Study of the Mechanical Properties of Al–Mg ADC6 Aluminum Alloy Produced by Unidirectional Casting Under Various Cooling Rates

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Abstract

To create the high strength and high ductility of Al–Mg based aluminum alloy (JIS–ADC6), ADC6 samples were produced by the unidirectional continuous casting (HMC). The HMC process was conducted with direct water cooling to melt ADC6, which can make fine microstructures and control crystal orientation. The cast samples were prepared under various cooling rates(CRs): 6.3, 34, and 62 K/s. The microstructure and crystal orientation of the samples were altered with CR. At CRs of 34 K/s and 62 K/s, the α-Al phases and intermetallic compounds, e.g., Mg_2Si and $Al_{15}(Fe,Mn)_3Si_2$, became finer and more spherical. The secondary dendrite arm spacing for the sample at 62 K/s was 8.7 μm—more than 70% smaller than the ADC6 sample (ingot) made by a gravity casting process. Notably, at a CR of 34 K/s, the crystal orientation was predominantly arranged with the(101) plane. Tensile properties—ultimate tensile strength(σ_{UTS}), 0.2% proof stress($\sigma_{0.2}$), and failure strain(ε_f)—varied with the CR. The tensile strength(σ_{UTS} and $\sigma_{0.2}$) consistently increased with increasing the CR. The improvement in the tensile strength resulted from the refined microstructures, such as the α-Al phase and intermetallic compounds. Similarly, the failure strain also increased with increasing CR, which was severely affected by the finer and more spherical intermetallic compounds. In this case, the ε_f value of the sample at 34 K/s was however slightly higher than that at 62 K/s, due to more uniformly organized crystal orientation, while their ductility was much higher than that of the gravity cast sample. The tensile properties in detail were further analyzed using their failure characteristics.

Keywords: Al–Mg alloy, heated mold continuous casting, mechanical property, microstructural characteristics, crystal orientation, fractography

1. Introduction

Because of environmental issues, improving the fuel efficiency of vehicles is receiving significant attention in our society. Aluminum alloys are widely considered to be highly advantageous materials due to their low density, exceptional recyclability, and superior corrosion resistance [1–8]. Recently, various aluminum alloys have been employed in our industries, such as Al–Cu, Al–Mn, Al–Si, and Al–Mg, and production amount of those aluminum alloys has increased as a replacement for steel. Al–Cu based alloys: duralumin, exhibit superior mechanical properties, and have been used in the manufacturing of aircraft components and sport goods, and Al–Mg based alloys are known for their excellent corrosion resistance, making them ideal for applications in ship components, including propellers, water pumps, and oil pans. These parts are typically produced by casting processes such as gravity casting (GC) and pressure diecasting (DC), due to the technical advantages, e.g., high productivity, creatable of complicated shape and thin shape. For the diecasting process, Al–Mg based alloys are widely employed engineering materials; however, the molten Al–Mg based alloy exhibits high viscosity, leading to cast defects, e.g., shrinkage porosity and surface cracks. These defects significantly degrade the mechanical properties. For example, the ultimate tensile strength (σ_{UTS}) of cast Al–Mg based alloys produced by GC or DC [9, 10] ranges from about 210 to 320 MPa, while the fracture strain (ε_{f}) ranges from about 5 to 12 %. These properties are considerably lower than those of medium carbon steel, such as SAE1045: σ_{UTS} = 550 MPa and ε_{f} = 15% [11]. In general, the mechanical properties are influenced by the grain size, eutectic phases, and intermetallic compounds, and casting defects [12–15].

To refine the microstructure of the related aluminum alloys, several researchers have explored the addition of alloying elements to Al–Mg based alloys. Zhang et al. [16] successfully refined the grain of Al–3.2Mg alloys by 0 to 1.6 wt% rare earth elements (Ce and La), resulting in an increase in tensile strength from 193 to 207 MPa. Zhou et al. [17] have also conducted to modify the α-Al phase formed with dendritic structure to fine equiaxed grains of as-cast Al–Mg based alloy by adding more than 0.4 wt% Sc. Another approach involves the application of ultrasonic vibrations during the casting process [18]. In this instance, ultrasonic vibration, applied to an Al–5Mg alloy using an ultrasonic generator, resulted in a reduction in grain size from approximately 2.8 to 0.12 mm.

Furthermore, several researchers have examined alternative casting technologies to reduce cast defects. Wan et al. [19] investigated the mechanical properties of Al–Mg based alloys produced by vacuum-assist die casting (VDC). They reported that the σ_{UTS} and ε_f (331 MPa and 4.4%, respectively) were at least 3% higher than those of conventional DC samples. From the fractographic observations, several pores were detected in the DC samples, whereas none were observed in the VDC samples. De Brito et al. [20] studied the effect of casting pressure (from 10 to 120 MPa) on the cast defects and tensile strength in cast Al–7Mg cylindrical ingots (ϕ 50 mm $\times$ 120 mm) produced by squeeze casting (SQ) method. Micro- and macroscale observations showed that common cast defects such as macrosegregation, micropores and oxide incrustations were significantly reduced in SQ samples, in contrast to GC ones. Furthermore, when the casting pressure exceeded 40 MPa, no visible cast defects were observed.

In an alternative way, a heated mold continuous casting (HMC) developed by Ohno [21] might be a suitable method to make fine grains with organizing lattice formation and reduce cast defects. The casting process is to heat the mold temperature maintained above the nucleation temperature of the primary α dendrite, and a unidirectional and rapid solidification through continuous water-cooling is carried out to produce cast components. Such a casting process could lead to fine grains with organized lattice formation and reduction in the porosity defects. To date, some researchers have investigated various HMC alloys, including Bi, Cu, Zn, Mg, and Al. Motoyasu et al. [22, 23] examined the material properties of single crystal Bi and Zn wires fabricated through the HMC process. Moreover, the fracture strain of Bi wire reached approximately 130% as the (001) plane is aligned perpendicular to the casting direction, whereas Zn wire at the $(1\bar{2}10)$ plane to the loading direction exhibited as high as ε_f = about 350%.

In addition, the HMC process has been used to create the cast aluminum alloys (Al–Si, Al–Si–Cu, and Al–Cu based alloys) [24–26]. Because of the fine grains, uniformly organized crystal orientation, and fewer cast defects, excellent mechanical properties are obtained. Although several aluminum alloys have been used in the HMC process, Al–Mg based alloys have not been created extensively until now. One of the reasons is due to the poor castability as mentioned above. Although Al–Mg based alloys have the high corrosion resistance, the mechanical properties are insufficient. However, the authors believe that the high mechanical properties of the Al–Mg alloys are creatable if the HMC process is used. Therefore, the aim of this work was to create cast Al–Mg based alloy via the HMC process under various casting conditions, where to determine the suitable casting condition, the HMC process was conducted at various casting speeds.

2. Experimental procedures

2-1. Sample preparation

In the present work, a commercial Al–Mg based aluminum alloy (JIS–ADC6) was used. Table 1 indicates the chemical composition of the aluminum alloy: Al–3.59Mg–0.47Mn–0.79Si–0.51Fe–0.02Cu in mass %. The cast samples were made by a heated mold continuous casting process to obtain better mechanical properties. In our previous studies, the high mechanical properties of Al–Si–Cu based aluminum alloys were created, compared to those alloys made by gravity casting and die-casting processes [15]. Figure 1 displays a schematic diagram of the HMC system, which was developed originally based on the following HMC concept: a unidirectional and rapid solidification through continuous water-cooling, and the mold temperature is maintained above the nucleation temperature of the primary α dendrite. As seen in Figure 1, the HMC system consists of mainly an electric furnace, a graphite crucible, a displacer block to maintain the pressure of the flowing molten metals, a heated graphite mold, a stainless-steel drawer bar and a drawer for withdrawal of the cast rod, e.g., ϕ5 mm × 500 mm. In the HMC process, the ADC6 alloy blocks were first melted at 945 K, which is about 35 K higher than its liquidus line. Prior to casting, flux treatment was applied to remove oxides and impurities. To make the cast rod, the drawer bar was employed to retract the melt ADC6 with surface tension at various speeds. Due to the higher melting point and viscosity of ADC6 compared to ADC12, the casting conditions must be carefully optimized. The retracted melt ADC6 was solidified by the direct water cooling immediately after retracting the melt ADC6. The retracting (casting) speed was set to 0.8, 1.9, and 2.7 mm/s (denoted as $HMC_{0.8}$, $HMC_{1.9}$ and $HMC_{2.7}$, respectively) to investigate the effect of cooling speed on the material properties, where the higher the cooling speed, the smaller the grain. Those casting speeds were selected on the basis of previous studies [15, 27], which identified 1.9 mm/s as an optimal speed for producing high-quality cast aluminum alloys, including ADC12 and AC4CH.

To understand the material properties of the HMC samples, ADC6 samples were further made by the two different methods: furnace cooling solidification in a crucible (FC) and gravity casting in a steel mold (GC). Figure 2 shows schematic illustrations of their sample preparation processes for (a) FC and (b) GC. The FC sample was prepared by melting approximately 0.2 kg of ADC6 alloy in a graphite crucible (ϕ50 mm × 70 mm). The molten ADC6 was then slowly solidified inside an electric furnace approximately 8 hours through furnace cooling. For the GC sample, a commercial ADC6 ingot, produced through gravity casting in atmospheric environment by Daiki Aluminium Industry Co., Ltd. was used. The dimensions of the ingot are approximately 600 mm × 90 mm × 40 mm. The solidification rate of the cast samples was measured using the secondary dendrite arm spacing (SDAS).

2-2. Mechanical testing

The hardness and tensile properties were examined. Micro-Vickers hardness measurements were performed with an indentation load of 9.8 N applied for 15 seconds. The tensile test was performed at room temperature using a screw-driven universal testing machine with a 50 kN capacity. A tensile load was applied to the test specimens at a rate of 1 mm/min until fracture. The test specimens were designed to be dumbbell-shaped with dimensions ϕ2.5 mm × 4 mm, made from the cast samples. A schematic diagram of the specimen is shown in Figure 3. Tensile properties were assessed based on the ultimate tensile strength (σ_{UTS}), 0.2% proof stress ($\sigma_{0.2}$) and strain to failure (ε_f). The tensile load and strain values were measured using a commercial load cell and strain gauge, respectively.

2-3. Microstructural analysis

Microstructural characteristics of the cast ADC6 samples were investigated using several analytical techniques, e.g., an optical microscope, a scanning electron microscope (SEM), an energy dispersive X-ray detector (EDS), an X-ray diffractometer (XRD), and electron back scattered diffraction pattern (EBSD) analysis. The sample surfaces for observation were polished to a mirror finish using colloidal silica for approximately 2 hours. EDS analysis was performed by the SEM (JIB-4500) with an acceleration voltage of 20 kV. XRD analysis was carried out using the X-ray diffractometer (Ultima IV) with Co Kα radiation. EBSD analysis was conducted with an acceleration voltage of 15 kV, a beam current of 13 nA, and a step size of 5 μm using the SEM (JSM-7001F).

3. Results and discussion

3-1. Microstructural characteristics

Figure 4 presents optical micrographs and SEM images of the ADC6 samples produced by the FC, GC, and HMC processes. All HMC samples were observed on surfaces perpendicular to the casting direction. As shown in Figure 4, the microstructural characteristics of all samples consist of an aluminum matrix (α-Al phase) and several eutectic phases, observed as light and dark gray regions. Figure 5 shows SEM images with elemental mappings, obtained by energy dispersive X-ray spectroscopy (EDS). The EDS analysis in Figure 5 revealed that these phases were mainly composed of Al, Mg, Si, Mn, and Fe. Furthermore, the XRD analysis identified the phases as the α-Al phase and the intermetallic compounds Mg_2Si and $Al_{15}(Fe,Mn)_3Si_2$, which are consistent with the previous reports [28–30]. However, the sizes and shapes of these phases varied among the samples. In the FC and GC samples, which solidified at lower cooling rates compared to that for the HMC samples, these phases were grown largely, while the greatly grown ones were obvious in FC. In the HMC samples, the high cooling $HMC_{2.7}$ samples seem to have relatively small phases. To quantitively evaluate the size of microstructure, the secondary dendrite arm spacing (SDAS) was measured for each sample. The SDAS values of all samples are summarized in Table 2, which were evaluated from the mean of more than 50 SDAS. From this measurement, the SDAS value for $HMC_{2.7}$ is as short as 8.7 μm, i.e., small grain, which decreases with increasing the cooling speed. In particular, largely grown grain (SDAS = 157.0 μm) is observed for FC. To verify the approximated cooling rate (CR), the Hall-Petch analysis was conducted using one of the related formulas: $CR = 2 \times 10^4\ SDAS^{-2.67}$ [25]. The CR values of all samples calculated are indicated in Table 2. From this estimation, the slower the CR, the larger the SDAS. In a previous study [31], the cooling rate (CR) of ADC12 using the HMC process at a casting speed of 2.7 mm/s was investigated, yielding a CR of 167 K/s. This value is approximately 2.5 times higher than that of ADC6-$HMC_{2.7}$ (62.0 K/s). This difference is attributed to the narrower nucleation temperature range of the primary α- dendrite of ADC12 (about 12 K) compared to that of ADC6 (about 50 K) [32, 33]. As a result, ADC12 solidifies more rapidly even under the same casting speed, which leads to the formation of finer grains.

Figure 6 presents the inverse pole figure (IPF) maps for representative FC, GC, and HMC samples, obtained by EBSD analysis. It is clear from the IPF maps that crystal orientation characteristics differed among the samples. It should be described that crystal orientation remains consistent within each colony, where multiple grains align with the same orientation. It is interesting to note that the crystal orientation in $HMC_{1.9}$ is predominantly organized with (101) plane over an extensive area, indicated with green-like color. This phenomenon may be affected by the organized and stable grain growth along the casting direction, as indicated by the striped patterns as enclosed by the dashed circles. In general, aluminum exhibits preferential crystal growth along the <100> direction (corresponding to (001) in IPF maps). However, when aluminum is alloyed with elements such as Mg, Cu, Zn, or Sm to form binary Al–X alloys, the preferential crystal growth direction of aluminum (α-Al) shifts from <100> to <110> (observed as (101) in IPF). A similar occurrence has been observed in other related aluminum alloys produced by HMC at 1.9 mm/s [34]. This phenomenon is considered to be primarily influenced by altered solid-liquid interfacial energy anisotropy arising from added solute elements, including the local composition near the solid–liquid interface [35–41]. Furthermore, in the present study, randomly oriented lattice structures were observed in $HMC_{0.8}$ and $HMC_{2.7}$, which could be affected by the variations in cooling rate.

Figure 7 shows the kernel average misorientation (KAM) maps corresponding to the

IPF maps (Figure 6). The relatively organized lattice regions in the $HMC_{1.9}$ sample shows the lower internal strain, as indicated in the KAM maps. In contrast, randomly oriented grain formations in other HMC samples induce higher strain, resulting in an increase in the KAM value.

3-2. Mechanical properties

Figure 8 shows the Vickers hardness (HV) data of the FC, GC and HMC samples. Figure 8a presents the measured HV for all samples, obtained from regions containing α-Al phases and various eutectic phases. Overall, the HMC samples showed a higher mean HV value than the FC and GC samples, while the data points for FC displayed significant scatter. Such variation in FC would be caused by the unrefined microstructure, which influenced the hardness of the intermetallic compounds mentioned and the α-Al phase. Specifically, the hardness value of the main intermetallic compounds and α-Al phase were as follows: Mg_2Si 2.1 GPa, $Al_{15}(Fe,Mn)_3Si_2$ 3.7 GPa, and α-Al phase: about 1.0 GPa [30]. On the other hand, for the HMC samples, HV increases with cooling speed, which the higher hardness observed in the $HMC_{2.7}$ sample. These differences could be attributed to grain size (or SDAS), based on the Hall-Petch relationship. Figure 8b shows the relationship between HV and SDAS. As seen, the HV value increases with decreasing SDAS.

Figure 9 shows representative tensile stress-strain curves for the FC, GC and all HMC samples. As seen, the stress-strain curves exhibit different tendencies depending on the cast samples. Based on these curves, the tensile properties are summarized in Figure 10: (a) ultimate tensile strength (σ_{UTS}), (b) 0.2% proof stress ($\sigma_{0.2}$) and (c) strain to failure (ε_f). From Figure 10a and 10b, HMC samples exhibited higher σ_{UTS} and $\sigma_{0.2}$ as the cooling rate increased, consistently much exceeding those of the FC and GC samples, following trends like the results of hardness. In $HMC_{2.7}$ sample, the σ_{UTS} and $\sigma_{0.2}$ values were 296.3 MPa and 155.6 MPa, respectively, which are more than 5% higher than those for the ADC6 samples made by conventional die-casting process (e.g., σ_{UTS}: 280 MPa and $\sigma_{0.2}$: 120 MPa [42]). These enhancements can also be attributed to their grain (or SDAS) refinement as described in the results of the hardness. From Figure 10c, the HMC samples also tended to exhibit higher ε_f compared to the FC and GC samples, whereas there is sometimes a trade-off relationship between strength and ductility. This may be attributed to the presence of relatively spherical and fine intermetallic compounds [31]. It is interesting to note that the $HMC_{1.9}$ sample exhibited a slightly high ε_f value, compared to that of the $HMC_{2.7}$ sample. The mean ε_f value of $HMC_{1.9}$ is about five times higher than that of the die-casting ADC6 samples [42]. Furthermore, in our previous work, the related Al–5Mg samples, made by HMC, had high ε_f value of about 30% [43]. Such high ductility is similar to that of the $HMC_{1.9}$ sample. The reason for the high ductility of $HMC_{1.9}$ is influenced by the relatively organized lattice formation as well as the fine grain as shown in Figure 6.

To understand the effect of grain size on the tensile properties, Hall-Petch analysis was carried out. Figure 11 shows the relationship between SDAS and tensile properties of the ADC6 samples. This analysis was conducted using the following tensile properties: (a) σ_{UTS}, (b) $\sigma_{0.2}$, and (c) ε_f. As seen, all tensile properties have a clear Hall-Petch relationship with a high correlation rate (R^2), i.e., the tensile properties increase with decreasing SDAS. Note that the R^2 value for ε_f was 0.82, which was slightly lower than those for σ_{UTS} and $\sigma_{0.2}$ (0.99 and 0.95, respectively), due to the irregularly formed microstructural characteristics, e.g., uniformly orientated lattice formation in only the $HMC_{1.9}$ sample.

3-3. Fractography

Figure 12 shows SEM images of the fracture surfaces of the FC, GC, and all HMC samples after tensile tests. Figure 13 shows micrographs of the cross-sections of the HMC samples adjacent to the fracture surfaces. As shown in Figure 12, the overall appearance of the fracture surfaces varied among the samples. Note that α-Al dendrite, created by the severe shrinkage of the sample, was widely detected in the FC sample. In contrast, no visible cast defects such as shrinkage porosity were observed on the fracture surfaces of any HMC samples, which was attributed to the directional solidification. From Figure 12, the fracture modes for all samples was devided with two different paterns. In the FC, GC and $HMC_{0.8}$ samples, a brittle failure mode, characterized by cleavage-like fractures, was dominant feature. From the cross-section of the $HMC_{0.8}$ sample, shown in Figure 13, the crack seems to propagate through large intermetallic compounds. In contrast, ductile failure characteristics were observed in the $HMC_{1.9}$ and $HMC_{2.7}$ samples although those are different mode. The shear slip was widely observed in the $HMC_{1.9}$; and $HMC_{2.7}$ samples shows both the shear slip and dimple failures. Such a change of the ductile fracture mode could be attributed to their microstructural characteristics, e.g., crystal orientation. In uniformly oriented lattice formation with (101) for $HMC_{1.9}$, shown in Figure 6, the Schmid factor is 0.408 for the slip system $(111)[\bar{1}10]$ under tensile loading along the [101] direction. Such high resolved shear stress may lead to the severe plastic deformation resulting in the shear failure. In contrast, in the case of randomly oriented lattice structures in $HMC_{2.7}$, the resolved shear stress on slip planes was heterogeneous, which resulted in rough fracture surfaces. Both ductile failure modes were verified from the micrographs of the cross-section of the fractured specimen, as shown in Figure 13. In the $HMC_{1.9}$ sample, the fracture occurred at an angle of approximately 45° to the tensile direction, in which plastic flow is seen near the shear failure. The $HMC_{2.7}$ sample showed several voids and cavities near and on the fracture suface, respectively. In this case, the cavity could be related to the dimple failure shown in Figure 12, due to their similar size (about 5–10 μm).

To further investigate the fracture mechanisms, SEM-EDS analysis was conducted on the fracture surfaces for all samples. Figure 14 shows EDS mapping with the elements of Al, Mg and Fe on the fracture surfaces. As seen, the Al, Mg, and Fe elements were detected in all samples with different intensities. In the FC, GC, and $HMC_{0.8}$ samples, Mg and Fe elements were widely distributed on the fracture surfaces, while those elements were detected in the small area of the $HMC_{1.9}$ and $HMC_{2.7}$ samples. From our microstructural observation, the Mg and Fe elements are associated to the intermetallic compounds Mg_2Si and $Al_{15}(Fe,Mn)_3Si_2$ [28–30]. It is reported in the previous study [30], the hardness values of those compounds are more than 2.1 GPa [30], which is more than twice as high as that for the Al matrix. Because of the high hardness of the intermetallic compounds, the failure in FC, GC, and $HMC_{0.8}$ could occur through the compounds, leading to the brittle failure mode. On the other hand, the fine intermetallic compounds were detected on the fracture surface in both $HMC_{1.9}$ and $HMC_{2.7}$: on the shear slip surface ($HMC_{1.9}$) and in the dimple ($HMC_{2.7}$).

4. Conclusions

Based on the experimental results discussed above, the following conclusions can be drawn:

1. The cast ADC6 alloys produced by FC, GC, and HMC processes consist of the α-Al phase

with intermetallic compounds Mg_2Si and $Al_{15}(Fe,Mn)_3Si_2$. The size and shape of the microstructures are altered depending on the CR. With increasing CR, the α-Al phases and the intermetallic compounds become finer and more spherical. At a CR of 62 K/s, the SDAS is 8.7 μm—more than 70% smaller than that of the ADC6 sample made by a gravity casting process.

2. In the HMC samples, a uniform crystal orientation is observed at a casting speed of 1.9 mm/s (CR = 34 K/s), which may be attributed to directional solidification. Alloy elements such as Mg may alter the solid–liquid interfacial energy anisotropy, leading to the formation of the (101) plane rather than the (100) plane. However, at other casting speeds, the crystal planes are randomly oriented, which is considered to result from the variations in cooling rate.
3. The tensile properties (σ_{UTS}, $\sigma_{0.2}$, and ε_f) of the cast ADC6 samples vary depending on the CR. Tensile strength (σ_{UTS} and $\sigma_{0.2}$) increases consistently with increasing CR. The enhancement in the tensile strength is attributed to the refined microstructures. Similarly, the ε_f also increases with CR, which is significantly affected by the refinement and spheroidization of intermetallic compounds. However, in the HMC samples, the ε_f at a casting speed of 1.9 mm/s (CR = 34 K/s) is slightly higher than that at 2.7 mm/s (CR = 62 K/s). The enhanced ductility is attributed to shear slip failure facilitated by the organized crystal orientation.

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Conflict of interest

The authors declare no conflict of interest.

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