

Impact of submicron TiB₂ particles on microstructure, casting performance, and mechanical properties of an Al-Cu alloy

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Abstract: Although the strengthening and grain refinement effects of TiB₂ particles on aluminum alloys have been extensively studied, their influence on casting behavior remains relatively underexplored. In this study, the influence of different addition amounts of submicron TiB₂ particles on the microstructure, casting performance, and mechanical properties of an Al-Cu (ZL205A) alloy was systematically investigated. The introduction of TiB₂ particles leads to significant grain refinement, transforming the microstructure from coarse grains to fine equiaxed grains by providing additional nucleation sites and inhibiting grain growth. SEM and TEM analyses reveal that the added submicron TiB₂ particles exhibit minimal effect on the distribution of intermetallic phases or precipitates. Casting performance, as evaluated by spiral fluidity and hot tearing tests, shows notable improvements with TiB₂ additions. At a TiB₂ content of 3wt.%, the fluidity length increases by 20%, and the hot tearing susceptibility coefficient decreases by 29%. These enhancements are mainly due to the refined grain structure and the formation of interdendritic bridging in TiB₂-reinforced alloys. However, the overall enhancement in casting properties shows little variation across the TiB₂ additions from 0.2wt.% to 3wt.%. Mechanical testing shows that the highest hardness and strength are achieved with a 1wt.% addition of TiB₂ particles, primarily attributed to refined grain size and reinforcement of the aluminum matrix. Based on these findings, a TiB₂ particle content of 1wt.% is recommended for optimizing both the casting performance and mechanical properties of the ZL205A alloy.

Keywords: TiB₂ particles; submicron particles; casting properties; strength; Al-Cu alloy

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1 Introduction

To meet growing economic and environmental demands, lightweight design strategies have become essential, driving the increased use of aluminum alloys in critical engine components, such as cylinder heads and engine blocks, that account for a substantial fraction of total engine mass^[1]. However, the pursuit of higher combustion temperatures and pressures to improve engine efficiency places severe thermodynamic and mechanical property requirements on conventional aluminum alloys. Among

these alloys, Al-Cu alloys are favored in engine manufacturing for their exceptional strength-to-weight ratio, high thermal conductivity, and excellent resistance to high-temperature degradation^[2, 3]. Yet, as engine designs push toward more extreme operating conditions, existing Al-Cu alloys face limitations in both mechanical performance and casting behavior. Consequently, further enhancements in strength and castability are essential to meet the rigorous operational standards of high-efficiency engines under extreme thermal and mechanical loads.

Ceramic particles such as TiB₂^[4, 5], Al₂O₃^[6, 7], and SiC^[8], known for their high strength and high elastic modulus, have garnered significant attention in materials science for their capability to effectively improve the microstructure and mechanical properties of Al-Cu alloys^[5, 8]. Although Al₂O₃ and SiC reinforcements have proven effective, a notable drawback is their tendency to form intermetallic compounds with the aluminum matrix, which can degrade the mechanical properties

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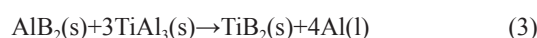
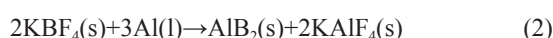
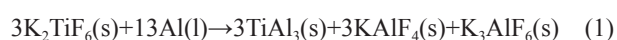
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and chemical stability of the composites^[9]. In contrast, TiB₂ particles exhibit a high degree of interface compatibility with the aluminum matrix, facilitating grain refinement during solidification. Additionally, TiB₂ particles have a high melting point (approximately 2,790 °C) and superior thermal stability^[10], ensuring minimal reaction with the aluminum matrix and preventing the formation of brittle intermetallic phases at the interfaces^[11]. These attributes make TiB₂ reinforcement a widely adopted approach for enhancing the performance of Al-Cu alloys.

The method used to introduce TiB₂ particles into aluminum alloys significantly influences the properties of the resulting materials. Two primary approaches are employed: ex-situ and in-situ techniques. In the ex-situ method, the aluminum matrix is melted, and pre-synthesized TiB₂ particles are introduced into the molten metal during stirring processes such as mechanical stirring, ultrasonic stirring, or electromagnetic stirring. However, this approach faces several challenges, including achieving homogeneous dispersion of the particles, ensuring their thermal stability during processing, and promoting strong interfacial bonding between the reinforced particles and the matrix^[9]. These challenges can adversely affect the mechanical properties of the composites. To overcome these limitations, the in-situ method has been developed. In this process, the reinforcement phase is synthesized directly within the matrix material through reactions such as direct melt oxidation, gas/melt reactions, or melt/salt reactions. Among these approaches, the melt/salt reaction method is widely used due to its cost-effectiveness and the ready availability of halide salts (K₂TiF₆). In this method, halide salts react with aluminum to form TiB₂ particles within the melt, leading to improved dispersion and interfacial bonding compared to the ex-situ approach. The reaction between halide salts and aluminum is given as below^[12, 13].



The incorporation of TiB₂ particles in Al-Cu alloys has been widely studied for its significant impact on microstructural evolution and mechanical properties. TiB₂ particles facilitate grain refinement by promoting the formation of fine equiaxed grains and inhibiting the growth of columnar structures^[14, 15]. Various theories, including the carbide-boride particle theory, duplex nucleation theory, and peritectic hulk theory, have been proposed to explain the grain refinement mechanism induced by TiB₂ particles^[16-19]. However, the precise mechanism remains uncertain. Beyond nucleation effects, the influence of nano-sized TiB₂ particles on precipitate distribution in Al-Cu alloys has also been investigated. Studies have demonstrated that introducing nano TiB₂ particles enhances the precipitation of θ' particles, leading to the coexistence of fine and dense θ' and θ' particles, thereby improving the mechanical properties of the alloys^[20, 21]. Additionally, TiB₂ particles have been observed to reduce porosity in aluminum alloys, contributing

to improved mechanical performance^[22]. Furthermore, TiB₂ particle reinforcement has been found to enhance the wear resistance of aluminum alloys. For instance, Mandal et al.^[23] reported that wear resistance of Al-4Cu alloy was improved with increasing TiB₂ content. Specifically, adding 10wt.% TiB₂ particles reduced the wear rate by approximately 1.3 times. However, research on the impact of submicron TiB₂ particles on precipitate evolution and casting properties of Al-Cu alloys remains limited.

In this work, the effects of varying contents of submicron TiB₂ particles on the microstructure, casting performance, and mechanical properties of the ZL205A alloy were systematically examined. Grain structure and precipitate distribution were analyzed through optical microscopy (OM) and transmission electron microscopy (TEM), respectively. Casting performance was evaluated through spiral fluidity and hot tearing tests. The mechanisms of TiB₂ particles influencing grain refinement and enhancing casting properties were analyzed. Furthermore, Brinell hardness and room-temperature tensile strength were examined to evaluate the effect of submicron TiB₂ particles on the mechanical properties of Al-Cu alloys. Based on the experimental results, an optimal TiB₂ particle content was identified.

2 Experimental procedures

2.1 Alloy preparation

The ZL205A Al-Cu alloy, composed of 5.16wt.% Cu, 1.06wt.% Mn, 0.49wt.% Mg, 0.37wt.% Ti, 0.08wt.% Fe, 0.07wt.% Si, and >0.03wt.% B, with the balance being Al, was selected as the base material in this study. It should be noted that the base alloy was pre-modified with a grain refiner Al-5Ti-1B. Submicron TiB₂ particles were introduced into the base alloy with mass fractions of 0.2%, 0.5%, 1%, and 3%, and the corresponding experimental alloys were designated as Base, 0.2wt.% TiB₂, 0.5wt.% TiB₂, 1wt.% TiB₂, and 3wt.% TiB₂, respectively. An Al-TiB₂ master alloy containing 20wt.% TiB₂, supplied by Dalian Ketian New Materials Co., Ltd., was applied in this work to achieve different amounts of TiB₂ particles-containing alloys. Its morphology was analyzed using scanning electron microscopy (SEM). The base alloy was melted at 780 °C, and varying amounts of Al-TiB₂ master alloy was introduced into the melt under continuous stirring for 10 min. After an additional holding period of 5 min and a subsequent refining treatment, ultrasonic processing was applied to the melt to ensure a uniform dispersion of TiB₂ particles prior to pouring. The melt was then poured into a steel mold at 730 °C. All specimens underwent the standard T6 heat treatment process, which comprised solution treatment at 535 °C for 10 h, water quenching, and artificial aging at 170 °C for 4 h, followed by air cooling.

2.2 Microstructure characterization

OM (ZEISS Axio Observer), SEM (QUANTA FEG 250), and TEM (FEI Talos F200X) were performed to examine the

microstructure characteristics of the experimental alloys. To reveal the grain structure, polished samples were etched with a mixed acid reagent consisting of 2.5% HF, 1.25% HNO₃, 1.25% HCl, and 95% water, followed by observation under OM. Grain size measurements were conducted utilizing ImageJ software, based on a minimum of 100 grains per sample. SEM, equipped with energy dispersive spectroscopy (EDS), was employed to detect intermetallic phases and examine fracture morphology after hot tearing tests. TEM analysis was carried out to examine the distribution of precipitates, with all bright-field (BF) TEM images taken along the <001> α -Al zone axis.

2.3 Casting performances characterization

The fluidity of ZL205A alloys with varying TiB₂ contents was measured using a spiral sand mold, as shown in Fig. 1(a), featuring three concentric spirals and a gating system composed of an external pouring channel and a short, tapered

sprue. Prior to casting, the sand molds were preheated at 120 °C for 3 h. Molten alloy was poured into the mold, and fluidity was quantified as the average length of metal flow along the spirals: a longer length indicates better fluidity.

Hot tearing susceptibility was assessed utilizing ring-shaped sand molds ($\Phi 188$ mm) fitted with interchangeable steel cores to produce 5 mm-thick rings with widths of 10, 12.5, 15, 17.5, and 20 mm, respectively, as shown in Fig. 1(b). Iron chills were placed adjacent to the ring-channel interface to direct solidification and induce cracking precisely at the end of solidification. The hot tearing susceptibility coefficient (HTSC) was determined utilizing the equation as follows^[24]:

$$HTSC = \frac{w}{188} \quad (4)$$

where w represents the critical welt-ring width at which cracking first appears. Lower HTSC values indicate low susceptibility to hot tearing.

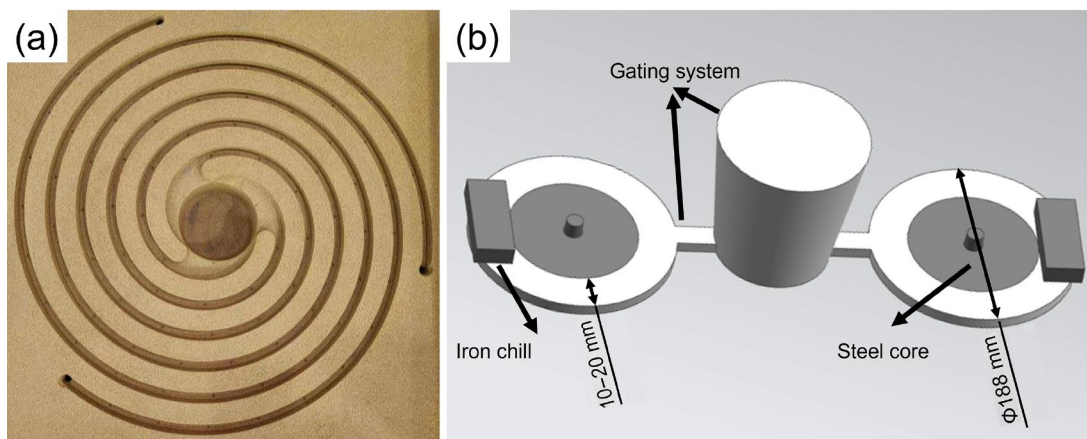


Fig. 1: Spiral sand mold (a) and schematic diagram of the crack-ring mold test setup (b)

2.4 Mechanical properties characterization

Brinell hardness of the experimental alloys was measured in accordance with GB/T 231.1-2018 standard utilizing a 320HBS-3000 testing machine. Measurements were performed with a 5 mm diameter steel ball under a load of 250 kgf. Three measurements were conducted for each sample, and the mean result was recorded.

The tensile tests were performed on T6-treated specimens at ambient temperature utilizing an MTS (E45.105) electronic universal testing machine, following the GB/T 228.1-2021 standard. The size of the tensile sample is shown in Fig. 2. The 0.2% offset yield strength (YS) was measured for all the experimental alloys. Each sample with varying TiB₂ contents was tested at least twice, and the average values were recorded.

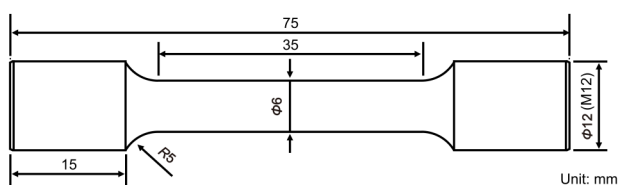


Fig. 2: Geometry and dimensions of the tensile sample

3 Results and discussion

3.1 Microstructure

Figure 3 presents the typical grain structures of all the tested alloys. Typical equiaxed grains are detected in all the tested alloys regardless of the TiB₂ particles content. The average grain sizes of all tested alloys were quantified, with the results presented in Fig. 4. The base alloy presents the largest grain size of 58 μ m. Upon the introduction of submicron TiB₂ particles, the value of grain size is markedly reduced, reaching 32.7 μ m at a TiB₂ content of 0.2wt.%, showing a 44% reduction. However, as the TiB₂ content increases from 0.2wt.% to 3wt.%, the grain size shows negligible changes, indicating the limited grain refinement effect of TiB₂ particles, consistent with the findings from previous study^[25].

The typical microstructure of the as-cast alloys was examined using SEM. As all the experimental alloys exhibit similar microstructural features, the alloy containing 0.2wt.% TiB₂ is presented as a representative example, as shown in Fig. 5. SEM-EDS analysis [Figs. 5(b-d)], supported by previous research^[26, 27], confirms the presence of Al₇Cu₂(Fe, Mn), Al₂Cu, and primary Al₃(Ti, Zr, V) phases. The Al₂Cu and

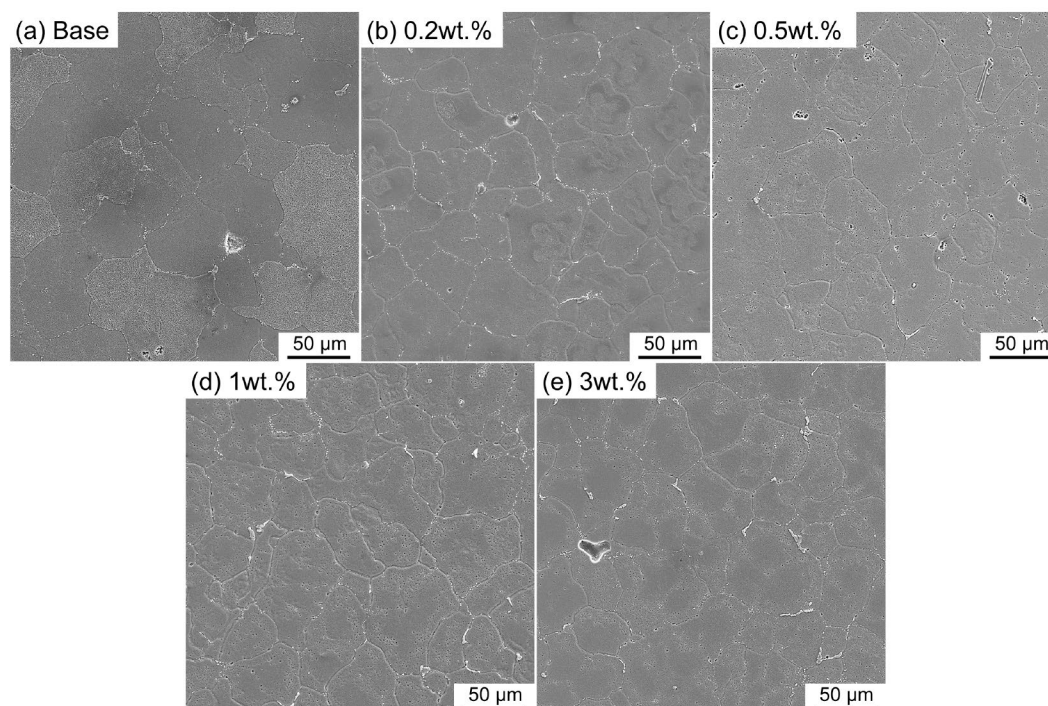


Fig. 3: Typical grain structures of experimental alloys with various contents of TiB_2 : (a) Base; (b) 0.2wt.%; (c) 0.5wt.%; (d) 1wt.%; (e) 3wt.%

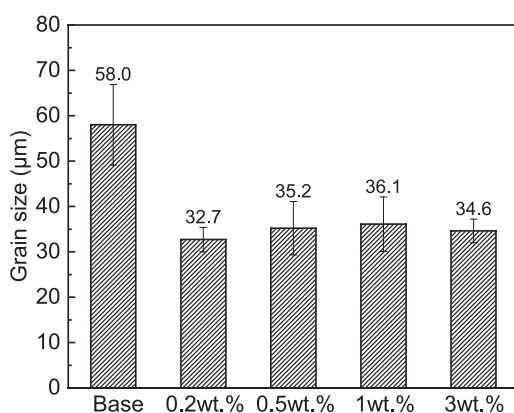


Fig. 4: Quantified average value of grain sizes of tested alloys with various TiB_2 contents

Fe-rich phases $[\text{Al}_7\text{Cu}_2(\text{Fe}, \text{Mn})]$ are found to predominantly distribute along the grain boundaries, whereas the $\text{Al}_3(\text{Ti}, \text{Zr}, \text{V})$ phases are randomly dispersed within the aluminum matrix.

To examine the distribution of TiB_2 particles, SEM-EDS mapping was conducted. Figure 6 shows the typical SEM-EDS mapping results for the 0.2wt.% TiB_2 alloy in its as-cast condition. A high-magnification image of the region outlined by the red rectangle in Fig. 6(a) is shown in Fig. 6(b), highlighting the spatial distribution of Ti elements. Due to the low atomic weight of boron, accurate detection via SEM-EDS is not feasible using the Quanta FEG 250 instrument. Thus, B element mapping is not presented here. As shown in Fig. 6(b), the majority of Ti elements are predominantly concentrated near the intermetallic phases. Since these intermetallic phases are primarily located along grain boundaries, it can be inferred that most submicron TiB_2 particles are distributed along the grain

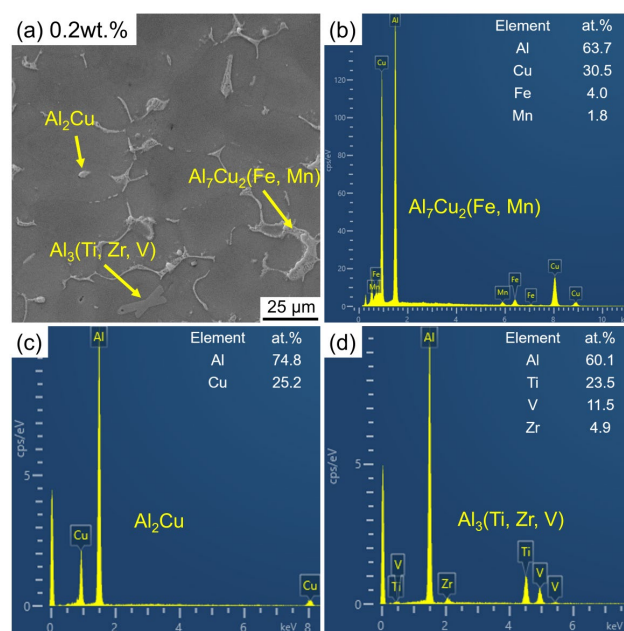


Fig. 5: Typical microstructure of as-cast 0.2wt.% TiB_2 alloy (a) and SEM-EDS analysis of $\text{Al}_7\text{Cu}_2(\text{Fe}, \text{Mn})$ (b), Al_2Cu (c), and $\text{Al}_3(\text{Ti}, \text{Zr}, \text{V})$ (d) intermetallic phases

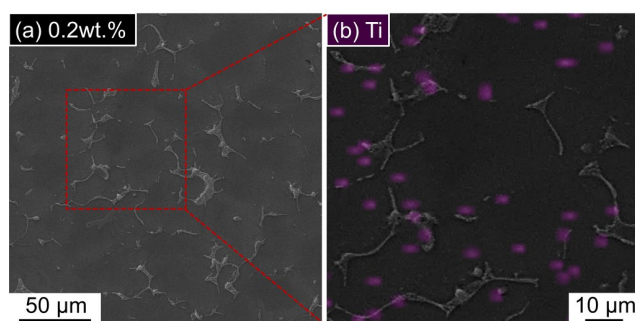


Fig. 6: SEM-EDS mappings of as-cast 0.2wt.% TiB_2 alloy

boundaries as well.

The typical microstructure of all the tested alloys following T6 heat treatment is shown in Fig. 7, observed using SEM in secondary electron mode. During solution treatment, most of the Al_2Cu phases dissolve into the aluminum matrix, while the Fe-rich $\text{Al}_7\text{Cu}_2(\text{Fe}, \text{Mn})$ and primary $\text{Al}_3(\text{Ti}, \text{Zr}, \text{V})$ phases remain largely unchanged due to their higher melting points. As revealed in Figs. 7(a)–(e), the Fe-rich $\text{Al}_7\text{Cu}_2(\text{Fe}, \text{Mn})$ phases are predominantly located along the grain boundaries. Besides, dark blocky features, highlighted within the green

rectangles and marked by orange arrows, are observed along the grain boundaries. SEM-EDS analysis confirms these features as TiB_2 particles, based on the measured Ti:B atomic ratio of approximately 1:2. This observation supports the inference that the majority of the submicron TiB_2 particles are located along or near the grain boundaries, which aligns well with the SEM-EDS mapping results presented in Fig. 6. Furthermore, the size of the TiB_2 particles was quantified using ImageJ software based on at least three SEM micrographs taken from different regions, revealing an average particle size of less than $0.65 \mu\text{m}$.

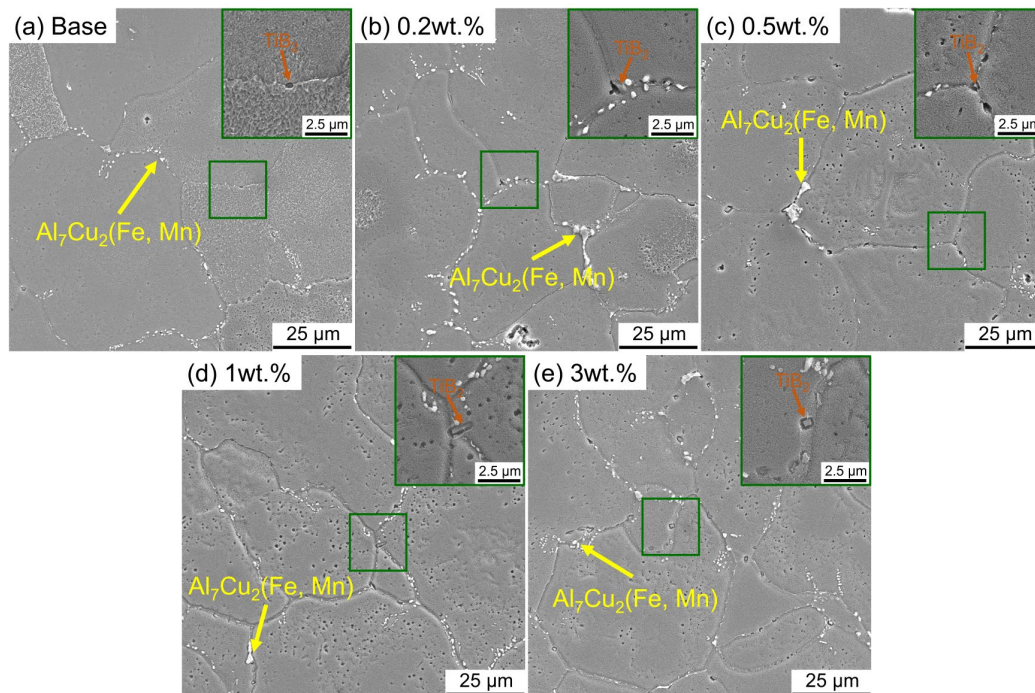


Fig. 7: SEM microstructures of T6-treated experimental alloys with various contents of TiB_2 : (a) Base; (b) 0.2wt.%; (c) 0.5wt.%; (d) 1wt.%; (e) 3wt.%. The inset images in the right corner showing the corresponding high-magnification micrographs of the region highlighted by green rectangles

Based on the microstructure characterization results, the effect of submicron TiB_2 particles on grain evolution is schematically depicted in Fig. 8. In the base alloy, the Al-5Ti-1B grain refiner promotes nucleation during solidification; however, its limited ability to restrict grain growth leads to rapid coarsening and a coarse as-cast microstructure. In contrast, in TiB_2 -containing alloys, the additional TiB_2 particles serve as extra nucleation sites, increasing the grain density per unit area. This constrained growth space refines the grain structure^[28]. Moreover, some TiB_2 particles are driven to the grain boundaries by the advancing solid/liquid interface, where they impede heat flow and solute movement, slowing down grain growth^[29]. Their presence at the boundaries also pins these interfaces, thereby increasing local surface tension and impeding interface movement, which further inhibit grain growth^[30]. As a result, the grain size is significantly refined, as confirmed by Fig. 3.

The BF TEM images observed through $\langle 001 \rangle_{\alpha\text{-Al}}$ zone axis, along with their corresponding selected area diffraction (SAD) patterns, are displayed in Fig. 9. Since all experimental alloys presented similar precipitate characteristics, representative TEM

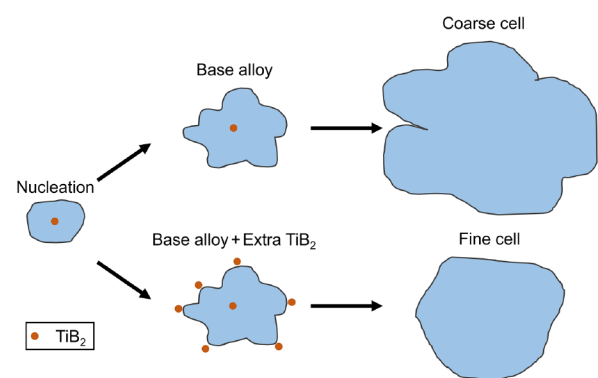


Fig. 8: Schematic graph of submicron TiB_2 particles on modification of grain structure

images of the 0.2wt.% and 1wt.% TiB_2 alloys are presented. Coexisting θ'' and θ' phases are found in the aluminum matrix for all the tested alloys, as highlighted by orange and blue arrows, respectively. These two types of precipitates can be distinguished by both size and contrast: θ'' particles are typically smaller (length $< 20 \text{ nm}$, width $\approx 2 \text{ nm}$), while θ' particles appear darker in BF-TEM images due to their higher copper content.

θ' has an Al_2Cu stoichiometry, while it is Al_3Cu for θ'' ^[31]. The presence of both θ'' and θ' phases was further confirmed by the characteristic diffraction spots and streaks in the SAD patterns (Fig. 9, insets), as indicated by white arrows^[32]. The thickness of the observed TEM foils in Figs. 9(a) and (b) is similar, and the size and number density of θ'' and θ' precipitates are consistent across all the alloys. This indicates that the introduction of submicron TiB_2 particles has a negligible effect on the precipitation behavior of the Al-Cu alloys.

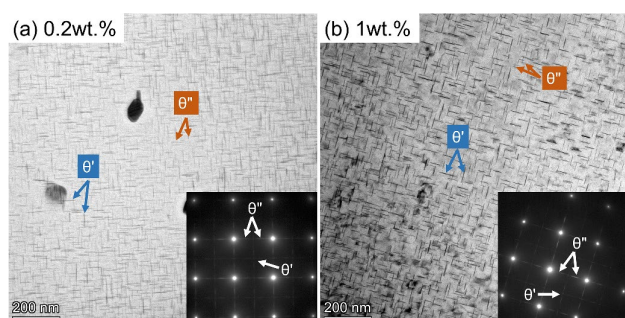


Fig. 9: BF TEM images of T6-treated 0.2wt.% TiB_2 (a) and 1wt.% TiB_2 (b) alloys

3.2 Casting properties

The spiral fluidity tests were performed for all the experimental alloys, and the results are displayed in Fig. 10, showing the average fluidity length. The addition of 0.2wt.% TiB_2 particles significantly increases the fluidity length from 523 mm to 617 mm, representing an 18% increment compared to the Base alloy. However, when further increases the TiB_2 particles content, enhancements in fluidity length become marginal, reaching a total increment of only 20% even at a TiB_2 content of 3wt.%. These results suggest that although the introduced submicron

TiB_2 particles can improve the fluidity property of ZL205A alloy, the efficiency of this enhancement diminishes at TiB_2 contents above 0.2wt.%.

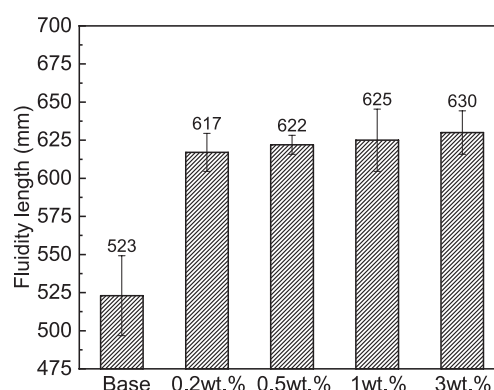


Fig. 10: Average fluidity length of experimental alloys

The hot tearing susceptibility tests were conducted, and the results are summarized in Table 1. It is evident that the introduced submicron TiB_2 particles reduced the critical width of crack ring from 17.5 mm to 12.5 mm. Notably, varying TiB_2 particle contents display a similar influence on improving the hot tearing resistance of ZL205A alloy, as all TiB_2 -containing alloys exhibit a critical crack ring width of 12.5 mm. Furthermore, the calculated HTSC of all the tested alloys based on Eq. (4) are presented in Fig. 11. A significant reduction in HTSC is noted with the addition of submicron TiB_2 particles, decreasing from 0.093 of the Base alloy to 0.066 of the TiB_2 -reinforced alloys, corresponding to a 29% reduction. The decreased critical width of crack ring and HTSC values collectively confirm the enhanced hot tearing resistance due to the presence of submicron TiB_2 particles.

Table 1: Hot tearing susceptibility test results of experimental alloys

Alloy	Welt-ring width (mm)					
	10	12.5	15	17.5	20	22.5
Base			With cracks	With cracks		
0.2wt.%						
0.5wt.%	With cracks	With cracks			Without cracks	Without cracks
1wt.%			Without cracks	Without cracks		
3wt.%						

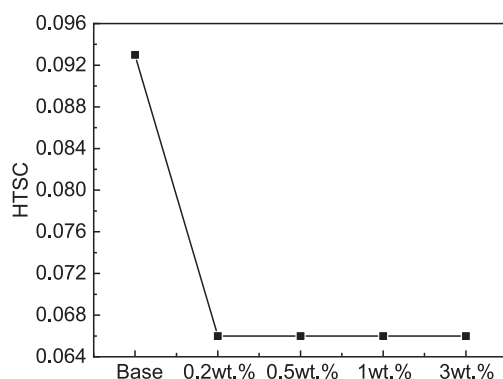


Fig. 11: Calculated HTSC of all the experimental alloys

The fracture morphology of the hot tearing samples from both the Base alloy and the 0.2wt.% TiB_2 alloy was observed using SEM, with typical images being shown in Fig. 12. The SEM analysis results reveal that nearly all the hot tearing fractures in this study propagate along grain boundaries, and no obvious brittle fracture facets are observed. Moreover, the fracture morphology of Base alloy displays relative coarser grain structures compared to the TiB_2 -containing alloys, as evidenced by larger cell sizes and obvious dendritic structures. This finding confirms that the addition of submicron TiB_2 particles modifies the grain morphology, as demonstrated in Fig. 3. In addition, interdendritic bridging is detected in the TiB_2 -containing alloys, as shown in Fig. 12(b).

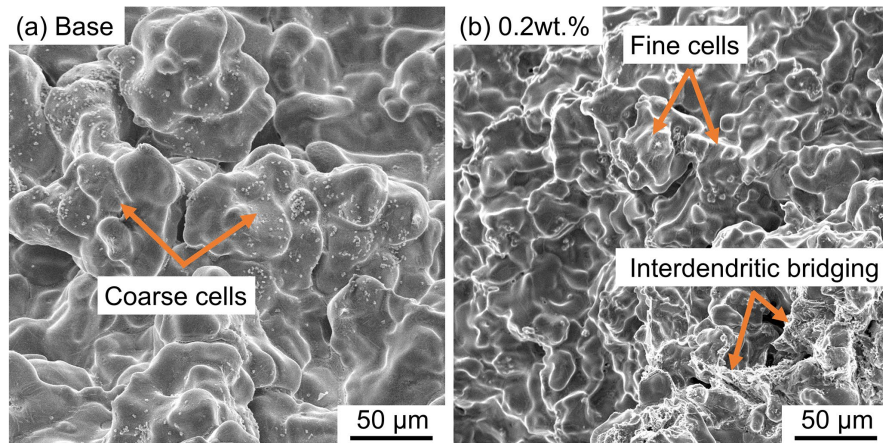


Fig. 12: Typical hot tearing fracture morphologies of Base alloy (a) and 0.2wt.% TiB₂ alloy (b)

During solidification, the presence of extra TiB₂ particles promotes heterogeneous nucleation of α -Al grains, thereby reducing the undercooling required for solidification. This reduction stabilizes the solidification front and delays the formation of rigid dendritic networks, allowing the melt to flow more freely for a longer period before extensive solidification occurs. Additionally, the refined and equiaxed grain structure in TiB₂-containing alloys provide a less tortuous path for the melt compared to the coarse structure of the Base alloy. This results in a longer fluidity length. Regarding hot tearing, the refined grain structure improves intergranular contact during late stages of solidification, facilitating interdendritic bridging. This bridging enhances the grain boundary binding force, which in turn increases the alloy's resistance to hot tearing. It can be concluded that grain size plays a critical role in influencing both fluidity and hot tearing resistance. As shown in Figs. 3 and 4, all TiB₂-containing alloys exhibit similar grain sizes. Therefore, within the investigated range, the variations in fluidity length and HTSC among different TiB₂ additions are relatively small.

3.3 Mechanical performance

Figure 13 presents the average Brinell hardness values of the experimental alloys. It is found that a small addition of submicron TiB₂ particles (0.2wt.%) has a negligible effect on hardness. However, as the content of submicron TiB₂ particles

elevates to 0.5wt.% and further to 1wt.%, the Brinell hardness displays a notable improvement, from 129.0 HB to 133.0 HB and 136.7 HB, corresponding to increments of 3.1% and 6.0%, respectively. When the TiB₂ particles content further increases to 3wt.%, the value of Brinell hardness drops significantly to 130.3 HB. Therefore, the 1wt.% TiB₂ alloy displays the highest value of Brinell hardness among all compositions tested.

Tensile tests were carried out at ambient temperature for all experimental alloys, and their average strength and elongation are displayed in Fig. 14. Consistent with the hardness testing results, the difference between the Base alloy and the 0.2wt.% TiB₂ alloy in ultimate tensile strength (UTS) is minimal. However, the YS of the 0.2wt.% TiB₂ alloy decreases slightly from 352 MPa of the Base alloy to 345 MPa, accompanied by a reduction in elongation from 8.5% to 8.3%. As the TiB₂ particle content increases up to 1wt.%, both UTS and YS are progressively enhanced, reaching peak values of 459 MPa and 376 MPa, respectively. Beyond this threshold, the addition of TiB₂ particles at a 3wt.% level leads to diminished mechanical properties, with UTS and YS declining to 446.5 MPa and 368 MPa, respectively. Therefore, the 1wt.% TiB₂ alloy attains the optimal balance of UTS and YS, representing increases of 3.6% and 6.8% relative to the Base alloy. In contrast, elongation exhibits a continuous decline with increasing TiB₂ content, falling from 8.5% of the Base alloy to 7.7% of the 3wt.% TiB₂ alloy. This reduction in elongation is primarily

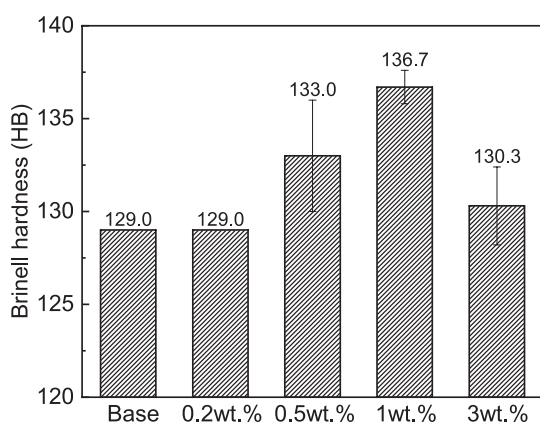


Fig. 13: Average Brinell hardness of test alloys

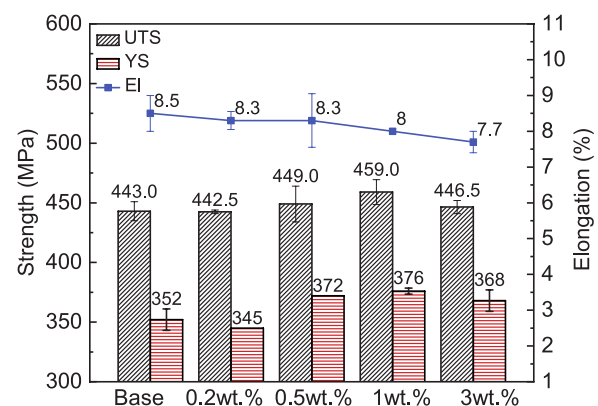


Fig. 14: Mechanical properties of T6-treated experimental alloys at ambient temperature

attributed to local stress concentrations and inhomogeneous deformation between the aluminum matrix and the submicron TiB_2 particles^[33, 34]. The strengthening mechanisms induced by the submicron TiB_2 particles are principally twofold: (1) the refinement of grain structure, which enhances strength in accordance with the Hall-Petch relationship^[35]; and (2) the reinforcement of the aluminum matrix by the submicron TiB_2 particles, which impedes dislocation motion, thereby further enhancing strength. Nonetheless, the overall mechanical properties of the TiB_2 -reinforced alloys critically depends on the volume fraction of TiB_2 particles: at very low concentrations (e.g., 0.2wt.%), the modest strengthening effect is outweighed by a loss in ductility, leading to a net decrease in strength; as the TiB_2 content increases, e.g., from 0.5wt.% to 1wt.%, the reinforcing benefits become more pronounced, resulting in an improvement in strength. However, at excessive levels (e.g., 3wt.%), the adverse effects of stress concentration and reduced ductility become dominant, ultimately leading to a reduction in overall strength.

4 Conclusions

The impact of varying submicron TiB_2 particle contents on the microstructure, casting performance, and mechanical properties of the ZL205A alloy was systemically investigated. The primary conclusions are summarized as follows:

(1) The introduced submicron TiB_2 particles act as effective nucleation sites and inhibitors of grain growth, leading to significantly finer and more equiaxed grain structures in the ZL205A alloy.

(2) SEM and TEM analyses indicate that the introduction of submicron TiB_2 particles do not significantly modify the microstructural constituents, with the intermetallic phases and precipitate distribution remaining largely unchanged.

(3) The addition of submicron TiB_2 particles enhances the casting performance of the ZL205A alloy by increasing fluidity and reducing hot tearing susceptibility. The introduction of 3wt.% TiB_2 particles achieves the most notable improvements, with an approximate 20% increase in fluidity length and a 29% reduction in HTSC. Nevertheless, the enhancement in casting properties exhibits no substantial difference among alloys containing 0.2wt.%–3wt.% TiB_2 .

(4) The highest Brinell hardness and tensile strength are achieved in the alloy containing 1wt.% TiB_2 , primarily due to the refined grain structure and effective particle reinforcement. Based on the observed improvements in both castability and mechanical properties, a TiB_2 content of 1wt.% is recommended for industrial application of the ZL205A alloy.

Conflict of interest

The authors declare that they have no known competing financial interests or personal relationships that could have appeared to influence the work reported in this paper.

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