Materials Research Express

PAPER

Refining and reinforcing effects of TiC-Al₂O₃/Al ribbons inoculant on Al–Si–Mg–Ti alloy

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Keywords: inoculant, ceramic nanoparticles, mechanical properties, microstructure

Abstract

In this study, TiC-Al₂O₃/Al ribbons inoculant were prepared by mechanical alloying, spark plasma sintering (SPS) and melt spinning. Adding 5vol.% TiC-Al₂O₃/Al ribbons to an Al–Si–Mg–Ti (ASMT) alloy melt significantly refined the α-Al grains from 28.5 μm to 14.9 μm. In addition, the morphology of the eutectic silicon were modified and optimized, changing from needle-like silicon to spherical silicon. Adding the TiC-Al₂O₃/Al ribbons significantly improved the mechanical properties of TiC-Al₂O₃/Al–Si–Mg–Ti (TAASMT) alloy, including its elastic modulus (EM), yield strength (YS), ultimate tensile strength (UTS) hardness, and elongation which were increased by 5.4%, 18.4%, 32.7%, 8.3%, and 62.3%, respectively. These increases in the mechanical properties could be attributed to the intracrystalline reinforcement of the Al₂O₃ and TiC nanoparticles, well-bonded particle/matrix interface, modification of the eutectic silicon phases, and synergistic effects of the Orowan strengthening, Coefficient of thermal expansion (CTE) mismatch strengthening, load transfer, grain refining strengthening, and geometrically necessary dislocations (GND) strengthening.

1. Introduction

Al–Si alloys are widely used in automobiles and other fields because of their low density, thermal stability, good corrosion, and wear resistance [1–4]. Al–Si alloys mainly consist of α-Al, Mg₃Si strengthening phases and eutectic silicon phases [5]. However, the tensile strength and ductility of an Al–Si alloy in the as-cast condition are low because of its coarse α-Al dendrites and needle-like eutectic silicon. This causes the stress concentration, which produces micro-cracks and seriously reduces the mechanical properties of Al–Si alloys and limiting their use in industry [6–9]. Therefore, in order to increase an alloy’s performance, attention has been given to refining the α-Al grains and modifying the eutectic silicon phases.

In early research, red phosphorus was added to an Al–Si alloy to modify the silicon phases. Adding red phosphorus to an Al–Si alloy generates AlP phase (Al[I] + P[I] → AlP[s]). AlP and Si both have cubic crystal structure, and AlP can act as a nucleation site for Si. The Si phase attaches to AlP and grows. Thus, coarse flake-like Si or needle-like Si is transformed into fibrous, but this can easily cause environmental pollution. Although sodium has a good modification effect, it has the disadvantage of easy segregation and a low absorption rate [10, 11]. Sr has a good modification effect on Al–Si alloys, does not pollute the environment, and solves the problem of the easy segregation of sodium. Sr is usually introduced by adding an Al-Sr master alloy to the aluminum melt. The addition of the Al-Sr master alloy does not generate any gas; therefore, it is harmless to the environment. The density of Sr is 2.54 g cm⁻³, and the density of Al is 2.7 g cm⁻³. Because they have very close densities, Sr can be evenly distributed in the Al melt without segregation, but Sr will increase the pinhole tendency of the alloy [12]. In contrast, sodium has a very low density. If sodium is added to aluminum, it floats on the surface of the aluminum liquid, resulting in segregation. Generally, ceramic particles can be added to Al–Si alloys to act as nucleants for grain refinement. Previous studies have proved that the addition of ceramic nanoparticles can control the growth of eutectic silicon during the solidification of Al–Si alloys [13–15].
eutectic silicon changes from a sheet to a block, and the eutectic silicon is re
re
alloying, spark plasma sintering, and melt spinning. Al2O3 and TiC nanoparticles were generated
TiC-Al2O3
agglomeration. This article discusses in detail the synthesis and preparation of TAASMT alloys and
Mn alloy using the stirring casting method
was used to ensure that the TiC and Al2O3 nanoparticles were uniformly dispersed in the alloy and avoid
environmental pollution and does not increase the pinhole tendency of the alloy
literature research on the use of the TiC-Al2O3
which was shown to have a good reinforcement effect on aluminum alloys
reinforced Al
its tensile strength and hardness were signi
hybrid ceramic particles
composites can be prepared by microwave sintering
high strength and good toughness and are widely used in the aerospace and automotive
mechanism of the TAASMT alloy were studied and analyzed. In addition, the microstructure, mechanical properties, and strengthening
properties because of their high surface energy and poor wettability with aluminum
transfer. However, ceramic nanoparticles easily agglomerate in the matrix and adversely affect the material
of the reinforcement phase decreases. When the size of the reinforcement particles reaches the nanoscale, they
According to the Orowan mechanism, the reinforcement effect on material becomes more signi
The size and distribution of the ceramic particles in HAMCs have significant impacts on their performances. According to the Orowan mechanism, the reinforcement effect on material becomes more significant as the size of the reinforcement phase decreases. When the size of the reinforcement particles reaches the nanoscale, they can improve the mechanical properties of HAMCs by reducing the stress concentration in the matrix and load transfer. However, ceramic nanoparticles easily agglomerate in the matrix and adversely affect the material properties because of their high surface energy and poor wettability with aluminum. Therefore, the preparation of aluminum matrix composites with uniformly distributed nanoparticles is a significant challenge. The in situ synthesis method has been shown to effectively solve the problem of the low wettability of ceramics nanoparticles with aluminum, and ceramic nanoparticles generated in situ may establish a well-bonded interface with α-Al. The cavitation effect of ultrasonic vibration treatment can enhance the wettability between ceramic nanoparticles and Al, and distribute the ceramic nanoparticles homogeneously. The wettability between the ceramic nanoparticles and aluminum matrix can be improved further improved by melt spinning. In addition, extensive research has shown that aluminum alloy can be strengthened by adding TiC, SiC, TiN, Al2O3, and other nanoparticles. Cai et al proposed the concept of composites inoculant, which was shown to have a good reinforcement effect on aluminum alloys. TiC–Al2O3/Al composites have high strength and good toughness and are widely used in the aerospace and automotive fields. TiC–Al2O3/Al composites can be prepared by microwave sintering and spark plasma sintering. Previous researchers have done extensive research on the use of the TiC–Al2O3/Al composites as an inoculant to refine Al–Si alloys. In this study, a TiC–Al2O3/Al composite inoculants was prepared by a combination of mechanical alloying, spark plasma sintering, and melt spinning. Al2O3 and TiC nanoparticles were adopted in situ, and TiC–Al2O3/Al ribbons were added to the ASMT alloy via ultrasonic vibration. The ultrasonic vibration method was used to ensure that the TiC and Al2O3 nanoparticles were uniformly dispersed in the alloy and avoid agglomeration. This article discusses in detail the synthesis and preparation of TAASMT alloys and TiC–Al2O3/Al ribbons inoculant. The microstructure and phase composition of the TiC–Al2O3/Al ribbons inoculant were also studied. In addition, the microstructure, mechanical properties, and strengthening mechanisms of the TAASMT alloy were studied and analyzed.

Table 1. Chemical composition of ASMT alloy.

| Element | Al | Si | Mg | Ti | Zn | Mn | Ni |
|---------|----|----|----|----|----|----|----|
| Wt%     | Bal.| 7.14 | 0.34 | 0.15 | 0.03 | 0.02 | 0.01 |

Figure 1. Preparation route of the TiC–Al2O3/Al composite ribbons (a) mechanical alloying, (b) SPS, (c) melt spinning.
2. Experimental procedures

The purpose of this study was to prepare TiC-Al₂O₃/Al ribbons inoculant with TiC and Al₂O₃ nanoparticles to refine and reinforce Al–Si–Mg–Ti alloy, thereby improving its mechanical properties. Table 1 lists the chemical compositions of the ASMT alloys. In this work, the TiC-Al₂O₃/Al ribbons were prepared with Al (3–15μm, with 99.5% purity) powder, TiO₂ (5–25μm, with 99.8% purity) powder, and graphene nanoplates (GNPs). GNPs have a lamellar structure with a high surface energy. In order to reduce the free energy, these GNPs have a tendency to attach to adjacent substances. They can also adhere to the surface layer of other material powders, where they prevent the agglomeration and growth of these powders; therefore, GNPs can not only be used as a carbon source, but can also be used as a dispersant to make other powders more dispersive. The particle distribution in the TiC-Al₂O₃/Al ribbons prepared from GNPs was uniform. The preparation route for the TiC-Al₂O₃/Al ribbons is illustrated in figure 1. First, Al powder, TiO₂ powder, and GNPs were placed in a 300mL steel cup, and 5–12mm steel balls were used as the grinding media. Ethanol was added to the steel cup as a dispersant. The steel cup was placed in the GN-2 high-energy ball mill and milling was conducted for 15h as shown in (figure 1(a), table 2 is the ball milling parameters). Then, using the SPS method (figure 1(b)), the uniformly mixed particles were poured into a graphite mold and sintered at 1000 °C and 50 MPa for 10 min to prepare the TiC-Al₂O₃/Al inoculant. Compared with traditional sintering methods and microwave sintering, SPS has the advantage of producing denser materials in a shorter time (10 min) under pressure. Because SPS has a very high heating rate, it can reduce the growth rate of grains, thereby limiting their growth [41]. The particle size of the inoculant prepared by this method was smaller than that of an inoculant prepared by the traditional sintering method; therefore, it had a better reinforcing effect on the aluminum alloy. Finally, the TiC-Al₂O₃/Al ribbons were prepared by melt spinning (figure 1(c)), and the TiC-Al₂O₃/Al inoculant was remelted by high-frequency induction heating and blown onto a copper roller at a speed of 2000rpm to produce TiC-Al₂O₃/Al ribbons.

To prepare the TAASMT alloy, the ASMT alloy was first heated to 760 °C and melted. Then, the TiC-Al₂O₃/Al ribbons (5 vol%) were added to the ASMT melt, and the temperature was maintained for 2 min. An ultrasonic probe that had been preheated to 760 °C was immersed in the ASMT alloy melt with the added TiC-Al₂O₃/Al ribbons, the ultrasonic generator was turned on, and ultrasonic vibration treatment was conducted for 30 s. Next, the ultrasonicated melt was quickly poured into a steel mold to prepare the TAASMT alloy. The ultrasonic vibration treatment uniformly dispersed the TiC and Al₂O₃ nanoparticles in the alloy. Wire-electrode cutting was used to cut off TAASMT alloy samples with a height of 10 mm. The samples were...
polished carefully with 400, 600, 800, 1000, 1500, and 2000 mesh sandpaper, and then performed by mechanical polishing. Finally, the TAASMT alloy was etched with a 0.5% HF solution to observe its microstructure.

The phase compositions of the TiC-Al2O3/Al ribbons were detected and analyzed using x-ray diffractometry (XRD, Bruker D8 Discover). The microstructures of the TiC-Al2O3/Al ribbons and TAASMT alloy were observed using optical microscopy and scanning electron microscopy (SEM, S4800). The chemical compositions of the TiC-Al2O3/Al ribbons were tested using energy dispersive spectrometer (EDS). The purpose of the EDS was to detect the elemental composition of the particles in the TiC-Al2O3/Al ribbons. Combined with the XRD results, this made it possible to preliminarily determine the particles in the TiC-Al2O3/Al ribbons. The crystallographic orientation relationship between the TiC particles, Al2O3 particles, and Al matrix was detected by high-resolution transmission electron microscopy (HRTEM, JEM-2100). The tensile properties of the TAASMT alloys were tested using a universal testing machine (SHT-5305).

**Figure 3.** (a) SEM images of TiC-Al2O3/Al composites ribbons, (b) EDS spectra of point A, (c) EDS spectra of point B.

**Figure 4.** TEM and HRTEM of TAASMT alloy (a) the TEM of Al2O3 particle, (b) the HRTEM of Area A in (a), (c) FFT pattern of zone B in (b), (d) Inverse FFT pattern of zone B.
3. Results and discussions

3.1. Phase composition and microstructure of TiC-Al2O3/Al ribbons

X-ray diffractometry (XRD) was used to determine the phase composition of the TiC-Al2O3/Al ribbons. The scanning speed of the XRD device was 6°/min, and the power was 4 kW. The XRD pattern of the TiC-Al2O3/Al ribbons is shown in figure 2. The TiC-Al2O3/Al ribbons mainly contained the α-Al, Al2O3, and TiC phases.

The SEM image and corresponding energy dispersive spectrometer (EDS) spectra of TiC-Al2O3/Al ribbons are shown in figure 3. It can be seen that particles in figure 3(a) have a size range of approximately 50–200 nm, and the nanoparticles are dispersedly distributed. Figure 3(b) shows the EDS spectrum of point A. The main elements at point A are Ti and C. Based on the XRD results, it could be inferred that this phase is TiC. Figure 3(c) shows the EDS spectrum of point B in figure 3(c), and the EDS results show that it is rich in Al and O. Based on the XRD results, it could be concluded that the phase is Al2O3.

3.2. Characterization of interface

Figure 4(a) shows the morphology of the Al2O3 particle in the TAASMT alloy, as determined using transmission electron microscopy (TEM). The Al2O3 particle had a size range of 100–150 nm, as shown in figure 4(a). Figure 4(b) shows an HRTEM image of area A. The interface was clean, and no impurity phases were observed. Figure 4(c) shows the Fast Fourier Transform (FFT) mode of zone B in figure 4(b). It can be seen that apart from the two sets of diffraction spots of Al and Al2O3, no other diffraction spots were found, which indicated that no other reactants were formed at the Al2O3/Al interface. Furthermore, there was a crystallographic orientation relationship between the Al2O3 and Al, (111)Al//(124)Al2O3, and the interplanar spacing measured in figure 4(a) were 0.2338 nm and 0.2315 nm respectively. The interplanar spacing of (111)Al and (124)Al2O3 were very close, which indicated that the Al2O3 and Al have good lattice matching relationships. Based on an edge-to-edge matching model [42], Al2O3 has a heterogeneous nucleation ability for α-Al. The interplanar spacing mismatch, f, is defined as follows [42]:

\[ f = \left| \frac{R_1 - R_2}{R_1} \right| \]

where R1 and R2 are the interplanar spacings of (111)Al and (124)Al2O3, respectively. The interplanar spacing mismatch between (111)Al and (124)Al2O3 is 0.98%; therefore Al2O3 and Al can form a nearly perfect coherent interface. Dislocations could be observed in the Inverse Fast Fourier Transform (IFFT) (figure 4(d)) at the Al/Al2O3 interface (zone B in figure 4(b)).
The TiC morphology of the TAASMT alloy is shown in figure 5(a); the TiC had a size range of 50–150 nm, as shown in figure 5(a). An HRTEM image of the Al/TiC interface of zone A is shown in figure 5(b). The lattice fringes of the two phases at the interface were clearly different, and the interface was clean and without any impurity phases. Figure 5(c) shows the result of the FFT in zone B in figure 5(b). It can be seen that there are only two sets of diffraction spots of Al and TiC. The crystallographic orientation difference measured by the digital micrograph software was 14.38°, as shown in figure 5(b), small-angle crystallographic orientation difference lead to good interface bonding between the α-Al and TiC. The interplanar spacing of (200)Al and (020)TiC were determined to be 0.2024 nm and 0.2164 nm, respectively, by calibrating the diffraction spots with the digital micrograph software and measuring the width of 20 crystal planes at high resolution and taking the average value. The interplanar spacing mismatch between (200)Al and (020)TiC is 6.46%. TiC can act as nucleants for grain refinement in Al alloys. Dislocation loops were observed in the IFFT (figure 5(d)) at the Al/TiC interface (zone B in figure 5(b)). Figures 5(e) and (f) show the EDS mapping of the TiC particles, where, the aggregation of Ti and C is obvious, which further proves that the particle in figure 5(a) is TiC.

3.3. Thermodynamic analysis of in situ TiC and Al2O3

Referring to the Al–TiO2–C system with excessive Al [36, 43], and based on the phase composition shown by the XRD test results, the reaction formulas of this experiment are as follows;

\[ 3\text{TiO}_2 + 13\text{Al} = 3\text{Al}_7\text{Ti} + 2\text{Al}_2\text{O}_3 \]  \tag{2} \]

\[ 4\text{Al} + 3\text{C} = \text{Al}_4\text{C}_3 \]  \tag{3} \]

\[ 3\text{Al}_7\text{Ti} + \text{Al}_4\text{C}_3 = 13\text{Al} + 3\text{TiC} \]  \tag{4} \]

The curve of the standard free energy values of formula (2)–(4) at different temperatures is shown in figure 6. It can be seen that at a reaction temperature of 1000 °C, the Gibbs free energy (ΔG) values of these three reaction formulas are all negative, which means that these three reactions can proceed spontaneously. To better understand the formation process of Al2O3 and TiC, a schematic diagram of their formation is shown in figure 7.
3.4. The microstructure of TAASMT alloy

Figure 8 shows metallographic images of the ASMT and TAASMT alloys. The refining effect of the TiC-Al_{2}O_{3}/Al ribbons on the ASMT alloy is evident. When TiC-Al_{2}O_{3}/Al ribbons were added, the average grain size of the TAASMT alloy decreased from 28.5 μm to approximately 14.9 μm (as shown in figure 8(b)). This grain refinement could be attributed to the \textit{in situ} formation of TiC and Al_{2}O_{3}, which can act as nucleants for the grain refinement of Al alloys [44–46]. Simultaneously, the ultrasonic vibration treatment of the melt, generated many cavitation bubbles in it. Under the action of ultrasonic waves, these cavitation bubbles rapidly expanded until they collapsed. The shock wave broke the dendrites that grew in the melt and inhibited grain growth [47]. Therefore, the ultrasonic vibration treatment was also a reason for the refinement. In addition, it can be observed from figure 8(a) that a large amount of eutectic silicon was segregated, which was not conducive to the

![Figure 8](image_url)

\textbf{Figure 8.} The metallographic images of (a) ASMT alloy, (b) TAASMT alloy.

![Figure 9](image_url)

\textbf{Figure 9.} SEM images of (a) ASMT alloy, (b) TAASMT alloy, (c) ASMT alloy in higher magnification, (d) TAASMT alloy in higher magnification.
Figure 10. Tensile stress-strain curves of ASMT alloy and TAASMT alloy.

Table 3. Mechanical properties of ASMT alloy and TAASMT alloy.

| Sample | EM (GPa) | YS (MPa) | UTS (MPa) | Hardness (HRB) | EI (%) |
|--------|----------|----------|-----------|----------------|--------|
| ASMT   | 82.2     | 82.4     | 153.1     | 95.4           | 6.1    |
| TAASMT | 86.6     | 97.6     | 203.2     | 103.3          | 9.9    |

Figure 11. SEM images of the fracture surface (a) ASMT alloy, (b) Enlarged view of the zone A, (c) TAASMT alloy, (d) Enlarged view of the zone B.
mechanical properties of the material. However, the eutectic silicon in the TAASMT alloy was distributed homogeneously.

The SEM images of the ASMT alloy are shown in figures 9 (a) and (c), (b) and (d) show SEM images of the TAASMT alloy. It can be seen that the grain size of the TAASMT alloy was significantly refined compared to that of the ASMT matrix, which was consistent with the experimental results shown in the metallographic images in figure 8. In addition, needle-like eutectic silicon phases were observed in the ASMT alloy, as shown in figure 9(c), which split the matrix and adversely affected the mechanical properties of the alloy. When TiC-Al2O3/Al ribbons were added, most of the needle-like silicon became spherical silicon, which showed that the TiC-Al2O3/Al ribbons could not only refine α-Al but also modify the eutectic to optimize its morphology. The modification of the eutectic Si could be attributed to the introduction of TiC particles via the addition of the TiC-Al2O3/Al ribbons. TiC and Si both have cubic crystal structures, and their lattice parameters are very close. TiC can act as an effective nucleation site for the silicon phase, making the silicon phase dependent on TiC for nucleation [48]. Simultaneously, the cavitation effect of the ultrasonic vibration treatment could also modify the eutectic silicon. When the cavitation bubbles imploded, a strong shock wave was generated, resulting in the fragmentation of the needle-like eutectic silicon. The heterogeneous nucleation effect of the TiC particles on the Si and the cavitation effect of the ultrasonic vibration treatment resulted in the modification and optimization of the eutectic Si [49].

3.5. Mechanical properties and strengthening mechanism

The EM is a measure of the ability of a material to resist elastic deformation. It refers to the stress required for a material to produce a unit of elastic deformation under the action of an external force. The YS is the limit at which a material yields, and it also refers to the stress that resists small amounts of plastic deformation. The UTS refers to the maximum tensile stress that a specimen can withstand until it breaks. Elongation refers to the length of a specimen after fracture minus the original length, divided by the original length, which can be used to measure the plasticity of the material. Figure 10 shows the stress-strain curves of the ASMT and TAASMT alloy. The mechanical properties of the ASMT alloy and TAASMT alloy are listed in table 3. Compared to the ASMT alloy, the EM, YS, UTS, hardness and elongation of the TAASMT alloy increased by 5.4%, 18.4%, 32.7%, 8.3%, and 62.3% respectively.

Figure 11 show the tensile fracture surface morphology of the ASMT and TAASMT alloy. Figure 11(a) shows the fracture morphology of the ASMT alloy, and figure 11(b) shows an enlarged view of zone A. It can be seen that there are a large number of tearing edges and cleavage planes in the ASMT alloy. This was a typical quasi-cleavage brittle fracture. Figure 11(c) shows the tensile fracture morphology of the TAASMT alloy, and figure 11(d) shows an enlarged view of zone B. It can be seen from the TAASMT alloy that the number of dimples increased significantly, and the number of tearing edges and cleavage planes decreased significantly. The existence of a large number of dimples is characteristic of a ductile fracture, which indicates that the TAASMT alloy had higher elongation. In addition, ultrasonic vibration has the effect of degassing and purifying the melt, reducing the casting defects of the alloy and avoiding the segregation of specific gravity, which leads to an increase in elongation.

The mechanical conditions of micro-cracks can be described by the following formula [50].

\[
\sigma_i = \frac{\sigma_m}{K} \sqrt{\frac{\Delta V}{V_i}} + \frac{1}{K} \sqrt{\frac{\gamma E_i}{d_s}}
\]  

(5)

where \( \sigma_i \) is the fracture tensile stress when the material appears micro-cracks, \( \sigma_m \) is the YS of the matrix, K is the stress concentration factor, \( \Delta V \) is the volume of plastic deformation caused by the matrix around the silicon, \( V_i \) is the average volume of Si, and E is the EM of Si, \( \gamma \) is the fracture surface energy of matrix, and \( d_s \) is the average size of eutectic Si. According to formula (5), \( \sigma_i \) is related to the value of stress concentration factor K. The value of stress concentration factor K is related to the aspect ratio of the Si. When the aspect ratio of the Si is larger, the value of K is larger, and the value of K is smaller when the Si is closer to a spherical shape. After adding the TiC-Al2O3/Al ribbons inoculant, most of the needle-like silicon was transformed into spherical silicon. There was still a small amount of needle-like silicon, and the aspect ratio of this needle-like silicon was not constant [50]. However, after adding the TiC-Al2O3/Al ribbons inoculant, the aspect ratio of the Si in the TAASMT alloy was smaller than that of the ASMT alloy; therefore, its K value was also smaller, and the stress required for the TAASMT alloy to generate micro-cracks was higher than that of the ASMT alloy.

The EM of the ASMT matrix was 82.2 GPa, and the EM of TAASMT alloy modified by the TiC-Al2O3/Al ribbons was 86.6 GPa, which was a 5.4% increase in the EM. The improvement in the EM could be attributed to the existence of high EM Al2O3 (375 GPa) and TiC (420 GPa) in the TAASMT alloy. The shear-lag model proposes that when a load acts on the alloy, it is will be transferred from the matrix to the reinforcing particles by the interface shear stress. According to the shear-lag theory, the EM of the TAASMT alloy can be described by the following formula [51, 52].
\[ E_c = E_n V_n \left[ 1 - \frac{\tanh (n.s)}{n} \right] s + (1 - V_n)E_m \]  

where \( E_n \) denotes the reinforced particles’ EM, \( V_n \) denotes the reinforced particles’ volume fraction, \( V_m \) denotes the volume fraction of the matrix, \( s \) denotes particles’ aspect ratio, and \( E_m \) is the elastic modulus of the matrix \( E_m = 82.2 \text{ GPa} \). A calculation showed that \( E_c = 85.7 \text{ GPa} \). This value was very close to the actual measured value of 86.6 GPa.

The increase in the YS of the TAASMT alloy could be attributed to (I) CTE mismatch strengthening (\( \Delta \sigma_{\text{CTE}} \)) caused by the mismatch in the coefficient of thermal expansion (CTE) between the matrix and the TiC and \( \text{Al}_2\text{O}_3 \) particles. (II) The geometrically necessary dislocations (GND) generated around the TiC and \( \text{Al}_2\text{O}_3 \) particles as a result of the plastic deformation mismatch between the matrix and the reinforcing particles (\( \Delta \sigma_{\text{GND}} \)). (III) Orowan strengthening (\( \Delta \sigma_{\text{Orowan}} \)) was caused by the restriction effect of the TiC and \( \text{Al}_2\text{O}_3 \) nanoparticles on dislocation movement. (IV) Grain refinement strengthening (\( \Delta \sigma_{\text{GRI}} \)) was produced by the addition of the TiC:\text{Al}_2\text{O}_3/Al ribbons, resulting in grain refinement. (V) A load transfer occurred from the Al matrix to the TiC and \( \text{Al}_2\text{O}_3 \) particles (\( \Delta \sigma_{\text{L}} \)).

The CTE values of the Al, TiC, and \( \text{Al}_2\text{O}_3 \) are 23.6 \( \times 10^{-6} \text{K}^{-1} \), 7.4 \( \times 10^{-6} \text{K}^{-1} \), and 8.8 \( \times 10^{-6} \text{K}^{-1} \), respectively. The CTE values of the Al, TiC, and \( \text{Al}_2\text{O}_3 \) were very different, which increased the dislocation density around the TiC and \( \text{Al}_2\text{O}_3 \) particles. In addition, because of the mismatch between the Al matrix and the TiC and \( \text{Al}_2\text{O}_3 \) particles, GND were generated around the TiC and \( \text{Al}_2\text{O}_3 \) particles to adapt to the plastic deformation mismatch between the Al matrix and the TiC and \( \text{Al}_2\text{O}_3 \) particles. As shown in figures 4(d) and 5(d), high-density dislocations and dislocation loops appeared in the Al/\( \text{Al}_2\text{O}_3 \) and Al/TiC interface and Al/TiC interface. The synergy between the dislocations and GND increased the strength of the alloy [53]. The radius of the CTE mismatch area can be described by the following formula [23, 54]:

\[ R_n = \frac{d_n}{2} \sqrt{\frac{B^2 + (1 - 2V_n)^2 + 16NB \frac{n^2}{G} + B(1 - 2V_n)}{4 \frac{n^2}{G}}} \]  

where \( d_n \) is the average size of the reinforcing particles, \( B \) and \( N \) are the CTE mismatch between the matrix and reinforcing particles, \( n \) is the shear strength of the matrix, and \( G \) is the shear modulus of the matrix (2.6 \( \times 10^4 \text{MPa} \)). The area of the CTE mismatch zone is related to the size of the reinforcing particles. The \( \text{Al}_2\text{O}_3 \) particles had sizes of approximately 100–150 nm, and the TiC particles had sizes of approximately 50–150 nm. The two size ranges for the reinforcing particles produced CTE mismatch areas of two different sizes, and their mismatch strengthening effects were superimposed on each other, which further hindered the movement of dislocations, thereby increasing the strength of the TAASMT alloy.

CTE mismatch strengthening can be described by the following formula [52]:

\[ \Delta \sigma_{\text{CTE}} = 4.33G.b \sqrt{\frac{\Delta T \Delta CTE V_n}{bd_n}} \]  

where \( b \) is the Burgers vector of the matrix (2.86 \( \times 10^{-10} \text{m} \)), \( \Delta T \) is the difference between the test temperature and room temperature, \( \Delta \text{CTE} \) is the difference between the CTE values of the Al and reinforcing particles, and \( \Delta \sigma_{\text{CTE}} \) was calculated to be 5.3 MPa.

The GND strength can be described by the following formula [55]:

\[ \Delta \sigma_{\text{GND}} = 0.4G \sqrt{\frac{\varepsilon.b \sqrt{V_n}}{d_n}} \]  

where \( \varepsilon \) denotes the plastic strain in the matrix. A calculation found that \( \Delta \sigma_{\text{GND}} \) was 9.3 MPa.

According to the Orowan mechanism, nanoscale \( \text{Al}_2\text{O}_3 \) and TiC can hinder dislocation movement. These dislocations could not pass directly through the nanoparticles but bent between them, generating dislocation loops and back stress, and thereby improving the strength of the TAASMT alloy. The enhancement effect of the Orowan mechanism can be described using the following formula [23, 55]:
A calculation shows that the YS contributed by Orowan strengthening was 2.1 MPa.

The Hall-Page formula can be used to calculate the contribution of grain refinement to the strength
\[ \Delta \sigma = K (d_1 \frac{1}{2} - d_2 \frac{1}{2}) \]

where \( K \) is the Hall-Page constant for Al (0.04 MPa.m\(^{1/2}\)) \[^{56}\] , \( d_1 \) is the grain size of the TAASMT alloy, and \( d_2 \) is the grain size of the ASMT alloy. It can be observed from the formula that finer grains lead to a greater contribution to the increase in the strength of the alloy. This is because with finer grains, there is a greater number of grains per unit volume, more grain boundaries, and a larger the for the grain boundaries. The grain boundaries can hinder the movement of dislocations to adjacent grains, and the same amount of deformation can be dispersed into more grains during deformation, reducing the stress concentration. Thus the alloy can withstand greater plastic deformation before fracture, thereby improving its strength. It was calculated that \( \Delta \sigma_{H} = 0.3 \) MPa.

Because of the well-bonded between the \( \text{Al}_2\text{O}_3 \) and TiC particles with the Al matrix, a load transfer could be carried out effectively from the Al matrix to the \( \text{Al}_2\text{O}_3 \) and TiC particles, and the contribution of this load transfer can be determined by the following formula \[^{52}\] 
\[ \Delta \sigma_{L} = \frac{s}{4} \sigma_{m} V_{n} \]

where \( s \) denotes the aspect ratio of the reinforcing particles, and for \( s = 1 \), a calculation shows that \( \Delta \sigma_{H} = 0.3 \) MPa.

The YS of the TAASMT alloy can be calculated as follows \[^{57}\] :
\[ \sigma_{c} = \sigma_{m} + \Delta \sigma_{\text{CTE}} + \Delta \sigma_{\text{GND}} + \Delta \sigma_{\text{Orowan}} + \Delta \sigma_{H} + \Delta \sigma_{L} \]

It was calculated that \( \sigma_{c} = 99.7 \) MPa. This value was very close to the actual measured value of 97.6 MPa. The contribution ratio of the various strengthening mechanisms to the yield strength of the TAASMT alloy is shown in figure 12.

In conclusion, the yield strength of the TAASMT alloy increased by 18.4% after adding the TiC-\( \text{Al}_2\text{O}_3 \)/Al ribbon inoculant. The contributions of the CTE mismatch strengthening, GND strengthening, Orowan strengthening, grain refinement strengthening, and load transfer to the YS of the TAASMT alloy were 30.6%, 53.7%, 12.1%, 1.8%, and 1.8%, respectively. Among these, the GND strengthening contributed the most to the improvement in the YS of the TAASMT alloy.

4. Conclusions

1. In this study, the TiC-\( \text{Al}_2\text{O}_3 \)/Al ribbons were successfully fabricated by mechanical alloying, spark plasma sintering, and melt spinning, and could be used as an inoculant for ASMT alloy.
2. The results showed that the TiC-Al2O3/Al ribbons had an excellent inoculation effect on the ASMT alloy. After adding TiC-Al2O3/Al ribbons, the grain size of the α-Al decreased from 28.5 μm to 14.9 μm. In addition, the morphology of the eutectic silicon was optimized, and changed from a needle-like to a spherical shape.

3. The mechanical properties of the TAASMT alloy were significantly improved by adding TiC-Al2O3/Al ribbons. Compared to the ASMT alloy, the EM, YS, UTS, hardness and elongation of the TAASMT alloy increased by 5.4%, 18.4%, 32.7%, 8.3% and 62.3% respectively.

4. There was an obvious interface and orientation relationship between the in situ Al2O3 and TiC particles with the α-Al. Al2O3 and TiC particles could act as nucleants for the grain refinement of ASMT alloys.

5. The improvements of the mechanical properties of the TAASMT alloy were due to the synergistic effects of CTE mismatch strengthening, GND strengthening, Orowan strengthening, grain refinement strengthening, and load transfer.

Acknowledgments

This work was financially supported by National Natural Science Foundation of China with No. 51871087 and Natural Science Foundation of Hebei Province in China with No. E2014202008 and No. E2016202406.

Data availability statement

The data generated and/or analysed during the current study are not publicly available for legal/ethical reasons but are available from the corresponding author on reasonable request.

CRediT authorship contribution statement

Lu Liu: Conceptualization, Writing—original draft. Chunxiang Cui: Methodology, Conceptualization, Project administration, and Supervision. Hongtao Geng: Analysis and characterization of mechanical properties. Yingguang Liang: Microstructure, Characterization. Sen Cui: Analysis of TEM image. Shichang Lei: Characterization. Shuo Zhang: Characterization.

Conflict of interest statement

We declare no competing interests.

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