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THE ROLE OF Si AT A LOWER LEVEL ON THE MECHANICAL PROPERTIES OF Al-BASED AUTOMOTIVE ALLOYS

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Abstract

The effect of silicon, at a lower level around 3.5 wt% on the tensile, notched impact, and fracture properties of Al-based alloys for the automotive industry was investigated by T6 heat treatment. The results show that the tensile strength in the aged alloys increases mainly due to the development of Al_2Cu and Mg_2Si precipitates within the Al-matrix. The alloy with higher Si content exhibits higher strength due to the formation of Si-rich precipitates which impede the dislocation movement and cause higher values of true strain with a change in tensile behavior. The ductility of the alloys decreases with the treatment temperature in favor of GP zones, β'', β', β-precipitates formation and the formation of \ Q'' phase and finally increases after the coarsening of the precipitates. Microstructural examination confirms the coarse grain boundary and the plate-like eutectic phase when Si is added. Fractography shows a small dimple structure originating from the Al dendrites and the crack is initiated through brittle Si-rich particles.

Keywords: *Automotive Al-alloy; T6 heat treatment; True stress-strain; Impact strength; Fractography*

1. Introduction

Many studies have been done with the Al-based automotive alloy due to its outstanding properties [1-3]. It is generally used in the automotive sector due to its low weight and high strength. It is well established that the major alloying elements such as Si play significant roles in the different properties like castability, hardness, wear, corrosion, as well as machinability of this automotive alloy. In most cases, this type of alloy contains of around 5 to 23 wt% Si depending on application requirements [4, 5]. It is clear from the literature of I. J. Polmear that the presence of small amounts of other elements such as Cu, Ni, Mg, Mn, Zn, and Fe improves the properties of such Al-Si alloys. However, the most demandable minor alloying elements such as Cu and Mg are commonly used not only for improving the mechanical properties but also in precipitation hardening [6]. For better performance of these alloys, Cu is added in the amount of around 2 wt % and Mg ranging from 0.3 to 1 wt% [7, 8]. Dissimilar types of elements and their amounts have a great impact on the different physical, mechanical, electrical, and chemical characteristics of the alloys. A small addition of alloying elements improves one property but may have unintentional effects on other properties. A thorough study was by earlier study regarding the formation of different intermetallics and their role on the properties of the Al-based alloy [9].

From the literature it is also inferred that around 7% Si, 12-13% Si and 18% Si namely hypoeutectic, eutectic, and hypereutectic Al-Si automotive alloys are used for light, medium, and heavy structures of engine applications [10-12]. The level of Si plays an important role by giving the dissimilarity in the properties of these alloys. However, there is not enough information about the basic tensile properties when a lower level of Si is added in this type of alloy. The present work was undertaken to study the heat treatment effects on the mechanical properties of Al-Si automotive alloys where Si is added in the amount of 3.5wt%. The alloy without additional Si was also investigated in order to have a better understanding of Si addition on tensile and impact strength, fracture, and microstructural behavior of the investigated alloys.

2. Experimental details

The main concern of this study was the role of Si at low levels on the mechanical properties of Al-based

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automotive alloys. Another alloy without additional Si was also investigated as a base alloy for comparison. Two alloys were prepared by melting the commercial purity aluminium, copper, magnesium and ingots of Al-50 wt% Si master alloy. During this process, a crucible made of clay-graphite was used in a natural gas-fired pit furnace. . The furnace temperature was maintained at 750±10°C throughout the melting process. Prior to casting the melt was homogenized by stirring and the pouring temperature was around 700°C. The temperature was monitored with the help of a non-contact digital laser temperature gun. Casting was done in mild steel mold sized 20 x 200 x 300 mm preheated at 250°C. The experimental alloys' chemical compositions were evaluated through a Shimadzu PDA 700 optical emission spectrometer, as reported in the Table 1.

Using a shaper machine, the oxide layers of the castings were removed from the surface and then the alloys were homogenized at 450°C for 12h. The samples were allowed to cool to room temperature for the internal stresses relieving. Following the heat treatment process, the samples were solution treated at 535 °C for 2h and quenched in salt ice water to obtain a supersaturated single-phase region. The tensile tests were performed at room temperature using an Instron testing machine with the strain rate of 10^{-3}/s. The specimens used were as per ASTM specification where the gauge length was 25 mm. For the impact test, standard-sized 10 × 10 × 55 mm specimens were used, having a V-shaped notch, 2 mm deep with a 45° angle. The samples were subjected to ageing isochronally from room temperature to 350°C for one hour. For all heat treatment purposes, an Electric Muffle Furnace JSMF-30T with a range of 900 ± 3ºC was used. Following ASTM E23, tensile and impact strength were determined where seven tests were considered for each data. The true stress-strain curves were drawn to the closest average value given by the experimental results. Advanced video extensometer was used to determine the Young's modulus of the specimens. For the optical metallographic characterization, the heat-treated samples were polished with different grit emery papers, as well as alumina in the conventional way and etched with Keller's reagent. Optical images were taken with a Versamet-II Microscope. The macroscopic view of the fracture surfaces caused by tensile testing was analyzed by JSM-5200 type Jeol Scanning Electron Microscope.

3. Results and Discussion

3.1. Tensile strength

The tensile test of the Alloy 1 and Alloy 2 at different isochronal ageing conditions was out with the strain rate of 10^{-3}/s. The results of ultimate tensile strength are plotted in Fig. 1. With the increase of ageing temperature, two consecutive peaks appear in the obtained tensile strength curves. Peaks are observed for both alloys but the intensities and positions of the peaks are different. In this process, the initial peak is associated with the GP zone formation and another one is connected to the metastable phases. Through the GP zone, the alloying elements form many fine solute-rich coherent clusters that are homogeneously ordered in the aluminium matrix. These clusters induce higher strains in the adjacent lattice and consequently strengthen the alloy by imposing friction against the slip movement [13, 14]. As the ageing temperature increases, semi-consolidated metastable precipitates begin to form in the matrix and resist the dislocation movement resulting in a certain strengthening effect. Before the formation of the metastable phase, the GP zones dissolve and lose their ability to resist dislocation movement, hence the poor strengthening effect [15].

In Alloy 1, strengthening occurs via the formation of fine Al_2Cu and Al_2CuMg second phase precipitates in the course of the heat treatment [16]. Similarly for Si containing Alloy 2 which is Al-Si-Cu-Mg alloy system Mg_2Si, Al_2Cu and $Al_5Cu_2Mg_8Si_6$, soluble phases are formed [17]. The additional intermetallic hardening phase Mg_2Si precipitates in the α-aluminium matrix. It also refines the α-Al grains and makes great interfacial bonding with the strengthening Mg_2Si particles, hence increasing the tensile strength. With the increase in the Si content, an increased formation of Mg_2Si particles follows, resulting in an increase in the dislocation density around the Mg_2Si particles during solidification; hence, the strength improves [18]. The addition of Si results in shifting the peak strength of the alloy under the ageing process as it controls the formation of GP zone, metastable phase, etc. It enhances nucleation heterogeneously which contributes to early ageing and higher strength due to the accelerated ageing kinetics [19]. Ageing at higher temperature, the strength of the alloys suffer due to precipitation, grain coarsening, and recrystallization effects known as overageing. These coarser particles reduce the pinning effect and dislocation movement more readily leading to lower tensile strength [20].

Table 1. *Average composition of elements by wt% from OES analysis for both alloys*

	Si	Cu	Mg	Fe	Ni	Pb	Zn	Al
Alloy 1	0.244	2.158	0.767	0.211	0.199	0.163	0.076	Bal
Alloy 2	3.539	2.309	0.784	0.273	0.217	0.166	0.083	Bal

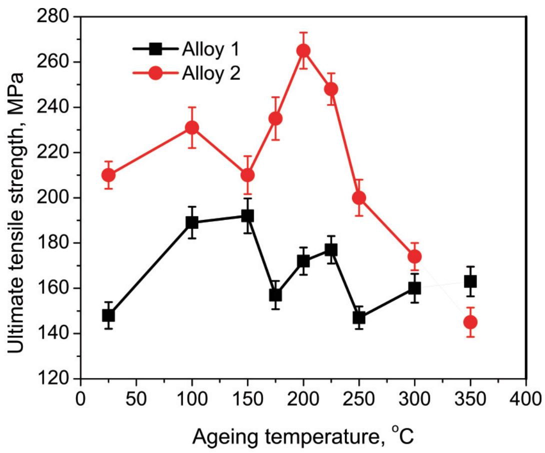

Figure 1. *The effect of isochronal ageing for one hour on the UTS of investigated alloys*

3.2. Percentage of elongation

The change in percent elongation of both alloys under the same ageing conditions is presented in Fig. 2. It is very much known that throughout the ageing process this type of Al-Si-Cu-Mg alloy forms a variety of phases like a supersaturated solid solution, GP zones, intermediate β′′, intermetallic β′, equilibrium β and the Q′′ phase [21]. These precipitation sequences are reflected in the form of ductility minima. Fine precipitates act as primary nucleation sites for microvoids. Hence, the materials lose the fracture resistance properties. Total elongation is also reduced due to the pinning effect of precipitated particles. An initial drop in percent elongation is related to the formation of the GP zones; additionally, another decrease in percent elongation is further detected that is caused by the formation of the metastable phases. It can be noted that the precursor is always dissolved during intermetallic formation. At this dissolution stage, the opposite phenomenon is observed with maximum ductility. The elongation increases rapidly since the particles start to coarsen in the final stage of ageing. These coarser particles lose the pinning effect hence allowing the dislocations to move more easily.

3.3. True stress-strain

Figures 3, 4, and 5 show the true stress and true strain curves of under-aged, peak-aged and over-aged alloys, respectively, while the test was conducted at above common strain rate of 10^{-3}/s. Solution treated alloys are here considered as being under aged condition, alloys aged at 200°C and 350°C for one hour are considered as peak aged and overaged conditions, respectively. The corresponding figure demonstrates some deviation between them. The base alloy have relatively higher ductility and therefore exhibits a higher slope of the true stress-strain curve than the alloy containing 3.5Si (Fig. 3). After the initial slope, alloy 1 has lower values of stress on the stress-strain curve in comparison to the alloy 2. Strengthening and loss of ductility in alloy 2 is due to the 3.5% Si addition. Therefore, alloy 2 has higher amount of intermetallics in the solid solution causing higher solution strengthening. At the peak-aged state the true stress and true strain curves for both alloys attain the lower slope as displayed in Fig. 4. At peak-aged condition the alloys contain the maximum fraction of fine precipitates. Additionally, during ageing, a higher amount of Si-rich fine precipitates is formed which create an additional obstruction to dislocation movement. Therefore, alloy 2 demonstrates higher tensile strength and a relatively lower slope. Due to the over-ageing, the coarsening of the fine precipitates takes place, causing easier movement of dislocation [22, 23].

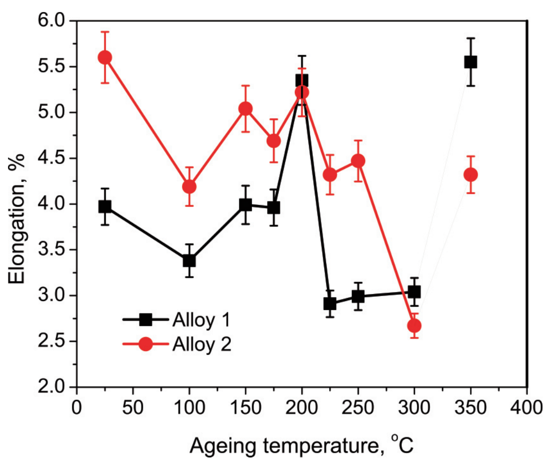

Figure 2. *The effect of isochronal ageing for one hour on the percentages of elongation of investigated alloys*

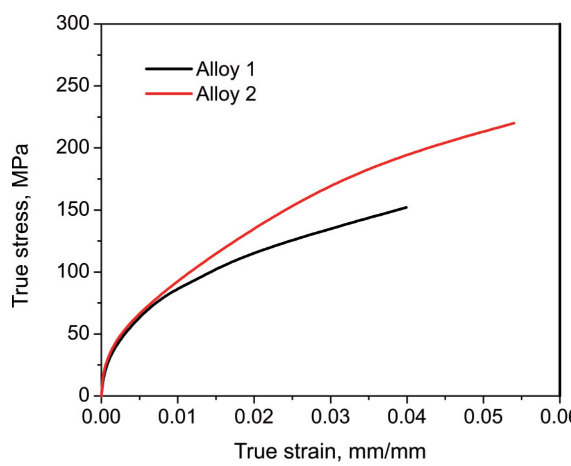

Figure 3. *The curves of true stress and true strain of investigated alloys without ageing treatment as under aged condition*

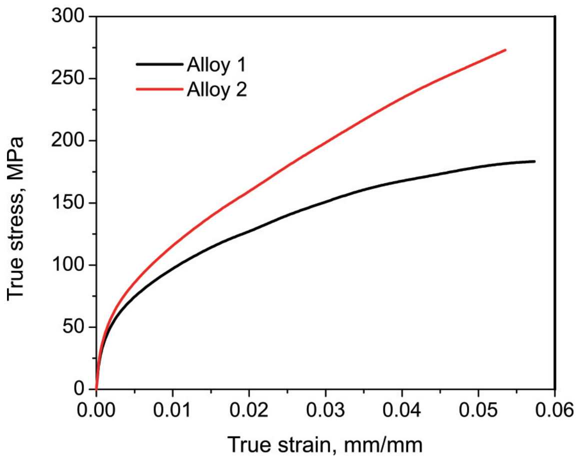

Figure 4. *The curves of true stress and true strain of investigated alloys under peak aged condition when aged at 200°C for one hour*

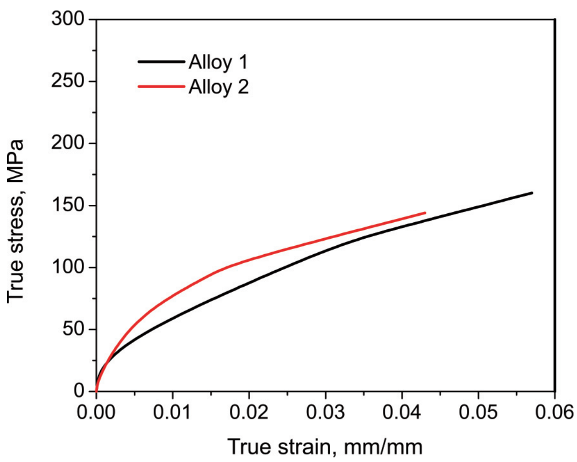

Figure 5. *The curves of true stress and true strain of investigated alloys under over aged condition when aged at 350°C for one hour*

The effect of ageing treatment and the role of 3.5% addition of Si on the Young's modulus of the experimental alloys are represented by a bar chart in Fig. 6. A small increased variation in Young's modulus between the alloys is observed at the under aged condition. It can be concluded that most of the alloying elements remain in solid solution after quenching so a higher amount of Si causes higher solution strengthening in alloy 2, evident in the Fig. 6. During ageing at 200°C for one hour which is the peak aged condition, Young's modulus values are highest where alloy 2 shows significantly higher values than base alloy 1. This can again be explained by the dislocation theory. At the peak aged condition alloys have the maximum amount of fine precipitates that strengthen the alloy. These precipitates slow down the dislocation motion and as a result, under load, the stress increases and the strain decreases [24, 25]. Thus, higher Young's modulus is displayed. Both the precipitates Al_2Cu and Mg_2Si play an active role for alloy 2 and lack of Mg_2Si for alloy 1 makes a difference to this behaviour. In the over aged condition (aged at 350°C for one hour), the behaviour of Young's modulus is quite different in comparison to the previous ones. In this case, Young's modulus in alloy 2 is lower than in alloy 1. At higher ageing treatments the precipitate becomes coarser. These coarse precipitates are not as useful as finely dispersed precipitates in resisting dislocation movement and therefore do not provide sufficient rigidity to the alloys. The lower modulus of alloy 2 can be attributed to the fact that Mg_2Si readily dissolves in the α(Al) matrix but not Al_2Cu through the additional ageing treatment because of the lower diffusion rate of Cu into Al than Mg and Si [26]. Consequently, the matrix strength is further reduced in the case of alloy 2. This phenomenon is also observed earlier for ultimate tensile strength, elongation and impact strength properties of Si-added alloy 2 compared to base alloy 1.

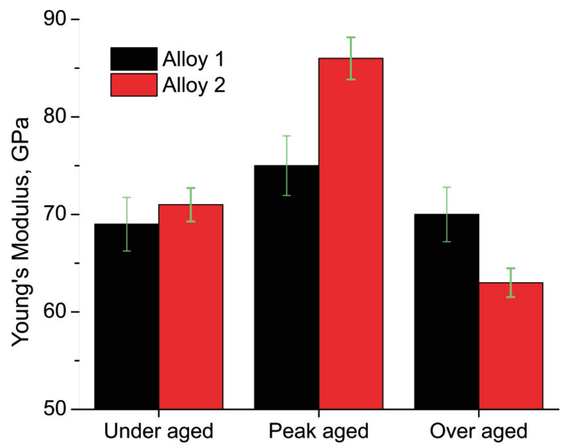

Figure 6. *The chart of Young's modulus of investigated alloys under different ageing conditions*

3.4. Impact toughness

Impact toughness characterization data obtained from both alloys under different ageing conditions are graphically presented in Fig. 7. It may be noted that the impact strength of both alloys sometimes increases and decreases with ageing temperature. However, at all ageing conditions alloy 2 exhibits lower impact toughness than base alloy (alloy 1). This observation can be attributed to the fact that in course of ageing treatment, the formation of precipitates reduces the impact energy and the dissolution of those phases increases the impact energy. Fine precipitates act as primary nucleation sites, due to which the fracture resistance of the alloy decreases. It is already been stated that GP zones, β″, β′, β and Q″ phase are formed during the ageing process. As the amount of Si increases, the impact strength decreases relatively, meaning that the alloy 2 contains [21, 27]. The impact strength significantly increases at a higher ageing temperature due to coarsening of the precipitates that leads to significant softening of the alloys.

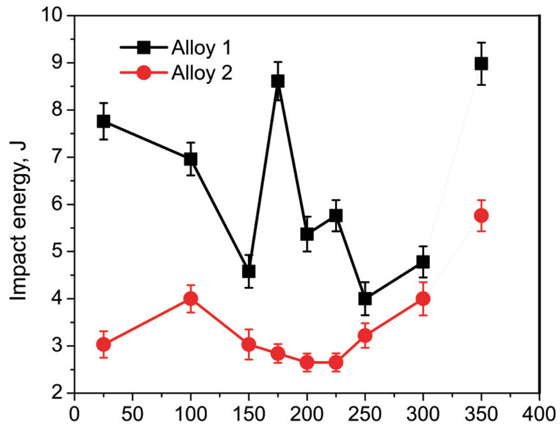

Figure 7. *The effect of isochronally ageing temperature for one hour on the impact energy of investigated alloys*

3.5. Optical microstructure

The optical microstructures of both alloys solution treated and aged at 200°C for 60 minutes are shown in Fig. 8. The microstructure of alloy 1 consists of α-Al phase with various intermetallic particles distributed intragranularly and at grain boundaries (8.a). The type and the number of second phases in the alloy have a huge influence on its properties. Addition of 3.5 wt% Si causes the appearance of eutectic phases in the microstructure of alloy 2 (8.b). The rough and elongated eutectic silicon separate the α-Al matrix. They are plate-like in nature and distribute eutectic Si across grain boundaries resulting in coarse grain boundaries. Due to ageing, different types of precipitates arise within the matrix but these fine precipitates cannot be exposed via this type of photography [28].

3.6. SEM observation

The influence of Si was investigated through the analysis of the fracture morphology of the samples of alloy 1 and alloy 2 by SEM (Fig. 9). Both alloys are aged at 200°C for an hour and simultaneously tensile tested using the strain rate10^{-3}/s. It is clear from the presented microstructures that higher concentration of Si has an impact on the mode of the fracture. The fracture surface of the base alloy shows a huge number of voids uniformly distributed in the alloy's matrix (Figure 9a). It has been previously observed that Al-Cu-Mg alloys undergo the most common fracture modes such as: intergranular fracture along grain boundaries. Several dimple morphologies and α-phase fracture of ductile nature are observed, which indicates a compound fracture [29, 30]. These dimples are formed

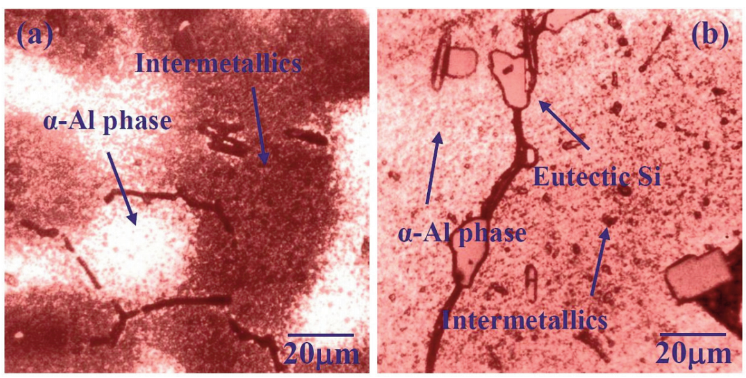

Figure 8. *Images of the optical microscopy of a) alloy 1 and b) alloy 2 aged at 200°C for one hour*

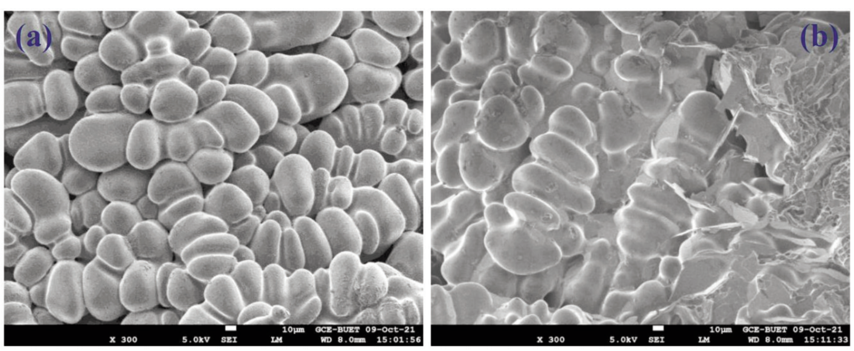

Figure 9. *The tensile fractography SEM images of (a) Alloy 1 and (b) Alloy 2 aged at 200°C for one hour*

through the Al dendrites and not as a result of a dimple rupture. Addition of 3.5%Si to the alloy, causes additional cracking because the Al-Si eutectic boundary widens the channels for microcracks (Figure 9b). Complementary scarring is also observed at fracture surfaces and grain boundaries due to the presence of higher Si-rich intermetallics in the alloy [31].

4. Conclusions

The following conclusions can be drawn from the conducted study of the mechanical properties of Al-based automotive alloys.

GP zones and other precipitated metastable phases cause the strengthening of the investigated alloys. The strain hardening is illustrated by two ageing peaks in the obtained UTS curves. The addition of Si improves the tensile strength and shifts the alloy aging peaks upwards due to the formation of Mg_2Si along with the Al_2Cu intermetallics.

The percent elongation and impact energy of the alloys decrease during ageing due to the formation of the GP zones, β′′, β′, β and Q phases but increase significantly after dissolution of the aforementioned precipitates and after the coarsening of the precipitates due to over-ageing at higher temperatures. A higher Si content leads to a larger amount of fine precipitates in the alloy resulting in lower elongation and notched impact strength especially in the peak aged condition.

The curves of alloy 2 appear higher than alloy 1 in all stress-strain curves, regardless of the ageing condition. This is due to the higher Si content , which further impedes dislocation movement, increases the tensile strength and decreases the slope of the true stress and true strain curves.

The values for Young's modulus are higher for alloy 2 in the underaged and most aged condition than the values of the base alloy. In the overagedcondition , the opposite phenomenon is observed. Lower values of Young's modulus are detected for alloy 2 compared to the base alloy. This is due to lower diffusion rates of Cu into Al than of Mg and Si into the alloy matrix.

The microstructures of the studied alloys contain the α-Al phase with various intermetallic particles distributed in the grains and at the grain boundaries. When an additional amount of Si is added to the alloy composition, a coarse, plate-like eutectic Si phase appears at the grain boundaries.

The fracture analysis with SEM micrographs shows that a higher Si content leads to additional crack propagation since brittle and plate-like Si-rich intermetallics are cleaved better.

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Author Contributions

Investigation, Data curation: Akib Abdullah Khan; Writing-original draft: Al-Kabir Hossain; Over all supervision: Mohammad Salim Kaiser.

Data availability

The data used to support the findings of this study are available from the corresponding author upon request.

Conflict of interest

The authors declare that they have no known competing financial interests or personal relationships that could have appeared to influence the work reported in this paper.

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ULOGA NIŽEG SADRŽAJA Si NA MEHANIČKE OSOBINE LEGURA NA BAZI Al ZA AUTOMOBILSKU INDUSTRIJU

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Apstrakt

Uticaj silicijuma na zatezana, udarna i lomna svojstva legura na bazi Al za automobilsku industriju ispitan je termičkom obradom T6. Rezultati pokazuju da se zatezna čvrstoća u ostarenim legurama povećava uglavnom zbog razvoja Al_2Cu i Mg_2Si precipitata unutar Al matrice. Legura sa većim sadržajem Si pokazuje veću čvrstocu zbog formiranja precipitata bogatih Si koji ometaju kretanje dislokacije i uzrokuju veće vrednosti prave deformacije sa promenom zateznog ponašanja. Duktilnost legura opada sa temperaturom obrade zbog formiranja GP zona, β'', β ', β-precipitatima i formiranjem \ Q'' faze nakon čega raste posle ogrubljivanja taloga. Mikrostrukturno ispitivanje potvrđuje granicu grubog zrna i eutektičku fazu u obliku ploča kada se doda Si. Fraktografija pokazuje malu strukturu udubljenja koja potiče od Al dendrita, a pukotina je inicirana lomljivim česticama bogatim Si.

Ključne reči: *Legure na bazi Al za automobilsku industriju; T6 termička obrada; Naprezanje; Udarna čvrstoća; Fraktografija*