu is to analyze the influence of nanoparticles on the TC of primary Si and eutectic Si.

Because nanoparticles lead to a refinement of primary Si, the TC increment $\Delta k$ of primary Si in SiCnps/Al–17Si–3.5Cu with SiCnps addition is shown as:

$$\Delta k = \frac{k_{Si}^*}{1 + \frac{2R_{Al-Si}k_{Si}}{d}f} - \frac{k_{Si}}{1 + \frac{2R_{Al-Si}k_{Si}}{d}f}$$  \hspace{1cm} (18)

where $R_{Al-Si}^*$ and $d$ are the interfacial thermal resistance between the primary Si and Al and the size of primary Si in SiCnps/Al–17Si–3.5Cu, respectively, and $R_{Al-Si}$ and $d_0$ are the interfacial thermal resistance between the primary Si and Al and the size of primary Si in Al–17Si–3.5Cu, respectively.
Because the effect of nanoparticles on the size of eutectic Si is not as significant as that of primary Si, the size of eutectic Si can be considered to be unchanged. The TC increment $\Delta k$ of eutectic Si in SiC$_{nps}$/Al–17Si–3.5Cu with the addition of SiC$_{nps}$ is determined by:

$$\Delta k = \frac{2k^2_{Si}(R_{Al-Si} - R^*_{Al-Si})D}{(D + 2R^*_{Al-Si}k_{Si})(D + 2R_{Al-Si}k_{Si})} \quad (19)$$

where $D$ is the size of eutectic Si.

Fig. 12 shows the TC increment of primary Si, eutectic Si and the matrix in SiC$_{nps}$/Al–17Si–3.5Cu. Fig. 12(a) shows that the effective TC of primary Si decreases with SiC$_{nps}$ addition, but is reduced less with an increase in SD. The increasing content of nanoparticles lead to an increase in the number of nanoparticles at the interface between the primary Si and Al, which in turn refined the primary Si. The refinement of primary Si and the high content of nanoparticles at the interface between primary Si and Al lead to the increase in interfacial thermal resistance. When the content of nanoparticles is high, the SD has a significant effect on the effective TC of the primary Si. The increase in SD leads to less of a reduction in the effective TC of primary Si. Therefore, compared with that of Al–17Si–3.5Cu, the effective TC of primary Si decreased. Fig. 12(b) shows that compared with Al–17Si–3.5Cu, the effective TC of the eutectic Si increased in SiC$_{nps}$/Al–17Si–3.5Cu. Because fewer nanoparticles were present on the surface of eutectic Si compared with those on the surface of primary Si, the effect of nanoparticles on the size of the eutectic Si was less significant compared with the primary Si. Furthermore, compared with Al–17Si–3.5Cu, because of a small number of nanoparticles on the surface of eutectic Si, the interfacial thermal resistance between Al and eutectic Si was reduced considerably. When the size of the eutectic Si was smaller and the SD was large, the increase in the effective TC of the eutectic Si, which was caused by the nanoparticles, was more significant. Thus, the effective TC of the eutectic Si was improved with the SiC$_{nps}$ addition. Fig. 12(c) shows that the TC of the matrix decreases with an increase in SiC$_{nps}$ content. When 3 wt% nanoparticles were added to the alloys, the content of nanoparticles in the matrix was less than 3%, because a large number of nanoparticles were distributed at the Si/Al interface. Therefore, compared with that of the matrix in Al–17Si–3.5Cu, the TC of the matrix in 3 wt% SiC$_{nps}$/Al–17Si–3.5Cu decreased by less than 3%. Because a small number of nanoparticles were added to the alloys, nanoparticles caused a limited decrease in the matrix TC. In addition, the above analysis shows that the TC of the alloys will decrease significantly when a large number of nanoparticles are added.

The improvement in the TC of SiC$_{nps}$/Al–17Si–3.5Cu resulted mainly from the eutectic Si. Compared with Al–17Si–3.5Cu, nanoparticles led to a decrease in the effective TC of primary Si and the matrix in SiC$_{nps}$/Al–17Si–3.5Cu. However, due to the small interfacial thermal resistance between the eutectic Si and Al, the effective TC of the eutectic Si increased significantly.

### Table 1 Material parameters for theoretical calculations

| Materials | TC (W m$^{-1}$ K$^{-1}$) | Density (g cm$^{-3}$) | Specific heat (J g$^{-1}$ K$^{-1}$) | Phonon velocity (m s$^{-1}$) | Young’s moduli (GPa) | Shear moduli (GPa) |
|-----------|--------------------------|------------------------|-------------------------------|-----------------------------|----------------------|-------------------|
| Al        | 237                      | 2.7                    | 0.895                         | 3595                        | 68                   | 26                |
| SiC       | 248                      | 3.21                   | 0.29                          | 11 600                      | 448                  | 192               |
| Si        | 148                      | 2.33                   | 0.7                           | 5860                        | —                    | —                 |
4.2 Analyses of tensile properties

The significant improvement of YS, UTS and EL of SiCnps/Al–17Si–3.5Cu compared with Al–17Si–3.5Cu resulted because of: (i) the refinement of primary Si. Nanoparticles at the interface between primary Si and Al restrict the growth of the primary Si, which leads to a refinement of primary Si. Numerous investigations have illustrated that crack initiation originates easily from primary Si and eutectic Si. Furthermore, cracks may originate from and propagate along the interface between the primary Si and Al. The refinement of primary Si can reduce or eliminate premature crack initiation and fracture. The quantitative relationship of the intrinsic fracture stress ($\sigma_f$) on the primary Si is given by Griffith:}

$$\sigma_f = \left( \frac{2E\gamma}{\pi C} \right)^{\frac{1}{2}}$$  \hspace{1cm} (20)

where $E$ is the Young's modules of the particles, $\gamma$ is the fracture surface energy and $C$ is the internal defect length. Because of its shorter internal defects compared with that in the coarse primary Si, the fracture stress of the refined primary Si is larger. In addition, the coarse primary Si tends to cause a stress concentration and then crack initiation. Hence, the refinement of primary Si can improve the tensile strength of SiCnps/Al–17Si–3.5Cu. (ii) Orowan strengthening effect. In addition to nanoparticle distribution at the interface between Si and Al, some nanoparticles were distributed uniformly in the eutectic Al. Nanoparticles can hinder the motion of dislocation and pine dislocations and enhance the tensile strength of SiCnps/Al–17Si–3.5Cu. Furthermore, the load transfer between the Al and nanoparticles provides a strengthening effect during the tensile process.

Fig. 12 Compared with Al–17Si–3.5Cu, TC increment of primary Si (a), eutectic Si (b) and the matrix (c) in SiCnps/Al–17Si–3.5Cu respectively.
5. Conclusions
SiC$_{\text{nps}}$/Al–17Si–3.5Cu, in conjunction with an addition of 1 wt%, 2 wt% and 3 wt% SiC$_{\text{nps}}$ was prepared in this study, and the effect of SiC$_{\text{nps}}$ on the microstructures, TC and tensile properties of SiC$_{\text{nps}}$/Al–17Si–3.5Cu was investigated.

SiC$_{\text{nps}}$ addition leads to the refinement of primary Si. When SiC$_{\text{nps}}$ is 3 wt%, the primary Si is refined significantly. Nanoparticles at the interface between the Si and Al hinder the growth of the primary Si, which leads to the refinement of primary Si.

The TC of SiC$_{\text{nps}}$/Al–17Si–3.5Cu increases with an increase in SiC$_{\text{nps}}$ content. Nanoparticles can reduce the interfacial thermal resistance between Si and Al and increase the effective TC of the Si. The improvement in the TC of SiC$_{\text{nps}}$/Al–17Si–3.5Cu results mainly from eutectic Si.

The YS, UTS and EL of SiC$_{\text{nps}}$/Al–17Si–3.5Cu increases with an increase in SiC$_{\text{nps}}$ content. When the SiC$_{\text{nps}}$ content is 3 wt%, the primary Si is refined significantly. Nanoparticles at the interface between the Si and Al hinder the growth of the primary Si, which leads to the refinement of primary Si.

The TC of SiC$_{\text{nps}}$/Al–17Si–3.5Cu increases with an increase in SiC$_{\text{nps}}$ content. Nanoparticles can reduce the interfacial thermal resistance between Si and Al and increase the effective TC of the Si. The improvement in the TC of SiC$_{\text{nps}}$/Al–17Si–3.5Cu results mainly from eutectic Si.

The YS, UTS and EL of SiC$_{\text{nps}}$/Al–17Si–3.5Cu increases with an increase in SiC$_{\text{nps}}$ content. When the SiC$_{\text{nps}}$ content is 3 wt%, the primary Si is refined significantly. Nanoparticles at the interface between the Si and Al hinder the growth of the primary Si, which leads to the refinement of primary Si.

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The YS, UTS and EL of SiC$_{\text{nps}}$/Al–17Si–3.5Cu increases with an increase in SiC$_{\text{nps}}$ content. When the SiC$_{\text{nps}}$ content is 3 wt%, the primary Si is refined significantly. Nanoparticles at the interface between the Si and Al hinder the growth of the primary Si, which leads to the refinement of primary Si.

Data availability
The raw data required to reproduce these findings are available to download from DOI: 10.17632/xgmh9x8yp8.2.

Conflicts of interest
There are no conflicts to declare.

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