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A study on the interaction between a Sr modifier and an Al-5Ti-1B grain refiner in an Al-7Si-0.35Mg casting alloy



Ji-Young Lee ^{a,b}, Jung-Moo Lee ^a, Kwang-Suk Son ^c, Jae-il Jang ^b, Young-Hee Cho ^{a,*}

^a Metallic Materials Department, Korea Institute of Materials Science, Changwon 51508, Republic of Korea

^b Division of Materials Science and Engineering, Hanyang University, Seoul 04763, Republic of Korea

^c Department of Materials Science and Engineering, Dong-A University, Busan 49315, Republic of Korea

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ABSTRACT

A Sr modifier and an Al-Ti-B grain refiner were simultaneously added to an Al-7Si-0.35Mg cast alloy and the interaction between Sr and Al-5Ti-1B was investigated. The combined additions of 100 ppm Sr and 500 ppm Ti produced both eutectic Si modification and α -Al grain refinement while increasing the nucleation frequency of Al-Si eutectic grains compared to an alloy with the addition of Sr alone. TEM analysis revealed that Al₂Si₂Sr phases existed with either TiB₂ or AlP and TiB₂ coupling particles. This suggests that pre-existing TiB₂ particles in the melt form prolific AlP, the most potent nuclei for Si, and thereby promote the nucleation of the eutectic Al-Si grains during solidification despite the deactivation of AlP consumed for the preceding formation of Al₂Si₂Sr.

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1. Introduction

Hypoeutectic Al-Si alloys are widely used for light metal casting in the industry due to their excellent castability and high specific strength. The as-cast structure generally contains coarse columnar primary α -Al and brittle flake-like eutectic Si. Therefore, grain refinement and eutectic modification are being attempted industrially by the addition of master alloys containing inoculant particles and by the addition of trace elements to further improve the mechanical properties [1–3]. Al-Ti-B system master alloys are best known for the grain refiner since TiB₂ particles coved with two-dimensional compound Al₃Ti in the master alloys can act as active nucleation sites for α -Al [4,5]. In addition, considering that both nucleation and growth affect the grain refinement, Ti, which has a high growth restriction factor, provides constitutional undercooling to restrict the grain growth and facilitate nucleation in addition to forming potent nucleants (TiB₂ and Al₃Ti) [6–8]. Modification of Si is commonly approached by minor addition of Sr that transforms the eutectic Si from a coarse flake-like morphology to a fine fibrous form. The modification phenomena have been extensively studied and the responsible mechanisms can be categorized according to the role of Sr in nucleation [9–14] and growth [14–18] of eutectic Si. The growth

E-mail address: y.cho@kims.re.kr (Y.-H. Cho).

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mechanism has been widely accepted as impurity induced twinning (IIT) and/or the twin plane re-entrant edge (TPRE) mechanisms, suggesting that Sr atoms [15,16,18] and/or Sr-rich clusters [17,19,20] are adsorbed at the growth steps of Si crystals and induce the alterations in the growth of Si by the formation of a high density of twins. Besides the Sr effect on Si growth, comprehensive studies have also examined Si nucleation and reported that Sr renders potent nucleation sites (i.e., AIP) inactive and thus significantly reduces the nucleation frequency of eutectic grains. Analytical methods such as transmission electron microscopy (TEM) [10,12], extended X-ray absorption fine structure (XFAS) combined with atom probe tomography (APT) [13] confirmed that the restricted nucleation of the eutectic grains is due to the Sr poisoning effect that potent nuclei for eutectic Si, AIP are preferentially consumed to nucleate Al₂Si₂Sr phases. It has been discussed that the nucleation mechanism itself is not capable of explaining the morphological change of eutectic Si from flake to fibrous forms. Nevertheless, the less prolific nucleation by a strong interaction of modifier elements with AIP may certainly account for the refinement of eutectic Si [11,21–23].

Commercial Al-Si-Mg casting alloys often contain a substantial amount of Ti in the form of Al-Ti-B master alloys for grain refinement while modification of eutectic Si is simultaneously approached with the Sr addition. There are however conflicts over the interactions between Sr and Al-Ti-B master alloys, which are either negative or negligible. It was reported that the addition of the Al-5Ti-1B master alloy exceeding 0.82 wt% to near eutectic Al-Si alloys

^{*} Corresponding author.

modified with Sr poisons the modifying effect of Sr on eutectic Si, and the poisoning event was proposed to be related to the interaction between Sr and Ti [24]. Instead of the interaction between Sr and Ti, a strong affinity of Sr to react with B in Al-5Ti-1B was suggested to partially decrease the grain refining effectiveness, and the formation of SrB₆ by the combined additions of Al-4B [25,26] and Al-1.5Ti-1.5B [27] with Sr reduced the degree of Si modification in hypoeutectic Al-Si alloys. Such negative interactions reported in the literature mainly came from combined additions of Sr and grain refiners with high amounts of B [25–29] and rarely focused on the interactions between Sr and Al-5Ti-1B which are most widely used.

Therefore, we investigated the influence of the combined additions of Sr and Al-5Ti-1B on a hypoeutectic Al-7Si-0.35Mg alloy where the amount of Al-5Ti-1B master alloy added in normal practice is ~500 ppm Ti. We characterized the solidification structures (α -Al grain, eutectic grain, and eutectic Si) in the alloys containing Sr and Ti either individually or together. We further examined the nucleation events that affect the nucleation frequency of eutectic grains by TEM, and we discuss their role in the interactions, particularly between Sr and Ti-B.

2. Experimental

Al-7 wt% Si-0.35 wt% Mg (A356) was used as a base alloy and was melted in an electric resistance furnace at 750 °C. The additions of Sr and Ti were made by adding Al-10 wt% Sr and Al-5 wt% Ti-1 wt% B master alloys, respectively. Degassing was performed on the melt using Ar-gas for 10 min and the melt was poured into a steel mold preheated to 200 °C at a cooling rate of 2 K/s. The chemical compositions of the alloys were analyzed by inductively coupled plasma (ICP) spectroscopy and are listed in Table 1. P. of which the amount was not precisely detected by ICP, is assumed to be sufficiently present in commercial alloys [30]. For inspection of the macrostructure of the quenched samples, two samples were simultaby submerging tapered neously taken stainless CUDS $(\Phi_{upper} \times \Phi_{bottom} \times height = 35 \times 25 \times 30mm)$ into the skimmed melt. One of the samples without a thermocouple was guenched in a water bath at room temperature halfway along the eutectic reactions whilst another with a thermocouple was held for monitoring cooling curves during solidification. Metallographic samples were cut from the middle of the samples and were prepared by a standard polishing procedure. The microstructures of the specimens were observed using an optical microscope (OM) and scanning electron microscope (SEM) equipped with an energy dispersive X-ray spectrometer (EDS). The grain sizes were determined by an electron backscatter diffraction (EBSD) pattern in conjunction with field emission gun SEM (FE-SEM). Nucleant particles were further analyzed by a field emission transmission electron microscope (FE-TEM) operated at 200 kV. Thin foil TEM samples containing the particles of interest were prepared by a focused ion beam (FIB).

3. Results

3.1. Solidification structure of the A356 alloys

Fig. 1 shows a Scheil simulation of the A356 alloy with the combined addition of Sr and Ti (ATS) predicting the solidification

 Table 1

 Chemical compositions of the investigated alloys (wt%)

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Elements	Si	Mg	Fe	Ti	Sr	Al
A (Base)	6.96	0.34	0.11	0.052	< 1 ppm	Bal.
AS2	6.80	0.34	0.12	0.052	170 ppm	Bal.
AS1	6.90	0.34	0.12	0.052	80 ppm	Bal.
AT	6.76	0.33	0.11	0.094	< 1 ppm	Bal.
ATS	6.69	0.33	0.11	0.094	80 ppm	Bal.

sequence using FactSage and FTlite database [31]. The addition of Ti in the form of a Al-5Ti-1B master alloy forms TiB_2 and $Al_5Si_{14}Ti_7$ at much above melting temperature (750 °C in this work), and thus can be present in the melt even before solidification. As illustrated in Fig. 1(a), trace P is likely to form AlP, followed by the serial formation of eutectic Si and Al_2Si_2Sr , whose reaction temperature ranges are nearly overlapped (see Fig. 1(b)). It should be also noted that the AlP can be partially transformed to Mg_3P_2 in the late stage of solidification, as shown in Fig. 1(b), possibly due to the presence of Mg in the A356 alloy.

Fig. 2 shows EBSD maps and optical micrographs of the A356 alloys with the addition of different amounts of Sr and Ti. EBSD maps exhibiting grains of the A356 alloys solidified at a cooling rate of 2 K/s confirm that the grain sizes with Sr up to 200 ppm (AS2 and AS1) are on the order of 300 μ m. In practice the A356 base alloy contains 500 ppm Ti, as shown in Fig. 3. With and without Sr, the addition of another 500 ppm Ti further reduced the grain size to ~200 μ m (Fig. 1(a)). The results also indicate that the additions of Sr and Ti to the A356 alloy work well for eutectic Si modification and grain refinement, respectively, while there was no apparent negative interaction between Sr and Ti with the combined additions (ATS), as shown in Fig. 1(d). The OM images in Fig. 2 indicate that, at all the addition levels, Sr can induce the modification of eutectic Si, and its interaction with Ti has no significant effect on the modification, with a slight increase in the aspect ratio of eutectic Si (see also Fig. 3).

Fig. 4 shows cooling curves of the A356 alloys, where the addition of either 100 or 200 ppm Sr decreases both the eutectic nucleation (T_n) and growth (T_g) temperatures of the eutectic Al-Si reaction. The characteristic temperatures of the eutectic Al-Si reaction, which were determined using the same method as reported in the literature [12], are listed in Table 2, and indicate that eutectic Si modification by Sr involves a decrease in the reaction temperatures by 5–10 °C. The amount of reduction appears to be less significant in the alloy with the combined additions of Sr and Ti (ATS).

Macrographs of all the quenched cup samples and the corresponding OM images are shown in Fig. 5. It is clear that the Srmodified alloys (AS2, AS1, and ATS) display circular eutectic grains in the interior of the specimen, along with a layer of eutectic grains nucleated at the wall, similarly to reports in previous studies [11,12]. However, in the absence of Sr, the unmodified alloy has no eutectic grains visible, as they were too small and numerous to be resolved in the macrographs (Fig. 5(c)). It has been well accepted that Sr deactivates the potency of AIP particles as nuclei for the eutectic Al-Si grains, by forming Al₂Si₂Sr intermetallics directly on the AIP prior to the eutectic reaction [12]. This may account for the significant decrease in the number of eutectic grains, particularly those nucleated in the interior of the sample, along with the coarsening of the eutectic grains with the increase in Sr amount from 100 to 200 ppm, as shown in Fig. 5(a) and (b).

However, it should also be noted that the nucleation frequency of eutectic grains is not only dependent on the Sr content, but also on the Ti (more specifically Al-5Ti-1B) content. As evidenced in the macrograph in Fig. 5(d), the combined additions of Sr and Ti to the A356 alloy effectively promote the nucleation of eutectic grains, increasing the number of eutectic grains with a further size reduction. Bearing in mind that the AIP nucleants can be poisoned by the Al₂Si₂Sr formation, such an increase in the number of eutectic grains by the combined additions may be related to the availability of AIP particles still active for eutectic nucleation.

3.2. Formation of the Al2Si2Sr phase

Fig. 6 shows typical Al₂Si₂Sr phases, which are found to form on the AlP in alloys with added Sr (e.g., AS2, AS1, and ATS). Considering the alloys with 100 ppm Sr, increasing the Ti amount up to 1000 ppm significantly increased the number density of the Al₂Si₂Sr phases



Fig. 1. Scheil simulations predicting (a) phases in the early stage of solidification and (b) the full solidification sequence of ATS (500 ppm Al-5Ti-1B + 100 ppm Sr) using FactSage with a FTlite database [31].



Fig. 2. EBSD maps and OM images exhibiting grains and microstructures, respectively, in the as-cast samples: (a) AS2 (200 ppm Sr); (b) AS1(100 ppm Sr); (c) AT (500 ppm Al-5Ti-1B) and (d) ATS (500 ppm Al-5Ti-1B + 100 ppm Sr).



Fig. 3. Grain size of $\alpha\text{-Al}$ and the aspect ratio of eutectic Si measured in the investigated A356 alloys.

from 12.5 μ m⁻² (for AS1) to 60.4 μ m⁻² (for AST) while reducing the size of each phase from 3 to 4 μ m to ~1 μ m (Fig. 6(a) and (b)). More interestingly, as shown in Fig. 5(d) and (e), the finer Al₂Si₂Sr phases in the alloy with 80 ppm Sr and 940 ppm Ti (ATS) were often found to contain multiphase particles. The EDS maps in Fig. 5(f) show that



Fig. 4. Cooling curves of the A356 alloys exhibiting eutectic Al-Si reactions.

a P-rich phase is in contact with Ti-rich particles, suggesting the coexistence of AlP and TiB_2 particles. However, no direct evidence of interactions between Sr and B nor the formation of SrB_6 was observed in this work.

Table 2

Characteristic temperatures of	f eutectic Al-Si	reactions	identified in	ו cooling cur	ves as o	lepicted	in Fig.	4.
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Temperature	AS2 (200 ppmSr)	AS1 (100 ppmSr)	AT (500 ppm Al-5Ti-1B)	ATS (500 ppm Al-5Ti-1B + 100 ppm Sr)
T _n	568.2	567.4	576.3	568.8
T _{min}	566.5	565.3	575.3	567.2
T_g	570.1	569.6	575.5	570.8

Most of the effort was therefore focused on the nucleation event of Al_2Si_2Sr , particularly in the alloy with both Sr and Ti (ATS). Fig. 7(a) shows a TEM micrograph exhibiting an Al_2Si_2Sr whose exterior part is physically in contact with the Si and TiB₂ phases. More interestingly, in the central region, aggregated particles can be clearly seen which are enriched with Ti, P, and Mg, as confirmed by EDS mapping (Fig. 7(b)). This may indicate the internal particles are possibly AlP or TiB₂. We will further discuss the potency of these internal particles for the nucleation of Al_2Si_2Sr based on the crystallographic properties, in order to understand the exact interactions between Sr and Al-5Ti-1B.

4. Discussion

It is well known that the potency of nucleant particles can be evaluated based on the crystallographic misfit between a nucleant substrate and a nucleated solid. One of the best- known theories to determine the effectiveness of a substrate in promoting heterogeneous nucleation is a planar disregistry (Turnbull and Vonnegut's theory) which considers the matching of both planes and the directions of the substrate and nucleating solid according to the equation [32,33]:

$$\delta \frac{(hkl)_s}{(hkl)_n} = \sum_{i=1}^3 \frac{|(d_{[uvw]_s}^i \cos \theta) - d_{[uvw]_n}^i|/d_{[uvw]_n}^i}{3} \times 100\%$$
(1)

where (*hkl*) is a low-index plane, [*uvw*] is a low-index direction, subscripts s and n are respectively the substrate and nucleated solid. It is accepted that a disregistry value of less than 20% can accommodate the lattice strain in the nucleated phase, promoting heterogeneous nucleation [32].

Another theory is an edge-to edge model that considers atomic matching along the close-packed directions on a low energy interface [34–36]. This model suggests that an orientation relationship

(OR) between the substrate and the nucleating solid can exist when the interatomic misfit is equal to or less than 10%, while the interplanar misfit is equal to or less than 6%. Based on nucleation theory, we will evaluate the nucleation potency for each case of OR and will discuss their role in the nucleation of Al2Si₂Sr and the solidification behavior of the A356 alloy.

4.1. Interactions between Sr and Al-5Ti-1B: the nucleation of Al2Si2Sr

As observed in the TEM micrographs in Figs. 7(a) and 8(a), the Tirich particles within the Al_2Si_2Sr phase (hexagonal, P $\overline{3}m1$ (139), a = b = 4.18 Å, c = 7.41 Å) are most likely TiB₂ (hexagonal, P6/mmm (191), a = b = 3.03 Å, c = 3.23 Å) and their sizes are in a range of 50 - 120 nm. It should be emphasized that the finer TiB₂ particles are in close contact with the surrounding Al₂Si₂Sr phases. The crystallographic evidence was further proved by investigating the atomic arrangement at the interfaces (squared areas in Fig. 8(a)). High resolution (HR) TEM images in Fig. 8(b) show the TiB₂ particles are well-wetted by the Al₂Si₂Sr. More interestingly, the selected area diffraction (SAD) patterns obtained by Fast Fourier transformation (FFT) reveal the crystallographic orientation relationship (OR) between the TiB₂ and Al_2Si_2Sr (see the inset of Fig. 7(b)): (1011)[0111] Al_2Si_2Sr // $(0001)[2\overline{1}\overline{1}0]$ TiB₂. It is also clear from Fig. 8(b) that the closely packed planes, Al_2Si_2Sr (1011) and TiB_2 (0001), are parallel to each other, suggesting that the TiB₂ itself can act as an epitaxial nucleation site for Al₂Si₂Sr. Based on the HR-TEM results and the simulated atomic arrangements in Fig. 8(c), the interatomic matching along the Al_2Si_2Sr [1012] and TiB_2 [0110] is ~4.5% (Table 3), while the planar disregistry is 2.3% at the interface.

Despite the difficulty of directly observing AlP nucleating Al₂Si₂Sr [12], it was often observed that the AlP are centrally located in the Al₂Si₂Sr phases, coexisting with O, possibly due to the active reaction with water during the conventional polishing process as shown in Figs. 6, 9(a) and (b). The EDS line scan results in Fig. 9(b) reveal that



Fig. 5. Macrostructures of quenched samples and OM images exhibiting eutectic Al-Si grains: (a) AS2 (200 ppm Sr); (b) AS1(100 ppm Sr); (c) AT (500 ppm Al-5Ti-1B) and (d) ATS (500 ppm Al-5Ti-1B + 100 ppm Sr).

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Fig. 6. Low voltage SEM images of Al₂Si₂Sr phases (arrowed) observed in (a) AS1 and (b) ATS alloys. For the ATS alloy, three different types of nucleation events were observed in the Al₂Si₂Sr phases: Al₂Si₂Sr nucleating on (c) AlP, (d) TiB₂ and (e) TiB₂ +AIP (f) EDS elemental maps of the squared region in (e).



Fig. 7. (a) Bright field TEM micrograph of the typical Al₂Si₂Sr observed in ATS and (b) the corresponding EDS elemental maps.



Fig. 8. (a) TEM image showing Al₂Si₂Sr co-existing with TiB₂ in ATS and (b) HR-TEM image showing the interface between the TiB₂ and Al₂Si₂Sr with the electron diffraction patterns having[0111] and [1120] zone axes, respectively. (c)Atomic distributions on the closely packed planes of (1011) for Al₂Si₂Sr and (0001) for TiB₂ with the planar disregistry calculation.

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Table 3

Crystallographic data and candidate ORs with interatomic misfits calculated between nucleants and nucleating solids.

Case	Phase	Crystal structure	Space group	Lattice parameter	Orientation Relationship (OR)	∆ (%)*
Case 1	TiB ₂	hexagonal	P6/mmm (191)	a = b = 3.03 Å c = 3.23 Å	(0001)TiB ₂ ($\overline{1}$ 011)Al ₂ Si ₂ Sr	4.55
	Al_2Si_2Sr	hexagonal	P3m1 (139)	a = b = 4.18 Å c = 7.41 Å	$[01\overline{1}0]$ TiB ₂ $[10\overline{1}2]$ Al ₂ Si ₂ Sr	
Case 2	AlP	Cubic	F43m (216)	a = 5.43 Å	(111) AIP (0001) Al ₂ Si ₂ Sr	7.80
	Al_2Si_2Sr	hexagonal	P3m1 (139)	a = b = 4.18 Å c = 7.41 Å	[Ī10] AIP // [Ī2Ī0] Al ₂ Si ₂ Sr	
Case 3	TiB ₂	hexagonal	P6/mmm (191)	a = b = 3.03 Å c = 3.23 Å	(0001)TiB ₂ (111)AIP	9.18
	AlP	Cubic	F43m (216)	a = 5.43 Å	[1010] TiB ₂ [011] AlP (111) AlP (0001) Al ₂ Si ₂ Sr	7.80
	Al_2Si_2Sr	hexagonal	P3m1 (139)	a = b = 4.18 Å c = 7.41 Å	[Ī10] AIP // [Ī2Ī0] Al ₂ Si ₂ Sr	

* lattice misfit (Δ)= $\frac{|d_s - d_N|}{d_N} \times 100\%$ d_s is the lattice parameter of the substrate nucleants, and d_n is the lattice parameter of nucleating solids

Mg is also present in the internal AlP particle, and indicate the partial transformation of AlP into Mg_3P_2 in the final stage of solidification, as predicted by the thermodynamic calculation (Fig. 1(b)). The nucleation of Al_2Si_2Sr is therefore likely to directly occur on the AlP, not Mg_3P_2 since the crystallographic mismatch values meet the nucleation criterion (planar disregistry, 7.9% and interatomic misfit, 7.8%). Atomic configurations of Al_2Si_2Sr and AlP on the closely packed planes are illustrated in Fig. 9(c) and suggest a candidate OR as $(0001)[\bar{1}2\bar{1}0]$ Al_2Si_2Sr // (111) $[\bar{1}10]$ AlP.

Another interface of the TiB_2 as squared in Fig. 10(a) was found to contact a partially crystallized Mg_3P_2 (Cubic, $Pn\bar{3}m$ (224), a = 5.92 Å). HR-TEM images with electron diffraction patterns in Fig. 10(b)

confirm an OR between TiB₂ and Mg₃P₂ as (002)[100]Mg₃P₂ // (0001) [2110] TiB₂. This suggests that AlP (Cubic, F43m (216), a = 5.43 Å) forming onto TiB₂ is possibly transformed into Mg₃P₂ during solidification. While the AlP, if present, is generally difficult to observe due to its high oxygen affinity, the phase transformation of the AlP to Mg₃P₂ during solidification may allow the phase entrapped within the Al₂Si₂Sr to be observed, as in Fig. 10(b).

Based on the experimental observations and crystallographic misfit, we therefore propose that three possible nucleation mechanisms could be associated with the Al₂Si₂Sr formation, which significantly reduce the subsequent nucleation events of eutectic Al-Si:





Fig. 9. (a) SEM image showing Al₂Si₂Sr co-existing with Mg₃P₂ or AlP in ATS and (b) EDS line scan across the phase in (a) showing the presence of Mg, P and O in the centrally located particle (c)Atomic distributions on the closely packed planes of (0001) for Al₂Si₂Sr and (111) for AlP with the planar disregistry calculation.

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Fig. 10. (a) TEM image showing Al_2Si_2Sr co-existing with TiB₂ in ATS and (b) HR-TEM image showing the interface between the TiB₂ and Mg_3P_2 with the electron diffraction patterns having[0111] and [100] zone axes, respectively. (c)Atomic distributions on the closely packed planes of (0001) for Al_2Si_2Sr and (1111) for AlP with the planar disregistry calculation.



Fig. 11. Schematics showing the solidification sequence of the A356 alloys with (a) Sr only and (b) the combined addition of Sr and Al-5Ti-1B, particularly focusing on the formation of Al₂Si₂Sr and eutectic grains.

Case 1. TiB_2 nucleates Al_2Si_2Sr prior to the eutectic Al-Si reaction (as observed in Fig. 8).

Case 2. AlP nucleates Al₂Si₂Sr prior to the eutectic Al-Si reaction (as reported in [12]).

Case 3. Serial nucleation events involving the nucleation of Al_2Si_2Sr by AlP formed on TiB₂ (AlP+TiB₂) prior to the eutectic Al-Si reaction (as observed in Fig. 10).

4.2. Interactions between Sr and Al-5Ti-1B: the solidification sequence of the A356 alloy

In the A356 alloy with both Sr and Al-5 T-1B (ATS), the increase in the frequency of Al-Si eutectic grains importantly indicates that the above three mechanisms are most likely to occur during solidification. Fig. 11 illustrates a schematic diagram of the solidification sequence in the A356 alloys with Sr and Al-5Ti-1B additions,

particularly focusing on the formation of Al₂Si₂Sr and eutectic grains. In the alloy containing only Sr (AS1 or AS2), AlP forms early and is mostly associated with the Al₂Si₂Sr formation. Fewer AlP particles are therefore available for the nucleation of eutectic grains and far fewer eutectic grains form in the interdendritic liquid (Fig. 11(a)). In the alloy with the combined additions of Sr and Al-5Ti-1B (ATS), TiB₂ particles are preferentially involved in the nucleation of α -Al and its refinement, while the remains are pushed ahead of the α -Al dendrite-liquid interface during solidification. The remaining TiB₂ particles, which are still prolific, are then partially consumed by the formation of AIP. Since the TiB₂, AIP and AIP+TiB₂ particles are active nucleants for Al₂Si₂Sr, the formation of Al₂Si₂Sr phases is likely to occur following the three mechanisms suggested above (Fig. 11(b)). This may increase the number of AIP particles available in the melt and thereby promote the nucleation of eutectic grains with reduced size.

5. Conclusions

In summary, the combined additions of Sr and Al-5Ti-1B to A356 alloys led to both eutectic Si modification and α -Al grain refinement while increasing the nucleation frequency of Al-Si eutectic grains compared to the alloy with the addition of Sr alone. TEM results with crystallographic matching confirmed that pre-existing TiB₂ particles in the Al-5Ti-1B master alloy are potent nucleant substrates for both AlP and Al₂Si₂Sr. We also observed that all of the three potent nucleants, TiB₂, AlP, and AlP+TiB₂, were likely to nucleate Al₂Si₂Sr prior to the eutectic Al-Si reaction. This may make more AlP (and AlP+TiB₂) particles available in the melt, and thus promote the nucleation of eutectic grains. Despite the increase in the frequency of the eutectic nucleation, the overall effect of Sr on the microstructure was rarely affected by the combined additions.

CRediT authorship contribution statement

Ji-Young Lee: Investigation, Visualization, Data curation and Writing – original draft. **Jung-Moo Lee:** Methodology, Supervision and Proof reading. **Kwang-Suk Son:** FIB and TEM experiment and Proof reading. **Jae-il Jang:** Supervision and Writing – reviewing. **Young-Hee Cho:** Conceptualization, Methodology, Writing – original draft, reviewing and editing, Supervision and Project administration.

Data Availability

Data will be made available on request.

Declaration of Competing 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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