Effect of Geometrical Parameters of Microscale Particles on Particle-Stimulated Nucleation and Recrystallization Texture of Al-Si-Mg-Cu-Based Alloy Sheets

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Abstract: The effects of the shapes (needle and round) and volume fractions (low and high) of microscale particles in Al-Si-Mg-Cu-based alloys on recrystallization behavior, texture evolution, mechanical properties, and formability are investigated. The recrystallized grain size decreases as the size and volume fraction of the particles decrease and increase, respectively, regardless of the particle shape. The investigated alloys with a relatively low volume fraction of 0.7 to 2.4 vol.% exhibit higher efficiency particle-stimulated nucleation (PSN) than alloys with a high volume fraction of 6.0 to 21.0 vol.%. This is because the interaction between the particles and dislocations cannot be greatly promoted when the volume fraction of the particles is large enough to form agglomerates. The sheets with round-shaped particles exhibit higher yield strength (YS) and elongation (EL) than sheets with needle-shaped particles. The improvement in YS is due to the combined effects of grain refinement and particle strengthening, and the EL is improved by reducing the probability of cracking at the tips of round-shaped particles. The sheets with round-shaped particles exhibit relatively higher average plastic strain ratio (r) and planar anisotropy (∆r) than the sheets with needle-shaped particles, owing to the development of Goss {110}<001> or rotated-Goss {110}<110> orientations.

Keywords: Al-Si-Mg-Cu alloys; microscale particle; recrystallization behavior; texture; mechanical properties

1. Introduction

There has been a considerable amount of research interest in Al alloys to help reduce the weight of automobiles and solve global environmental problems. In particular, heat-treatable Al-Si-Mg-Cu alloy sheets have received great attention for manufacturing the outer bodies of automobiles, such as hoods, side doors, and roof panels, owing to their low specific density, reasonable corrosion resistance, comparatively good formability, and high strength after the paint bake cycle [1,2]. However, the formability of Al-Si-Mg-Cu alloy sheets is lower than that of conventional steel sheets, making it difficult and expensive to manufacture automotive parts with complex shapes [3].

The recrystallization behavior and texture evolution, which are mainly determined by particle-stimulated nucleation (PSN), significantly affect the mechanical properties and formability of the final gauge sheet [4]. The highly strained regions in the vicinity of microscale particles serve as preferred nucleation sites for recrystallization, forming fine recrystallized grains with random orientations [5]. Numerous studies have shown that various microscale particles, such as eutectic Si, Mg2Si, Al7Cu2Fe, Al3CuMg, and...
β-Al(FeMnCr)Si particles, promote PSN to refine the recrystallized grains [6–8]. The microscale particles with a size suitable for PSN (>1 μm) easily induce non-uniform stored energy that drives the nucleation of recrystallized grains. When the average sizes of the microscale particles are similar, the increase in the volume fraction of microscale particles promotes the occurrence of PSN. Zhang et al. [8] reported that the Mn content increased the volume fraction of microscale β-Al(MnCrFe)Si particles, which could promote the formation of recrystallized grains. Particles with a relatively large aspect ratio have been reported to induce a large strain gradient compared to particles with a small aspect ratio [9,10], but their effect on recrystallization behavior has not been discussed in detail. Since geometrical parameters of the particles, such as the shape, size, and volume fraction, have complex effects on recrystallization behavior, texture evolution, mechanical properties, and formability, this requires further studies to clearly understand the complex effects of and expand the applicability of Al-Si-Mg-Cu alloy sheets.

In this study, four Al-Si-Mg-Cu-based alloy sheets with different shapes (needle and round) and volume fractions (low and high) of microscale particles were prepared by controlling the fabrication methods and process conditions. The effects of the shape, size, and volume fraction of the particles on recrystallization behavior, texture evolution, mechanical properties, and formability were systematically investigated.

2. Materials and Methods

Two methods of twin-roll casting (TRC) and gravity casting (GC) were used to prepare alloys to obtain round- and needle-shaped microscale particles, respectively, as shown in Figure 1. The TRC strip, whose width and thickness were 1380 mm and 5.6 mm, respectively, was manufactured by Choiil Aluminum Co. (Gyeongsan-si, Korea) with a casting speed of 850 mm/min at 710 °C. The GC alloy was fabricated by adding alloying elements to the melt at 740 °C, followed by casting into a steel mold with a thickness of 10 mm. The GC alloy was hot-rolled up to 5.6 mm at 350 °C to achieve the same total reduction ratio in cold rolling as the TRC alloy. The chemical composition of the fabricated alloys was determined using optical emission spectrometry (SPECTROCHECK, Ametek Co., Berwyn, PA, USA). Table 1 presents the chemical compositions of the TRC and GC alloys. The alloys with a relatively low and high volume fractions of microscale particles were obtained by providing process conditions with and without homogenization heat treatment (at 540 °C for 12 h), respectively. Accordingly, the investigated alloys were named NH, NL, RH, and RL according to the shapes (needle and round) and volume fractions (high and low) of the microscale particles. The alloys with a thickness of 5.6 mm were cold-rolled for up to 1 mm with a reduction of 20%. The cold-rolled sheet was recrystallized at 540 °C for 30 min and then quenched with water.

Figure 1. Schematic of the manufacturing method of NH, NL, RH, and RL sheets. The prepared GC and TRC alloys were each made into sheets using two different processes: (a) rolling without homogenization heat treatment and (b) rolling after homogenization heat treatment at 540 °C for 12 h.
Table 1. Chemical compositions (in wt.%) of the Al-Si-Mg-Cu alloy sheets.

| Samples | Si   | Mg  | Cu | Fe  | Mn  | Cr  | Al   |
|---------|------|-----|----|-----|-----|-----|------|
| NH      | 1.51 | 0.40| 0.30| 0.18| 0.10| 0.05| Bal. |
| NL      |      |     |     |      |     |     |      |
| RH      | 1.53 | 0.41| 0.29| 0.16| 0.09| 0.06|      |
| RL      |      |     |     |      |     |     |      |

The distribution and composition of the microscale particles were analyzed using a combination of scanning electron microscopy (SEM, JEOL JSM-7001F, Tokyo, Japan) and energy dispersive X-ray spectrometry (EDS, AMETEK Octane plus, Mahwah, NJ, USA). All specimens for SEM analysis were polished with up to 4000-grit SiC paper and then carefully polished using buffing and abrasion (Alumina suspension, 0.05 µm, Allied High Tech Products, Inc., Rancho Dominguez, CA, USA) with water. The crystallographic orientations of the specimens were evaluated using electron backscattered diffraction (EBSD, EDAX-TSL, Draper, UT, USA). The inverse pole figure (IPF) maps were observed with a step size of 3.5 µm. The orientation distribution functions (ODFs) were analyzed using orientation imaging microscopy (OIM) analysis software (TSL OIM analysis 7.3, EDAX Inc., Mahwah, NJ, USA). The software ImageJ (version 1.41o, National Institutes of Health, Bethesda, MD, USA) was used to calculate the size and volume fraction of the microscale particles. The average size and volume fraction of the microscale particles were calculated in three planes: rolling direction (RD), transverse direction (TD), and normal direction (ND).

Tensile tests were performed using an Instron-type tensile machine (Unitech™, R&B, Daejeon, Korea) with a strain rate of $10^{-3}$ s$^{-1}$ at 25 °C. Tensile specimens according to ASTM E8 were machined from the fabricated Al sheets. The yield strength (YS, 0.2% offset yield stress), ultimate tensile strength (UTS), elongation (EL), average plastic strain ratio ($r$), and planar anisotropy ($\Delta r$) of the recrystallized sheets in the three directions of 0°, 45°, and 90° with respect to the RD were evaluated as the average of the three measurements.

3. Results and Discussion

Figure 2a–f shows the SEM images and corresponding EDS results in the RH and RL sheets, respectively. Figure 2a–c show the needle-shaped $\beta$-AlFeSi and skeleton-shaped eutectic Mg$_2$Si particles, which are consistent with the microstructures reported in other studies [11,12]. The Fe element has high solubility in liquid Al but extremely low solubility in solid Al, unintentionally forming Fe-containing intermetallic compounds [13]. The $\beta$-AlFeSi particles are mainly formed in Al-Si-Mg-based alloys due to Fe impurities in the raw Al with a purity of 99.8% [13,14]. After homogenization heat treatment, the majority of the eutectic Mg$_2$Si particles were dissolved in Al matrix, but there are polygonal-shaped Si particles (Figure 2e) and round-shaped $\alpha$-Al(FeCrMn)Si particles (Figure 2f). According to the reported results [15–17], a homogenization heat treatment promotes the transformation of needle-shaped $\beta$-AlFeSi particles into smaller round-shaped $\alpha$-Al(FeCrMn)Si particles. Kuipers et al. [18] reported that the $\beta$-AlFeSi particles were transformed into particles denoted as $\alpha$-Al(FeMn)$_2$Si, $\alpha$-Al$_3$(FeMn)$_2$Si, and $\alpha$-Al$_2$(FeMn)$_3$Si. After homogenization heat treatment, the volume fraction of $\alpha$-Al(FeCrMn)Si particles is relatively larger than that of $\beta$-AlFeSi particles [17,19]. Kuipers et al. [18,20] suggested that $\alpha$-(FeMn)Si nucleated on the basal face of $\beta$-AlFeSi particles; $\beta$-AlFeSi particles partially dissolve and then $\alpha$-Al(FeMn)Si particles grow by consuming released Mn content. In addition, an increase of homogenization temperature can induce greater transformation of $\beta$-AlFeSi particles into $\alpha$-Al$_3$(FeMn)$_2$Si particles [16]. A high temperature can accelerate the dissolution of $\beta$-AlFeSi particles, promoting the formation of $\alpha$-Al(FeCrMn)Si particles [21].
The average particle sizes measured for the NH, NL, RH, and RL alloys are 10.0, 9.1, 7.9, and 7.9 μm, respectively. The volume fractions of Fe-containing intermetallic compounds and eutectic Mg2Si particles are shown in Figure 3b. The volume fractions of Fe-containing intermetallic compounds in NH, NL, RH, and RL alloys are 1.36, 1.09, 0.76, and 0.71 vol.%, respectively. The volume fractions of eutectic Mg2Si particles in the NH and RH alloys are 1.02 and 0.87 vol.%, respectively. The RH and RL alloys possess a relatively small size and volume fraction compared to the NH and NL alloys, owing to the presence of eutectic Mg2Si particles.

Figure 5a shows the size distribution of the microscale particles before cold rolling. The average particle sizes measured for the NH, NL, RH, and RL alloys are 10.0, 9.1, 3.6, and 2.9 μm, respectively. The volumefractions of the Fe-containing intermetallic compounds and eutectic Mg2Si particles are shown in Figure 3b. The volume fractions of Fe-containing intermetallic compounds in NH, NL, RH, and RL alloys are 1.36, 1.09, 0.76, and 0.71 vol.%, respectively. The volume fractions of eutectic Mg2Si particles in the NH and RH alloys are 1.02 and 0.87 vol.%, respectively. The RH and RL alloys possess a relatively small size and volume fraction compared to the NH and NL alloys, owing to the rapid solidification of the TRC process. After homogenization heat treatment (NL and RL alloys), the majority of the eutectic Mg2Si particles are dissolved in the Al matrix because of the high diffusivity of Mg in Al [25].

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Figure 3. SEM images of (a–c) NH and (d–f) NL alloys with a thickness of 5.6 mm before cold rolling: (a,d) RD, (b,e) ND, and (c,f) TD.

Figure 4. SEM images of (a–c) RH and (d–f) RL alloys with a thickness of 5.6 mm before cold rolling: (a,d) RD, (b,e) ND, and (c,f) TD.

Figure 6a–d show the IPF maps and ODFs of the NH, NL, RH, and RL sheets recrystallized at 540 °C for 30 min, respectively. In the IPF maps, the grain boundaries are determined using a 15° misorientation criterion and are represented by solid lines. The black dots (indicated by black arrows) with a confidence index less than 0.1 correspond to regions where the system cannot specify any orientation, indicating the microscale particles. The NH and NL sheets exhibit a larger number of particles than the RH and RL sheets because of the relatively large size and volume fraction of the particles. The average grain sizes of the NH, NL, RH, and RL recrystallized sheets are 26.7, 55.2, 19.1, and 29.1 μm, respectively. The mechanism by which the recrystallized grain size varies significantly with the shape, size, and volume fraction of the microscale particles will be described later. In
the ODFs, the strong cube $\{001\}<100>$ orientation, commonly developed in the NL and RL sheets, results from the weak PSN, owing to the relatively low volume fraction of the particles. The RH and RL sheets develop Goss $\{110\}<001>$ and rotated-Goss $\{110\}<110>$ orientations, respectively.

Figure 6. IPF maps and ODFs in the (a) NH, (b) NL, (c) RH, and (d) RL sheets recrystallized at 540 °C for 30 min. The microscale particles of the NH and NL sheets are crushed and aligned parallel to the RD, as indicated by the black arrows.
Figure 7 shows the engineering stress–strain curves of recrystallized sheets in the RD. Although the NH sheet exhibits a higher UTS than the other sheets, the highest YS is shown by the RH sheet. The RH sheet displays the discontinuous yielding phenomenon (Piobert–Lüders effect) due to fine recrystallized grains. This phenomenon, in which solute atoms act as obstacles to the dislocation movement, is more frequently encountered in fine-grained alloys [14].

The mechanical properties and formability of the recrystallized sheets corresponding to tensile directions of 0°, 45°, and 90° are summarized in Table 2. The RH and RL sheets exhibit a relatively higher YS than the NH and NL sheets, respectively, owing to the combined effects of grain boundary and particle strengthening. The grain boundary strengthening cannot significantly improve the strength in Al-Si-Mg-Cu alloys, owing to the relatively low Hall–Petch constant ($\sigma_0 = 5.5$ MPa and $k = 40$ MPa $\mu m^{1/2}$ [26]). In addition, the average EL of the NH and NL sheets is relatively lower than that of the RH and RL sheets because the needle-shaped particles easily cause the formation of cracks, owing to the stress concentration at the tips. The texture evolution and hence the $r$-value is highly dependent on the shape, size, and volume fraction of the microscale particles because the texture develops from the competition of nucleation at the cube bands, high-angle grain boundaries, and near-microscale particles [27,28]. The RH and RL sheets exhibit relatively higher $r$ and $\Delta r$ values than the NH and NL sheets, respectively, owing to the development of Goss {110}<001> or rotated-Goss {110}<110> orientations [29,30]. The Goss {110}<001> orientation is nucleated from high-angle boundaries formed by the strong interaction between dislocations and round-shaped particles with narrow inter-particle spacing [31,32]. In addition, the NH and RH sheets exhibit higher $r$ and lower $\Delta r$ values than the NL and RL sheets, respectively. This is because the more frequent occurrence of PSN in NH and RH sheets produces large numbers of grains with various orientations [33–35]. Thus, the maximum intensities of the NH and RH sheets in the ODFs are relatively smaller than those of the NL and RL sheets (Figure 6), which indicates the development of a relatively random texture. Furthermore, the NH and RH sheets show a rotated-cube {001}<110> orientation nucleated near microscale particles, whereas the NL and RL sheets show a strong cube {001}<100> orientation nucleated at the cube bands, owing to the weak occurrence of PSN [4,28,36]. The rotated-cube {001}<110> orientation exhibits a relatively higher $r$ and lower $\Delta r$ compared to the cube {001}<100> orientation [37].
Table 2. Recrystallized grain size, YS, UTS, EL, r, and ∆r of the NH, NL, RH, and RL recrystallized sheets corresponding to the tensile directions of 0, 45, and 90° with respect to the RD.

| Samples | Tensile Direction (°) | Grain Size (µm) | YS (MPa) | UTS (MPa) | EL (%) | r | ∆r |
|---------|-----------------------|-----------------|----------|-----------|--------|---|----|
| NH      | 0                     | 26.7            | 78.8     | 219.9     | 29.9   | 0.63 | 0.66 | −0.006 |
|         | 45                    |                 | 75.8     | 226.5     | 30.1   | 0.66 |                |
|         | 90                    |                 | 78.4     | 224.6     | 29.2   | 0.68 |                |
| NL      | 0                     | 55.2            | 61.8     | 210.9     | 32.0   | 0.61 | 0.58 | −0.07 |
|         | 45                    |                 | 66.4     | 211.4     | 28.9   | 0.55 |                |
|         | 90                    |                 | 66.3     | 216.6     | 29.6   | 0.63 |                |
| RH      | 0                     | 19.1            | 80.4     | 210.7     | 32.2   | 0.58 | 0.69 | −0.22 |
|         | 45                    |                 | 80.8     | 207.9     | 32.5   | 0.80 |                |
|         | 90                    |                 | 81.1     | 211.4     | 29.6   | 0.58 |                |
| RL      | 0                     | 29.1            | 73.7     | 210.8     | 31.9   | 0.91 | 0.66 | 0.29 |
|         | 45                    |                 | 74.2     | 211.3     | 31.9   | 0.51 |                |
|         | 90                    |                 | 70.5     | 205.1     | 29.8   | 0.70 |                |

The relationship between the recrystallized grain size and ratio of the size (d) to volume fraction (f) of the microscale particles is shown in Figure 8. Mikhaylovskaya, et al. [38] reported that the recrystallized grain size (D) is proportional to the d/f value, and it can be expressed as:

\[ D = k \left( \frac{d}{f} \right) + b, \]  

(1)

where b is the theoretical minimum of the recrystallized grain size. k (slope of the linear relationship) is a constant related to the sensitivity on which the recrystallized grain size depends when the size and volume fraction of the microscale particles are changed. A relatively high k value indicates that small changes in the d/f values significantly alter the recrystallized grain size. The d/f values for the NH, NL, RH, and RL alloys are 4.21, 8.35, 2.21, and 4.08, respectively, which are linearly related to their recrystallized grain sizes regardless of the particle shape. Consequently, the k value of 6.0 obtained in this study is larger than the k values of 0.4 to 0.8 reported in Mikhaylovskaya’s study [38], which can be explained by the volume fraction of microscale particles. A large volume fraction of particles with a range of 6–21 vol.% can reduce the efficiency for PSN because the particles are more likely to agglomerate than maintain a proper distance. However, in this study, a relatively small volume fraction of the particles ranging from 0.7 to 2.4 vol.% can improve the efficiency for the occurrence of PSN (i.e., high k value). More specifically, it can be demonstrated by considering the concept of inter-particle spacing. The inter-particle spacing (λ) is estimated using the following equation [39]:

\[ \lambda = \frac{2}{3}d \left( \frac{1}{f} - 1 \right). \]  

(2)

As the particle size decreases and their volume fraction increases, the inter-particle spacing decreases. The narrow inter-particle spacing causes strong interactions between the particles and dislocations by the Orowan looping model, forming a large strain gradient. Regions with large strain gradients near the microscale particles serve as preferred nucleation sites for recrystallization, refining the recrystallized grains [40,41]. However, when the volume fraction of microscale particles is large enough to form agglomerates, the interaction between the particles and dislocations (i.e., the occurrence of PSN) cannot be greatly promoted. Therefore, in this study, the Al-Si-Mg-Cu alloys with a low volume fraction in the range of 0.7 to 2.4 vol.% have a large k value of 6.0, owing to the high-efficiency PSN by uniformly distributed particles. Consequently, the k value can significantly vary when the range of volume fractions of the particles is different, even in a similar alloy system, as shown in Figure 8.
Al-4.8Si-8.2Mg and Al-14Cu-7Fe alloys were reported by Mikhaylovskaya, et al. [38].

According to a previous study [43], a higher dislocation density is accumulated at the tip of rod-shaped Al_{2}Cu particles than at that of round-shaped Al_{2}Mn particles, leading to strong lattice rotation. In addition, the particles with a large aspect ratio form a larger strain gradient than the particles with a small aspect ratio [44]. However, in this relationship, the shape of the microscale particles in the initial state has little effect on the recrystallization behavior because the needle-shaped β-AlFeSi particles with extremely brittle properties are crushed and distributed during cold rolling (Figure 6a,b). Before the stress concentration at the tip of the needle-shaped particles exceeded a certain level, the particles were crushed to release the stress concentration. Consequently, the ratio of the size to volume fraction of the microscale particles is proportional to the recrystallized grain sizes regardless of particle shape.

4. Conclusions

The ratio of the size to volume fraction of the microscale particles is proportional to the recrystallized grain sizes. However, the shape of the particles does not affect the relationship significantly because the needle-shaped β-AlFeSi particles are crushed and distributed during cold rolling. As the microscale particle size decreases and their volume fraction increases, the narrower inter-particle spacing causes strong interactions between particles and dislocations, reducing the recrystallized grain size. When the volume fraction of microscale particles is large enough to form agglomerates, the interaction between the particles and dislocations cannot be greatly promoted. Therefore, the Al-Si-Mg-Cu alloys with a relatively low volume fraction ranging from 0.7 to 2.4 vol.% exhibit a large k value of 6.0, owing to the high-efficiency PSN by uniformly distributed particles. The RH and RL sheets exhibit higher average YS and EL than the NH and NL sheets, respectively. In addition, the RH and RL sheets exhibit relatively higher r and Δr values than the NH and NL sheets. This is because the Goss [110]<001> or rotated-Goss [110]<110> orientations develop at high-angle grain boundaries formed by the narrowly spaced round-shaped particles. The NH and RH sheets with a large volume fraction of the particles exhibit fine

![Figure 8. Dependence of the recrystallized grain size on d/f in the different Al alloys. Results for Al-4.8Si-8.2Mg and Al-14Cu-7Fe alloys were reported by Mikhaylovskaya, et al. [38].](image-url)
recrystallized grains with random orientations, owing to the more frequent occurrence of PSN, resulting in an increase and decrease in \( \gamma \) and \( \Delta r \), respectively.

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**References**

1. Yuan, L.; Guo, M.; Dong, X.; Zhuang, L. Design, Evolution, Formation and Effect Mechanism of Coupling Distributed Soft and Hard Micro-Regions in Al–Zn–Mg–Cu–Fe Alloys with High Formability. *Mater. Sci. Eng. A* 2022, 855, 143951. [CrossRef]
2. Yuan, L.; Guo, M.; Habraken, A.M.; Duchene, L.; Zhuang, L. Extremely Improved Formability of Al–Zn–Mg–Cu Alloys Via Micro-Domain Heterogeneous Structure. *Mater. Sci. Eng. A* 2022, 837, 142737. [CrossRef]
3. Mahabunphachai, S.; Koç, M. Investigations on Forming of Aluminum 5052 and 6061 Sheet Alloys at Warm Temperatures. *Mater. Des.* 2010, 31, 2422–2434. [CrossRef]
4. Humphreys, F.J.; Hatherly, M. *Recrystallization and Related Annealing Phenomena*; Elsevier: Amsterdam, The Netherlands, 2012.
5. Huo, W.T.; Shi, J.T.; Hou, L.G.; Zhang, J.S. An Improved Thermo-Mechanical Treatment of High-Strength Al–Zn–Mg–Cu Alloy for Effective Grain Refinement and Ductility Modification. *J. Mater. Process. Technol.* 2017, 239, 303–314. [CrossRef]
6. Wu, Y.; Liao, H.; Liu, Y.; Zhou, K. Dynamic Precipitation of Mg\(_2\)Si Induced by Temperature and Strain During Hot Extrusion and Its Impact on Microstructure and Mechanical Properties of near Eutectic Al–Si–Mg–V Alloy. *Mater. Sci. Eng. A* 2014, 614, 162–170. [CrossRef]
7. Dumont, D.; Deschamps, A.; Brechet, Y. On the Relationship between Microstructure, Strength and Toughness in Aa7050 Aluminum Alloy. *Mater. Sci. Eng. A* 2003, 356, 326–336. [CrossRef]
8. Zhang, G.W.; Nagaumi, H.; Han, Y.; Xu, Y.; Parish, C.M.; Zhai, T.G. Effects of Mn and Cr Additions on the Recrystallization Behavior of Al-Mg-Si-Cu Alloys. In *Materials Science Forum*; Trans Tech Publications Ltd.: Bäch, Switzerland, 2017.
9. Schäfer, C.; Song, J.; Gottstein, G. Modeling of Texture Evolution in the Deformation Zone of Second-Phase Particles. *Acta Mater.* 2009, 57, 1026–1034. [CrossRef]
10. Kim, J.; Shin, S.; Lee, S. Correlation between Microstructural Evolution and Corrosion Resistance of Hypoeutectic Al–Si–Mg Alloy: Influence of Corrosion Product Layer. *Mater. Charact.* 2022, 193, 112276. [CrossRef]
11. Kumar, L.; Jang, J.C.; Yu, H.; Shin, K.S. Effects of Cr and Ti Addition on Mechanical Properties and Thermal Conductivity of Al–Si–3Mg Die-Casting Alloys. *Met. Mater. Int.* 2022, 1–11. [CrossRef]
12. Kim, H.Y.; Park, T.Y.; Han, S.W.; Lee, H.M. Effects of Mn on the Crystal Structure of A-Al (Mn, Fe) Si Particles in A356 Alloys. *J. Cryst. Growth* 2006, 291, 207–211. [CrossRef]
13. Zhang, G.W.; Nagaumi, H.; Han, Y.; Xu, Y.; Parish, C.M.; Zhai, T.G. Effects of Mn and Cr Additions on the Recrystallization Behavior of Al-Mg-Si-Cu Alloys. In *Materials Science Forum*; Trans Tech Publications Ltd.: Bäch, Switzerland, 2017.
14. Kim, H.Y.; Park, T.Y.; Han, S.W.; Lee, H.M. Effects of Mn on the Crystal Structure of A-Al (Mn, Fe) Si Particles in A356 Alloys. *J. Cryst. Growth* 2006, 291, 207–211. [CrossRef]
15. Wang, M.; Guo, Y.; Wang, H.; Zhao, S. Characterization of Refining the Morphology of Al–Fe–Si in A380 Aluminum Alloy Due to Ca Addition. *Processes* 2022, 10, 672. [CrossRef]
16. Alvarez-Antolin, F.; Asensio-Lozano, J.; Cofiño-Villar, A.; Gonzalez-Pociño, A. Analysis of Different Solution Treatments in the Transformation of B-Alfesi Particles into A-(Femn) Si and Their Influence on Different Ageing Treatments in Al–Mg–Si Alloys. * Metals* 2020, 10, 620. [CrossRef]
17. Kumar, S.; Grant, P.; O’Reilly, K.A.Q. Evolution of Fe Bearing Intermetallics During Dc Casting and Homogenization of an Al-Mg-Si Al Alloy. *Mettall. Trans. A* 2016, 47, 3000–3014. [CrossRef]
18. Kuijpers, N.C.W.; Vermolen, F.J.; Vuik, K.; van der Zwaag, S. A Model of the B-Alfesi to A-Al (Femn) Si Transformation in Al-Mg-Si Alloys. *Mettall. Trans. A* 2003, 44, 1448–1456. [CrossRef]
19. Wang, M.; Xu, W.; Han, Q. Effect of Heat Treatment on Controlling the Morphology of Alfesi Phase in A380 Alloy. *Int. J. Met.* 2016, 10, 516–523. [CrossRef]
20. Kuipers, N.C.W.; Vermolen, F.J.; Vuik, C.; Koenis, P.T.G.; Nilsen, K.E.; van der Zwaag, S. The Dependence of the B-Alfesi to A-Alfesi Transformation Kinetics in Al-Mg-Si Alloys on the Alloying Elements. Mater. Sci. Eng. A. 2005, 394, 9–19. [CrossRef]

21. Chen, X.-M.; Dong, Q.-P.; Liu, Z.-G.; Wang, X.-N.; Zhang, Q.-Y.; Hu, Z.-R.; Nagaumi, H. Fe-Bearing Intermetallics Transformation and Its Influence on the Corrosion Resistance of Al-Mg-Si Alloy Weld Joints. J. Mater. Res. Technol. 2020, 9, 16116–16125. [CrossRef]

22. Verma, A.; Kumar, S.; Grant, P.; O’Reilly, K. Influence of Cooling Rate on the Fe Intermetallic Formation in an Aa6063 Al Alloy. Mater. Sci. Eng. A. 2013, 555, 274–282. [CrossRef]

23. Belmares-Perales, S.; Castro-Román, M.; Herrera-Trejo, M.; Ramírez-Vidaurri, L.E. Effect of Cooling Rate and Fe/Mn Weight Ratio on Volume Fractions of A-Alfesi and B-Alfesi Phases in Al–7.3 Si–3.5 Cu Alloy. Met. Mater. Int. 2008, 14, 307–314. [CrossRef]

24. Mahmoud, M.G.; Mosleh, A.O.; Pozdniakov, A.V.; Khalifa, W.; Mohamed, M.S. Characterization of the Solidification Behavior, Microstructure and Mechanical Properties of Aluminum Alloy 6063 with Samarium Addition. J. Alloys. Compd. 2022, 929, 167234. [CrossRef]

25. Ammar, H.R.; Samuel, A.M.; Doty, H.W.; Samuel, F.H. Premium Strength and Optimum Quality in Al-Si-Mg/Al-Si-Mg-Cu Cast Alloys Using Two Different Types of Molds. Int. J. Met. 2022, 16, 1347–1362. [CrossRef]

26. Kamikawa, N.; Huang, X.; Tsuji, N.; Hansen, N. Strengthening Mechanisms in Nanostructured High-Purity Aluminium Deformed to High Strain and Annealed. Acta Mater. 2009, 57, 4198–4208. [CrossRef]

27. Liu, W.; Man, C.-S.; Raabe, D.; Morris, J. Effect of Hot and Cold Deformation on the Recrystallization Texture of Continuous Cast Aa 5052 Aluminum Alloy. Scr. Mater. 2005, 53, 1273–1277. [CrossRef]

28. Wen, Z.; Liu, Y.Y.; Jia, Z.H.; Zhao, P.Z.; Zhang, Q.Z.; Liu, Q. Study of Texture in 6016 Aluminum Alloy During Processing. In Materials Science Forum; Trans Tech Publications Ltd.: Bäch, Switzerland, 2017.

29. Kim, H.-K.; Kim, H.-W.; Cho, J.-H.; Lee, J.-C. High-Formability Al Alloy Sheet Produced by Asymmetric Rolling of Strip-Cast Sheet. Mater. Sci. Eng. A. 2013, 574, 31–36. [CrossRef]

30. Sidor, J.; Miroux, A.; Petrov, R.; Kestens, L. Microstructural and Crystallographic Aspects of Conventional and Asymmetric Rolling Processes. Acta Mater. 2008, 56, 2495–2507. [CrossRef]

31. Wakeel, A.; Huang, T.L.; Wu, G.L.; Mishin, O.V.; Huang, X. Development of a Strong Goss Texture During Annealing of a Heavily Rolled Aa 1.3% Mn Investigated by Means of Precipitation Behavior and Mechanical Properties of Al–Mg–Si Alloy. Acta Mater. 2013, 75, 1273–1277. [CrossRef]

32. Perez-Prado, M.T.; Gonzalez-Doncel, G.; Ruano, O.A.; McNELLEY, T.R. Texture Analysis of the Transition from Slip to Grain Boundary Sliding in a Discontinuously Recrystallized Superplastic Aluminum Alloy. Acta Mater. 2001, 49, 2259–2268. [CrossRef]

33. Doherty, R.; Hughes, D.; Humphreys, F.; Jonas, J.; Jensen, D.; Kassner, M.; King, W.; McNelley, T.; McQueen, H.; Rollett, A. Current Issues in Recrystallization: A Review. Mater. Sci. Eng. A. 1997, 238, 219–274. [CrossRef]

34. Engler, O.; Hirsch, J.; Lücke, K. Texture Development in Al-1.8 Wt% Cu Depending on the Precipitation State—II. Recrystallization Textures. Acta Metall. Mater. 1995, 43, 121–138.

35. Perez-Prado, M.T.; Gonzalez-Doncel, G.; Ruano, O.A.; Mcnelley, T.R. Texture Analysis of the Transition from Slip to Grain Boundary Sliding in a Discontinuously Recrystallized Superplastic Aluminum Alloy. Acta Mater. 2001, 49, 2259–2268. [CrossRef]

36. Huo, W.; Hou, L.; Lang, Y.; Cui, H.; Zhuang, L.; Zhang, J. Improved Thermo-Mechanical Processing for Effective Grain Refinement of High-Strength Aa 7050 Al Alloy. Mater. Sci. Eng. A. 2015, 626, 86–93. [CrossRef]

37. He, H.; Yi, Y.; Huang, S.; Guo, W.; Zhang, Y. Effects of Thermomechanical Treatment on Grain Refinement, Second-Phase Particle Dissolution, and Mechanical Properties of 2219 Al Alloy. J. Mater. Process. Technol. 2020, 278, 116506. [CrossRef]

38. Humphreys, J.; Bate, P. Gradient Plasticity and Deformation Structures around Inclusions. Scr. Mater. 2003, 48, 173–178. [CrossRef]

39. Engler, O.; Kong, X.W.; Lücke, K. Influence of Precipitates on the Microstructure and Texture During the Rolling of Al-Cu and Al-Mn Single Crystals with Rolling Texture Orientations. Philos. Mag. A 2001, 81, 543–570. [CrossRef]

40. Engler, O.; Yang, P.; Kong, X. On the Formation of Recrystallization Textures in Binary Al-1.3% Mn Investigated by Means of Local Texture Analysis. Acta Mater. 1996, 44, 3349–3369. [CrossRef]