The effect of solution treatment on aging behavior and mechanical properties of AA2024-TiB₂ composite synthesized by semi-solid casting

Semegn Cheneke¹ · D. Benny Karunakar¹

Received: 29 July 2019 / Accepted: 17 October 2019 / Published online: 26 October 2019
© Springer Nature Switzerland AG 2019

Abstract

AA2024 reinforced with varying amounts of titanium boride (TiB₂) was developed using stir rheocasting to study the effect of solution heat treatment on aging behavior and mechanical properties. The developed specimens were subjected to solution heat treatment at 490 °C for 1, 2.5, and 5 h, followed by artificial aging at 190 °C for an aging duration of 1–12 h. The metallurgical characterization was done by x-ray diffraction, optical microscopy, and field emission scanning electron microscope and mechanical characterization by Vickers hardness tester and universal testing machine. Microstructural observation showed the formation of globular grains and fairly uniform particle distribution. Up on peak aging, the AA2024/2 wt% TiB₂ composite has shown better improvement in yield strength (YS = 230 MPa), tensile strength (UTS = 342 MPa), and Vickers hardness (VH = 147) amongst other developed specimens. The improvement in mechanical properties of AA2024/2 wt% TiB₂ composite up on peak aging was nearly 83%, 77%, and 20%, respectively, over their as-cast states. However, the percent elongation has decreased by 51% for the AA2024/2 wt% TiB₂ composite against its as-cast state. The optimum solution heat treatment time was found to be 2.5 h that led to the fastest aging kinetics and peak hardness. Due to incomplete dissolution of precipitates at a solution treating time less than 2.5 h, the hardness decreased, and the aging kinetics decelerated. The aging kinetics further accelerated with increasing solution heat treating time beyond 2.5 h, but the hardness decreased due to grain growth of the matrix.

Keywords AA2024 · Stir rheocasting · Mechanical properties · Solution heat treatment · Microstructure · Aging

1 Introduction

Al alloys and Al-based metal matrix composites have attracted the attention of researchers and material engineers due to their wide applicability in the areas of aerospace and automobile industries [1]. Particularly, 2xxx series Al alloys and their composites have been studied extensively due to ease of formability, high strength to weight ratio, and age hardenability, etc., [1].

For synthesizing Al and Al-based composites, various methods such as liquid state, solid state, and in situ methods have been developed. Among them, the liquid casting

(stir casting) is the most commonly used method due to its simplicity. Moreover, it offers economic benefits compared with other fabrication processes (one-third to half of the competitive methods), and it is suitable for high volume production [2, 3]. However, the development of composites using the liquid casting method has its drawbacks. The mixing of the ceramic particles and the matrix is difficult due to a lack of wettability between them.

Furthermore, several defects, such as porosity formation, interfacial reactions, and non-homogeneous reinforcement distribution, arise [4]. In the liquid casting route, there is always a density difference between the

✉ D. Benny Karunakar, bennyfme@iitr.ac.in | ¹Mechanical and Industrial Engineering Department, Indian Institute of Technology Roorkee, Roorkee 247667, India.

reinforcement and the matrix. As a result, the particles sink or float depending on their relative density to the liquid matrix density, causing a non-uniform reinforcement distribution. Therefore, due to poor particle dispersion, agglomeration occurs, which leads to the formation of high porosity content in the composite [5, 6]. There are several methods reported to improve the wettability between the particle and matrix. Some of them are the addition of wetting agents [7] and fluxes [8], oxidation [9], preheating [10], and coating [11] of the particles. Another cost-effective method reported for improving wettability is to develop the composite in the semi-solid temperature region of the alloy [4–6, 12–14]. The developed alternative method is called semi-solid metal (SSM) casting/com-pocasting/stir rheocasting.

SSM casting is one of the cost-effective manufacturing processes for synthesizing lightweight metals such as aluminum. It offers enhanced metallurgical and mechanical properties compared to the conventional casting method [15]. The defects like voids, cracks, porosity, and others can be minimized by refining the grains of the slurry where refinement could be achieved by processing in the mesh region [15].

There are two variations of processing methods in the SSM region of the alloy based on the slurry state during the casting step. They are named as semisolid–semisolid (SS), where the slurry is in a partially liquid state and the semi-solid-liquid (SL), where the slurry is in the liquid state during casting [4, 16]. Some researchers [17, 18] used the SL technique to develop Al-based composites, probably due to the ease of operation. But it was reported that reheating the slurry to a higher temperature leads to rapid grain coarsening and a change in microstructure from globular to dendritic structure [19]. As a result, the intended properties of the final cast product could be deteriorated. Therefore, SSM casting should be performed at a temperature that maintains the globular grain structure and high enough to maintain the fluidity for the incorporation of the particles during mixing (stirring) step. In SSM casting, the agitation process breaks down the dendritic arms, and it is possible to maintain the fluidity until the solid fraction reaches about 60% [5, 20]. In SSM, it is generally claimed that 30–50% solid is needed in the slurry [21].

The properties of Al alloys can be improved by suitable processing methods and heat treatment conditions. Besides, the incorporation of ceramic reinforcements in the base matrix enhances the properties and alters the aging behavior of the composites. The addition of ceramic particles in the Al base matrix has altered the precipitation behavior of the composites compared with the unreinforced alloys [22–24]. It was reported that the aging kinetics of the composites accelerated due to the addition of reinforcement particles [25–28]. However, some

researchers also reported that it decelerated, or there is a very little alteration in the aging kinetics [29, 30]. Cottu et al. [31] reported that the aging kinetics of Al–Cu–Mg matrix reinforced with 10 wt% of SiC fibers altered due to the presence of the particles. It is due to the difference in coefficients of thermal expansion (CTE) between the particles and the matrix, which induce strain filed between their interfaces. Thomas and King [32] studied the precipitation behavior of AA2124 reinforced with SiC composite developed by powder metallurgy. They reported that the particles facilitated the nucleation of the S' precipitate, which leads to reaching the peak hardness in a short time.

Dutta and Bourell [33], and Appendino et al. [34] investigated AA6061/SiC composite and reported that precipitation sequence of the base matrix and that of the composite is similar, but they observed an alteration in aging kinetics. The accelerated aging kinetics was due to an increase in solute diffusivity and an increase in dislocation density. However, Pal et al. [35] reported that the aging kinetics of Al–Cu–Mg/SiC composites decelerated. It was due to inadequate dislocation density, lower vacancy concentration, and extensive interfacial segregation of alloying elements. A similar report was made for Al alloys/Al₂O₃ composites [36, 37]. The aging kinetics of alloys and composites depends on the type and proportion of reinforcements, heat treatment, alloy composition, and other process parameters [35].

Most of the researches surveyed above aimed at investigating the influence of particles on the aging kinetics of the composites. In the current work, the effect of solution heat treatment followed by artificial aging on aging behavior and mechanical properties of stir rheocast AA2024 reinforced with TiB₂ composite was investigated. The appropriate heat treatment condition was optimized. Accordingly, the peak-aged properties of the developed samples were improved significantly, and comparison was made with their as-cast counterparts. The metallurgical analysis was also conducted to identify the main precipitate formed in the composite. Accordingly, the main precipitate identified was CuAl₂. Finally, the fracture mode of the unreinforced and TiB₂ reinforced AA2024 composite was analyzed. The study will provide an experimental basis for Al-based metal matrix composites/metal matrix composites.

2 Materials and methods

2.1 Materials

The raw material used for the experiment in the current study is aluminum AA2024 alloy. It was received in the form of a block from the supplier, Suresh Metals, Mumbai, India,

Table 1 Chemical composition of 2024 aluminum alloy

Cu	Mg	Mn	Si	Fe	Others	Al
3.9730	1.3390	0.4950	0.0811	0.0686	0.0546	Bal

Table 2 Characteristics of the TiB₂ reinforcement particle

Reinforcement	Density [g cm ⁻³]	Size	Melting point [°C]
TiB ₂	4.38	– 325 mesh	3225

Fig. 1 DTA curve of the alloy

with the chemical composition as given in Table 1. The reinforcement material used in this research is TiB₂ powder supplied by Sigma-Aldrich, and its properties are given in Table 2.

2.2 Methods

The liquidus and solidus temperature of the AA2024 are 645 °C and 550 °C, respectively, obtained from the differential thermal analysis (DTA) curve shown in Fig. 1. The alloy was prepared in the form of powder for DTA analysis. Heating and cooling of the sample were done at a rate of 10 °C/min in an argon atmosphere.

Initially, the AA2024 alloy was melted in a graphite crucible at a temperature of 720 °C using a vertical muffle furnace. The melt was isothermally held for 20 min for complete homogenization. After fully mixing the alloy, 1 wt% of hexachloroethane (C₂Cl₆) was added for slag-removing and degassing. 1 wt% of mg was also added into the slurry for improving wettability between the particles and the base matrix.

Next, the furnace temperature was adjusted to a semi-solid temperature of 640 °C to keep the slurry in a

Table 3 Compositions of the composite samples developed

No.	Compositions (wt%)	Designation
1	AA2024 + 0 wt% TiB ₂	RC-1
2	AA2024 + 2 wt% TiB ₂	RC-2
3	AA2024 + 4 wt% TiB ₂	RC-3
4	AA2024 + 6 wt% TiB ₂	RC-4

semi-solid state. According to the Scheil equation, this corresponds to a 40% solid fraction, Eqs. (1) and (2) [38],

$$f_L = (T_M - T) / (T_M - T_L)^{1/(1-K)} \quad (1)$$

$$f_S = (1 - f_L) \quad (2)$$

where f_L is the weight fraction of liquid in the liquid–solid zone, f_S is the weight fraction of solid in the liquid–solid zone, T_M is the melting point of pure aluminum, T is the selected semi-solid temperature of the alloy, T_L is the liquidus temperature of the alloy and k is constant.

The preheated reinforcement was added to the slurry at proportions of 2 wt%, 4 wt%, and 6 wt% of the total metal being melted. The reinforcement was preheated at a temperature of 300 °C for 2 h to remove the moisture content. Then, the slurry was stirred for about 10 min at 500 RPM. For each sample, approximately 300 g of the alloy was melted. Sample without reinforcement at the same semi-solid temperature, with the same procedure above, was also prepared for comparison purposes.

After holding the slurry for 10 min, it was manually transferred to the split type die, which was preheated at 300 °C for 1 h. The developed AA2024 and TiB₂-reinforced composites were solution treated at 490 °C for 1 h, 2.5 h, and 5 h and then quenched. All the solution treated samples were subjected to artificial aging at 190 °C for different time durations of 1 h, 3 h, 6 h, 9 h, and 12 h.

Table 3 shows the alloy and the reinforcement weight percentages used to develop the composites along with their designations. Figure 2 shows the experimental setup. The morphology of the TiB₂ particle used as reinforcement in this research is shown in Fig. 3.

2.3 Microstructure characterization

Metallographic samples were cut and prepared for microstructural analysis. The surfaces of the specimens

Fig. 2 A schematic of the stir rheocasting setup used for developing the composite

Fig. 3 SEM image of TiB₂ particles used in the development of the composite

were polished gradually using 320–1200 grit SiC emery papers, followed by velvet cloth with the Al₂O₃ suspension on a disk polisher. Finally, the samples were etched using Keller's reagent (2.5% HNO₃, 1.5% HCl, 1% HF, and 95% H₂O by volume). Microstructural characterization was done using field emission scanning electron microscope (FE-SEM), equipped with energy-dispersive spectroscopy (EDS) (Make-FEI Model-FE-SEM QUANTA 200 FEG, Netherlands), and optical microscope (OM), (Model: TL LED, Germany). X-ray diffraction (XRD) (Make-Bruker, Model-D8-Advance, Germany) analysis with CuKα ($k = 1.5418 \text{ \AA}$) was also carried out. Xpert Highscore plus-Philips XRD software was used for confirming the

formation of phases and intermetallic compounds in the developed composites.

Differential thermal analysis (DTA) was carried out to determine the liquidus and solidus temperatures of the alloy. The alloy was prepared in the form of powder for DTA analysis. The instrument used for thermal analysis was Make-EXSTAR (Model-SII 6300 EXSTAR). Heating and cooling of the sample were done at a rate of 10 °C/min in an argon atmosphere.

2.4 Testing of mechanical properties

2.4.1 Tensile test

The miniature samples of both unreinforced and reinforced composite tensile samples were machined according to ASTM E8–16a [39]. The miniature specimen dimension for the current experiment is (total length = 50 mm and gauge length = 18 mm). The tensile tests were conducted on an Instron tensile testing machine (Model-5982, Make: Instron, USA) to evaluate the yield strength (YS), ultimate tensile strength (UTS), and elongation to failure (% EL). The loading rate used to perform the test was 0.5 mm/min. The 0.2% proof stress, UTS, and % EL were extracted from the stress–strain curves.

2.4.2 Hardness test

Hardness measurements were performed to understand the aging kinetics. Microhardness test was conducted on Micro-Vickers Hardness Tester (Model: MHV-s-AUTO), at a 100 g load and dwelling time of 10 s. The test was performed after polishing the samples with a fine-grained

emery paper. At least five readings were taken randomly, and the average was reported.

3 Results and discussion

3.1 Microstructural characterization

3.1.1 XRD analysis

The XRD analyses of the samples are given in Fig. 4. It was conducted to confirm the formation of intermetallic phases and particle distribution in the base matrix. Al and Mg_2Si peaks were observed in the unreinforced rheocast sample (0 wt% TiB_2). TiB_2 , Al_3Ti , AlB_2 , and $MgAl_2O_4$ peaks were observed in the rheocast composite samples reinforced with 2 wt% of TiB_2 . As the percentage of the reinforcements increased from 2 to 6 wt%, the intensity of the peaks also increased. It shows the existence and distribution of reinforcement particles and other phases in more concentration. The weak intermetallic phases like Mg_2Si , Al_3Ti , and AlB_2 , which were confirmed by XRD phase analysis, affect the final property of the composite. It was reported that they act as preferential sites for the initiation of cracks and their propagation [40].

3.1.2 OM and SEM analysis

During SSM casting, stirring the slurry in the mesh region homogenizes the temperature distribution. The homogenized temperature distribution and high cooling rate affect the final quality of the cast product. It provides minimized micropores, dispersion of eutectic nuclei, spheroidization of the primary phase, and grain refinement [41]. In conventional (liquid) casting, it is obvious that the grains are dendritic in structure, whereas the developed rheocast

samples have a spherical fine grain structure, as shown in Fig. 5. The optical micrograph of the unreinforced AA2024 alloy shown in Fig. 5a has a spherical grain structure. It can be observed that the microstructures of the composites, as shown in Fig. 5b, c, are globular in structure and were better refined compared to the unreinforced alloy. However, beyond 6 wt% of TiB_2 reinforcement in the matrix, the grains were not refined as expected due to the agglomeration of the particles, as shown in Fig. 5d. The grain sizes calculated using the linear intercept method were 30 μm , 21 μm , 22 μm , and 24 μm for the unreinforced rheocast alloy, 2 wt% TiB_2 reinforced composite, 4 wt% TiB_2 composite, and 6 wt% TiB_2 composite, respectively. The results show that the grains were refined significantly after the addition of TiB_2 particles compared to the unreinforced alloy.

In addition to the TiB_2 particles, stirring also has further refined the grain structure of the rheocast composite samples, which contributed to an improvement in the mechanical properties. The globular microstructure obtained, as clearly seen in the figures, shows that the rheocasting technique followed was effective in developing a better composite. As a result, ductility, strength, and hardness were improved significantly, which will be discussed later.

The FE-SEM microstructural features of the unreinforced and reinforced rheocast samples are shown in Fig. 6. As shown clearly in Fig. 6a, the grain structure of the unreinforced rheocast sample consists of oxide inclusions and micropores. Figure 6b shows the grain structure of the composite sample reinforced with 2 wt% TiB_2 . The particles are uniformly distributed in the matrix, as shown. The FE-SEM image of the sample with 4 wt% TiB_2 reinforcement is shown in Fig. 6c, and its grain refinement level is better than the first two samples (Fig. 6a, b). The distribution of the reinforcement in the matrix is also fairly uniform. However, particles clustered together in some portions of the structure around which voids and micropores were observed. Figure 6d shows the microstructural feature of the sample with a reinforcement ratio of 6 wt%. As the percentage of the reinforcement increased in the matrix, the tendency of the agglomeration of the particles also increased. Due to the agglomeration, the anticipated particle distribution within the matrix has not been achieved properly. It can be seen that the grains become coarser compared with that of the other reinforced samples. Even though it is a very good refining agent, in this case, the TiB_2 particles resulted in unsuccessful refinement due to agglomeration. Micropores and voids were also noticed in the developed samples. The large voids which were formed around the agglomerates diminish the mechanical properties of the composite.

Fig. 4 XRD pattern of the developed rheocast samples

Fig. 5 OM images of AA2024/TiB₂ rheocast composite **a** RC-1, **b** RC-2, **c** RC-3, **d** RC-4

The FE-SEM images of the magnified areas of the grains and the grain boundaries (GB) with EDS analysis are given in Fig. 7. The images of the magnified GB area of the developed unreinforced sample are shown in Fig. 7a. Large peaks of Al, Cu, and Mg were observed on the spectrum taken from selected area 1 in the inset of image 'a' as shown in Fig. 7b. Their approximate elemental compositions are also given. The magnified GB area of the sample with 2 wt% of TiB₂ reinforcement is shown in Fig. 7c. The spectrum of the EDS analysis of the selected area on the sample shows Ti, Al, and B peaks. Their compositions are also provided, as shown in Fig. 7d. Figure 7e shows the magnified areas of the sample with 4 wt% of TiB₂ reinforcement. As indicated in the EDS spectrum, as shown in Fig. 7f, large Ti, Al, and B peaks were observed. The magnified grain area on the microstructure shows the formation of voids around the agglomerates, which affect the mechanical properties. The FE-SEM image taken from the magnified area of the 6 wt% TiB₂ reinforcement is also shown in Fig. 7g. However, in this case, the particles were clustered together, as shown. The Ti and B peaks are shown on the EDS spectrum, as shown in Fig. 7h. As shown clearly, very large voids were formed around the cluster, which deteriorated the mechanical properties of the composite

to a great extent. The particles clustered are TiB₂ agglomerates, which were confirmed by the EDS analysis.

Figure 8 shows the magnified FE-SEM images of some intermetallics formed in the composites. The approximate elemental compositions analyzed using EDS are also provided. Figure 8a shows the magnified FE-SEM images of grain area showing TiB₂ particles. Figure 8b shows the EDS spectrum of spot 1 in the inset of image 'a', with approximate compositions of Ti and B elements. As shown clearly, the large peaks are Ti and B elements. Figure 8c shows the magnified FE-SEM images of grain area showing Mg₂Si in small concentration. The Mg and Si minor peaks were identified, as shown in Fig. 8d. The EDS spectrum of selected area 1 in the inset of image 'c', with approximate compositions of Mg and Si elements, is also provided. It shows the availability of very lesser amounts of Mg and Si elements. Figure 8e shows the magnified FE-SEM images of the grain area showing Al₃Ti. The EDS spectrum of spot 1 in the inset of image 'e', with approximate compositions of Al and Ti elements, is also shown in Fig. 8f. It can be noticed that large peaks of Al and Ti elements were observed on the spectrum, which confirmed the formation of Al₃Ti intermetallic.

Fig. 6 FE-SEM images of AA2024/TiB₂ rheocast composite **a** RC-1, **b** RC-2, **c** RC-3, **d** RC-4

The intermetallic compound Mg₂Si exhibits low density and high hardness that improves the properties of Al- and Mg-based MMCs [42–45]. But the major problem of this intermetallic is its brittleness that affects ductility depending on its size [43–45]. Generally, various researchers used Mg₂Si intermetallic as reinforcement for MMCs due to the improved mechanical properties it offers. Therefore, the Mg₂Si intermetallic formed in the present research was beneficial to mechanical properties. However, the AlB₂ and Al₃Ti intermetallics degrade the mechanical properties.

Figure 9 shows the FE-SEM images of AA2024 reinforced with 2 wt% TiB₂ composite, solution treated at 490 °C for different time durations (1–5 h). Precipitates can be observed in the structure. Figure 9a shows the existence of more undissolved precipitates (bright phase). The XRD pattern shown in Fig. 10 has confirmed the existence of this undissolved CuAl₂ precipitates. Figure 9d shows the microstructure area under consideration for EDS analysis of the

precipitate. Figure 9e shows the approximate elemental compositions of this precipitate, showing the enrichment of Cu solute in the structure. It can be concluded that solution treatment for 1 h is not sufficient for a complete dissolution of the precipitates as the undissolved precipitates were observed in much more amount. Figure 9b shows the FE-SEM image of the sample solution treated at 2.5 h. In this case, a few undissolved precipitates were observed in the structure (bright phases), compared to Fig. 9a. Figure 9c shows the FE-SEM image of the sample solution treated at 5 h with very few undissolved precipitates in the structure (bright phases), compared to Fig. 9a, b). The XRD result of the unaged and peak-aged RC-2 sample is shown in Fig. 10.

Fig. 7 FE-SEM images of unreinforced/reinforced AA2024 matrix rheocast composite samples reinforced with **a** 0 wt% TiB₂, **b** EDS analysis of selected area 1 in inset of image 'a', **c** 2 wt% TiB₂, **d** EDS analysis of selected area 1 in inset of image 'c', **e** 4 wt% TiB₂, **f** EDS analysis of selected area 1 in inset of image 'e', **g** 6 wt% TiB₂, **h** EDS analysis of selected area 1 in inset of image 'g'

3.2 Mechanical properties

3.2.1 Hardness

Figure 11 shows the hardness of the developed alloys and composites solution treated at different time durations as a function of aging time. It can be noticed that there is a large scatter in the microhardness of the composites. It could be due to the contact between the indenter and the TiB₂ particles in the subsurface locations and the inhomogeneous distribution of the precipitates. The unreinforced alloy, as shown in Fig. 11b, solution treated at 490 °C for 2.5 h, reached a peak hardness of 120 HV after aging for about 6 h. With the addition of TiB₂, the peak hardness has increased significantly. From Fig. 11a–c, it can be noticed that the time to reach the peak hardness decreased with an increase in the solution treatment time.

With an increase in the amount of TiB₂ up to 2 wt%, the peak hardness has increased from 120 to 147 VH (82% increase) for solution treated sample at 490 °C for 2.5 h. The time to reach the peak hardness is the same and was not studied independently for both unreinforced and reinforced composite samples in the current research. Therefore, we can only conclude that the presence of a higher content of particles in the matrix enhanced the hardness. The increase in the hardness due to the presence of TiB₂ particles was attributed to the possible nucleation of the Al₂Cu precipitates. Additionally, the difference in CTE between the Al matrix and TiB₂ causes a strain field between the particles and the matrix, which leads to an increase in dislocation density [46]. These dislocations can also act as nucleating sites during precipitation, which would enhance the hardness of the composites.

Generally, the hardening mechanisms of aluminum alloys can be described by the Hall–Petch hardening, solid solution hardening, and Orowan hardening [47, 48]. The grain refinement due to the TiB₂ particles and the distribution of fine TiB₂ particles in the matrix contributed to the enhancement in the hardness of the developed composite samples. Furthermore, the refinement was achieved by processing the composite in the semi-solid temperature region of the alloy. In the case of the solutionized composites, in addition to the Hall–Petch hardening and hardening due to particles, the hardness is also governed by solid solution strengthening. The hardness and tensile properties of the as-cast and peak-aged composites and

the base alloys were compared and compiled in Table 4. It is observed that there is a significant increase in tensile properties for the peak-aged condition.

3.2.2 Tensile properties

The YS of the unreinforced alloy increased from 109 (as-cast condition) to 182 MPa (peak-aged condition, solution treated at 490 °C for 2.5 h). Whereas, for the reinforced alloy, which exhibited maximum strength (RC-2), the YS increased from 120 to 230 MPa. The UTS of the unreinforced base alloy increased from 195 MPa (as-cast condition) to 254 MPa (peak-aged condition, solution treated at 490 °C for 2.5 h). Whereas, for the RC-2, the UTS increased from 285 to 342 MPa. However, the % EL of the unreinforced base alloy decreased from 5.2 (as-cast condition) to 3.4% (peak-aged condition, solution treated at 490 °C for 2.5 h) and for the RC-2, it has decreased from 12.9 to 6.2%.

The increase in the YS and UTS for peak-aged base alloy is 60% and 52%, respectively, and 77% and 83% for the RC-2, against their as-cast conditions. However, the % EL has decreased by 34 and 51% for the unreinforced alloy and RC-2, respectively, compared with their as-cast conditions. The YS increment is attributed to grain refinement level achieved due to TiB₂ particles and processing route. Figure 12 shows the image of fractured specimens after the tensile test. Figure 13 shows the stress–strain diagram of the peak-aged alloy and composite specimens.

The addition of TiB₂ particles altered the microstructure and enhanced the mechanical properties of Al alloys [49–51]. TiB₂ particles modified the dendritic structure of AA7075 [49] and also refined the grain structure of Mg2Si/Al composite [51]. In the current research, as shown under Sect. 3.1.2, the morphology of the grains was globular, and the grains were refined significantly after the addition of TiB₂ particles. As shown in Table 4, significant enhancement in mechanical properties such as microhardness, YS, and UTS was achieved after adding 2 wt% TiB₂.

Further increase in the particles resulted in lesser mechanical properties due to agglomeration and increased porosity in the composite. The increased porosity was due to particle clusters that hinders the molten metal flow, and gas entrapment during the stirring stage. Similarly, the addition of TiB₂ particles beyond 1.5 vol% in AA356 composite exhibited lesser mechanical properties. The mechanical properties of AA2024/TiB₂ composite enhanced significantly compared to the unreinforced alloy. Similarly, for Mg2Si/Al composite, the mechanical properties such as modulus of elasticity, YS, and % EL were enhanced significantly, compared to the unreinforced composite [51].

Fig. 8 Magnified FE-SEM images showing some intermetallics **a** TiB_2 , **b** EDS spectrum of spot 1 in inset of image 'a', with approximate compositions of Ti and B elements, **c** Mg_2Si , **d** EDS spectrum of selected area 1 in inset of image (c), with approximate compo-

sitions of Mg and Si elements, **e** Al_3Ti , **f** EDS spectrum of spot 1 in inset of image (e), with approximate compositions of Al and Ti elements

3.3 The effect of solution heat treatment on aging behavior

The change in hardness vs. time for the unreinforced and reinforced samples solution treated at 490 °C for the duration of 1 h, 2.5 h, and 5 h, respectively, is shown in Fig. 11a–c. The time to reach the peak hardness has decreased (the aging kinetics accelerated) for the rheocast samples, which were solution treated for 2.5 h. It can be supposed that a complete dissolution of a precipitate

is impossible for the rheocast samples, which were solution treated for 1 h due to insufficient solution treatment time. It was proved by observing the remained precipitates on the samples treated for 2.5 h (Fig. 11b), which shows a lesser amount of precipitates in comparison with the first, treated for 1 h (Fig. 11a). We can conclude that the lower aging kinetics of the composite treated for 1 h is due to the undissolved precipitates. It can lead to alloying elements reduction in the base matrix, which would reduce

Fig. 9 SEM images of RC-2 solution treated at 490 °C for **a** 1 h, **b** 2.5 h, **c** 5 h, and **d** EDS analysis of the precipitates

Fig. 10 XRD pattern of unaged and peak-aged RC-2

the supersaturation of solutes [35], reducing the precipitates, which cause slower aging kinetics.

For solution treated rheocast samples at 2.5 h, there is a considerable amount of time, which is enough for the precipitates to dissolve. Therefore, an increase in the aging kinetics was observed. From Fig. 11a, b, it can be observed that the peak hardness has increased for both unreinforced and reinforced samples (solution treated at 2.5 h), compared with those solution treated for 1 h. It was due to more precipitates in the sample treated at 2.5 h. In the case of samples solution treated at 1 h, the undissolved precipitates are more. It decreased the amount of hardening precipitates formation during aging, compared to the aging treatment made after solution treatment at 2.5 h.

The hardness of unaged unreinforced and reinforced composite samples treated at 2.5 h decreased, compared

Fig. 11 Microhardness of the developed alloys and composite as a function of aging time, solution treated at **a** 1 h, **b** 2.5 h, and **c** 5 h

Table 4 Mechanical properties of the alloy and the composites

No.	Designation	Mechanical properties							
		As-cast				Peak aged			
		UTS (Mpa)	YS (Mpa)	Hardness (HV)	% EL	UTS (Mpa)	YS (Mpa)	Hardness (HV)	% EL
1	RC-1	195	109	99	5.2	254	182	120	3.4
2	RC-2	285	120	116	12.9	342	230	147	6.2
3	RC-3	236	116	109	7.35	311	214	141	4.5
4	RC-4	180	108	107	5.36	215	146	135	2.9

with the unaged samples treated at 1 h. The decrease in hardness was due to an increase in grain size and a decrease in the undissolved precipitates for the rheocast samples treated at 2.5 h. Therefore, we can conclude that

an increase in solution treatment time (without aging) contributes to more dissolution of precipitates and accelerates the aging kinetics. However, the hardness of the alloys and composite decreased. From Fig. 11a–c, we observed that

Fig. 12 Fractured specimens after the tensile test

Fig. 13 Stress-strain diagram of peak-aged alloy and composite samples

the time required to reach maximum hardness for both the base alloy and composite samples was 9, 6, and 3 h for the samples solutionized at 490 °C for the duration of 1, 2.5, and 5 h, respectively. In the current work, the aging behavior of the unreinforced and reinforced composite was not studied independently. However, it was generally reported that the aging kinetics of the composites to be faster than that of the base matrix [1], due to the generation of high dislocation at the interface of the matrix and reinforcement particles. Dislocation formed between the interfaces acts as heterogeneous nucleation sites for the precipitates, which subsequently contribute to aging kinetics acceleration. When quenching the reinforced alloy from a high temperature (solution treatment temperature), the mismatch strain evolves due to the difference in CTE between the reinforcement and the base matrix.

We can calculate this thermal strain using the following equation Eq. (3) [35]:

$$\varepsilon_{th} = \Delta\alpha\Delta T \quad (3)$$

where $\Delta\alpha$ is the difference between the CTE of the reinforcement and the base matrix, and ΔT is the change in temperature.

Due to strains formed by thermal difference between the particles and the base alloy, plastic deformation evolves, which generates a higher dislocation density. This dislocation density (ρ) can be calculated using the following equation Eq. (4) [35]:

$$\rho = \frac{BV_f\varepsilon}{bt(1-f)} \quad (4)$$

where B =constant (4 for reinforcement aspect ratio ∞ , 12 for reinforcement aspect ratio 1), ε =thermal strain [Eq. (3)], V_f is the particle volume fraction, b is the Burgers vector, t is the smallest dimension of the reinforcement. During aging treatment, it was reported that the dislocations formed due to thermal mismatch are the sites for heterogeneous nucleation of the precipitates increasing the aging kinetics [52].

3.4 Fractography analysis

The failure mode of AA2024/2 wt% TiB₂ composite, solution treated at 1 h and 2.5 h (unaged and peak-aged conditions), was evaluated by FE-SEM and shown in Fig. 14. Figure 14a shows the fractured image of the unaged tensile test specimen solution treated at 1 h. As seen on the microstructure, a small number of large dimples, which shows a typical ductile fracture mode, as well as brittle fracture mode, were observed. A closer observation shows more undissolved precipitates, as described earlier in Fig. 9a. A large number of fine and deeper dimples were observed on the fractured surface of the peak-aged sample, as shown in Fig. 14b. The dimples show a typical ductile fracture mode observed after aging. Figure 14c, d shows the fractured image of unaged and peak-aged sample, respectively, with an extended solution treatment time to 2.5 h. When we closely look at the image, small dimples were observed on the fractography of the unaged sample. Micro-crack was also observed around the precipitates. The concentration of the precipitate in this sample is small compared to the sample solution treated at 1 h (Fig. 14a). It is in a good agreement with Fig. 9b, which shows that the solution treatment time is enough to dissolve the remaining precipitates. Large and fine dimples, as well as shallow dimples, were observed on the fracture surface of peak-aged sample, as shown in Fig. 14d. The fracture mode, in this case, is ductile as well. The ductility in the developed samples retained due to aging at

Fig. 14 SEM images showing fracture surfaces of RC-2, solution treated at 490 °C for 1 h **a** (unaged), **b** solution treated at 490 °C for 1 h (peak aged), **c** solution treated at 490 °C for 2.5 h (unaged), and **d** solution treated at 490 °C for 2.5 h (peak aged)

optimum conditions. This is confirmed by observing the fractured surfaces of the aged samples, which shows more number of dimples (Fig. 14b, d), compared to the unaged samples (Fig. 14a, c). The evolution of fine dimples in the peak-aged condition Fig. 14b, d shows the evolution of fine grains, compared to the unaged specimen [53]. Generally, at the optimum peak-aged conditions, the fracture mode of the samples shows the ductile type of fracture as confirmed by the formation of more number of large and fine dimples compared to the unaged conditions.

Generally, after adding TiB_2 particles to the base alloy, the mechanical properties of AA2024, such as hardness, YS, and UTS, have been considerably enhanced. Furthermore, the heat treatment applied to age hardenable AA2024 enhanced the mechanical properties. In this study, optimized solution heat treatment at 490 °C for 2.5 h followed by artificial aging at 190 °C for 6 h substantially improved mechanical properties such as hardness, YS, and UTS but reduced %EL. The primary precipitate found in the age-hardened AA2024 alloy was CuAl_2 , as analyzed by XRD and EDS. The optimum amount of precipitates observed in the samples, optimized at solution treatment time followed by artificial aging, contributed to strengthening mechanisms.

Previous researches focused on the effect of the reinforcement particles on the aging behavior of Al-based metal matrix composites. This study, however, provided a comprehensive idea about the effect of solution heat treatment on the aging behavior of Al-based composites. Accordingly, solution treatment time was optimized to provide an adequate amount of precipitates in the composites that contribute to enhancing strength. Therefore, the primary benefit of the strategy followed in this study was to provide a thorough understanding of the heat treatment conditions that would improve the mechanical properties of Al-based composites.

4 Conclusions

In this research, AA2024/ TiB_2 composite was successfully developed by stir rheocasting. The developed specimens were subjected to solution heat treatment followed by aging, and the following conclusions were drawn from the study.

- The FE-SEM microstructure showed a fairly uniform distribution of the TiB_2 particles in the composite up to 4 wt%, but the particles agglomerated in the case of the 6 wt% TiB_2 -reinforced composite.
- The presence of the TiB_2 particles in the matrix improved the mechanical properties of the developed composites.
- The optical microscopy image showed the microstructure of the developed composite is non-dendritic, which was attained by processing in the semi-solid temperature region of the alloy. Therefore, this suggests that the processing route followed was effective for synthesizing aluminum matrix composites.
- The optimum solution heat treatment time was 2.5 h for both the unreinforced and the reinforced matrix that led to the fastest aging kinetics and peak hardness. Increasing solution heat treatment time has accelerated the aging kinetics of the developed alloy and composites, but hardness decreased after the optimum solution heat treatment time. The observed difference in age hardening behavior after solution treatment at different time durations was attributed to the amount of the solutes in solid solution after solution treatment.
- The main precipitate observed in the peak-aged condition for the analyzed AA2024/2 wt% TiB_2 composite was CuAl_2 precipitate.
- The yield strength of the unreinforced base alloy increased from 109 (as-cast condition) to 182 MPa (peak-aged condition, solution treated at 2.5 h). Whereas, for the reinforced alloy, which exhibited maximum strength (AA2024/2 wt% TiB_2), the yield strength increased from 120 (as-cast condition) to 230 MPa (peak-aged condition, solution treated at 2.5 h).
- The ultimate tensile strength of the unreinforced base alloy has increased from 195 (as-cast condition) to 254 MPa (peak-aged condition, solution treated at 2.5 h). Whereas, for the reinforced alloy, which exhibited maximum strength (AA2024/2 wt% TiB_2), the ultimate tensile strength increased from 285 to 342 MPa (solution treated at 2.5 h).

Compliance with ethical standards

Conflict of interest There is no conflict of interest.

References

1. Abarghouie SM, Reihani SS (2010) Aging behavior of a 2024 Al alloy-SiCp composite. *Mater Des* 31:2368–2374. <https://doi.org/10.1016/j.matdes.2009.11.063>
2. Vembu V, Ganesan G (2015) Heat treatment optimization for tensile properties of 8011 Al/15% SiCp metal matrix composite using response surface methodology. *Def Technol* 11:390–395. <https://doi.org/10.1016/j.dt.2015.03.004>
3. Nagaraj N, Mahendra KV, Nagaral M (2018) Investigations on mechanical behaviour of micro graphite particulates reinforced Al-7Si alloy composites. In: IOP conference series: mater sci eng 310:012131. IOP Publishing
4. Akhlaghi F, Lajevardi A, Maghanaki HM (2004) Effects of casting temperature on the microstructure and wear resistance of compocast A356/SiCp composites: a comparison between SS and SL routes. *J Mater Process Technol* 155:1874–1880. <https://doi.org/10.1016/j.jmatprotec.2004.04.328>
5. Ghandvar H, Farahany S, Idris MH (2018) Effect of wettability enhancement of SiC particles on impact toughness and dry sliding wear behavior of compocasted A356/20SiCp composites. *Tribol Trans* 61:88–99. <https://doi.org/10.1080/10402004.2016.1275902>
6. Hoziefa W, Toschi S, Ahmed MMZ, Morri A, Mahdy AA, Seleman MES, Atlam A (2016) Influence of friction stir processing on the microstructure and mechanical properties of a compocast AA2024- Al_2O_3 nanocomposite. *Mater Des* 106:273–284. <https://doi.org/10.1016/j.matdes.2016.05.114>
7. Kumar BA, Murugan N (2012) Metallurgical and mechanical characterization of stir cast AA6061-T6-AlNp composite. *Mater Des* 40:52–58. <https://doi.org/10.1016/j.matdes.2012.03.038>
8. Kalaiselvan K, Murugan N, Parameswaran S (2011) Production and characterization of AA6061- B_4C stir cast composite. *Mater Des* 32:4004–4009. <https://doi.org/10.1016/j.matdes.2011.03.018>
9. Urena A, Martinez EE, Rodrigo P, Gil L (2004) Oxidation treatments for SiC particles used as reinforcement in aluminum matrix composites. *Compos Sci Technol* 64:1843–1854. <https://doi.org/10.1016/j.compscitech.2004.01.010>
10. Sahin Y (2003) Preparation and some properties of SiC particle reinforced aluminum alloy composites. *Mater Des* 24:671–679. [https://doi.org/10.1016/S0261-3069\(03\)00156-0](https://doi.org/10.1016/S0261-3069(03)00156-0)
11. Ramesh CS, Keshavamurthy R, Channabasappa BH, Ahmed A (2009) Microstructure and mechanical properties of Ni-P coated Si3N4 reinforced Al6061 composites. *Mater Sci Eng A* 502:99–106. <https://doi.org/10.1016/j.msea.2008.10.012>
12. Gladston JAK, Sheriff NM, Dinaharan I, Raja Selvam JD (2015) Production and characterization of rice husk ash particulate reinforced AA6061 aluminum alloy composites by compocasting. *T Nonferrous Metals Soc China (English Edition)* 25:683–691. [https://doi.org/10.1016/S1003-6326\(15\)63653-6](https://doi.org/10.1016/S1003-6326(15)63653-6)
13. Raja Selvam JD, Robinson Smart DS, Dinaharan I (2013) Microstructure and some mechanical properties of fly ash particulate reinforced AA6061 aluminum alloy composites prepared by compocasting. *Mater Des* 49:28–34. <https://doi.org/10.1016/j.matdes.2013.01.053>
14. Gladston JAK, Dinaharan I, Sheriff NM, Selvam JDR (2017) Dry sliding wear behavior of AA6061 aluminum alloy composites reinforced rice husk ash particulates produced using compocasting. *J Asian Cer Soc* 5:127–135. <https://doi.org/10.1016/j.jascer.2017.03.005>
15. Agarwal M, Srivastava R (2016) Influence of solid fraction casting on the microstructure of aluminum alloy 6061. *Mater Manuf Process* 31:1958–1967. <https://doi.org/10.1080/10426914.2015.1127956>
16. Sajjadi SA, Torabi Parizi M, Ezatpour HR, Sedghi A (2012) Fabrication of A356 composite reinforced with micro and nano Al_2O_3 particles by a developed compocasting method and study of its properties. *J Alloys Compd* 511:226–231. <https://doi.org/10.1016/j.jallcom.2011.08.105>
17. Ourdjini A, Chew KC, Khoo BT (2001) Settling of silicon carbide particles in cast metal matrix composites. *J Mater*

- Process Technol 116:72–76. [https://doi.org/10.1016/S0924-0136\(01\)00843-3](https://doi.org/10.1016/S0924-0136(01)00843-3)
18. Samuel AM, Gotmare A, Samuel FH (1995) Effect of solidification rate and metal feedability on porosity and SiC/Al₂O₃ particle distribution in an Al–Si–Mg (359) alloy. *Compos Sci Technol* 53:301–315. [https://doi.org/10.1016/0266-3538\(95\)00001-1](https://doi.org/10.1016/0266-3538(95)00001-1)
 19. Giroto FA, Albingre LO, Quenisset JM, Naslain RO (1987) Rheocasting Al matrix composites. *JOM* 39(11):18–21. <https://doi.org/10.1007/BF03257530>
 20. Mazahery A, Shabani MO (2013) Microstructural and abrasive wear properties of SiC reinforced aluminum-based composite produced by compocasting. *T Nonferrous Metals Soc China (English Edition)* 23:1905–1914. [https://doi.org/10.1016/S1003-6326\(13\)62676-X](https://doi.org/10.1016/S1003-6326(13)62676-X)
 21. Xing B, Hao Y, Li YD, Ma Y, Chen TJ (2013) Microstructure control of AZ31 alloy by self-inoculation method for semisolid rheocasting. *T Nonferrous Metals Soc China (English Edition)* 23:567–575. [https://doi.org/10.1016/S1003-6326\(13\)62501-7](https://doi.org/10.1016/S1003-6326(13)62501-7)
 22. Chen KC, Chao CG (1995) Effect of d alumina fibers on the aging characteristics of 2024-based metal-matrix composites. *Metall Trans A* 26:1035–1043. <https://doi.org/10.1007/BF02670599>
 23. Bekheet NE, Gadelrab RM, Salah MF, Abdel-Azim AN (2002) The effects of aging on the hardness and fatigue behavior of 2024 Al alloy/SiC composites. *Mater Des* 23:153–159. [https://doi.org/10.1016/S0261-3069\(01\)00072-3](https://doi.org/10.1016/S0261-3069(01)00072-3)
 24. Nieh TG, Karlak RF (1984) Aging characteristics of B4C-reinforced 6061 aluminum. *Scripta Metall* 18:25–28. [https://doi.org/10.1016/0036-9748\(84\)90083-8](https://doi.org/10.1016/0036-9748(84)90083-8)
 25. Christman T, Suresh S (1988) Microstructural development in an aluminum alloy–SiC whisker composite. *Acta Metall* 36:1691–1704. [https://doi.org/10.1016/0001-6160\(88\)90236-2](https://doi.org/10.1016/0001-6160(88)90236-2)
 26. Suresh S, Christman T, Sugimura Y (1989) Accelerated aging in cast Al alloy–SiC particulate composites. *Scripta Metall* 23:1599–1602. [https://doi.org/10.1016/0036-9748\(89\)90136-1](https://doi.org/10.1016/0036-9748(89)90136-1)
 27. Mandal A, Maiti R, Chakraborty M, Murty BS (2004) Effect of TiB₂ particles on the aging response of Al–4Cu alloy. *Mater Sci Eng A* 386:296–300. <https://doi.org/10.1016/j.msea.2004.07.026>
 28. Lu L, Lai MO, Su Y, Teo HL, Feng CF (2001) In situ TiB₂ reinforced Al alloy composites. *Scripta Mater* 45:1017–1023. [https://doi.org/10.1016/S1359-6462\(01\)01128-9](https://doi.org/10.1016/S1359-6462(01)01128-9)
 29. Li W, Long JP, Jing S, Shen BL, Gao SJ, Tu MJ (2003) Aging characteristics of short mullite fiber reinforced Al–4.0Cu–1.85 Mg metal matrix composite. *J Mater Eng Perform* 12:19–22. <https://doi.org/10.1361/1059949037>
 30. Janowski GM, Pletka BJ (1995) The effect of particle size and volume fraction on the aging behavior of a liquid-phase sintered SiC/aluminum composite. *Metall Mater Trans A* 26:3027–3035. <https://doi.org/10.1007/BF02669659>
 31. Cottu JP, Coudere JJ, Viguier B, Bernard L (1992) Influence of SiC reinforcement on precipitation and hardening of a metal matrix composite. *J Mater Sci* 27:3068–3074. <https://doi.org/10.1007/BF01154120>
 32. Thomas MP, King JE (1994) Comparison of the aging behavior of PM 2124 Al alloy and Al–SiC p metal-matrix composite. *J Mater Sci* 29:5272–5278. <https://doi.org/10.1007/BF01171535>
 33. Dutta I, Bourell DL (1989) A theoretical and experimental study of aluminum alloy 6061–SiC metal matrix composite to identify the operative mechanism for accelerated aging. *Mater Sci Eng A* 112:67–77. [https://doi.org/10.1016/0921-5093\(89\)90345-6](https://doi.org/10.1016/0921-5093(89)90345-6)
 34. Appendino P, Badini C, Marino F, Tomasi A (1991) 6061 aluminum alloy–SiC particulate composite: a comparison between aging behavior in T4 and T6 treatment. *Mater Sci Eng A* 135:275–279. [https://doi.org/10.1016/0921-5093\(91\)90575-8](https://doi.org/10.1016/0921-5093(91)90575-8)
 35. Pal S, Mitra R, Bhanuprasad VV (2008) Aging behavior of Al–Cu–Mg alloy–SiC composites. *Mater Sci Eng A* 480:496–505. <https://doi.org/10.1016/j.msea.2007.07.072>
 36. Daoud A, Reif W (2002) Influence of Al₂O₃ particulate on the aging response of A356 Al-based composites. *J Mater Process Technol* 123:313–318. [https://doi.org/10.1016/S0924-0136\(02\)00103-6](https://doi.org/10.1016/S0924-0136(02)00103-6)
 37. Abdel-Azim AN, Shash Y, Mostafa SF, Younan A (1995) Aging behavior of 2024–Al alloy reinforced with Al₂O₃ particles. *J Mater Proc Tech* 55:140–145. [https://doi.org/10.1016/0924-0136\(95\)01946-4](https://doi.org/10.1016/0924-0136(95)01946-4)
 38. Flemings MC (1991) Behavior of metal alloys in the semi-solid state. *Metall Trans A* 22:957–981. <https://doi.org/10.1007/BF02661090>
 39. ASTM E8/E8 M, Standard Test Methods for Tension Testing of Metallic Materials, E8/E8 M, 2016, pp 1–30. (http://dx.doi.org/10.1520/E0008_E0008M-16A)
 40. Chen F, Chen Z, Mao F, Wang T, Cao Z (2015) TiB₂ reinforced aluminum based in situ composites fabricated by stir casting. *Mater Sci Eng A* 625:357–368. <https://doi.org/10.1016/j.msea.2014.12.033>
 41. Natori K, Utsunomiya H, Tanaka T (2017) Improvement in formability of semi-solid cast hypoeutectic Al–Si alloys by equal-channel angular pressing. *J Mater Process Technol* 240:240–248. <https://doi.org/10.1016/j.jmatprotec.2016.09.022>
 42. Lu L, Thong KK, Gupta M (2003) Mg-based composite reinforced by Mg₂Si. *Compos Sci Technol* 63:627–632. [https://doi.org/10.1016/S0266-3538\(02\)00203-8](https://doi.org/10.1016/S0266-3538(02)00203-8)
 43. Zhang J, Fan Z, Wang YQ, Zhou BL (2000) Microstructural development of Al–15 wt% Mg₂Si in situ composite with mischmetal addition. *Mater Sci Eng A* 281:104–112. [https://doi.org/10.1016/S0921-5093\(99\)00732-7](https://doi.org/10.1016/S0921-5093(99)00732-7)
 44. Li C, Wu YY, Li H, Liu XF (2011) Morphological evolution and growth mechanism of primary Mg₂Si phase in Al–Mg₂Si alloys. *Acta Mater* 59:1058–1067. <https://doi.org/10.1016/j.actamat.2010.10.036>
 45. Li J, Zhao G, Wu S, Huang Z, Lü S, Chen Q, Li F (2019) Preparation of hybrid particulates SiCnp and Mg₂Si reinforced Al–Cu matrix composites. *Mater Sci Eng A* 751:107–114. <https://doi.org/10.1016/j.msea.2019.02.076>
 46. Varma SK, Ponce J, Solis M, Andrews S, Salas D (1996) The control of grain size and distribution of particles in a (6061 alloy m/ (Al₂O₃) P composite by solutionizing treatment. *Metall Mater Trans A* 27:2023–2034. <https://doi.org/10.1007/BF02651951>
 47. Wilkes DMJ, Jones H (1999) Structure and properties of rapidly solidified Al-rich Al–Mn–Si alloys Part I Melt spun ribbons. *J Mater Sci* 34:735–747. <https://doi.org/10.1023/A:1004516728379>
 48. Wang Y, Zhang Z, Wang W, Bian X (2004) Microstructural evolution and microhardness of a melt-spun Al–5Ti–1B alloy during annealing. *Mater Sci Eng A* 366:17–24. <https://doi.org/10.1016/j.msea.2003.08.055>
 49. Rajan HM, Ramabalan S, Dinaharan I, Vijay SJ (2013) Synthesis and characterization of in situ formed titanium diboride particulate reinforced AA7075 aluminum alloy cast composites. *Mater Des* 44:438–445. <https://doi.org/10.1016/j.matdes.2012.08.008>
 50. Akbari MK, Baharvandi HR, Shirvanimoghaddam K (2015) Tensile and fracture behavior of nano/micro TiB₂ particle reinforced casting A356 aluminum alloy composites. *Mater Des* 66:150–161. <https://doi.org/10.1016/j.matdes.2014.10.048>
 51. Du R, Yuan D, Li F, Zhang D, Wu S, Lü S (2019) Effect of in situ TiB₂ particles on microstructure and mechanical properties of Mg₂Si/Al composites. *J Alloy Compd* 776:536–542. <https://doi.org/10.1016/j.jallcom.2018.10.301>

52. Guo J, Yuan X (2009) The aging behavior of SiC/Gr/6061 Al composite in T4 and T6 treatments. *Mater Sci Eng A* 499:212–214. <https://doi.org/10.1016/j.msea.2007.11.142>
53. Kumar A, Ghosh S, Mula S (2016) A Simultaneous improvement of strength, ductility, and corrosion resistance of Al2024 alloy processed by cryoforging followed by aging. *Mater Sci Eng A* 651:774–785. <https://doi.org/10.1016/j.msea.2015.11.032>

Publisher's Note Springer Nature remains neutral with regard to jurisdictional claims in published maps and institutional affiliations.