Effect of Added Gas-Atomized Al-Si/SiCp Composite Powder on the Sinterability and Mechanical Properties of Alumix 431 fabricated by Hot-Pressing Process

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Abstract: The aim of this study is to establish an optimized hot-press sintering condition for aluminum matrix composites (AMCs) by analyzing their sintered microstructure and mechanical properties. The fabricated AMCs in this study consisted of a blend of commercial Ecka Alumix 431 powder, and Al-Si/SiCp powder manufactured by gas atomizer (Al-Si/SiCp). The Al-Si/SiCp powder was added to the Alumix 431 powder as a binder. The blended powder was hot-pressed and sintered and its properties were analyzed based on three variables: the blending ratio of each powder, sintering time and sintering temperature. The MgZn2 phases in the Alumix 431 powder were finely and homogeneously distributed when the sintering temperature was increased. The liquid phase of the Al-Si alloy facilitated densification of the Alumix 431 particles and promoted a reduction in pores in the sintered body. Dispersion of the MgZn2 phases, degree of porosity and densification of the fabricated AMCs affected the ultimate tensile strength (UTS) of the sintered body at room temperature (RT). The highest UTS of the fabricated AMCs was 242MPa at RT. The optimum sintering conditions were: a mixing ratio of Al-Si/SiCp to Alumix 431 powders of 5 to 95 wt%, a sintering time of 30 min and a sintering temperature of 610 ℃. Wear behavior of the specimens prepared with the optimized sintering conditions included stick-slip, abrasive, adhesive and delamination phenomena for each test condition, under correlated vertical load and linear speed.

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Keywords: composites, sintering, mechanical properties, wear, scanning electron microscopy (SEM)

1. INTRODUCTION

Recently, the application of lightweight materials has been actively discussed and studied in the vehicle industry, as efforts to increase fuel efficiency continue. In particular, aluminum and its alloys have been considered as substitutes for iron and its alloys because of aluminum’s low density, high workability and corrosion resistance [1,2].

Meanwhile, the powder metallurgy (PM) technique is one of the effective methods used to mass-produce parts, from simple shapes to intricate patterns. The PM technique can also be used in the production process of aluminum alloys with near-net shapes, even though these have a complicated shape. As a result, the PM technique is drawing attention in the engineering field as a high potential to reduce manufacturing costs [1-3].

7xxx series aluminum alloys, Al-Zn-Mg alloys, have been widely used in automotive parts and aerospace components because they have higher mechanical strength than those of other Al-based alloys. These properties originate from solid solution hardening with Mg and Zn elements and precipitation hardening by metastable η’ (MgZn2) phase and stable η(MgZn2) phase transformed by metastable η’ (MgZn2) phase to the main matrix in the Al-Zn-Mg alloy [2,4]. However, Al-Zn-Mg alloys have limited application in parts which require high wear resistance and strength, because while these alloys have high strength they have low wear resistance. When Al-Zn-Mg alloys are manufactured by PM with near-net shape, there are difficulties in densification, because they form a sturdy oxide film, and the intrinsic high strength of these powder alloys result in poor sinterability.
Table 1. Chemical composition of Al-Si/SiC\textsubscript{p} and Alumix 431 powder.

<table>
<thead>
<tr>
<th></th>
<th>wt%</th>
<th>Al</th>
<th>Si</th>
<th>Zn</th>
<th>Mg</th>
<th>Cu</th>
<th>Sn</th>
<th>SiC</th>
</tr>
</thead>
<tbody>
<tr>
<td>Al-Si/SiC\textsubscript{p}</td>
<td>Bal.</td>
<td>9</td>
<td>-</td>
<td>0.5</td>
<td>0.2</td>
<td>-</td>
<td>-</td>
<td>20vol%</td>
</tr>
<tr>
<td>Alumix 431 (alloy powder)</td>
<td>Bal.</td>
<td>-</td>
<td>11.8</td>
<td>5</td>
<td>3.4</td>
<td>0.52</td>
<td>-</td>
<td>-</td>
</tr>
<tr>
<td>(Al powder)</td>
<td>99.9</td>
<td>-</td>
<td>-</td>
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</tbody>
</table>

The oxide films which are formed prevent diffusion of the Al-Zn-Mg alloy powder between interfaces of Al-Zn-Mg alloy powder during solid state sintering, and the high strength of the Al-Zn-Mg alloy limits its plastic deformation [5,6].

Quite a number of studies have been carried out in efforts to improve the sinterability of Al-Zn-Mg alloy powders, using different powder mixtures. Different compositions, or the addition of different elements, can provide better compactibility and sinterability in Al-Zn-Mg alloy powders [7,8]. Unfortunately, these methods can also result in a reduction in mechanical properties [9,10].

As a result, to simultaneously enhance the mechanical properties and sinterability of Al-Zn-Mg alloy powders, combined mixtures with binders and reinforcements are required. Suitable binders need to have both stable viscosity and high wettability, as well as a softening point that is lower than that of the Al-Zn-Mg alloy powder [3]. Adding a binder to the Al-Zn-Mg alloy powder can facilitate the diffusion of atomic elements in the Al-Zn-Mg alloy powder during the sintering process. It is necessary for the Al-Zn-Mg alloy powder and binder material to react in the matrix, and also to form a liquid phase.

An added reinforcement can promote even load shearing by inducing homogeneous distribution in the main matrix of the Al-Zn-Mg alloy powder. For example, it was previously reported that SiC reinforcement promoted an increase in the strength and the wear resistance of a ductile Al-based matrix [11-13]. If the wear resistance, mechanical properties and the sinterability of Al-Zn-Mg alloys can be improved by adding SiC as a reinforcement, the resulting high wear resistance alloys could be utilized in automobile components which have complicated shapes, such as race rings, sprockets, chain rings, gear parts, and etc.

However, the fabrication of metal matrix composites (MMCs) using SiC reinforcements also has a couple of problems, including the formation of clusters of SiC particles and pores, due to poor wettability between the metal matrix and the SiC reinforcements during the sintering process. The formation of these defects is a major cause of reduced mechanical properties in MMCs.

To address these issues, a number of methods have been studied to achieve homogeneous dispersion of the SiC particles in the Al-based matrix, including in-situ powder metallurgy by gas atomization, and surface treatment of the reinforcements, etc. [14-17].

In this study, an Al-Si/SiC\textsubscript{p} composite powder with homogeneously distributed SiC particles in Al-Si alloy powder was prepared fabricated by in-situ gas-atomizing. This Al-Si/SiC\textsubscript{p} composite was used as a binder to enhance the sinterability, wear and mechanical properties of Al-Zn-Mg alloy powder. The microstructure and mechanical properties of hot-pressed specimens (sintered body or specimen), which were composed of the Al-Si/SiC\textsubscript{p} and Al-Zn-Mg alloy powders, were investigated. Based on the observed mechanical properties, the optimized sintering condition was determined. In addition, the wear properties of the optimized specimens were analyzed based on vertical load and linear speed conditions.

2. EXPERIMENTAL PROCEDURES

2.1. Specimen preparation

The Al-Si/SiC\textsubscript{p} was fabricated by the gas atomizing method with 80% N\textsubscript{2} and 20% O\textsubscript{2} gas under 2MPa pressure. Al-Si alloy was melted at 900 °C to obtain sufficient fluidity, and then 20 vol% of 10-20 μm SiC particles were added. The commercial Al powder (Alumix 431, ECKA Granules GmbH, Germany) was manufactured by water atomization. The chemical composition of the Alumix 431 is shown in Table 1. The powders produced by the various methods were mixed in a turbulent mixer at 45 rpm speed for 24 hours at a mixing ratio which is shown in Table 2.

The mixed powder was sintered by the hot-pressing
Table 2. Variables of the hot-press sintering condition.

<table>
<thead>
<tr>
<th>Mixing Ratio</th>
<th>Sintering Temperature (℃)</th>
<th>Sintering Time (min)</th>
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<tr>
<td>Alumix 431: Al-Si/SiCp (wt%)</td>
<td>560</td>
<td>15</td>
</tr>
<tr>
<td>100 : 0</td>
<td>570</td>
<td>30</td>
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<td>95 : 5</td>
<td>580</td>
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<td>90 : 10</td>
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<td>610</td>
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<tr>
<td>85 : 15</td>
<td>600</td>
<td>620</td>
</tr>
<tr>
<td>80 : 20</td>
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Fig. 1. Microstructures of the Alumix 431 powder by OM (a), FE-SEM (b) and EDS results (c) and in-situ type the Al-Si/SiCp powder by OM (d), FE-SEM (e) shows EDS results (f).

Table 3. The chemical compositions of SUJ2.

<table>
<thead>
<tr>
<th>Elements</th>
<th>Fe</th>
<th>C</th>
<th>Si</th>
<th>Mn</th>
<th>Cr</th>
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<td>Composition (wt%)</td>
<td>Bal.</td>
<td>0.95-1.10</td>
<td>0.15-0.35</td>
<td>0.20-0.50</td>
<td>1.30-1.60</td>
</tr>
</tbody>
</table>

The microstructures of the in-situ type Al-Si/SiCp and the Alumix 431 powders are shown in Fig. 1. The Al-Si/SiCp specimens consisted of a primary Al phase, which can be seen as the white region, SiC particle-phases with irregular features, and the Al-Si eutectic-like structure-phases in Fig. 1(a). The cross-sectional morphology of the in-situ type Al-Si/SiCp shown in 1(a) appeared to have a dendritic structure.

The SiC particles exist inside of the primary Al phases. The Al-Si eutectic-like structure-phases (biphases) matrix and pores are not observed at the interface area between the biphases matrix and the SiC particles, in Fig. 1(a) and 1(b). The Al-Si/SiCp fabricated by the gas atomizing method was...
Table 4. Qualitative EDS result of alloy powder in the Alumix 431 powder.

<table>
<thead>
<tr>
<th>at%</th>
<th>Al</th>
<th>Zn</th>
<th>Mg</th>
<th>Cu</th>
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<tr>
<td>1</td>
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<td>2</td>
<td>91.65</td>
<td>3.70</td>
<td>3.59</td>
<td>1.06</td>
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</table>

controlled to prevent shrinkage-contraction stress between the molten hypoeutectic Al-Si alloy and the SiC particles by applying a rapid solidification rate. It consequently had advantages, such as powder homogeneity and good packing properties [3,18,19].

Figure 1(c) shows the results of the EDS mapping, and Fig. 1(b) for the Al-Si/SiC indicates that Si, Al and the SiC particles exist in the Al-Si/SiC. The EDS results agree with the microstructures seen in the OM and SEM images (Fig. 1(a) and 1(b)).

The Alumix 431 powder has two types of powder. One type is pure Al particles and the other is Al-Zn-Mg alloy type particles with a chemical composition of Al-11.8Zn-5Mg-3.4Cu-0.52Sn, as shown in Table 1. The white phase in the alloy powder is a $\text{Mg}_2\text{Al}_3(\text{Mg, Zn})_{49}$ phase, which is a nonequilibrium phase formed by the rapid solidification rate, as shown in Figure 1(e) and (f) and in the qualitative EDS results of point 1 and 2 shown in Table 4 [5,20]. The $\text{Mg}_2\text{Al}_3(\text{Mg, Zn})_{49}$ phase, which is known to be a metastable intermetallic compound, is transformed to a stable phase by heat treatment. The stable phases, such as the MgZn$_2$ intermetallic compounds, help improve the mechanical properties when they are distributed in the matrix [21].

Figure 2 presents the XRD patterns of the Al-Si/SiC, Alumix 431 powder and the mixed powder (Al-Si/SiC$_p$: Alumix 431 ratio = 50:50). All of the detected XRD peaks are Al, Si, and SiC phases, from the Al-Si/SiC$_p$, and Mg$_2$Al$_3$(Mg, Zn)$_{49}$ and Al phases from the Alumix 431 powder. These peaks are in good agreement with the reference from the JCPDS database No. 04-0787 for Al, No.772111 for Si, No. 190029 for Mg$_2$Al$_3$(Mg, Zn)$_{49}$ and No. 040756 for SiC, respectively. These results were well matched with the observed microstructures of each powder as shown in Fig. 1. In addition, it had no effect to make solid-state-reacted phase during mixing process.

Figure 3 shows the DTA results of the mixed powder between 430 °C to 640 °C. One exothermic reaction and three endothermic reactions appeared, and were divided into 4 regions by the existing phases. The mixed powder is composed of Mg$_2$Al$_3$(Mg, Zn)$_{49}$, $\alpha$-Al, primary Al, SiC$_p$ and Al-Si eutectic phases. The 1st peak is a dissolution peak of the Mg$_2$Al$_3$(Mg, Zn)$_{49}$ phase in the Alumix 431 powder in the matrix [5]. Region I consists of $\alpha$-Al, the remaining Mg$_2$Al$_3$(Mg,
Fig. 4. X-ray diffraction patterns for the hot-press sintered body (a) mixing ratio at 610 °C during 30 minutes, (b) temperature for 30 minutes with mixing ratio (Al-Si/SiCp: Alumix 431) of 5:95 and (c) time at 610 °C with mixing ratio (Al-Si/SiCp: Alumix 431) of 5:95.

Fig. 5. Microstructures of the sintered body manufactured at 610 °C during 30 minutes with each mixing ratio (Al-Si/SiCp: Alumix 431) by FE-SEM, (a) Alumix 431 (b) 5:95, (c) 10:90, (d) 15:85 and (e) 20:80.

Zn)49, primary Al, SiC phases and the Al-Si eutectic phase of Al-Si/SiCp. The 2nd peak indicates the transition reaction from the metastable Mg₃2(Al, Zn)₄9 phase to the stable MgZn₂ phase, and region II is composed of the remaining Mg₃2(Al, Zn)₄9 and the stable MgZn₂ [22]. The metastable Mg₃2(Al, Zn)₄9 phase remained at high temperature because there was the phase transition time and insufficient time for the atoms in the powder to diffuse into the matrix and by heating rate 10 °C/min. The 3rd peak indicates the solid-liquid phase transition of the Al-Si eutectic-like phase in region III. Additionally, the 4th peak is the melting temperature of the alloy powder in the Alumix 431 powder. Finally, region IV consists of the liquid phase of the primary Al, α-Al, and Al-Si eutectic-like structural phases [22,23]. The range of the sintering temperature is determined by the formation temperature of each liquid phase in the DTA mixed powder result.

Figure 4 (a) shows the XRD results of the sintered bodies with the addition of the Al-Si/SiCp. The intensity of the MgZn₂ phase peak decreases with the increasing addition of Al-Si/SiCp. Additionally, the apparent phases in the XRD pattern are in good agreement with the DTA results for all mixing ratios of the Al-Si/SiCp and Alumix 431 powders. It is thought that the Mg and Zn atoms are dissolved and diffused into the pure Al and the Al-Si/SiCp particles by the diffusion driving force as the temperature is increased. Figure 4 (c) presents the XRD pattern of the sintered bodies, which were changed by the various sintering times. When the sintering time was increased from 15 minutes to 30 minutes, dissolution of the MgZn₂ phase occurred in the surrounding
Fig. 6. Microstructures of the sintered body manufactured at each temperature during 30 minutes with mixing ratio (Al-Si/SiCp: Alumix 431) 5:95 by FE-SEM (a) 560 °C, (b) 570 °C, (c) 580 °C, (d) 590 °C, (e) 600 °C, (f) 610 °C and (g) 620 °C.

particles. Overall, a fraction of the metastable phase, Mg32(Al, Zn)49, is unchanged in each condition. This phase is known as a quasicrystalline phase. The Mg32(Al, Zn)49 phase crystallizes into a more stable Frank-Kasper phase (Mg32(Al, Zn)49) at approximately 200-300 °C [24,25].

Figure 5 shows the microstructures of the manufactured sintered bodies according to the ratio of Al-Si/SiCp. The MgZn2 phases have formed inside all of the specimens, and are finely dispersed as white particles. It is thought that the Mg and Zn atoms had sufficient driving-force to diffuse into the bulk of the sintered body manufactured at 610 °C for 30 minutes. This is because dissolution of the Mg32(Al, Zn)49 phase in the matrix occurs at 470 °C, while the MgZn2 phase forms at 510 °C, based on the DTA results in Figure 3.

The MgZn2 phase with the dendritic structure in the matrix has the effect of enhancing the material’s strength, by precipitation and solution hardening in the matrix [9]. In Figure 5, the distribution of the MgZn2 phase and the densification of the sintered bodies are clearly different, depending on the Al-Si/SiCp ratio. In particular, in the matrix with 5 wt% Al-Si/SiCp, the number of pores have decreased significantly, in comparison with those of the sintered Alumix 431 matrix, and with the other wt% Al-Si/SiCp for the same sintering conditions, as shown in Fig. 5.

This phenomenon at the sintering temperature of 610 °C can be attributed to the formation of an Al-Si liquid phase with a low flow-viscosity, because the eutectic temperature of the Al-Si alloy is 577 °C. As a result, this liquid phase of the Al-Si alloy contributes to the increase in densification by increasing the diffusion path of the Mg and Zn atoms. On the other hand, increasing the ratio of Al-Si/SiCp above 5 wt% tends to increase the amount of pores, as shown in Figure 5 (c), (d) and (e). This is because the volume transition of the Al-Si alloy is from liquid to solid state during the cooling process. Additionally, Mg2Si phases can be observed around the Al-Si/SiCp. It is thought that as the Mg2Si phase is formed in the Al-Mg-Si alloy by the ternary reaction, the molten Al-Si alloy reacts with the Mg atoms due to diffusion and the negative bond enthalpy at 610 °C [27,28]. Based on the microstructure, the optimized ratio of Al-Si/SiCp was estimated to be 5 wt% in the Alumix 431 based on observation of the amount of pores and the degree of densification.
Figure 6 shows that the microstructures of the sintered bodies depend on the sintering temperature condition. The observed prior grain boundary and a lot of pore between interface of powder are produced by incomplete densification in the temperature region from 560 °C to 570 °C, as shown in Figure 6 (a) and (b). As the sintering temperature increases from 580 °C up to 610 °C, the densification becomes well developed, with a reduction in the number of pores and structural refinement of the MgZn₂ phase, as shown in Figure 6 (c), (d), (e) and (f). When the sintering temperature reaches 620 °C, large pores are formed around the Al-Si/SiCₚ.

Because there are two mechanisms, one is remained trace of the boundary surface junction between each powder and another is a newly formed pore by thermal contraction in the sintered body. These phenomena are accelerated by the formation of the liquid phase of the Al-Si alloy. Notably, almost all of the observed pores are distributed around the SiCₚ, and this is the result of the high thermal conductivity and low thermal expansion of the SiCₚ. Additionally, the large pores which are formed at a sintering temperature of 620 °C are caused by the phase transition of the Al-Si alloy from liquid to solid state during the cooling process.

In the case of the MgZn₂ phase change behavior, the distribution of the MgZn₂ phase is limited to the Alumix 431 region in the initial sintering condition in Fig. 6 (a) and (b). Figure 7 shows the dependence of the microstructures in the sintered body on the sintering time. After a sintering time of 15 minutes, it has not fully densified around the Al-Si/SiCₚ or Alumix 431 particles. In addition, the MgZn₂ phase has not spread into the matrix in Fig. 7 (a). On the other hand, Fig. 7 (b), (c) and (d) shows similar densification after 30 minutes of sintering time. The MgZn₂ phase is finely and uniformly dispersed for all of the sintering time conditions. The results of the observed microstructures are in good agreement with the phase analysis results exhibited by the XRD pattern in Fig. 4 (c). It is thought that the Al-Si alloy became liquefied, but the atoms of Mg and Zn did not have enough time to disperse into the surrounding particles. As the sintering time passes 30 minutes, it seems that the atoms of Mg, Zn and Al had enough diffusion time, and the diffusion rate of the atoms for densification and dispersion of precipitation of the MgZn₂ phase at 610 °C.

At the elevated sintering temperature, the MgZn₂ phase begins to disperse in the Al and Al-Si/SiCₚ regions, as can be seen in Fig. 6(c), (d), (e), (f) and (g). The threshold temperature for the diffusional transformation of the MgZn₂ phase seems to be 610 °C, as shown in Fig. 6(f). The phenomena results from the increase densification between the particles and diffusion path of the atoms in the particle with the enough driving force for dispersing by formed liquid phase of the Al-Si alloy at high temperature. Then, during the cooling process after sintering, the MgZn₂ phase starts to finely precipitate throughout the sintered body. These observations are in good agreement with the DTA results in Fig. 3.

Figure 7 shows the dependence of the microstructures in the sintered body on the sintering time. After a sintering time of 15 minutes, it has not fully densified around the Al-Si/SiCₚ or Alumix 431 particles. In addition, the MgZn₂ phase has not spread into the matrix in Fig. 7 (a). On the other hand, Fig. 7 (b), (c) and (d) shows similar densification after 30 minutes of sintering time. The MgZn₂ phase is finely and uniformly dispersed for all of the sintering time conditions. The results of the observed microstructures are in good agreement with the phase analysis results exhibited by the XRD pattern in Fig. 4 (c). It is thought that the Al-Si alloy became liquefied, but the atoms of Mg and Zn did not have enough time to disperse into the surrounding particles. As the sintering time passes 30 minutes, it seems that the atoms of Mg, Zn and Al had enough diffusion time, and the diffusion rate of the atoms for densification and dispersion of precipitation of the MgZn₂ phase at 610 °C.

Figure 8 shows the UTS of the sintered bodies for each
sintering condition. The UTS of the sintered body with 5 wt% Al-Si/SiC<sub>p</sub> is higher than those of the sintered body with Alumix 431 and the other wt% of Al-Si/SiC<sub>p</sub> in the Alumix 431 matrix. The UTS for the various ratios of Al-Si/SiC<sub>p</sub> greater than 5 wt% reveals the tendency of decrease which has lower level of the UTS than that of the sintered Alumix 431 with no added Al-Si/SiC<sub>p</sub> in Fig. 7(a).

Based on the microstructure of the specimen sintered at 610 °C for 30 minutes, the liquid phase of the Al-Si alloy contributes to the densification and dispersion of the MgZn<sub>2</sub> phase throughout the matrix for each ratio of Al-Si/SiC<sub>p</sub>. But the stress is easily concentrated around the formed pores due to the thermal shrinkage produced by the phase transformation in the matrix.

Moreover, the Mg<sub>2</sub>Si phase formed at the interface of each the Al-Si/SiC<sub>p</sub> which has the weak interatomic bonding force. The Mg<sub>2</sub>Si phase is known to be an intermetallic compound like the MgZn<sub>2</sub> phase, and the finely dispersed Mg<sub>2</sub>Si phase contributes to the enhancement of the mechanical properties [29,30]. However, the non-dispersed or coarse Mg<sub>2</sub>Si phase with the Chinese script shape leads to a reduction in the mechanical properties [30]. As a result, the formation of the liquid phase due to the addition of Al-Si/SiC<sub>p</sub> contributes to the densification and dispersion of the MgZn<sub>2</sub> phase.

However, as the ratio of Al-Si/SiC<sub>p</sub> increases, the UTS decreases due to the concentration of applied stress at the pores and the Mg<sub>2</sub>Si phase. Thus, the proper mixing ratio of Al-Si/SiC<sub>p</sub> was determined to be 5 wt%. This was due to the UTS, which was affected by the densification and amount of Mg<sub>2</sub>Si in the sintered body, as shown in Fig. 5.

As the sintering temperature was increased from 560 °C up to 610 °C, the UTS of the sintered body showed a tendency to increase, but it decreased at 620 °C, as shown in Fig. 8 (b). As previously observed in the XRD pattern shown in Fig. 4 (a), the Mg and Zn atoms in Figure 6 were dissolved around particles, and the MgZn<sub>2</sub> phases were finely precipitated in the matrix by the increasing sintering temperature. These helped to enhance the strength of the sintered body by interrupting the transfer of dislocation. This process is known as dispersion-strengthening and solution strengthening. Meanwhile, the decrease in the UTS at 620 °C is due to the concentration of stress on the large pores formed by the phase transformation, as mentioned in Fig. 6 (g).

The UTS of the sintered body for each sintering time at RT is shown in Fig. 8 (c). As the sintering time increases from 15 minutes to 30 minutes, the UTS increased 2 times. As the sintering time increases to more than 30 minutes, the UTS does not change significantly, as shown in Fig. 8(c). Low UTS resides in the remaining prior grain boundary between the powders shown in Fig. 7(a) because these boundaries allow crack propagation to more easily occur in the microstructure, as shown in Fig. 7(a).

Additionally, the MgZn<sub>2</sub> phase did not disperse in the sintered body because it had insufficient diffusion time. In contrast, as the sintering time grew to more than 30 minutes, the Mg and Zn atoms had enough driving force to cause diffusion precipitation and dispersion in the matrix. The UTS of sintered Alumix 431 with Al-Si/SiC<sub>p</sub> are about 80MPa and 40MPa higher than the UTS of Alumix 431 which was sintered (non-hot-pressed) at 600 °C and 620 °C. It is thought that these results were influenced by the formation of the Al-Si liquid phase and reinforcement of the SiC<sub>p</sub> in Al-Si/SiC<sub>p</sub> [32].

Based on these results, in this study the sintering condition was optimized at 610 °C for 30 minutes at a ratio of 5 wt%
Table 5. Qualitative EDS result of fractured surface of the sintered body manufactured at 610 °C during 30 min with mixing ratio (Al-Si/SiCp: Alumix 431) 5:95

(a) at% Al Si Zn Mg Cu Sn C O

<table>
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<tr>
<td>Zn</td>
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<td>Cu</td>
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<td>C</td>
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(b) at% Al Si Zn Mg Cu Sn C O

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<td>02.42</td>
</tr>
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Fig. 10. Wear rates of the sintered body with the optimized sintering condition under each wear condition.

A fractured surface of the sintered body manufactured by the optimized sintering condition has both a transgranular fracture pattern and an intergranular fracture pattern, as shown in Fig. 9 (a). The images in Fig. 9 (b), (c) and (d) are those of Fig. 9 (a) observed at higher magnification, which shows a section of the transgranular fracture pattern. Fig. 9 (c) and (d) shows fractographies of the Alumix 431 area and of the Al-Si/SiCp area, respectively.

Table 5 shows the results of a qualitative analysis for each point in Fig. 9 (c) and (d). Point ① in Fig. 9 (c) exhibits ductile fracture behavior with dimple patterns, and point ② in Fig. 9 (c) shows brittle fracture behavior with cleavage patterns. These fracture behaviors are connected with the relatively ductile Al matrix and the fine MgZn2 phase in the results of Table 5. In addition, point ① in Fig. 9 (d) shows brittle fracture behavior with the cleavage patterns of the SiC particle, and point ② indicates traces of ductile fracture with dimple patterns in the Al-Si alloy, as shown in Table 5. These phenomena indicate that the dissolution of Mg and Zn atoms and the precipitation of the MgZn2 phases in the sintered body lead to an increase in the strength of the sintered body by solid-solution hardening and the precipitation hardening effect. In addition, the SiC particles in the sintered body play an effective role in increasing the strength as a reinforcement. It was verified in this study that the MgZn2 phases and the SiC particles in the sintered matrix have a strong relationship with the enhanced strength of the sintered body against exterior stress.

Figures 10, 11 and 12 each show a separate analytic approach to measuring the wear behavior of the sintered body fabricated with the optimized sintering condition: wear rates, morphological track formation and observation of the tracks cross-section, respectively.

Figure 10 shows changes in the wear rate of the sintered body prepared with the optimized sintering condition. When the vertical load is increased from 5N to 10N, the wear rate at 0.1 m/s linear speed increases, but it is significantly reduced at 20N. At 0.2 m/s linear speed for each vertical load, the wear rate increases along with the increasing vertical load.

Figure 11 shows the wear tracks on the surface of the sintered bodies with the optimized sintering condition under the proposed wear condition. The width of the wear tracks were increased by the plastic deformation on the surface of the specimens, which was produced by increasing the vertical load and the linear speed [33].

In addition, adhesive and abrasive wear behaviors were coincidentally observed in the wear tracks for each condition. Abrasive wear can represent either a two-body abrasion or a three-body abrasion [34]. These wear behaviors were observed in all wear test conditions in this study. The two-body abrasion behavior occurred at the beginning stage. Then the three-body abrasion developed following the inflow of brittle particles, such as debris between the counter ball and the specimen.

Also, the trace of groove and ploughing can be observed on the wear track of the specimen in Fig. 11 [35,36]. A stick-slip phenomenon was produced by the difference between the
Fig. 11. Wear surface of the sintered body at 610 °C during 30 minutes with mixing ratio (Al-Si/\text{SiC_p}: Alumix 431) of 5:95 under each condition (a) 5N with 0.1 m/s, (b) 10N with 0.1 m/s, (c) 20N with 0.1 m/s, (d) 5N with 0.2 m/s, (e) 10N with 0.2 m/s and (f) 20N with 0.2 m/s.

kinetic friction force and the static friction force on the sintered bodies in all vertical load conditions at 0.1 m/s linear speed in Fig. 11(a), (b) and (c). The debris is formed by abrasion or delamination of the specimen. It became stacked on the wear track and introduced a discontinuous change of friction force between the counter ball and the surface of the specimen by disturbing the movement of the counter ball. Finally, it led to the stick-slip phenomenon.

In addition, the spacing of the stick-slip sections grew longer as the vertical load was increased at the 0.1 m/s linear speed. This was caused by the increasing amount of plastic deformation on the surface, as the vertical load and frictional heat increased. And, this reduced the difference between the friction forces because the spacing of the stick-slip sections is getting longer.

At 0.2 m/s linear speed, the stick-slip phenomenon wasn’t observed in Fig. 11(d), (e) and (f). It is likely that the amount of debris remaining on surface of the track was decreased by centrifugal force at the 0.2 m/s linear speed.

On the other hand, the amount and the size of delamination which occurred at the surface of the specimen was increased when the vertical load and the linear speed were increased, as seen in Fig. 11(a), (b), (d), (e) and (f). The large delamination was partially generated by the partial concentrated stress, which exceeded the yield strength of the deformed surface under 5N and 10N of vertical load. Additionally, at 0.2 m/s linear speed, the stress was more rapidly concentrated on the deformed surface of the specimen than at 0.1 m/s linear speed. This led to large delamination by fatigue stress [33]. Meanwhile, the size and the amount of delamination decreased at 20N of vertical load at 0.1 m/s linear speed, as shown in Fig. 11(c).

It is thought that the adhesive behavior was mainly produced by the frictional heat and the applied load. In these results, the wear rate at 20N of vertical load was significantly smaller than at 5N and 10N of vertical load.

Figure 12 shows the cross section of the wear track for each condition. Oxides or oxide layers of O, Zn, Mg and Al were observed with each condition in Fig. 12. As vertical load and the linear speed increased, the oxides developed into an oxide layer due to the applied load and the frictional heat. In addition, as the vertical load increased, the oxide layer became thicker, as seen in Fig. 12(c), because of the active diffusion of atoms in the sintered bodies, which contain oxides in the matrix due to the high frictional heat.

This oxide layer, seen in Fig. 12(c), has a greater hardness than the matrix, and thus contributes to the reduction in the wear rate [37]. The oxide layer that formed on the surface protected the specimen against wear damage, then the oxide layer was worn down by the continuous loading, shown in Fig. 12(a), (b), (c) and (d).

On the other hand, the oxides and the matrix were largely
Fig. 12. Morphological cross-section images of the wear tested specimens at each condition (a) 5N with 0.1 m/s, (b) 10N with 0.1 m/s, (c) 20N with 0.1 m/s, (d) 5N with 0.2 m/s, (e) 10N with 0.2 m/s and (f) 20N with 0.2 m/s.

Delaminated in the previous results in Fig. 11. As the vertical load was increased to 10N and 20N at 0.2 m/s linear speed, cracks could be observed in the sub-surface of the specimen, as in Fig. 12(e) and (f). It is thought that the nucleation and propagation of cracks is generated by fatigue under the repeated loading. Notably, the cracking occurs at the interface between the matrix and the oxides, which have different hardnesses [38,39]. Additionally, this delamination led to an increase in the wear rate in Fig. 10. The wear depth can be observed in the cross section, and is shown in Table 6.

Based on these results, the mechanisms of the wear behaviors mentioned earlier in Figs. 11 and 12 were connected to the wear rate in Fig. 10.

### Table 6. Wear depth at each condition; the vertical load and the linear speed.

<table>
<thead>
<tr>
<th>Linear Speed (m/s)</th>
<th>Vertical Load (N)</th>
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<tbody>
<tr>
<td></td>
<td>5</td>
</tr>
<tr>
<td>0.1</td>
<td>47.98</td>
</tr>
<tr>
<td>0.2</td>
<td>40.87</td>
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</table>

Delaminated in the previous results in Fig. 11. As the vertical load was increased to 10N and 20N at 0.2 m/s linear speed, cracks could be observed in the sub-surface of the specimen, as in Fig. 12(e) and (f). It is thought that the nucleation and propagation of cracks is generated by fatigue under the repeated loading. Notably, the cracking occurs at the interface between the matrix and the oxides, which have different hardnesses [38,39]. Additionally, this delamination led to an increase in the wear rate in Fig. 10. The wear depth can be observed in the cross section, and is shown in Table 6.

Based on these results, the mechanisms of the wear behaviors mentioned earlier in Figs. 11 and 12 were connected to the wear rate in Fig. 10.

### 4. CONCLUSION

The sintering condition of Al-Si/SiCp was optimized using the measured UTS, and analyses of microstructure dependence on three variables: the Al-Si/SiCp / Alumix 431 ratio, the sintering temperature and the sintering time. The wear properties of the sintered body manufactured using the optimized sintering condition was then observed for various vertical loads and linear speed.

Al-Si/SiCp consists of SiC particles, primary Al and Al-Si eutectic-like structures, and the Alumix 431 powders consist of pure Al and an Al-Zn-Mg alloy powder which contains the Mg$_{32}$(Al, Zn)$_{49}$ phase.

The highest UTS was observed in the sintered body containing 5 wt% of added Al-Si/SiCp and sintered at 610 °C for 30 minutes, due to dispersion of the MgZn$_2$ phase and densification produced by the formation of the liquid phase of the Al-Si alloy. On the basis of these results, the optimized sintering condition for the 5 wt% Al-Si/SiCp was determined to be a sintering temperature of 610 °C for 30 minutes.

In the fractography of the sintered body manufactured by the optimized sintering condition, intergranular fractures and transgranular fractures occurred together. Ductile fracture behavior was exhibited in the matrix of the sintered body and brittle fracture behavior was observed at the SiC particles and the MgZn$_2$ phases of the sintered body.

In wear tests at 0.1 m/s of linear speed, oxides were formed and the stick-slip phenomenon occurred. In addition, as the vertical load increased, the spacing of the stick-slip sections and the thickness of the oxide layer increased. When the vertical load was increased to a 0.2 m/s linear speed, the
oxide layer, and the nucleation and propagation of cracks were observed in the sub-surface. These wear behaviors corresponded to the wear rate.

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REFERENCES