



The Effect of Heat Treatment on the Mechanical Properties and Oxidation Resistance of AlSi10Mg Alloy

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Abstract

In this study, the microstructure, hardness, tensile strength, and dry wear properties of the cast AlSi10Mg alloy as well as the effects of processing parameters on the oxidation behavior of this alloy were investigated. In this context, AlSi10Mg (wt%) alloy was produced by gravity die casting method, and then, air cooling, quenching, and T6 treatment were applied. The microstructure of the alloy was investigated using an optical microscope, SEM, and EDS. Additionally, thermogravimetric analysis (TGA) was used to determine the effect of processing parameters on the oxidation behavior of the samples. The friction and wear properties of the alloy were investigated using a ball-on-disk wear test device. It was founded that the as-cast AlSi10Mg alloy's microstructure contained phases of α (Al), Si, and β -Mg₂Si. It has been observed that the microstructure of the cast AlSi10Mg alloy gradually turns into a spherical form when quenching and air-cooling heat treatments are applied. Also, T6 heat treatment led to further spheroidization of Si particles, formation of β -Mg₂Si precipitates, and elimination of the Chinese script morphology of the β -Mg₂Si phase. It was observed that the yield strength of the AlSi10Mg alloy in the as-cast state decreased negligibly compared to the air-cooled state, while there was an increase of up to 70% in the aging-treated condition. The results show that the tensile strength of the as-cast AlSi10Mg alloy increased by 115% after aging. Additionally, it was found that the AlSi10Mg alloy's wear resistance increased with the aging process.

Keywords AlSi10Mg · Wear properties · Heat treatment · Microstructure

1 Introduction

Aluminum alloys are one of the most widely used materials after steel in automotive manufacturing, aerospace, and other fields due to their low density, high specific strength, and good thermal and electrical conductivity [1–3]. It is well known that even with all of these advantages, binary aluminum alloys are not suitable for all applications. Therefore, various methods are applied to increase the strength and

ductility properties of aluminum alloys used in industrial applications. The most important of these methods is element addition to the primary aluminum phase [4–7]. Aluminum's ability to form compounds with other chemical elements has led to the development of various alloys (including Al–Cu, Al–Si–Mg, Al–Mg–Zn, Al–Mg, and Al–Si) [8–10]. Aluminum–silicon alloys are currently the most widely used of these alloy types. Because of their exceptional qualities, including high specific strength, low thermal expansion coefficient, high fluidity, and high resistance to corrosion and wear, Al–Si alloys are extremely important in the automotive industry [5–10]. These alloys contain elements like copper and magnesium, which dissolve in the α -Al phase and form precipitates [10–13]. The θ (CuAl₂) phase consisting of Cu and Al and the β -Mg₂Si phase consisting of Mg and Si are the main phases observed from these alloys. Addition of more than 1% of Mg results in a volume fraction of the β -Mg₂Si phase [10–13]. The hard and brittle β -Mg₂Si phase in the microstructure reduces the tensile strength of the alloy and causes crack tendency. For this reason, the magnesium content of aluminum–silicon alloys is kept below 1%

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[10–13]. It has been determined that the microstructure of the Al–10Si–Mg alloy in as-cast state consists of α -Al phase, needle silicon particles, and skeletal Mg_2Si phase [5–8, 11–16]. After applying an air-cooling heat treatment to the Al–10Si–Mg alloy, it was observed that the Mg_2Si skeletal phase was largely eliminated. Subsequently, the quenching and aging heat treatments caused the needle silicon particles to become spherical. In industrial applications, many alloying elements are added to aluminum–silicon alloys [14–17]. These additive elements can be listed as copper, magnesium, and zinc, respectively. 1–4 (wt%) copper additive forms the θ (CuAl_2) phase in the α -Al phase and increases the strength and hardness of the alloy [18–23]. In addition, zinc improves the strength properties of the alloy as it dissolves in the α -Al phase. The addition of magnesium to this alloy causes the formation of the Mg_2Si phase in the matrix, which increases the alloy's wear resistance [18–23]. Studies have shown that adding up to 1% Mg to the Al–Si alloy results in the creation of the β - Mg_2Si phase, alters the morphology of silicon particles, and converts the plate-like AlFeSi phase into a more ordered π - $\text{Al}_8\text{Mg}_3\text{FeSi}_6$ phase. It has also been reported that these changes increase the alloy's strength [20–22]. In another study, it was determined that the addition of more than 1.5% Mg increased the ratio of β - Mg_2Si and π phases in the microstructure [15, 16, 24, 25]. These phases reduce the load-bearing ability of the alloy and form internal cracks, negatively affecting the tensile properties [15, 16, 24, 25].

There are numerous studies in the literature on the production of Al–10Si–Mg alloys. However, in these studies, the effect of process conditions on the microstructure and mechanical properties of the alloys was investigated [24–27]. Based on the literature studies, it was concluded that applying different heat treatments to the Al–10Si–Mg alloy did not fully reveal its friction and wear properties. Therefore, the aim of this study is to investigate the effect of heat treatment on the microstructure, strength, wear behavior, and oxidation resistance of Al–10Si–Mg alloy and to determine optimum data for tribological applications.

2 Experimental

The initial Al10SiMg alloy used in this study was fabricated by using gravity die casting method. The chemical composition of the alloy was detected by the spectral analysis method (Table 1).

For microstructural investigations, alloy samples were subjected to sanding, polishing, and etching processes. Etching was performed by electro-etching with fluoroboric acid using the Struers Lectropol device. Al10SiMg alloy was melted at 700 °C using an electrical resistance furnace. The melted alloy was mixed for 5 min and then poured into a mild steel mold at 690 °C. The dimensions of the casting mold are

Table 1 Chemical composition of Al–10Si–Mg alloy

Alloys	Chemical composition (%wt)				Density (g/cm ³)
	Al	Si	Mg	Other	
Al–10Si–Mg	88.8	9.77	0.385	1.045	2.6360

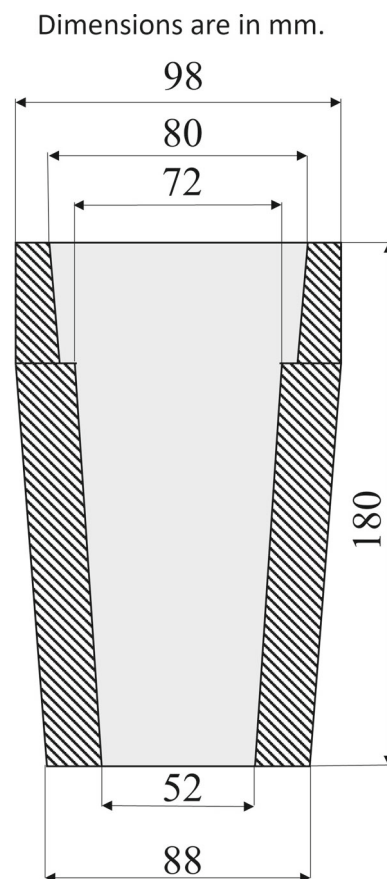


Fig. 1 Shape and dimensions of the casting mold

180 mm long and conical, with a lower area of 52 mm and an upper area of 72 mm. The technical drawing of the casting mold is shown in Fig. 1. The spectral analysis method was used to ascertain the alloy's chemical composition as soon as solidification was finished. To determine the alloy's properties, it was subjected to 2 hours of solution heat treatment at 560 °C, followed by 5 hours of quenching and aging at 180 °C [28, 29]. The microstructures of the alloy were investigated using a Zeiss Axio Observer A1m optical microscope. Density was determined by dividing the mass of the alloy by the sample volume. Microhardness values of the alloy were measured with the STRUERS DURAMIN A/S DK-2750 device, and Brinell hardness values were determined with the Innovatest Nemesis 9000 Universal Hardness Tester. Average values were obtained by taking at least five measurements

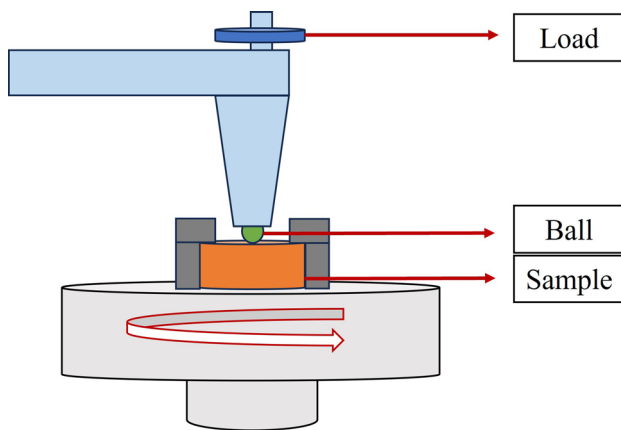


Fig. 2 Schematic of the wear test device

from each sample in the microhardness and hardness tests. The yield and tensile strengths of the alloy were measured by making at least five measurements at a deformation rate of $5 \times 10^{-4} \text{ s}^{-1}$ using the MTS Criterion Model 45 universal testing device in accordance with ASTM:E8 standards.

The friction and wear properties of the alloy were investigated using a UTS Design Tribology wear device, based on ball-on-disk type and meeting ASTM: G99 standards. The schematic image of the friction and wear test device is given in Fig. 2. 100Cr6 steel balls were used to abrade the samples. In friction and wear tests, the test speed and load were 0.16 m/s and 5 N, respectively, and the sliding distance was between 250 and 1500 m. Each sample that was tested for friction and wear was cleaned in an ultrasonic cleaning device using a mixture of alcohol + trichlorethylene. The mass loss of the worn samples was measured with a balance with a sensitivity of $\pm 0.01 \text{ mg}$. The volume loss was detected by dividing the measured mass losses by the density of the alloy. SEM images and EDS analysis of the sample surfaces were investigated using the Zeiss Evo LS10 device. Thermogravimetric analysis (TGA) is used to determine the oxidation behavior of samples. PerkinElmer TGA4000 was used for TGA analysis. The experimental procedure of the TGA was specified as 30–650 °C temperature, 20 °C/min heating rate and air.

3 Results and Discussion

3.1 Microstructure

Figure 3 shows the microstructure of the as-cast Al–10Si–Mg alloy comprised of α -Al and Mg_2Si phases and Si fragments. The Si particles in the alloy's microstructure were found to be needle-like and block-shaped appearance (Fig. 3a), whereas the Mg_2Si phase appeared to be skeletal, as shown in Fig. 3b. It was shown that the Si particles in the microstructure of the

alloy exhibited a needle-like and block-shaped appearance, while the Mg_2Si phase exhibited a skeletal appearance. It is known that needle-shaped Si particles in the alloy are formed as a result of eutectic transformation, and block-shaped Si particles are formed directly spontaneously during solidification [30–32]. When the solidification behavior of Al–10Si–Mg alloys in the casting state is examined, the first $\alpha(\text{Al})$ nucleus is formed when the molten alloy begins to cool. As the temperature decreases, the $\alpha(\text{Al})$ phase begins to grow. As the chemical composition of the Al–10Si–Mg alloy approaches the eutectic point, silicon, and magnesium react to form $\beta\text{-Mg}_2\text{Si}$, and solidification is completed [26, 27]. The formation of the $\beta\text{-Mg}_2\text{Si}$ phase reduces silicon concentration in molten solution. This prevents the growth of silicon particles and causes the formation of finer silicon particles [23–27]. There is no important change in the microstructure of the air-cooled Al–10Si–Mg alloy samples compared to the as-cast condition (Fig. 3b). However, it was determined that there was a relative increase in the $\alpha(\text{Al})$ phase ratio in the internal structure of the air-cooled sample. It is thought that this situation may be related to the continuous withdrawal of the Mg_2Si phase from the liquid metal during solidification.

The microstructure of the Al–10Si–Mg alloy in the quenched and aged state shows that the silicon particles become relatively spherical and the $\beta\text{-Mg}_2\text{Si}$ phase disappears (Fig. 3c, d). However, it was observed that there were more primary silicon particles in the quenched alloy and these particles tended to cluster. The reason why the primary silicon particles in the quenched state of the Al–10Si–Mg alloy is higher than in the cast state of the alloy may be due to the growth of needle-shaped eutectic silicon particles during the dissolution and quenching process. The disappearance of the Mg_2Si phase in the structure can be explained by the high diffusion rate of magnesium in the $\alpha(\text{Al})$ phase [26, 27]. During aging, Mg and Si atoms are withdrawn from solution, forming unstable fine Mg_2Si precipitates [23–27]. On the other hand, it has been stated in other studies that silicon atoms display a faster diffusion rate than Al atoms [14–19, 26, 27]. This causes the needle-shaped silicon particles to break down and turn into a relatively spherical form and be distributed homogeneously throughout the structure.

3.2 Mechanical Properties

The yield and tensile strength, hardness, and microhardness curves of the as-cast, air-cooled, quenched, and aged Al–10Si–Mg alloy samples are given in Fig. 4, respectively. Moreover, Table 2 shows numerical values of mechanical properties of the Al–10Si–Mg alloy samples. It is seen that the hardness and microhardness values of the alloy exhibit similar changes with yield and tensile. Additionally, as shown in Fig. 4, samples that were allowed to cool in air demonstrated a slight decrease for all values, whereas samples



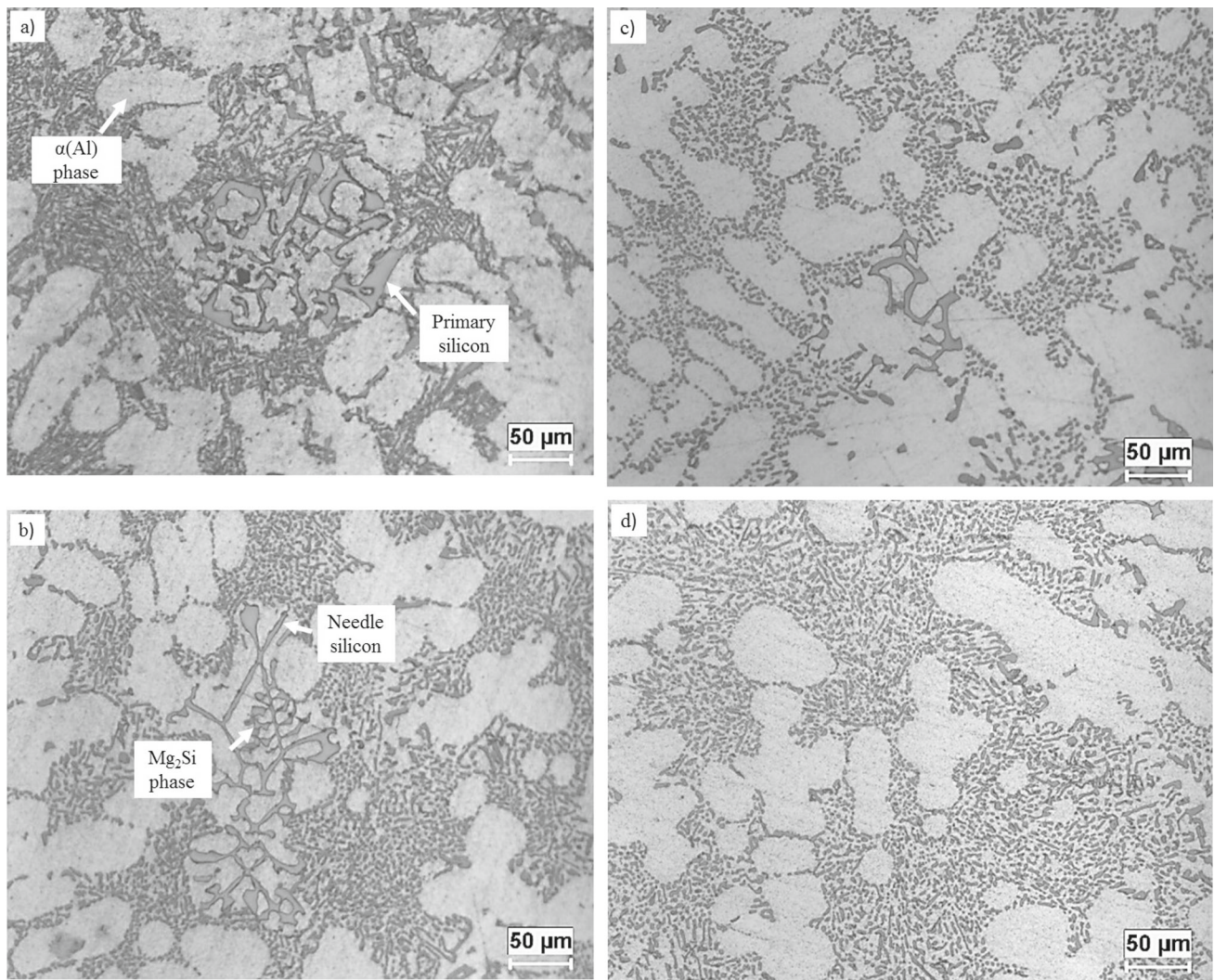
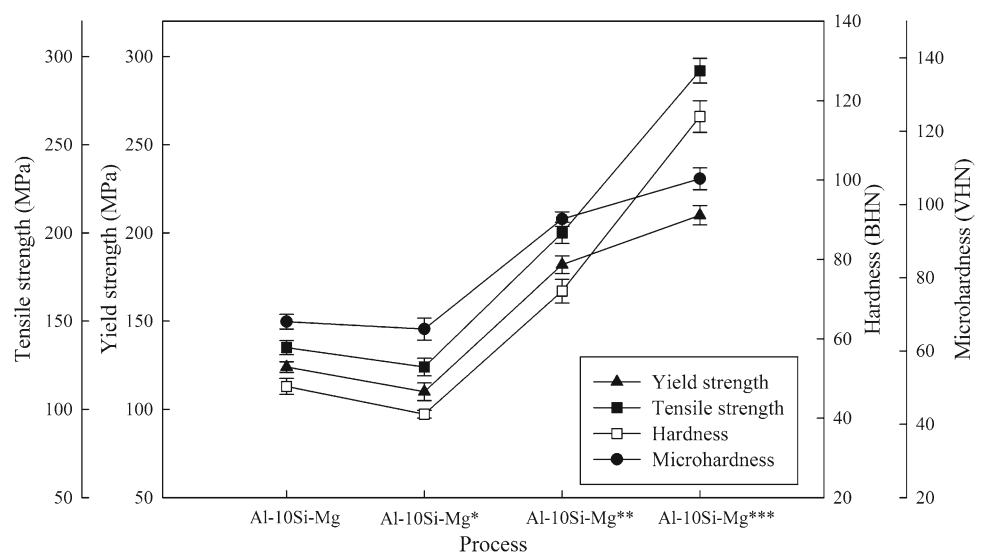


Fig. 3 Microstructure of Al-10Si-Mg alloy: **a** as-cast, **b** air-cooled, **c** quenched, and **d** aged treatment state

Fig. 4 The change of the tensile, yield strength, hardness, and microhardness of Al-10Si-Mg alloy samples



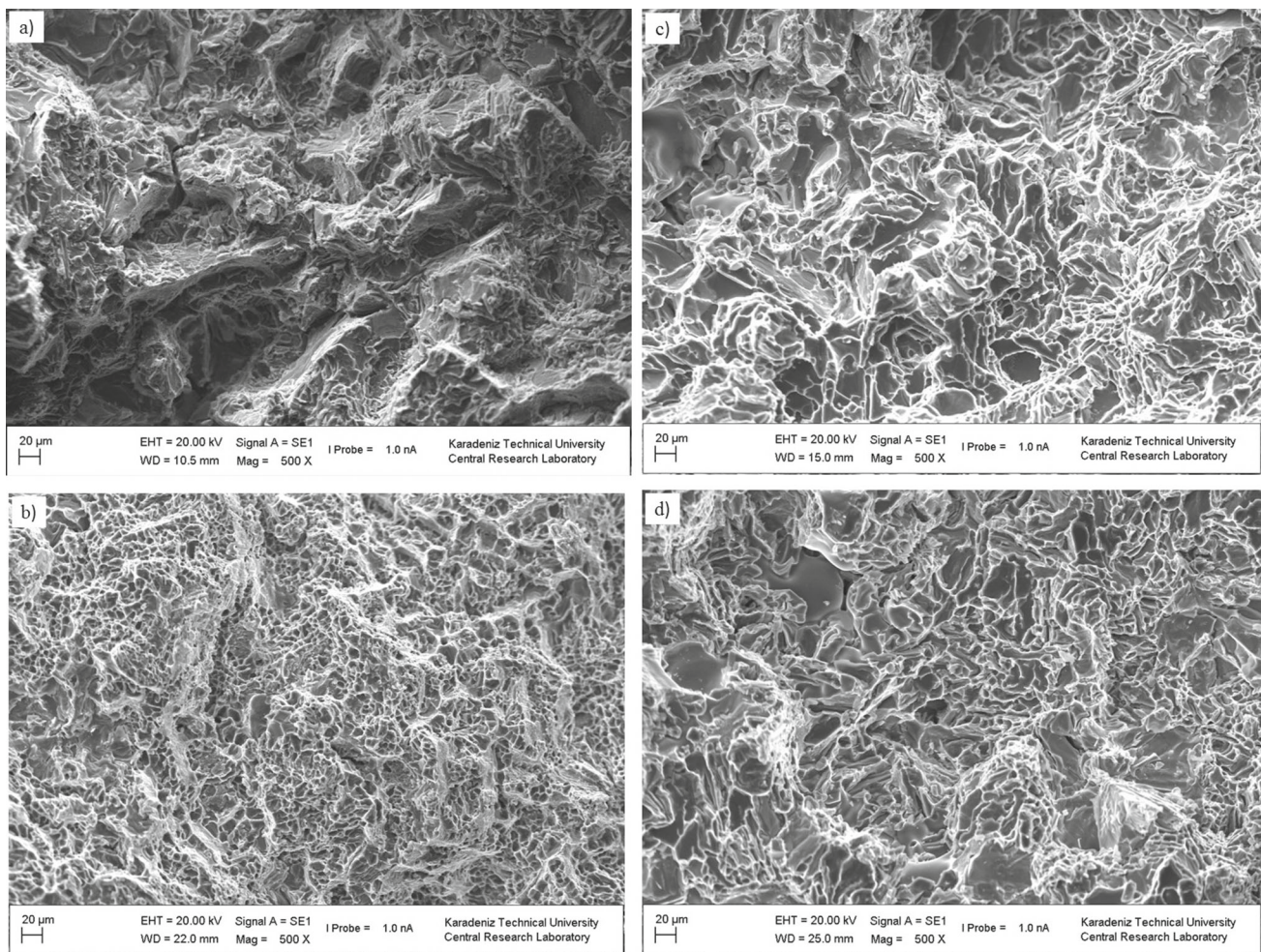


Fig. 5 SEM photographs of fracture surfaces of the alloys with different heat treatments: **a** as-cast, **b** air-cooled, **c** quenched, and **d** aged treatment state

Table 2 Yield, tensile strengths, hardness, and microhardness values of Al-10Si-Mg alloy samples

Alloys	Mikrohardness (VHN)	Hardness (BHN)	Yield strength (MPa)	Tensile strength (MPa)
Al-10Si-Mg	68 ± 2	48 ± 2	124 ± 3	135 ± 4
Al-10Si-Mg*	66 ± 3	41 ± 1	110 ± 5	124 ± 5
Al-10Si-Mg**	96 ± 2	72 ± 3	182 ± 5	200 ± 6
Al-10Si-Mg***	107 ± 3	116 ± 4	210 ± 5	292 ± 7

(*) Air-cooled treatment, (**) quenched treatment, and (***) aged treatment

that were quenched and aged demonstrated a continuous increase. These changes can be explained by examining the microstructures of the Al-10Si-Mg alloy in different heat treatments [14–19]. In contrast to the cast state, the amount of primary Si in the Al-10Si-Mg alloy decreases when it is air-cooled. Considering the load-carrying effect of primary silicon particles in the microstructure, it is expected that there will be some decrease in hardness and strength as this effect decreases. In the quenched and aged states of the alloy, the

relatively spherical homogeneous distribution of silicon particles within the microstructure increases the hardness and strength of the alloy samples. In addition, the disappearance of the Mg_2Si phase, which has a skeletal phase appearance in the microstructure, reduces the crack tendency and increases the strength values [23–27]. In addition, it was observed that after quenching, the primary silicon content decreased and there was a relative increase in the amount of $\alpha(Al)$ in the microstructure (Fig. 3c). The presence of primary silicon



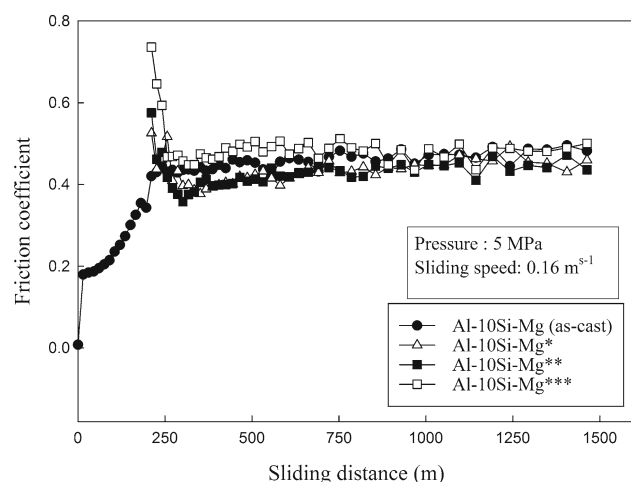


Fig. 6 The change of friction coefficient of as-cast, air-cooled, quenched, and aged state Al-10Si-Mg alloy samples

in the alloy caused a partial increase in the yield strength (Fig. 4). The aging process of the sample resulted in a partial improvement in microhardness due to the growth of the $\alpha(\text{Al})$ phase in its internal structure, but it also significantly increased the hardness values (Fig. 4b).

Fracture surfaces of alloys, which subjected to various heat treatments, are shown in Fig. 5. It was observed that the fracture surfaces of the alloy comprise of a small amount of cleavage planes (CP) and mostly tear ridges (TR). It should be noted that the CP consists of hard phases with silicon and magnesium, while the tear ridges have soft $\alpha(\text{Al})$ phases. Similar fracture surface characteristics were determined in all alloys subjected to different heat treatments. However, the transformation of silicon particles in the microstructure of quenched and aged samples into a relatively spherical shape causes more cleavage planes on the fracture surfaces (Fig. 5c, d).

3.3 Wear Behavior

Figure 6 shows the variation in the Al-10Si-Mg alloy's friction coefficient values depending on the sliding distance in both the as-cast and heat-treated states. The friction coefficient of the alloy initially increased sharply for each sample and reached a relatively equilibrium state. This situation can be explained by the tribological behavior of the materials [33–35]. At the beginning of friction and wear experiments, the contact surface between the ball and the sample is point-like gave rise to high pressures. The adhesion bonds formed between the contact surfaces create resistance to sliding, causing the friction coefficient to increase suddenly [33–35]. As friction continues on the sample surfaces, the contact area increases and reaches equilibrium after a certain period of

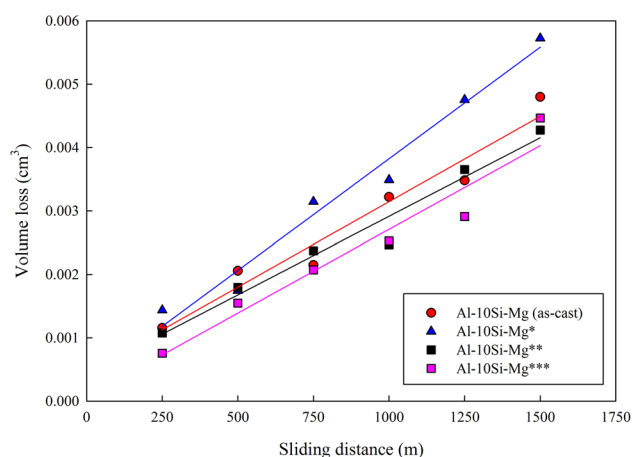


Fig. 7 The change of volume loss of as-cast, air-cooled, quenched, and aged treatment state Al-10Si-Mg alloy samples by sliding distance

time. However, particles broken off from the material surface during friction enter between the ball and the sample surface. This causes a variable coefficient of friction.

Figure 7 shows the volume loss values for the Al-10Si-Mg alloy in its cast and heat-treated states as a function of sliding distance. From this Fig. 7, it can be noted that the volume loss increases linearly with increasing sliding distance for each case. It was determined that the alloy that was subjected to aging heat treatment, quenched, cast, and cooled in air exhibited the highest wear resistance, respectively. This observation can be explained depending on the microstructure of the alloy, its hardness, and Archard's equation [31–33, 36–39]. As mentioned above, when heat treatment is applied to the Al-10Si-Mg casting alloy, silicon particles turn into a relatively spherical shape and spread homogeneously into the microstructure. This increases the hardness in the microstructure of the alloy and improves the wear resistance due to the load-carrying ability of hard and brittle silicon particles. Moreover, according to Archard's equation, volume loss increases as sliding distance increases and decreases as hardness and strength increase [36–39]. Since the hardness and strength increase when heat treatments are applied to the alloy, volume loss is expected to increase as the sliding distance increases [36–39].

The change in volume loss value of Al-10Si-Mg alloy in both cast- and heat-treated states depending on pressure is given in Fig. 8. This alloy samples exhibited a gradually increasing volume loss characteristic in all cases, and it was determined that the lowest volume loss was the sample with aging heat treatment. This observation can be explained based on hardness and strength. Hard and brittle silicon particles dispersed homogeneously and relatively spherically within the microstructure have a load-bearing effect, causing an increase in wear resistance [40–43]. In addition, the needle-like and block-shaped silicon particles in



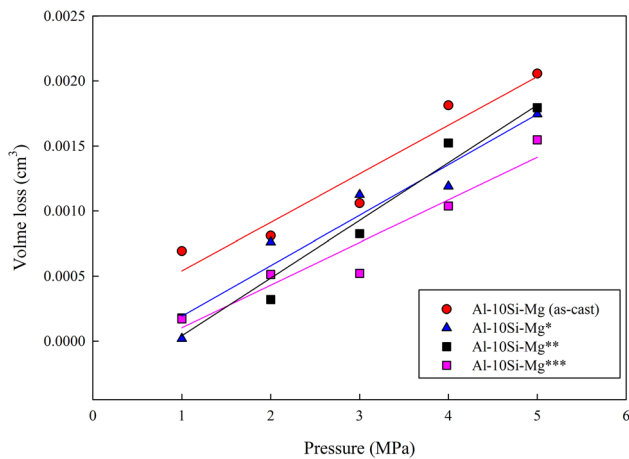


Fig. 8 The variation of volume loss of as-cast, air-cooled, quenched, and aged state Al-10Si-Mg alloy samples by pressure

the microstructure of the as-cast and air-cooled samples may have created a notch effect and caused the particles to break off due to the increasing pressure from the sample surface.

This situation negatively affects the wear resistance of cast and air-cooled samples compared to quenched and aged samples.

SEM images showing the wear surfaces of the as-cast and heat-treated states of the Al-10Si-Mg alloy, the particles morphologies obtained from the wear tests, and the ball surface used in the wear tests in the as-cast state are given in Figs. 9, 10, and 11, respectively. Since the wear surfaces, debris pictures and ball surface pictures in different heat-treated states of the alloy are similar, representative as-cast alloy pictures are given. As a result of friction and wear tests, it is observed that the wear surfaces of the sample consist of scratches and delamination layers (Fig. 9). It can be seen that the ball surface is relatively covered with wear material, and the wear particles comprise of large and small particles. The EDS analysis results of the materials adhering to the sample and ball surfaces and the debris are given as a table within the SEM images. According

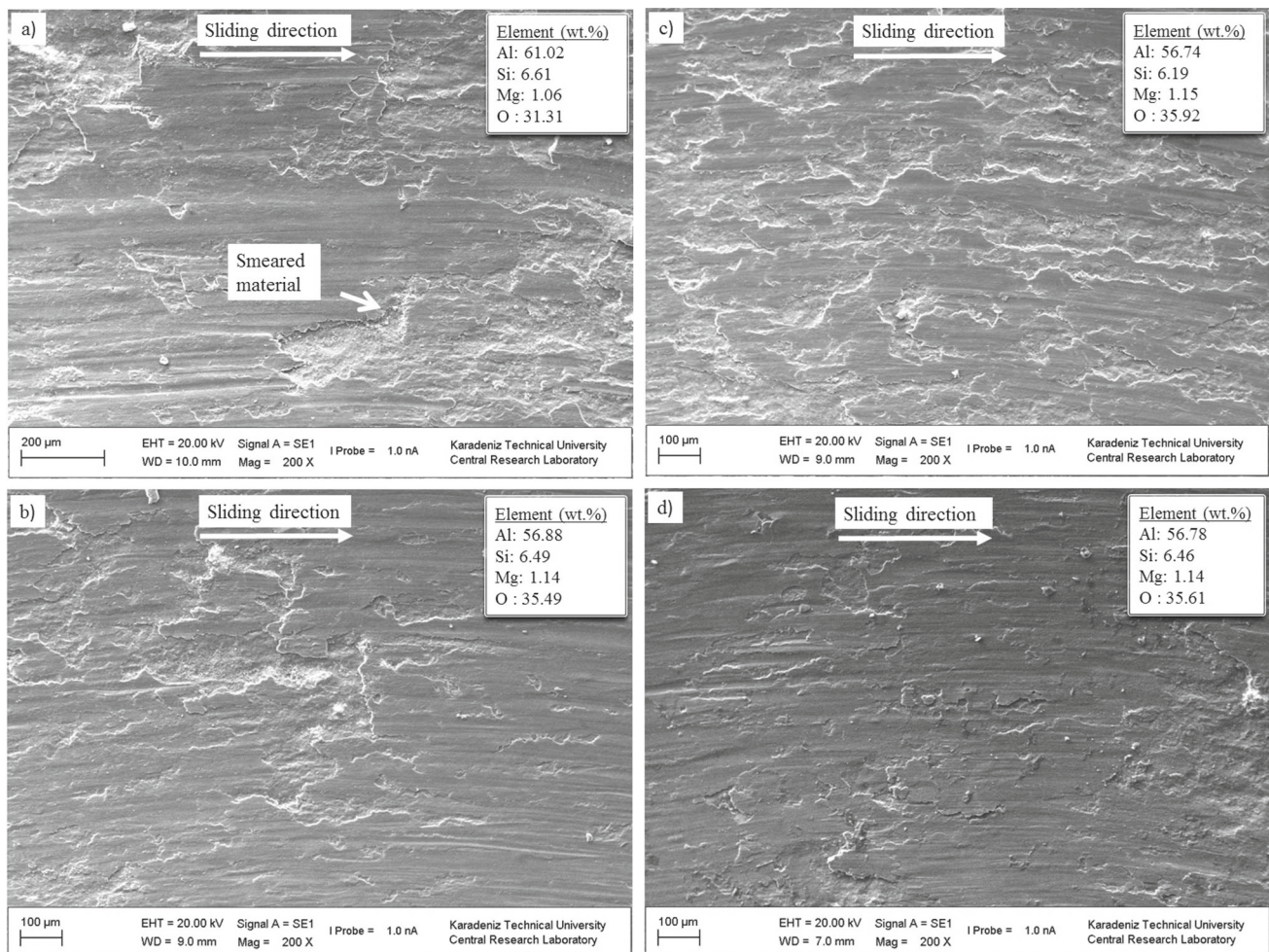


Fig. 9 SEM images with EDS analysis of worn surface of the alloy: **a** as-cast, **b** air-cooled, **c** quenched, and **d** aged treatment



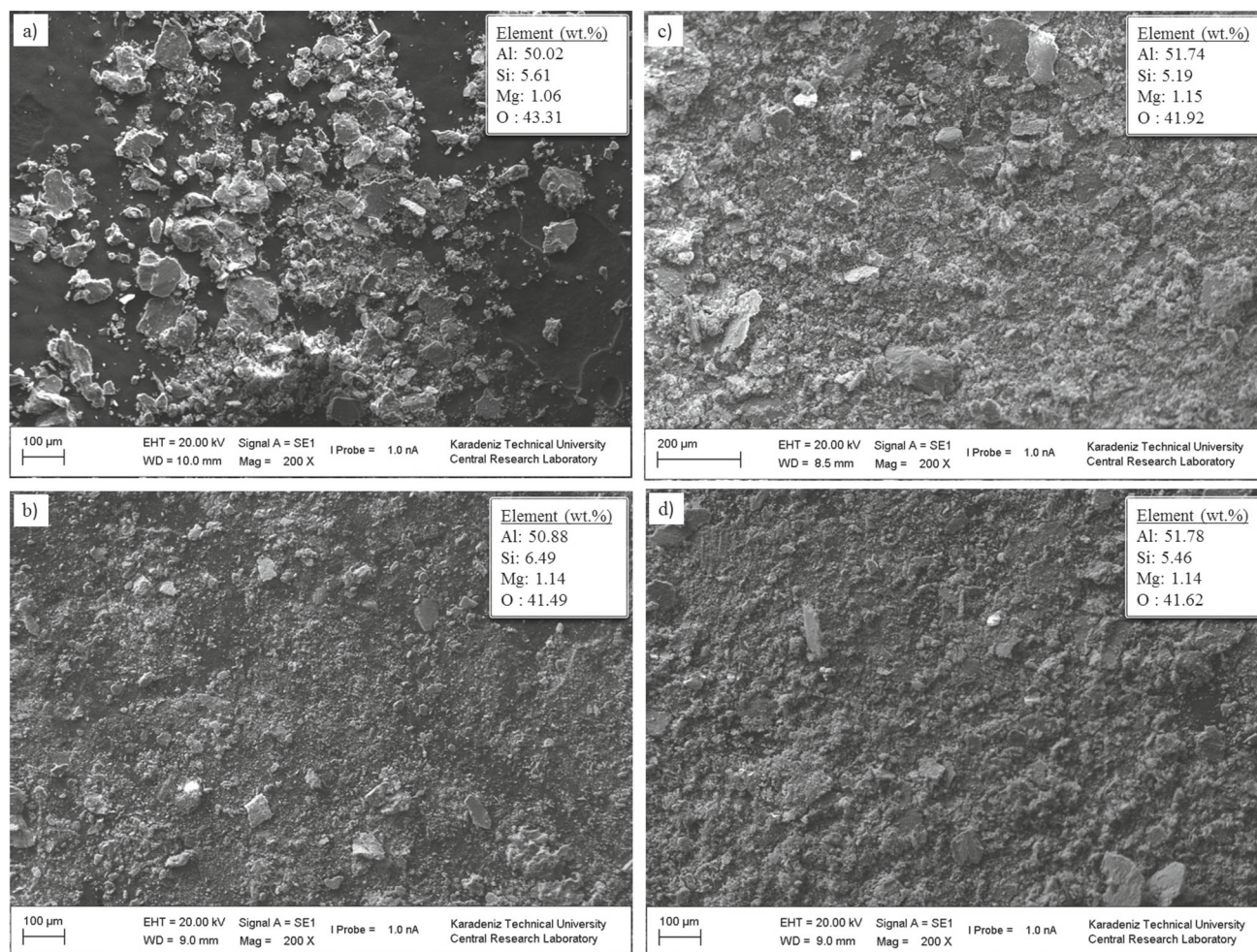


Fig. 10 SEM images with EDS analysis of debris of the alloy: **a** as-cast, **b** air-cooled, **c** quenched, and **d** aged treatment

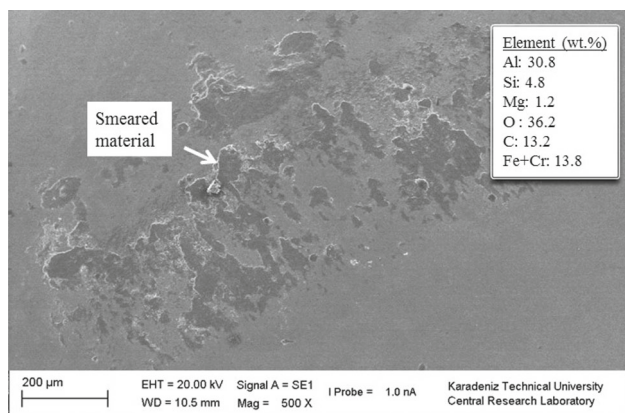


Fig. 11 SEM images and EDS analysis of ball surface after wear test of 1500 m for just as-cast Al-10SiMg alloy

to this table, it was determined that the chemical composition consists of the elements aluminum, silicon, magnesium, iron, chrome, and oxygen. These examinations can be associated with the microstructures of the Al-10Si-Mg alloy

in its as-cast and heat-treated states. The microstructure of Al-10Si-Mg alloy consists of soft α (Al) matrix, hard silicon particles, and Mg_2Si phase [26, 27, 30–32]. The hard and brittle Si particles are easier to smear onto the sample surface thanks to their superior load-carrying abilities. On the contrary, since the α (Al) phase is soft, it is more smeared to the ball surface. In friction wear tests, initially point contact and high-pressure cause particles to break off from the sample surface. Some of these particles adhere to the sample surface and some to the ball surface, causing an adhesion layer to form on the surface. This adhesion layer becomes brittle as oxidation increases. Thus, delamination layers are formed. In addition, hard and brittle particles within the microstructure of the alloy cause cracks on the wear surfaces. These cracks progress under the influence of oxidation and break off from the sample surface. Some of the broken pieces move away from the sample surface under the effect of centrifugal force, while the other part turns into dust between the ball and the sample surface.

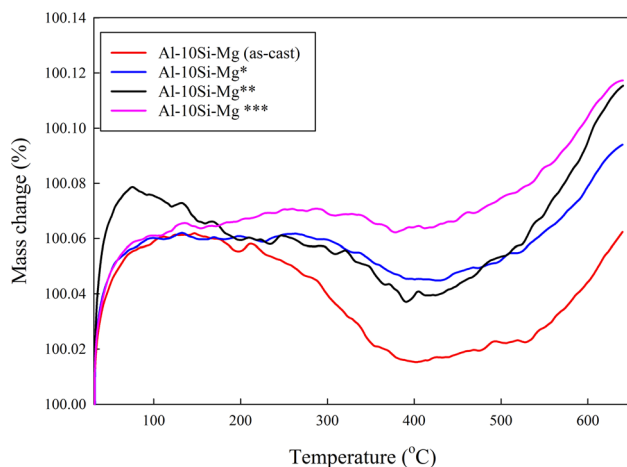


Fig. 12 TGA curves of the samples of Al-10Si-Mg alloy samples

3.4 Oxidation

Figure 12 shows the results of thermogravimetric analysis of Al10SiMg alloys subjected to casting, air cooling, quenching, and aged treatment processes, respectively. As can be seen in Fig. 12, no significant mass increase was observed in the TGA experiments of all test specimens up to 650 °C. As a result of the oxidation experiments, the highest mass increase was 0.12%. This can be attributed to the Al_2O_3 layer formed on the Al surface with increasing temperature. Oxide layers formed on the metal surfaces can be evaluated as protective and non-protective oxide layers. The most concrete expression related to this observation can be explained with the Pilling–Bedworth ratio. The Pilling–Bedworth ratio is a rule used to determine the oxide films formed by metals as protective or non-protective and was presented by N.Pilling and E.Bedworth. The formula for the Pilling–Bedworth ratio is given in Eq. 1 [44, 45].

$$\text{Pilling–Bedworth Ratio} = \frac{M_{\text{oxide}} \times d_{\text{metal}}}{n \times M_{\text{metal}} \times d_{\text{oxide}}} \quad (1)$$

where M is the molecular mass, n is the metal atom per one oxide molecule, d is the density.

Calculating the Pilling–Bedworth ratio for the oxidation of aluminum;



It is known that the molar mass of aluminum is about 26.98 g/mol and molar mass of Al_2O_3 is about 101.96 g/mol. Density of aluminum is 2.70 g/cm³, and density of Al_2O_3 is 3.95 g/cm³ [46]. Accordingly, the Pilling–Bedworth ratio for

aluminum was calculated using Eq. 3:

$$\text{Pilling–Bedworth Ratio of Aluminum} = \frac{M_{\text{Al}_2\text{O}_3} \times d_{\text{Al}}}{n \times M_{\text{Al}} \times d_{\text{Al}_2\text{O}_3}} \quad (3)$$

$$\begin{aligned} \text{Pilling–Bedworth Ratio of Aluminum} \\ = \frac{101.96 \text{ g/mol} \times 2.7 \text{ g/cm}^3}{2 \times 26.98 \text{ g/mol} \times 3.95 \text{ g/cm}^3} = 1.29 \end{aligned} \quad (4)$$

If the Pilling–Bedworth ratio is less than 1, the oxide film formed by the metal is very thin and has no protection. If the Pilling–Bedworth ratio is greater than 2, the oxide film is extremely thick and easily disintegrates. However, if the Pilling–Bedworth ratio is greater than 1 and less than 2, the oxide film formed by the metal is protective. When the Pilling–Bedworth ratio is calculated for the oxide layer formed by aluminum metal, a value of approximately 1.29 is reached. This proves that aluminum forms a protective oxide layer [45]. According to the literature studies, it is understood that the oxidation mechanism for aluminum proceeds very slowly up to a temperature of approximately 550 °C and that there is a not very high oxidation between 550 and 600 °C. This is due to the protective oxide layer formed by aluminum, as explained by the Pilling–Bedworth ratio. In other words, aluminum undergoes oxidation with increasing temperature, but the progress of this oxidation is prevented by the amorphous oxide layer (Al_2O_3) formed on the aluminum surface. Thus, aluminum can remain resistant to oxidation almost up to its melting temperature [47, 48].

In the present study, the effects of the production of AlSiMg10 alloy by casting method and subsequent secondary treatments on oxidation resistance were investigated. In this context, it is understood that all test specimens have a very low oxidation rate, i.e., very high oxidation resistance, in accordance with the literature studies (Fig. 12). In addition, the relative effects of microstructural changes due to casting and subsequent secondary processing can also be understood from the TGA results. When the wt% change is examined up to a temperature value of approximately 400 °C, there is a small mass decrease for all samples. This is a similar situation to that encountered in the literature and in previous studies. Researchers have interpreted it as a situation encountered due to the removal of water or other residues from the structure. Park et al. observed a decrease in the wt% values of Al up to about 600 °C in their study, which they interpreted as the removal of moisture and impurities absorbed by the sample during the experiment by thermal experiment and thus a decrease in mass [49, 50].

The cross-sectional investigation of the Al-10Si-Mg (as-cast) and Al-10Si-Mg*** coded sample is shown in Fig. 13. Figure 13 shows that the oxidized surface contains nearly three times more oxide than the polished surface. The reason



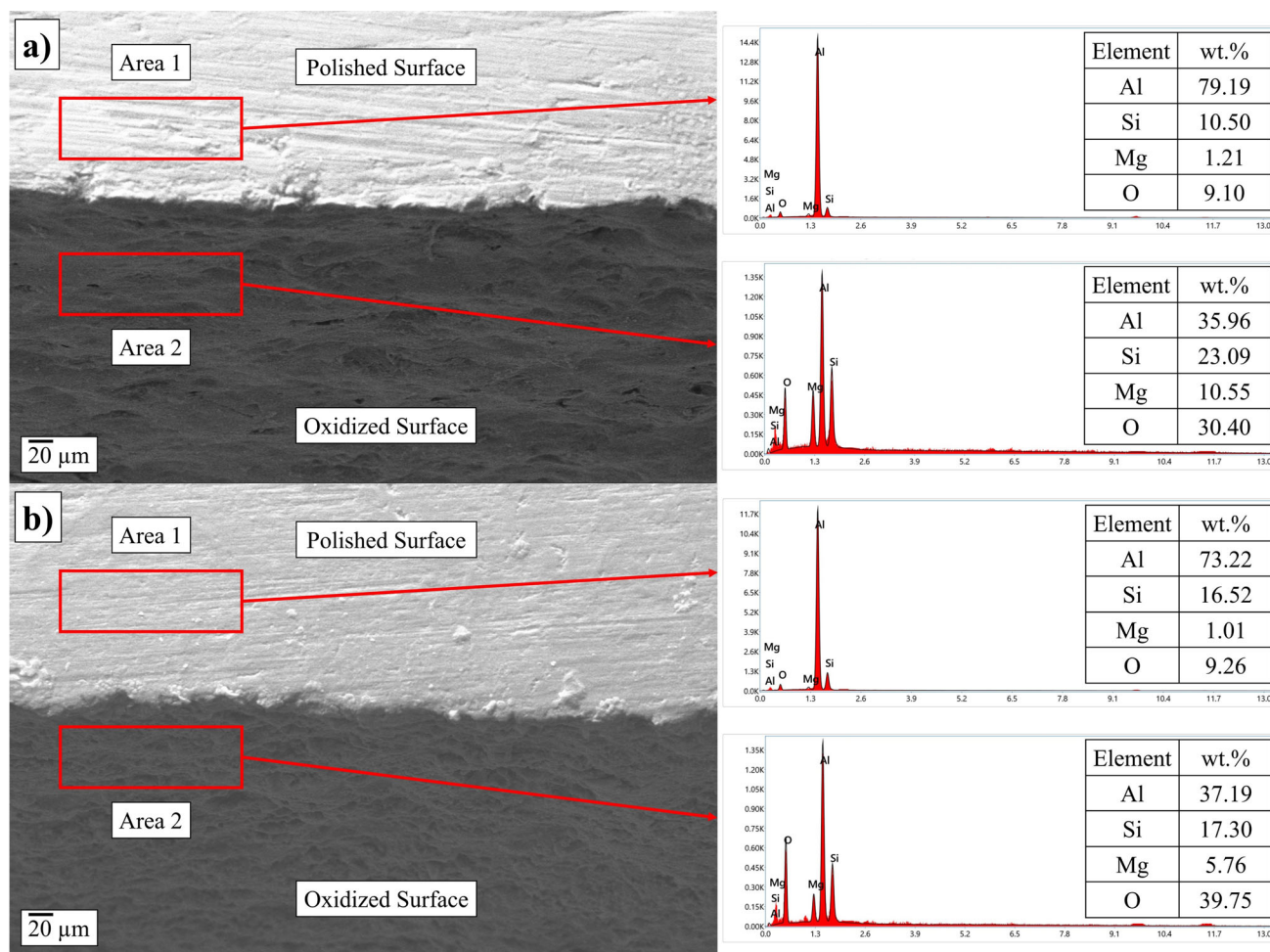


Fig. 13 Cross-sectional investigation and EDS analysis of the Al-10Si-Mg (as-cast) (a) and Al-10Si-Mg*** coded samples (b)

for the presence of oxide on the polished surface is that Al can easily form a protective oxide layer on its surface even at room temperature in an open atmosphere. In addition, when the EDS spectrum analysis of the surfaces and the wt% values of the elements are examined, it is clear that the amount of oxide is still lower than the amount of Al even on the oxidized surface in parallel with the TGA results. This confirms that the oxide layer formed by Al with O is a protective oxide layer.

4 Conclusions

1. The microstructure of the cast Al-10Si-Mg alloy consists of α (Al), eutectic silicon, primary silicon, and β -Mg₂Si phases. Due to the aging heat treatment, the silicon particles in the Al-10Si-Mg alloy microstructure become comparatively spherical. Additionally, heat treatment leads to the removal of β -Mg₂Si's Chinese script morphology.

2. Heat treatment led to an approximate 60% increase in hardness, a 70% increase in yield strength, and a 115% increase in tensile strength for the as-cast Al-10Si-Mg alloy.

3. The fracture surfaces of the cast and heat-treated alloy consist of cleavage planes and tear ridges. There are relatively more cleavage planes in the heat-treated state of the alloy. This reveals that the heat-treated alloy exhibits a more brittle behavior.

4. Adhesion, delamination, and cracks were observed on all wear surfaces of the Al-10Si-Mg alloy, both as-cast and heat-treated. It was determined that heat treatment increased the alloy's wear resistance.

5. In the friction and wear tests, it was observed that the variation of the friction coefficient with the sliding distance showed a sudden increase at the beginning of the test and reached a steady state.

6. It was determined that the volume loss values of the alloy increased linearly with the sliding distance and pressure values and the highest wear resistance was reached in the Al-10Si-Mg alloy with aging heat treatment.



5. No significant weight change was observed with increasing temperature for all alloy types until temperatures close to the melting temperature. This result is due to the high oxidation resistance of Al-based alloys. On the other hand, as the melting temperature approached, the oxidation resistance decreased due to the loss of effectiveness of the protective surface layer and the beginning of the transition of the material to the liquid phase.

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Data Availability The authors confirm that all data supporting the findings of this study are available within the article. The data reported in this study are available upon request by contact with the corresponding author.

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References

- Basavakumar, K.G.; Mukunda, P.G.; Chakraborty, M.: Influence of melt treatments on sliding wear behavior of Al–7Si and Al–7Si–2.5Cu cast alloys. *J. Mater. Sci.* **42**, 7882–7893 (2007). <https://doi.org/10.1007/s10853-007-1633-7>
- Tiryakioğlu, M.: Si particle size and aspect ratio distributions in an Al–7%Si–0.6%Mg alloy during solution treatment. *Mater. Sci. Eng. A* **473**, 1–6 (2008). <https://doi.org/10.1016/j.msea.2007.03.044>
- Zhang, L.Y.; Jiang, Y.H.; Ma, Z.; Shan, S.F.; Jia, Y.Z.; Fan, C.Z.; Wang, W.K.: Effect of cooling rate on solidified microstructure and mechanical properties of aluminium–A356 alloy. *J. Mater. Process. Technol.* **207**, 107–111 (2008). <https://doi.org/10.1016/j.jmatprotec.2007.12.059>
- Clarke, J.; Sarkar, A.D.: Wear characteristics of as-cast binary aluminium–silicon alloys. *Wear* **54**, 7–16 (1979). [https://doi.org/10.1016/0043-1648\(79\)90044-9](https://doi.org/10.1016/0043-1648(79)90044-9)
- Sarkar, A.D.; Clarke, J.: Friction and wear of aluminium–silicon alloys. *Wear* **61**, 157–167 (1980). [https://doi.org/10.1016/0043-1648\(80\)90120-9](https://doi.org/10.1016/0043-1648(80)90120-9)
- Jasim, K.M.; Dwarakadasa, E.S.: Wear in Al–Si alloys under dry sliding conditions. *Wear* **119**, 119–130 (1987). [https://doi.org/10.1016/0043-1648\(87\)90102-5](https://doi.org/10.1016/0043-1648(87)90102-5)
- Dwivedi, D.K.: Adhesive wear behaviour of cast aluminium–silicon alloys: overview. *Mater. Des.* **31**, 2517–2531 (2010). <https://doi.org/10.1016/j.matdes.2009.11.038>
- Ye, H.: An overview of the development of Al–Si-alloy based material for engine applications. *J. Mater. Eng. Perform.* **12**, 288–297 (2003). <https://doi.org/10.1361/105994903770343132>
- Bai, B.N.P.; Biswas, S.K.: Characterization of dry sliding wear of AlSi alloys. *Wear* **120**, 61–74 (1987). [https://doi.org/10.1016/0043-1648\(87\)90133-5](https://doi.org/10.1016/0043-1648(87)90133-5)
- Camicia, G.; Timelli, G.: Grain refinement of gravity die cast secondary AlSi7Cu3Mg alloys for automotive cylinder heads. *Trans. Nonferrous Metals Soc. China (Engl. Ed.)* **26**, 1211–1221 (2016). [https://doi.org/10.1016/S1003-6326\(16\)64222-X](https://doi.org/10.1016/S1003-6326(16)64222-X)
- Sjölander, E.; Seifeddine, S.: The heat treatment of Al–Si–Cu–Mg casting alloys. *J. Mater. Process. Technol.* **210**, 1249–1259 (2010). <https://doi.org/10.1016/j.jmatprotec.2010.03.020>
- Han, Y.; Samuel, A.M.; Doty, H.W.; Valtierra, S.; Samuel, F.H.: Optimizing the tensile properties of Al–Si–Cu–Mg 319-type alloys: role of solution heat treatment. *Mater. Des.* **58**, 426–438 (2014). <https://doi.org/10.1016/j.matdes.2014.01.060>
- Farkoosh, A.R.; Pekguleryuz, M.: Enhanced mechanical properties of an Al–Si–Cu–Mg alloy at 300°C: effects of Mg and the Q-precipitate phase. *Mater. Sci. Eng. A* **621**, 277–286 (2015). <https://doi.org/10.1016/j.msea.2014.10.080>
- Zuo, L.; Ye, B.; Feng, J.; Kong, X.; Jiang, H.; Ding, W.: Effect of Q-Al5Cu2Mg8Si6 phase on mechanical properties of Al–Si–Cu–Mg alloy at elevated temperature. *Mater. Sci. Eng. A* **693**, 26–32 (2017). <https://doi.org/10.1016/j.msea.2017.03.087>
- Yang, H.; Ji, S.; Yang, W.; Wang, Y.; Fan, Z.: Effect of Mg level on the microstructure and mechanical properties of die-cast Al–Si–Cu alloys. *Mater. Sci. Eng. A* **642**, 340–350 (2015). <https://doi.org/10.1016/j.msea.2015.07.008>
- Sjölander, E.; Seifeddine, S.: Artificial ageing of Al–Si–Cu–Mg casting alloys. *Mater. Sci. Eng. A* **528**(7402), 7409 (2011). <https://doi.org/10.1016/j.msea.2011.06.036>
- Tebib, M.; Ajersch, F.; Samuel, A.M.; Chen, X.G.: Solidification and microstructural evolution of hypereutectic Al–15Si–4Cu–Mg alloys with high magnesium contents. *Metall. Mater. Trans. A Phys. Metall. Mater. Sci.* **44**, 4282–4295 (2013). <https://doi.org/10.1007/s11661-013-1769-9>
- Taghiabadi, R.; Fayege, A.; Pakbin, A.; Nazari, M.; Ghoncheh, M.H.: Quality index and hot tearing susceptibility of Al–7Si–0.35Mg–xCu alloys. *Trans. Nonferrous Metals Soc. China (Engl. Ed.)* **28**, 1275–1286 (2018). [https://doi.org/10.1016/S1003-6326\(18\)64783-1](https://doi.org/10.1016/S1003-6326(18)64783-1)
- Jiang, B.; Xu, S.; Xu, H.Y.; Hu, M.L.; He, Y.J.; Ji, Z.S.: Effect of Mg addition on microstructure and mechanical properties of Al–Si–Cu–Fe alloy with squeeze casting. *Mater. Res. Express* (2019). <https://doi.org/10.1088/2053-1591/ab5cc6>
- Alemdag, Y.; Beder, M.: Microstructural, mechanical and tribological properties of Al–7Si–(0–5)Zn alloys. *Mater. Des.* **63**, 159–167 (2014). <https://doi.org/10.1016/j.matdes.2014.06.006>
- Alemdag, Y.; Beder, M.: Effects of zinc content on strength and wear performance of Al–12Si–3Cu based alloy. *Trans. Nonferrous Metals Soc. China (Engl. Ed.)* **29**, 2463–2471 (2019). [https://doi.org/10.1016/S1003-6326\(19\)65154-X](https://doi.org/10.1016/S1003-6326(19)65154-X)
- Alemdag, Y.; Beder, M.: Effects of Mn content on microstructure, mechanical and dry sliding wear properties of eutectic Al–Si–Cu alloy. *Metals Mater. Int.* **26**, 1811–1819 (2020). <https://doi.org/10.1007/s12540-019-00572-0>
- Nemri, Y.; Gueddouar, B.; Benamar, M.E.A.; Sahraoui, T.; Chiker, N.; Hadji, M.: Effect of Mg and Zn contents on the microstructures and mechanical properties of Al–Si–Cu–Mg alloys. *Int. J. Met.* **12**, 20–27 (2018). <https://doi.org/10.1007/s40962-017-0134-y>
- Liu, H.; Zhao, G.; Liu, C.M.; Zuo, L.: Effects of magnesium content on phase constituents of Al–Mg–Si–Cu alloys. *Trans. Nonferrous*



- Metals Soc. China (Engl. Ed.) **16**, 376–381 (2006). [https://doi.org/10.1016/S1003-6326\(06\)60064-2](https://doi.org/10.1016/S1003-6326(06)60064-2)
25. Huang, Z.L.; Wang, K.; Zhang, Z.M.; Li, B.; Xue, H.S.; Yang, D.Z.: Effects of Mg content on primary Mg₂Si phase in hypereutectic Al–Si alloys. *Trans. Nonferrous Metals Soc. China (Engl. Ed.)* **25**, 3197–3203 (2015). [https://doi.org/10.1016/S1003-6326\(15\)63956-5](https://doi.org/10.1016/S1003-6326(15)63956-5)
 26. Kang, H.J.; Park, J.Y.; Choi, Y.S.; Cho, D.H.: Influence of the solution and artificial aging treatments on the microstructure and mechanical properties of die-cast Al–Si–Mg alloys. *Metals (Basel)* (2022). <https://doi.org/10.3390/met12010071>
 27. Cheng, W.; Liu, C.Y.; Huang, H.F.; Zhang, L.; Zhang, B.; Shi, L.: High strength and ductility of Al–Si–Mg alloys fabricated by deformation and heat treatment. *Mater Charact* (2021). <https://doi.org/10.1016/j.matchar.2021.111278>
 28. Ghanbari, D.; Kasiri Asgarani, M.; Amini, K.; Gharavi, F.: Influence of heat treatment on mechanical properties and microstructure of the Al2024/SiC composite produced by multi-pass friction stir processing. *Meas. J. Int. Meas. Confed.* **104**, 151–158 (2017). <https://doi.org/10.1016/j.measurement.2017.03.024>
 29. Albiter, A.; León, C.A.; Drew, R.A.L.; Bedolla, E.: Microstructure and heat-treatment response of Al-2024/TiC composites. *Mater. Sci. Eng. A* **289**, 109–115 (2000). [https://doi.org/10.1016/S0921-5093\(00\)00900-X](https://doi.org/10.1016/S0921-5093(00)00900-X)
 30. Kori, S.A.; Murty, B.S.; Chakraborty, M.: Development of an efficient grain refiner for Al–7Si alloy and its modification with strontium. *Mater. Sci. Eng. A* **283**, 94–104 (2000). [https://doi.org/10.1016/S0921-5093\(99\)00794-7](https://doi.org/10.1016/S0921-5093(99)00794-7)
 31. Gupta, M.; Ling, S.: Microstructure and mechanical properties of hypo/hyper-eutectic Al–Si alloys synthesized using a near-net shape forming technique. *J. Alloys Compd.* **287**, 284–294 (1999). [https://doi.org/10.1016/S0925-8388\(99\)00062-6](https://doi.org/10.1016/S0925-8388(99)00062-6)
 32. Birol, Y.: Semisolid processing of near-eutectic and hypereutectic Al–Si–Cu alloys. *J. Mater. Sci.* **43**, 3577–3581 (2008). <https://doi.org/10.1007/s10853-008-2565-6>
 33. Li, X.Y.; Tandon, K.N.: Mechanical mixing induced by sliding wear of an Al–Si alloy against M2 steel. *Wear* **225–229**, 640–648 (1999). [https://doi.org/10.1016/S0043-1648\(99\)00021-6](https://doi.org/10.1016/S0043-1648(99)00021-6)
 34. Li, X.Y.; Tandon, K.N.: Microstructural characterization of mechanically mixed layer and wear debris in sliding wear of an Al alloy and an Al based composite. *Wear* **245**, 148–161 (2000). [https://doi.org/10.1016/S0043-1648\(00\)00475-0](https://doi.org/10.1016/S0043-1648(00)00475-0)
 35. Biswas, S.K.: Some mechanisms of tribofilm formation in metal/metal and ceramic/metal sliding interactions. *Wear* **245**, 178–189 (2000). [https://doi.org/10.1016/S0043-1648\(00\)00477-4](https://doi.org/10.1016/S0043-1648(00)00477-4)
 36. Prasad, B.K.: Effects of silicon addition and test parameters on sliding wear characteristics of zinc-based alloy containing 37.5% aluminium. *Mater. Trans. JIM* **38**, 701–706 (1997). <https://doi.org/10.2320/matertrans1989.38.701>
 37. Prasad, B.K.; Venkateswarlu, K.; Modi, O.P.; Jha, A.K.; Das, S.; Dasgupta, R.; Yegneswaran, A.H.: Sliding wear behavior of some Al–Si alloys: role of shape and size of Si particles and test conditions. *Metall. Mater. Trans. A Phys. Metall. Mater. Sci.* **29**, 2747–2752 (1998). <https://doi.org/10.1007/s11661-998-0315-7>
 38. Zhang, G.H.; Zhang, J.X.; Li, B.C.; Cai, W.: Characterization of tensile fracture in heavily alloyed Al–Si piston alloy. *Prog. Nat. Sci. Mater. Int.* **21**, 380–385 (2011). [https://doi.org/10.1016/S1002-0071\(12\)60073-2](https://doi.org/10.1016/S1002-0071(12)60073-2)
 39. Alexopoulos, N.D.; Stylianou, A.; Campbell, J.: Dynamic fracture toughness of Al–7Si–Mg (A357) aluminum alloy. *Mech. Mater.* **58**, 55–68 (2013). <https://doi.org/10.1016/j.mechmat.2012.11.005>
 40. Lasa, L.; Rodriguez-Ibabe, J.M.: Wear behaviour of eutectic and hypereutectic Al–Si–Cu–Mg casting alloys tested against a composite brake pad. *Mater. Sci. Eng. A* **363**, 193–202 (2003). [https://doi.org/10.1016/S0921-5093\(03\)00633-6](https://doi.org/10.1016/S0921-5093(03)00633-6)
 41. Anasyida, A.S.; Daud, A.R.; Ghazali, M.J.: Dry sliding wear behaviour of Al–12Si–4Mg alloy with cerium addition. *Mater. Des.* **31**, 365–374 (2010). <https://doi.org/10.1016/j.matdes.2009.06.007>
 42. Hutchings, I.; Shipway, P.: *Tribology: Friction and Wear of Engineering Materials*, 2nd edn., p. 1–388. Butterworth-Heinemann, Oxford (2017)
 43. Hekmat-Ardakan, A.; Liu, X.; Ajersch, F.; Chen, X.G.: Wear behaviour of hypereutectic Al–Si–Cu–Mg casting alloys with variable Mg contents. *Wear* **269**, 684–692 (2010). <https://doi.org/10.1016/j.wear.2010.07.007>
 44. Pilling, N.B.; Bedworth, R.E.: The oxidation of metals at high temperature. *J. Inst. Metals* **29**, 529–582 (1923)
 45. Xu, C.; Gao, W.: Pilling–Bedworth ratio for oxidation of alloys. *Mater. Res. Innov.* **3**, 231–235 (2000). <https://doi.org/10.1007/s100190050008>
 46. Ducheyne, P.: *COMPREHENSIVE biomaterials*, Vol. II, p. 1–4858. Elsevier, Amsterdam (2017)
 47. Coker, E.N.: The oxidation of aluminum at high temperature studied by thermogravimetric analysis and differential scanning calorimetry. Sandia Report, Sandia National Laboratories, pp 1–22 (2013)
 48. Trunov, M.A.; Schoenitz, M.; Dreizin, E.L.: Effect of polymorphic phase transformations in alumina layer on ignition of aluminium particles. *Combust. Theory Model.* **10**, 603–623 (2006). <https://doi.org/10.1080/13647830600578506>
 49. Park, K.; Kim, D.; Kim, K.; Cho, S.; Kwon, H.: Behavior of intermetallic compounds of Al–Ti composite manufactured by spark plasma sintering. *Materials (Basel)* **12**, 331 (2019). <https://doi.org/10.3390/ma12020331>
 50. Mfusi, B.J.; Mathe, N.R.; Popoola, P.A.I.; Tshabalala, L.C.: Influence of stress relieving thermal cycles on AISI10Mg specimens produced by selective laser melting. *IOP Conf. Ser. Mater. Sci. Eng.* **655**, 012027 (2019). <https://doi.org/10.1088/1757-899X/655/1/012027>

