Original Article

Effects of TiC$_{0.5}$N$_{0.5}$ nanoparticles on the microstructure, mechanical and thermal properties of TiC$_{0.5}$N$_{0.5}$/Al-Cu nanocomposites

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In this work, TiC$_{0.5}$N$_{0.5}$/Al-Cu nanocomposites with high mechanical and thermal performance were successfully synthesized via ultrasonic assisted casting route. The effects of TiC$_{0.5}$N$_{0.5}$ nanoparticle on the microstructure, mechanical and thermal properties of nanocomposites were systematically investigated from micro to nano scale. The microstructural analysis reveals that the TiCN nanoparticles in the matrix mainly exhibit two distribution patterns, i.e. intergranular and intragranular distribution. The experimental results show that the increased addition level can lead to the more refined and homogeneous microstructure and enhanced mechanical and thermal properties. The effects of ultrasonic treatment (UT) and NP addition on the microstructure evolution were discussed and it was found that the enhanced nucleation under ultrasonication and NP-induced growth restriction were the main reasons for the refinement of primary α-Al dendrite and Al$_3$Cu phase. In addition, the 2 vol.% TiC$_{0.5}$N$_{0.5}$/Al-Cu nanocomposite can exhibit the optimum mechanical and thermal properties with its hardness, ultimate tensile strength, yield strength, elongation increased by 30%, 26%, 30% and 46% and coefficient of thermal expansion (CTE) decreased by 6.2%, respectively as compared to the matrix alloy.

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1. Introduction

The development of particulate-reinforced aluminum matrix composites (AMCs) has great practical significance in automotive, aerospace and electronic industries due to their high specific strength and modulus, low thermal expansion coefficient, desirable wear resistance as well as ease of fabrication [1–3]. So far, a myriad of ceramic particles including Al$_2$O$_3$, SiC, TiB$_2$ and TiC [4–7] have been utilized to reinforce AMCs. Of them, TiCN is a promising reinforcement for AMCs because it is considered as a solid solution for FCC TiN and FCC TiC and has the advantages of both sides such as high melting point, hardness and modulus, and low thermal expansion coefficient [8]. As such, TiCN can impart enhanced mechanical properties to the matrix alloy. Furthermore, studies show that TiCN has low interfacial energy and desirable wettability with aluminum [9]. More importantly, unlike the conventional...
reinforcing particles like SiC and TiC, which are thermodynamically unstable and tend to react with Al to form the water-soluble, brittle Al$_4$C$_3$ causing the interfacial corrosion and the embrittlement of the composites. TiCN has high thermal and chemical stability and no or little harmful reactions may occur at the Al/TiCN interface. Therefore, a strong interfacial bonding may be obtained between TiCN and the Al matrix, which may induce the improved load transfer efficiency.

Nowadays, the fabrication of AMCs can be divided into two main categories. one is the solid-state route e.g. powder metallurgy [10], the other is the liquid-state route e.g. stir casting [11]. The former is usually relatively complicated in process and cost-inefficient. Also, as the sintering is usually performed at high temperatures, not only are the undesirable reactions between the matrix and particles likely to occur to form brittle intermetallics, but also the grain coarsening may be induced leading to the degradation of the composite properties. By contrast, the latter is capable of mass-producing AMCs with complex shapes and limitless sizes and thus taken as a cost-efficient technique for AMC fabrication [12,13]. It involves the incorporation of particles into the molten melt. In this case, the particles are inclined to agglomerate and form clusters owing to the presence of attractive van der Waals forces, especially when the particle size is in the nanoscale range [14]. The non-uniform distribution of reinforcements can lead to the crack initiation and propagation compromising the reinforcement effect. In addition, the poor wettability between particles and the matrix due to the presence of the oxide layer on liquid metal can lead to the weak interfacial bonding strength and thus ineffective load transfer [15].

To address these problems, ultrasonic assisted casting technique has been developed in the past few decades. Ultrasonic treatment (UT) has been successfully applied to produce AMCs [16–18]. Many studies [19,20] demonstrate that the introduction of high-intensity ultrasound into the melt during solidification can induce the nonlinear cavitation and acoustic streaming, which are effective in disrupting Al oxide films, enhancing the matrix-particle wettability and dispersing them uniformly in the matrix. Li et al. [16] suggested that UT is an effective approach to disperse SiC nanoparticles (NPs) in Al-5 wt.%Cu composites and found that after UT for 5 min, the NPs were uniformly distributed in the composites. Liu et al. [17] stated that the uniform distribution of in situ TiB$_2$ particles in Al-4.5 wt.%Cu composites had been obtained via UT, and the average size of in situ TiB$_2$ was found to be around 100 nm. Tsunekawa et al. [18] found that the wettability between the oxide particles and the matrix could be improved using UT. Although many researchers have investigated the effect of various ceramic particles on the microstructure and properties of AMCs produced by UT, a very limited study is focused on the development of TiCN nanoparticle reinforced aluminum matrix composites [16–18,20]. In particular, the effects of TiCN nanoparticles on the microstructure, mechanical and thermal properties of Al-Cu alloys remain unclear. The mechanisms underlying the microstructure and property evolution need to be clarified.

In the present work, TiC$_{0.5}$N$_{0.5}$/Al-Cu nanocomposites were fabricated via ultrasonic assisted casting route. The effects of NP content and distribution on the microstructure, mechanical and thermal properties of Al-Cu nanocomposites were systematically examined from micro to nano scale. The mechanisms of microstructure and property evolution of nanocomposites were investigated as well.

2. Experimental

The commercial-purity (CP) Al (99.7 %) and Al-50 wt. %Cu master alloy, as supplied commercially by Jiangsu Taicheng metal material manufacturing Co., Ltd, were used for the preparation of Al-4.5 wt. %Cu matrix alloy. TiC$_{0.5}$N$_{0.5}$ nanoparticles, as supplied commercially by Shanghai st-nano science and technology Co., Ltd, were adopted as particle reinforcements, which have an average size of 80 nm and quadrilateral morphology. The synthesis of TiC$_{0.5}$N$_{0.5}$/Al-Cu nanocomposites was performed using ultrasonic assisted casting route. Firstly, the Al-4.5Cu matrix alloy was melted in an alumina crucible by an electric resistance furnace. To prepare the nanocomposites with different NP contents, 0, 0.5, 1.0, 2.0 and 2.5 vol.% TiC$_{0.5}$N$_{0.5}$ NPs were wrapped with thin foils and fed into the melts, respectively. The UT system shown schematically in [21] was employed to disperse TiC$_{0.5}$N$_{0.5}$ NPs. The ultrasonic treatment was conducted at 993 K for 20 min in an inert atmosphere of argon gas. The power and frequency of ultrasonic generator were set as 2.8 kW and 20 kHz, respectively. After completing the NP dispersion, the ultrasonic probe was lifted out of the melt, and the melt was then cast into a cylindrical permanent mold (25 mm in diameter and 150 mm in height) preheated to 623 K.

The cast samples were ground, polished, and then etched. The etched samples were imaged using a polarized light optical microscope. The α-Al dendrite size was determined via the linear intercept method, as described by ASTM E112-10. The JEOL JSM-7600F FE-SEM coupled with energy dispersive X-ray spectra (EDX) was applied to characterize the nanocomposite microstructure and the distribution of NPs. The ion-beam-milled TEM foils produced using a Gatan Precision Ion Polishing System (PIPS, Gatan 691) were examined in a JEOL 2100 TEM at 200 kV. HRTEM analysis was performed using a JEOL 2100 F FEGTEM at 300 kV.

Tensile tests were performed at ambient temperature using Zwick/Roell Z100 testing machine. It was carried out three times for each sample and three readings were averaged. Vickers hardness of each sample was measured by a FIE-VM50 PC Vickers hardness tester under a load of 5 kg. The hardness of each sample is an average of at least five readings. The coefficient of thermal expansion (CTE) of the nanocomposites was determined from room temperature to 200 °C at a heating rate of 5 °C/min by a Dilatometer Linseis L75 thermomechanical analyzer.

3. Results and discussion

3.1. Effect of NPs on the microstructural evolution

Fig. 1 presents the anodized micrographs of the matrix alloy and the nanocomposites with different NP additions. The microstructure of the matrix alloy is characterized by coarse equiaxed dendritic grains with an average size of
approximately 246 μm as observed in Fig. 1(a). An increment in NP addition level can lead to the formation of much finer-equiaxed dendrites. There is an obvious trend that the dendrite size is remarkably decreased with NP content. When the NP content reaches 2 vol.%, a pronounced transition from coarse to fine equiaxed grain structure occurs throughout the matrix, and the dendrite size is significantly refined to 38 μm by approximately 85% as compared to the matrix alloy.

The X-ray diffraction patterns of the matrix alloy and the nanocomposites with various NP contents are illustrated in Fig. 2. From Fig. 2(a), it is clear that the matrix alloy consists of primary α-Al and intermetallic Al₂Cu phases. Fig. 2(b) to (d) exhibit the presence of TiC₀.₅N₀.₅ phases in the matrix, in addition to α-Al and Al₂Cu phases. As for TiC₀.₅N₀.₅, the peaks of (111) and (200) to which 2θ of 36° and 42° correspond, respectively, indicate that the diffraction intensity increases with increasing NP additions. The XRD results reveal that TiC₀.₅N₀.₅ nanoparticles have been successfully incorporated into the matrix alloy via UT.

Fig. 3 shows the typical eutectic structure of the matrix alloy and the nanocomposites. The eutectic microstructure of the matrix alloy is composed of eutectic α-Al and Al₂Cu, as shown in Fig. 3(a). The Al₂Cu phases usually precipitate along the grain boundary at the end of solidification and thus form the network distribution in the matrix. With increasing the NP content, Al₂Cu phases are gradually refined and exhibit a discontinuous network distribution when the content reaches 2.0 vol.% (Fig. 3(d)).

3.2. NP distribution patterns

The SEM images of 2.0 vol.% TiC₀.₅N₀.₅/Al-Cu nanocomposite are illustrated in Fig. 4. From Fig. 4(a), it can be observed that Al₂Cu phases are distributed along the α-Al grain boundary. Fig. 4(b) shows that a portion of NPs are uniformly distributed in the matrix though several NPs agglomerate together to form submicron NP clusters. Considering that the matrix does not contain Ti element, Ti peaks in the inset of Fig. 4(b) may confirm the intragranular distribution of TiC₀.₅N₀.₅ nanoparticles. In addition, NPs are also observed to be distributed along the α-Al grain boundary coexisting with Al₂Cu phase as shown in Fig. 4 (c) and (d). The intergranular distribution of NPs can be...
Fig. 3 – Optical microstructures of (a) Matrix alloy; (b) 0.5 vol.%; (c) 1.0 vol.%; (d) 2.0 vol.%.

Fig. 4 – SEM images and EDX analysis of 2 vol.% TiC$_{0.5}$N$_{0.5}$/Al-Cu nanocomposite.
further confirmed from Fig. 5. From Fig. 5(a) and (b), it is obvious that a large proportion of NPs accumulates on the surface of the Al_{2}Cu phase precipitated along the grain boundary. The SEM images combined with EPMA element mappings verify a NP layer existing along the grain boundary.

Whether NPs can be pushed or captured by the solidification fronts (SF) depends primarily on the interaction between NPs and the advancing SF during solidification. A number of theories [22-24] have been developed to account for the physics of particle-SF interaction. In the kinetics criterion [24], there exists a critical velocity for the particle pushing/capture transition, derived by balancing the repulsive and attractive forces acting on the particles, below which particles are pushed, and above which particle capture occurs. Our previous studies [25] reveal that under routine solidification conditions, the solidification rate is generally quite low, far less than the critical velocity. Therefore, the TiC_{0.5}N_{0.5} nanoparticles are highly likely to be pushed by the growing interfaces.

In this work, when the solidification temperature fell below the liquidus temperature, the primary α-Al began to nucleate and grow. In the meanwhile, a large amount of NPs uniformly dispersed in the melt via UT were pushed to the grain boundaries by α-Al dendrite fronts. At the last stage of solidification, the precipitation of Al_{2}Cu phases along the grain boundary resulted in the accumulation of NPs on their surface. On the other hand, many NPs are also observed inside α-Al grains. This may be attributed to the increased melt viscosity and the high local solidification rate. Studies [26] show that the melt viscosity increases with the volume fraction of NPs. With increasing the NP content, the enhanced viscous force on NPs exerted by high viscosity can act against the pushing by SF and facilitate the NP capture. Additionally, the local dendrite tip velocity may exceed the critical velocity near the mould wall because of high local cooling rates, which enables the NP capture by SF.

3.3 TEM analysis

The TEM analysis of the intergranular NPs in the nanocomposite with 2 vol.% NP additions is given in Fig. 6. Fig. 6(a) exhibits the assembly of NPs onto the grain boundary, which leads to the formation of NP layer. The selected area electron diffraction (SAED) pattern in the inset of Fig. 6(b) indicates that these NPs have a face-centered cubic (FCC) structure with the unit cell of NaCl-type, which is in accordance with TiCN identified by XRD analysis. Fig. 6(c) further shows that these NPs are either attached to each other or packed closely together. The SAED taken by tilting to the <001>_{NP} zone axis and the inverse Fourier transformation (IFT) of a nanoparticle are shown in the insets of Fig. 6(d). The crystallographic informa-
Fig. 6 – (a) TEM bright-field image showing the intergranular NPs; (b) and (c) HRTEM images of NPs under different magnifications in the same area; (d) HRTEM image of the NP/Al interface. The insets in (b) and (d) are the selected area electron diffraction (SAED) pattern of NP clusters and a nanoparticle, and the Inverse Fourier transformation (IFFT) of this nanoparticle, respectively.

...tion provided in SAED and IFFT is consistent with TiC₀.₅N₀.₅, whose interplanar spacing of (020) is 0.2143 nm. Fig. 6(d) is the HRTEM image showing the interface between the matrix and the nanoparticle at the grain boundary. From Fig. 6(d), it is clearly visible that no crystallographic orientation is found between the intergranular NP and the matrix, meaning that the NP has an incoherent interface with the α-Al.

Fig. 7 presents the intragranular NPs in the nanocomposite with 2 vol.% NP additions. As shown in Fig. 7(a) and (b), these NPs are uniformly distributed in the interior of α-Al grain. Likewise, the crystallographic information provided in SAED and IFFT (Fig. 7(c)) confirms the presence of TiC₀.₅N₀.₅ nanoparticle in the matrix. Distinct from the intergranular NPs, which exhibit random orientation with the growing solid, the intragranular ones are supposed to show a specific orientation relationship with the solid. As shown in Fig. 7(d), the NP-matrix interface is smooth and clean, at which no intermediate phase is observed. More significantly, the TiC₀.₅N₀.₅ nanoparticle is coherent with the Al matrix, and a specific orientation relationship between TiC₀.₅N₀.₅ and α-Al is obtained as \( \langle 200 \rangle _{\text{Al}} / \langle 011 \rangle _{\text{Al}} \)/\( \langle 020 \rangle _{\text{NP}} / \langle 011 \rangle _{\text{NP}} \). The atomically smooth and coherent interface can give high interfacial bonding between the TiC₀.₅N₀.₅ nanoparticle and the matrix, and thus improved load transfer efficiency. On the other hand, the well-defined OR can be used to evaluate the lattice misfit between α-Al and TiC₀.₅N₀.₅. By calculation, the lattice misfit is estimated to be around 5.85%. This small misfit implies the high potency of TiC₀.₅N₀.₅ as heterogeneous nucleation site for α-Al. It is noted that, based on classic nucleation theory, only one nucleation site is available for an individual grain. Consequently, only a limited fraction of intragranular NPs, which are typically located at the centers of grains, can be involved in the nucleation events.

3.4. Mechanisms of microstructure evolution

Based on the experimental results above, the refinement of α-Al dendrites can be explained from the perspective of heterogeneous nucleation and the growth restriction. First of all,
the cavitation induced pressure can change the equilibrium melting point, which is given by the Clapeyron equation [27]

$$\Delta P = \frac{L \Delta T}{T_m \Delta V}$$

where $\Delta P$ is the increased pressure, $L$ the latent heat, $T_m$ the melt point, $\Delta V$ the variation in volume and $\Delta T$ the undercooling. In this work, $\Delta P$ is 1000 atm [27], $L$ is $3.88 \times 10^5$ J/kg [28], $T_m$ is 918.15 K and $\Delta V$ is $3.08 \times 10^{-5}$ m$^3$/kg [19]. By calculation, the corresponding undercooling is estimated to be about 7.3 K. It means that a local undercooling of 7.3 K can be obtained under ultrasonication. According to the free growth model [29], for a given particle size, the critical undercooling for grain initiation on the particle is fixed. Above it, the nuclei can continue to grow and below which the free growth of the crystal is stifled. The undercooling for free growth $\Delta T_{fg}$ and the particle diameter $d$ is given by [29]

$$\Delta T_{fg} = \frac{4\sigma}{3K d}$$

where $\sigma$ is the solid-liquid interfacial energy and $\Delta S_n$ is the entropy of fusion per unit volume. It can be seen from Eq. 2 that the undercooling for grain nucleation is inversely proportional to the diameter of nucleating particles. As the undercooling is increased, progressively smaller particles become centres for free growth, which occurs as soon as the required undercooling is reached. For the NPs with an average size of 80 nm, the undercooling required for their activation is determined to be 7 K ($\sigma = 158$ mJ/m$^2$ [29]; $\Delta S_n = 1.112 \times 10^6$ J/K m$^3$ [30]). Therefore, it is reasonable that the undercooling in the liquid is sufficiently high to activate a fraction of NPs. These activated NPs can become potent nucleants for the heterogeneous nucleation of $\alpha$-Al resulting in the dendrite refinement. Moreover, the NPs pushed by SF can form a NP layer covering on the growing dendrite to inhibit the solute diffusion and restrict the grain growth. The reduced growth rate means the less latent heat released from the growing solid, and in turn the increased undercooling available before recalescence, which promotes the nucleation efficiency. Overall, the enhanced nucleation efficiency and potency lead to the refinement of $\alpha$-Al dendrites.

For the refinement of $Al_2Cu$ phase, there may be two main reasons. One is the refinement of $\alpha$-Al dendrites. As the NP addition level is increased, the $\alpha$-Al dendrite is refined and the number of grain boundary is increased. With the precipitation of $Al_2Cu$ phases along the grain boundary, $Al_2Cu$ phases becomes small and thin. The other one is the NP-induced restriction effect. As observed in Fig. 5, a large number of NPs accumulate onto the growing interface in the form of either nanoclusters or discrete nanoparticles, forming a NP layer covering the $Al_2Cu$ phases to restrict the phase growth. Apart from the growth restriction, the intergranular NPs may have influence on the precipitation of $Al_2Cu$. Specifically, these NPs can form a NP network along the grain boundary. As the $Al_2Cu$ phases precipitate along the grain boundary at the last stage of solidification, the NP network has a strong pinning effect to significantly restrict, if not completely prevent, the precipitation of $Al_2Cu$ phases, which leads to the incontinuous network distribution of $Al_2Cu$ phases as shown in Fig. 3(d).

### 3.5. Effect of NPs on the mechanical and thermal properties

The main purpose of microstructure design of nanocomposites is to improve their mechanical properties. The effects of
NP content on the microhardness and the tensile properties are evaluated for all nanocomposite samples. Fig. 8 illustrates the variations of the microhardness and tensile properties of TiC$_{0.5}$N$_{0.5}$/Al-Cu nanocomposites with various NP addition levels.

From Fig. 8(a), it is evident that the microhardness is continuously increasing with the NP content. The microhardness of the matrix alloy is 89.2 HV. When the addition level reaches 2.5 vol.%, the microhardness of the nanocomposite is increased up to about 121 HV by 35.6 % as compared to the matrix alloy. Based on the microstructural analysis, the intragranular NPs are uniformly distributed in the interiors of α-Al dendritic grains, whereas the intergranular ones are located at the grain boundaries. The intragranular NPs can act as rigid barriers to dislocation movement through the matrix and offer great resistance to local plastic deformation. Also, the NP network formed by intragranular NPs can be effective in inhibiting the local deformation of the matrix, thus resulting in high hardness. In addition, a refined grain size indicates an increased number of grain boundaries, which in turn leads to the enhanced grain boundary strengthening and resistance to crack propagation, thus facilitating improvement in hardness.

Fig. 8(b) shows the comparison of tensile properties among the matrix alloy and the nanocomposites with varying NP additions. Unlike the variation trend in microhardness, there is an evident trend of increasing strength and ductility of the nanocomposites with increasing the NP content until 2 vol.% NP addition, then a slight drop occurs in the tensile properties. The ultimate tensile strength (UTS), yield strength (YS) and elongation (EI%) of the matrix alloy are 216 MPa, 162 MPa and ~11 %, respectively. After 2.0 vol.% NP addition, the UTS of the nanocomposite is increased to 272 MPa, which is 26 % higher than that of the matrix alloy. Yield strength shows the same response to the addition level as the UTS. The addition of up to 2 vol.% NPs can lead to the greatest improvement in the YS of the nanocomposite, which is increased by 30 % as compared to the matrix alloy. Instead of degrading the EL as reported in the literature [31,32], there is a marked improvement in the ductility of 2.0 vol.% Al-TiC$_{0.5}$N$_{0.5}$ nanocomposite and the obtained EL can reach up to 16 %.

The enhancement in strength may be due to several strengthening mechanisms. Firstly, the grain refinement can increase the density of grain boundary that can immobilize the dislocation during deformation and thus the strength of nanocomposites [33]. Furthermore, the coefficient of thermal expansion mismatch between the matrix and NPs can produce a large number of dislocations. Their immobilization by the intragranular NPs can lead to the enhanced strength [34]. Another reason is attributed to the interfacial bonding strength between the NPs and the matrix. During plastic deformation, a strong interfacial bonding can promote the load transfer ability from the matrix to NPs leading to a more efficient strengthening of the nanocomposite [35].

In terms of AMCs, the improvement in ductility is more interesting than that in strength. In fact, particle reinforced metal matrix composites can usually exhibit high strength. However, as the addition level is raised, particles tend to agglomerate together. Because particle agglomerates can not only result in the formation of microvoids but also induce high stress concentration during plasticity deformation. The degradation of ductility is likely to occur, especially when it comes to high addition levels [36]. The enhanced ductility of 2.0 vol.% Al-TiC$_{0.5}$N$_{0.5}$ nanocomposite may be mainly attributed to the grain refinement and the improved dispersibility of NPs. Firstly, grain boundary can act as an effective barrier to the crack propagation through the matrix. The increased number of grain boundaries due to grain refinement can enhance the resistance to crack propagation and thus the ductility of the matrix. Additionally, UT is an effective technique to disperse the NPs in the melt. Under ultrasonication, large NP agglomerates can be fragmented into small ones and then dispersed to discrete NPs or submicron NP clusters as shown in Fig. 4(b). The improved dispersibility of NPs by UT can help to reduce the stress concentration and hinder the crack initiation, which in turn enhances the ductility.

Fig. 9 shows the coefficient of thermal expansion (CTE) of the nanocomposites as a function of the addition level. The CTE of the matrix alloy was measured to be 22.7 × 10$^{-6}$/K. The addition of TiC$_{0.5}$N$_{0.5}$ NPs can effectively reduce the CTE of the nanocomposites as compared with the matrix alloy. When the addition level reaches 2.0 vol.%, the CTE of the nanocomposite is reduced to 21.3 × 10$^{-6}$/K. The Turner model is used to determine the CTE of nanocomposites.

$$\alpha = \frac{\alpha_m K_m V_m + \alpha_p K_p V_p}{K_m V_m + K_p V_p}$$

(3)

where $\alpha_m$ and $\alpha_p$ are the CTE of the matrix and particles; $K_m$ and $K_p$ are the bulk modulus of the matrix and particles; $V_m$ and $V_p$ are the volume fraction of the matrix and particles. The
CTE and bulk modulus of TiCN are $7.4 \times 10^{-6}/K$ and 194 GPa, respectively. The calculated value of CTE for the nanocomposite with 2.0 vol.% NP addition is about $21.88 \times 10^{-6}/K$, which is in a very good agreement with the experimental value. The marked decline in CTE of the nanocomposite is mainly due to the incorporation of TiC$_{0.5}$N$_{0.5}$ reinforcements with high thermal stability. Moreover, the strong interfacial bonding between the matrix and the intragranular NPs can inhibit the matrix thermal expansion due to the effective thermal stress transfer. The experimental results suggest that the prepared nanocomposites can exhibit enhanced thermal and dimensional stability at elevated temperatures.

### 4. Conclusions

In summary, the TiC$_{0.5}$N$_{0.5}$/Al-Cu nanocomposites were successfully fabricated through ultrasonic assisted casting route. With increasing the addition levels, the primary $\alpha$-Al dendrites and Al$_2$Cu phase are refined markedly. The microstructural analysis reveals that the nanoparticles in the matrix mainly exhibit two distribution patterns, i.e. intergranular and intragranular distribution. The intergranular NPs have random orientation with the matrix while the intragranular ones are coherent with the matrix. Experimental and theoretical results demonstrate that the enhanced nucleation under ultrasonication and NP-induced growth restriction are the two major factors in refining the primary $\alpha$-Al dendrite and Al$_2$Cu phase. Because of the addition of TiCN nanoparticles with desirable mechanical and thermal properties and the refined and homogeneous microstructure, the prepared nanocomposites can exhibit superior mechanical and thermal properties over the matrix alloy at various addition levels. Moreover, the 2 vol.% TiC$_{0.5}$N$_{0.5}$/Al-Cu nanocomposite has optimal mechanical and thermal properties with its hardness, ultimate tensile strength, yield strength, elongation increased by 30 %, 26 %, 30 % and 46 % and CTE decreased by 106.2 %, respectively as compared to the matrix alloy.

### Conflicts of interest

The authors declare no conflicts of interest.

### Data availability

Data will be made available on request.

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### Appendix A. Supplementary data

Supplementary material related to this article can be found, in the online version, at doi: https://doi.org/10.1016/j.jmrt.2019.12.037.

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