Research Article

Mechanical Performance of Zr-Containing 354-Type Al-Si-Cu-Mg Cast Alloy: Role of Additions and Heat Treatment

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In this article, the volume fraction of intermetallic compounds in Zr-containing 354-type Al-Si-Cu-Mg alloys, characteristics of eutectic Si particles, and tensile, hardness, and impact properties have been evaluated with varying Ni and Mn contents and combination. The results revealed that additions of Ni and Mn in different amounts and combinations increased the volume fraction of intermetallic compounds in the tailored alloys, compared to the base alloy (cf. 12.21% for 4% Ni-containing alloy with 2.5% for base alloy), producing a significant effect on the mechanical performance. The proposed additions enhanced the mechanical performance of the alloys, namely, the ambient- and elevated-temperature tensile properties, hardness values, and impact properties. For the Mn-containing alloys, the improvement in properties was attributed to the formation of sludge particles in the form of blocky α -Al₁₅(Fe,Mn)₃Si₂ alongside the script-like α -iron phase that resisted crack propagation. The precipitation of Ni-bearing phases such as Al₉FeNi, Al₃CuNi, and Al₃Ni in the Ni-containing alloys improved the mechanical properties through hindering cracks propagation. Interestingly, addition of 0.75 wt.% Mn to the base alloy proved to be competitive in strength values to the addition of 2 and 4 wt.% Ni, and better in terms of ductility values. The investigations showed that the variations in hardness and impact values follow the same trend as variations in the percentage volume fraction of intermetallic compounds, i.e., maximum property value is associated to the alloy with highest volume fraction of intermetallic compounds. Furthermore, the impact properties showed higher dependency on Al₂Cu phase particles rather than the eutectic Si particles.

1. Introduction

In the automotive industry, Al-Si-Cu-Mg 354-type alloys are widely used in engine components owing to their excellent strength and hardness values, though, at some sacrifice of ductility and corrosion resistance. These alloys are very responsive to heat treatment in light of the presence of both copper and magnesium [1–4]. The 354-type Al-Si-Cu-Mg alloy is considered as an optimum candidate for the manufacture of multiple parts and components in the automotive and aerospace industries, including engine cooling fans, crankcases, high-speed rotating parts, structural aerospace components, timing gears, rocker arms, and many others [5–9]. However, during service, these alloys are subjected to elevated temperatures higher than 190°C; this high temperature instigates instability, coarsening, and/or dissolution of

the major strengthening phases such as θ' (Al₂Cu), β' (Mg₂Si), and S' (Al₂CuMg). Consequently, the resulting microstructures are not favorable for maintaining the mechanical performance at elevated temperatures [10, 11].

Many studies have been carried out in the past decade on how to maintain the mechanical properties of aluminum alloys at service temperatures that exceed 200°C. Among these, the addition of small amounts of transition metals was found to be a promising approach to maintain the mechanical properties of aluminum alloys at temperatures of up to 300°C [7, 12–16]. The idea is based on the formation of secondary fine heat-resistant Al₃M dispersoids, where “M” is a transition element such as Zr, Ni, and Mn [10, 12]. The selection process of the transition elements to be added to Al-alloys is a key factor in achieving the intended objective.

Knipling et al. [17] have introduced four criteria that have to be satisfied in the selection process of alloying elements in order to obtain castable, precipitation-strengthened aluminum alloys with both high stability and strength at elevated temperatures. These criteria state that the alloying element must

- (i) produce a suitable strengthening phase (precipitates)
- (ii) have a low solid-solubility in aluminum at the aging temperatures involved
- (iii) have a low diffusivity in aluminum
- (iv) preserve the alloy capability to be conventionally solidified

Zirconium has one of the lowest diffusion rates in aluminum in comparison to other transition elements [18]. Addition of Zr in the range of 0.1 to 0.3 wt.% to aluminum-based alloys leads to the formation of fine metastable $L1_2$ -structured Al_3Zr precipitates that have a very low lattice parameter mismatch with the Al matrix [19–21]. These Al_3Zr precipitates are noticeably stable and resist coarsening during heating; thus, the addition of Zr satisfies the four criteria proposed by Knipling et al. [17].

Similar to the outstanding behavior of Ni-based alloys at high homologous temperature owing to the existence of the Ni_3Al phase, researchers anticipated that a similar trend in behavior could be achieved in Al-based alloys by developing the Al_3Ni phase, which is analogous to the Ni_3Al (γ') phase upon the addition of Ni to Al alloys. This trialuminide phase (Al_3Ni) is thought to enhance the mechanical performance at elevated temperatures [12]. On the other hand, the addition of Mn to Fe-containing aluminum alloys is a practice commonly used to neutralize the negative effects of Fe. Manganese can modify the morphology and the type of Fe-intermetallic phases, which usually exist in aluminum cast alloys [22–24]. The addition of Mn will promote the formation of the less harmful α -iron $AlFeMnSi$ phase with Chinese script-like morphology, which will, in turn, improve the overall mechanical properties of Al alloys [6, 25, 26].

Based on the above arguments, Zr was added to the 354 alloy used in this study to form the base or reference alloy, and other elements (Ni and Mn) were subsequently added individually or in combination to study their mutual effect with Zr on the mechanical properties of 354 alloy at room and elevated temperatures and to achieve an appropriate modified chemistry of 354-type (Al-Si-Cu-Mg) alloys, which could enhance the overall mechanical performance of this category of alloys, as well as to resist the softening at elevated temperature during service.

2. Experimental Procedure

The alloys prepared for this work have been tailored from 354-type Al-Si-Cu-Mg alloy through adding Zr as a common alloying element along with the addition to Ni and/or Mn in different amounts and combinations. Alloy codes and respective addition are as follows:

- (1) Alloy M1S: 354 + 0.3 wt.% Zr (base alloy)
- (2) Alloy M2S: Alloy M1S + 2 wt.% Ni

(3) Alloy M3S: Alloy M1S + 0.75 wt.% Mn

(4) Alloy M4S: Alloy M1S + 4 wt.% Ni

(5) Alloy M5S: alloy M1S + 2 wt.% Ni + 0.75 wt.% Mn

The base alloy was selected based on its improved mechanical performance that was previously reported in the same research group [7, 27]. The chemical composition of the alloys under investigation is listed in Table 1.

The 354 alloy ingots were cut, dried, and melted in a 70-kg capacity SiC crucible using an electric resistance furnace. The melt was kept at a temperature of $800 \pm 5^\circ\text{C}$. The addition of master alloys was carried out instantly before starting the degassing process in order to ensure homogeneous mixing of additives during degassing. After successful degassing, the melt was carefully skimmed to remove the oxide layers from the melt surface. The melt was then poured into the preheated permanent mold of interest using a preheated pouring cup with a ceramic foam filter (15 ppi) in order to avoid entrance of inclusions and oxide films into the mold. Each permanent mold employed was preheated at 450°C in order to remove all traces of moisture from the mold.

An ASTM B-108 type permanent mold was used to prepare castings from which the standard tensile test bars were obtained with a gauge diameter of 12.7 mm. For preparing unnotched impact test bars, star-like mold was used to produce the impact test bars according to the ASTM E23 standard. The impact test bars have a square cross-sectional area of $10 \times 10 \text{ mm}^2$ and a length of 55 mm. For the hardness test bars, L-shaped mold was used to produce an L-shaped casting. After cutting off the feeding head, each casting was cut to produce three rectangular bars, which were subsequently machined to the final geometry ($35 \times 30 \times 80 \text{ mm}$) of the hardness test bars.

For the alloys investigated (i.e., M1S through M5S), the test bars of the five alloys were heat treated according to the procedures and parameters listed in Table 2.

Tensile testing at ambient temperature was carried out using an MTS servo-hydraulic mechanical testing machine at a strain rate of $4 \times 10^{-4} \text{ s}^{-1}$ till the point of fracture. Five test bars for each alloy/condition were tested, and the average values of ultimate tensile strength (UTS), 0.2% offset yield strength (YS), and percentage elongation to fracture (% El) were reported. An Instron Universal mechanical testing machine was used to carry out the tensile testing at elevated temperature (250°C), using the strain rate of $4 \times 10^{-4} \text{ s}^{-1}$. The testing was carried out at 250°C after holding the test bar for 15 min at the testing temperature in order to homogenize the temperature of the sample to 250°C throughout. The test sample was kept unmounted from one side inside the heating chamber during the holding process to avoid compressive stresses that might arise from the expansion of the bar, and then it was mounted from the other side and kept at the testing temperature for another 15 min. Five test bars were used for each alloy composition/condition studied, and the average values of UTS, YS, and % El were reported.

Hardness measurements were carried out on the prepared hardness test bar samples. A Rockwell hardness tester and F scale were employed using a 1/16-inch steel ball

TABLE 1: Chemical composition of the alloys investigated in this study.

Alloy codes	Chemical analysis (wt.%)									
	Elements									
	Si	Cu	Mg	Fe	Ti	Zr	Ni	Mn	Sr	Al
M1S	8.5	1.76	0.55	0.12	0.2	0.32	<0.1	0.01	0.022	Bal.
M2S	8.4	1.7	0.60	0.14	0.21	0.33	1.9	0.01	0.018	Bal.
M3S	8.6	1.8	0.55	0.11	0.20	0.33	<0.1	0.75	0.020	Bal.
M4S	8.6	1.8	0.67	0.12	0.22	0.29	4	0.01	0.019	Bal.
M5S	8.6	1.8	0.60	0.15	0.20	0.29	1.9	0.76	0.018	Bal.

TABLE 2: Heat treatment procedures and parameters applied to alloys studied.

Heat treatment procedures and parameters			
Heat treatment	Solution treatment	Quenching	Aging
SHT*	495°C for 5 h	Warm water (60°C)	NA
T5 temper	NA	NA	180°C for 8 h
T6 temper	495°C for 5 h	Warm water (60°C)	180°C for 8 h

SHT: solution heat treatment.

indenter and a load of 60 kgf. Ten measurements were made per sample, and the average value was reported as the Rockwell hardness value of that alloy sample/condition.

A computer-aided instrumented SATEC SI-1 Universal Impact Testing Machine was used to carry out the impact testing. The instrument and the attached data acquisition system provide the total absorbed energy (E_t) of the sample during the impact test. Five samples for each alloy/condition were tested, and the average value of the total energy obtained was reported.

Samples for metallographic observation were sectioned from the test bars, mounted, and polished using standard polishing procedures. The polished samples were examined using an Olympus PMG3 optical microscope connected to a Clemex Vision PE image-analysis system. A scanning electron microscope (SEM, JEOL JSM-6480LV) equipped with energy dispersive X-ray spectrometer (EDS) was used to investigate the intermetallic compounds and the fracture surfaces.

3. Results and Discussion

3.1. Microstructural Characterization

3.1.1. Intermetallic Compounds. The volume fraction (%) of intermetallic compounds observed in as-cast and as-quenched tensile bars is presented in Table 3. Figure 1 compares backscattered electron (BSE) images of all alloys in the as-cast (left) and as-quenched (right) conditions. The backscattered images shown to the right in Figure 1 demonstrate clearly the reduction in the volume fraction of the intermetallic compounds in the as-quenched samples.

TABLE 3: Volume fractions (%) of undissolved intermetallic compounds in the matrix of as-cast and as-quenched alloys.

Volume fraction (%)		Alloy code				
		M1S	M2S	M3S	M4S	M5S
As-cast	Average	2.51	6.17	4.34	12.21	8.79
	SD	0.41	0.56	0.36	0.77	0.79
SHT	Average	1.11	5.54	3.64	9.60	7.68
	SD	0.28	0.61	0.16	0.65	0.52

For the as-cast condition, it is obvious that the addition of Ni and Mn in different amounts and combinations (i.e., alloys M2S through M5S) significantly increases the volume fraction of existing phases compared to the base alloy (cf. 2.5% for alloy M1S and 12.21% for alloy M4S). Alloy M4S, which contains 4 wt.% Ni, shows an excessive increase in volume fraction in comparison to alloys M2S, M3S, and M5S. This substantial increase may be attributed to the formation of Ni-containing phases such as Al_3CuNi , Al_9FeNi , and Al_3Ni in addition to the phases commonly observed in other alloys such as the Q-phase, Al_2Cu , Mg_2Si , and Fe-containing phases. Other Ni-containing alloys, i.e., M2S and M5S, contain almost the same phases; however, the structure of alloy M4S uniquely comprises the eutectic $\text{Al-Al}_3\text{Ni}$ structure, as shown in Figures 1(g) and 1(h). This eutectic structure is believed to increase the overall volume fraction of intermetallic compounds in alloy M4S.

The addition of 0.75 wt.% Mn to the base alloy, i.e., alloy M3S, doubles the volume fraction in the as-cast condition and may be ascribed to the formation of $\alpha\text{-Al}_{15}(\text{Fe,Mn})_3\text{Si}_2$ phase in script-like and sludge morphologies [28]. In addition, the presence of the $\alpha\text{-Al}_{15}(\text{Fe,Mn})_3\text{Si}_2$ phase that does not dissolve with solution heat treatment would explain the nearly three times higher volume fraction observed in the as-quenched condition for alloy M3S compared to the base alloy M1S, similar to the observations of Elgallad [19].

As may be seen from Table 3 and Figure 1, applying solution treatment reduces the volume fraction (%) of intermetallic compounds owing to the dissolution of the Al_2Cu phase, and the partial dissolution of other phases such as Q- $\text{Al}_5\text{Mg}_8\text{Cu}_2\text{Si}_6$, Mg_2Si , Al_3CuNi , $\beta\text{-Al}_3\text{FeSi}$, $\pi\text{-Al}_8\text{Mg}_3\text{FeSi}_6$, and Al_9FeNi [7].

3.1.2. Characteristics of Eutectic Silicon Particles. The alloys studied, M1S through M5S, were modified by adding 200 ppm of strontium (Sr). Therefore, it is expected that the eutectic silicon particles will be modified to a large extent in the as-cast condition in all alloys. The morphology of eutectic silicon particles in the as-cast and as-quenched alloy samples is displayed in the optical micrographs shown in Figure 2, while the corresponding average Si particle characteristics are listed in Table 4; the average values were obtained from measurements of 20 fields per alloy sample/condition.

The optical micrographs shown in Figure 2 for the as-cast alloy samples, on the left, reveal that the majority of silicon particles are fully modified; nevertheless, partially modified silicon particles may still be observed in these

FIGURE 1: Continued.

FIGURE 1: Backscattered electron images for as-cast (left) and solution heat-treated (right) conditions of the alloys studied: (a, b) M1S (base alloy), (c, d) M2S (2 wt.% Ni), (e, f) M3S (0.75 wt.% Mn), (g, h) M4S (4 wt.% Ni), and (i, j) M5S (2 wt.% Ni + 0.75 wt.% Mn).

FIGURE 2: Continued.

FIGURE 2: Optical micrographs at 500X showing the morphology of the eutectic silicon in the alloys studied: (left) as-cast conditions and (right) solution heat-treated conditions: (a) as-cast alloy M1S, (b) SHT alloy M1S, (c) as-cast alloy M2S, (d) SHT alloy M2S, (e) as-cast alloy M3S, (f) SHT alloy M3S, (g) as-cast alloy M4S, (h) SHT alloy M4S, (i) as-cast alloy M5S, and (j) SHT alloy M5S.

TABLE 4: Characteristics of eutectic silicon particles in as-cast and solution heat-treated conditions of the alloys studied.

Alloy code and condition	Particle area (μm^2)		Particle length (μm)		Aspect ratio		Roundness (%)		Sphericity (%)	
	Av	SD	Av	SD	Av	SD	Av	SD	Av	SD
M1S-AC	2.85	4.39	2.67	2.48	1.96	1.33	0.49	0.21	0.75	0.28
M1S-SHT	3.71	4.71	2.60	2.10	1.71	1.49	0.58	0.19	0.85	0.21
M2S-AC	3.53	7.60	2.83	3.29	1.96	2.44	0.50	0.23	0.74	0.30
M2S-SHT	5.67	8.08	3.27	2.95	1.79	1.27	0.56	0.20	0.81	0.23
M3S-AC	1.92	3.10	2.14	2.22	1.92	2.43	0.51	0.22	0.77	0.28
M3S-SHT	5.35	6.04	3.12	2.32	1.69	2.92	0.61	0.17	0.86	0.19
M4S-AC	3.11	6.61	2.65	3.30	2.02	3.58	0.50	0.25	0.73	0.32
M4S-SHT	3.99	5.53	2.68	2.43	1.86	2.65	0.56	0.22	0.82	0.23
M5S-AC	3.87	6.88	3.12	3.43	2.12	6.27	0.48	0.22	0.70	0.30
M5S-SHT	5.70	7.44	3.52	3.20	1.95	3.60	0.54	0.20	0.78	0.25

micrographs to a certain extent. The existence of these partially modified silicon particles could be a result of the high Mg content, ~ 0.6 wt.%, of the alloys. It has been reported by Joenoes and Gruzleski [29] and Dunn and Dickert [30] that the presence of Mg and copper can change the microstructure from a fully modified structure into a partially modified one due to the formation of $\text{Mg}_2\text{Sr}(\text{Si}, \text{Al})$ and Al-Cu-Sr phases, which will result in reducing the amount of available strontium (Sr) to achieve the required degree of modification of the eutectic silicon particles. In the present

case, however, since the contents of Sr, Cu, and Mg are kept constant in the alloys investigated, other explanations are mandatory to explain the variations in the modification level of the eutectic Si particles in these alloys.

As per the micrographs shown in Figures 2(e) and 2(g), alloys M3S ($354 + 0.02$ wt.% Sr + 0.3 wt.% Zr + 0.75 wt.% Mn) and M4S ($354 + 0.02$ wt.% Sr + 0.3 wt.% Zr + 4 wt.% Ni) appear to contain less amounts of partially modified eutectic silicon. This observation can be ascribed to the existing phases in the microstructure of the two alloys, which may

contribute to variations in the free content of silicon and/or strontium. Such variations may lead to a more efficient modification in the case of a lower Si/Sr ratio and vice versa.

On one hand, a high amount of the copper in alloy M4S (containing 4% Ni) is consumed in forming Al_3NiCu phase [11], and thus, the possibility of forming the Al-Cu-Sr phase reported by Joenoes and Gruzleski [29] will be reduced. Consequently, more Sr will not be consumed to form the Al-Cu-Sr phase, and thus, better modification would be expected for Ni-containing alloys, particularly those with higher Ni-content as in alloy M4S. On the other hand, the addition of 0.75 wt.% Mn in alloy M3S changes some of the $\beta\text{-Al}_5\text{FeSi}$ iron phase into Chinese script-like $\alpha\text{-Al}_{15}(\text{Fe}, \text{Mn})_3\text{Si}_2$ phase and sludge particles. The mutual existence of the two Fe-based phases, i.e., $\beta\text{-Al}_5\text{FeSi}$ and $\alpha\text{-Al}_{15}(\text{Fe}, \text{Mn})_3\text{Si}_2$, will lower the silicon content (i.e., reduce the Si phase) in alloy M3S to an extent, which will allow a better modification level compared to that attained for the rest of the alloys studied, as depicted in Figure 2(e).

Thermal modification is yet another effective way to alter the morphology of eutectic silicon particles. It is evident from the micrographs shown in Figure 2 on the right that the solution treatment changes the fibrous interconnected eutectic silicon particles detected in the as-cast condition into globular particles with rounded edges. The evolution of the morphology of the eutectic silicon particles from fibrous to globular in these alloys is a direct result to the combined effect of solution-heat treatment and strontium modification, as previously stated by Chen et al. [31] and Yuying et al. [32].

It is evident from the data presented in Table 4 that the average Si particle area increases after solution treatment at 495°C for 5 hours for all the alloys. Additionally, the solutionizing treatment produces a noticeable improvement in the spheroidization of the Si particles concomitant with enhancements in the roundness values, as can be seen qualitatively from the micrographs shown in Figure 2. The increase in sphericity and roundness values would produce a corresponding decrease in the aspect ratio as seen in Table 4.

3.2. Mechanical Performance

3.2.1. Ambient-Temperature Tensile Testing. In the present study, the sole addition of ~0.3 wt.% Zr to the 354-type Al-Si-Cu-Mg cast alloy (i.e., the base alloy M1S) in the as-cast condition enhances the ambient-temperature strength values of the Zr-free 354 alloy (alloy A) used in previous investigations in the same group by Hernandez-Sandoval [16], by ~26 MPa (UTS) and 40 MPa (YS), respectively. These enhancements in the strength values are accompanied by a limited reduction in the alloy ductility (~0.054%). For the solution heat-treated condition, on the other hand, the UTS and ductility values of the base alloy M1S remain virtually constant at ~300 MPa and ~6.3%, respectively, while the yield strength increases by ~33 MPa compared to alloy A in the work of Hernandez-Sandoval [16]. The improved strength values of Zr-containing Al-Si-Cu-Mg alloy

emphasize the role of Zr addition in enhancing the ambient-temperature tensile properties.

Figure 3 shows the ambient-temperature tensile properties obtained for the alloys under investigation. In the as-cast condition, the base alloy M1S (354 + 0.3 wt.% Zr) shows the lowest UTS and YS values. As-cast strength values of the other alloys, i.e., M2S through M5S, show enhancements of 14–25% and 16–26% in UTS and YS, respectively, compared to the base alloy (M1S). However, the ductility values of alloys M2S through M5S show inconsistent variations with respect to the ductility of the base alloy M1S; these variations comprise enhancements in case of alloy M3S, deteriorations in the case of alloys M4S and M5S, and almost unchanged ductility value in the case of alloy M2S.

The probable explanations for the enhanced strength values of alloy M3S are the formation of favorable phases, which are advantageous to the strength values, and the highly refined as-cast Si particles (Table 4) as a result to the addition of 0.75 wt.% Mn to the base alloy (M1S). It is well established that the addition of Mn neutralizes to some extent the deleterious effect of iron impurities by transforming the detrimental needle-like $\beta\text{-Al}_5\text{FeSi}$ phase into the less harmful $\alpha\text{-Al}_{15}(\text{Fe}, \text{Mn})_3\text{Si}_2$ phase, which appears in either script-like form or as blocky sludge particles, or both. While there is no doubt about the positive effect of the script-like α -phase on strength and ductility values, there is much debate, however, on the effect of the sludge particles on the tensile properties either they are harmful, as regularly believed [28, 33, 34], or favorable to the tensile properties [7, 35, 36]. Garza-Elizondo [7] has reported that the presence of these hard particles in the soft $\alpha\text{-Al}$ matrix may contribute in enhancing the tensile properties through the development of more uniformly distributed stresses within the matrix.

For the Ni-containing alloys, enhancements in the strength values can be correlated to the formation of Ni-containing intermetallic compounds such as Al_9FeNi , Al_3CuNi , and Al_3Ni , which can hinder the propagation of cracks; similar findings are reported in previous studies [11, 16]. Increasing the Ni content from 2 wt.% to 4 wt.% inversely affects the tensile properties at room temperature. It is believed that the precipitation of higher volume fractions of the acicular Al_9FeNi and Al_3CuNi phases in the as-cast alloy M4S (4 wt.% Ni) is responsible for the deterioration of the tensile properties at room temperature.

Apparently from Figure 3, the base alloy M1S proves to be very responsive to solution heat treatment because it shows an increase of 40 MPa and 8 MPa in the as-cast UTS and YS values, respectively, after solutionizing at 495°C for 5 hours. In addition, the ductility of the solutionized M1S alloy also shows an increase of ~5% over the ductility value obtained for the as-cast condition. On the other hand, the other alloys, i.e., M2S through M5S, show small improvements, in the range of 2–10 MPa for UTS values, and a noticeable reduction in YS values (18–36 MPa) compared to their strength values in the as-cast condition. The ductility values of alloys M2S through M5S increase after solution treatment. The enhanced ductility values after solution heat treatment, in particular, are related to the changes in the eutectic silicon morphology. Coarse acicular silicon particles

FIGURE 3: Variation in average (a) UTS, (b) YS, and (c) % El values of the alloys studied in as-cast, solution heat-treated (SHT), and T5- and T6-treated conditions obtained at ambient temperature.

serve as crack initiators, which is the case in the as-cast condition; whereas more spherical Si particles with rounded edges and decreased aspect ratios are obtained in the solution heat-treated case.

Figures 3(a) and 3(b) show that the strength values of the base alloy M1S are 289 MPa (UTS)/259 MPa (YS) and 342 MPa (UTS)/325 MPa (YS) in the T5- and T6-treated conditions, respectively. Alloys M2S and M3S exhibit the highest strength among the alloys studied with values of 300 MPa (UTS)/277 MPa (YS) and 315 MPa (UTS)/279 MPa (YS), respectively, in the T5-treated condition; and 362 MPa (UTS)/352 MPa (YS) and 357 MPa (UTS)/355 MPa (YS) in the T6-treated condition, respectively. The strength values (UTS and YS) of alloys M1S, M2S (M1S + 2 wt.% Ni), and M3S (M1S + 0.75 wt.% Mn) show distinct enhancements after applying the T6 treatment in comparison to the as-cast and as-solutionized conditions. In contrast, alloys M4S (M1S + 4 wt.% Ni) and M5S (M1S + 2 wt.% Ni + 0.75 wt.% Mn) exhibit a very limited enhancement in UTS (~15 MPa) after T6 heat treatment with reference to their strength in the as-cast and as-quenched conditions. Additionally, from Figure 3(c), it can be seen that the ductility values reduce considerably after T6 treatment when compared to the as-cast ductility values of the respective alloys. Ductility values of the alloys studied are limited in general due to the high silicon content ~9wt.% and the presence of both Mg and Cu [30].

The enhanced strength values of alloy M2S, particularly for the T6-treated condition, can be attributed to the presence of Ni-containing phases such as Al_3Ni , Al_9FeNi , and Al_3CuNi , in addition to the fine precipitates formed after applying T6 treatment [37, 38]. With respect to alloy M3S, the improved strength values following T5- and T6-heat treatments can be ascribed to (i) the presence of $\alpha\text{-Al}_{15}(\text{Fe, Mn})_3\text{Si}_2$ in the form of blocky hard particles, (ii) the precipitated strengthening Cu- and Mg-containing dispersoids, and (iii) the probable formation of Al_6Mn fine dispersoids in the presence of a high Mn content of 0.75 wt.% [39, 40]. The presence of these Mn dispersoids significantly improves the yield and ultimate strength values without sacrificing the ductility [39, 41, 42].

The presence of a high nickel content in T6-treated alloy M4S (4 wt.%) results in deteriorating the ambient-temperature tensile properties as can be inferred from Figure 3; similar observations have been noted by other authors [7, 10, 43–46]. This deterioration in the strength values is a direct result of the high content of Ni which, in turn, will consume some of the available copper in the alloy to form the detected Al_3CuNi phase. As a result, this will reduce the amount of copper available for strengthening through precipitation hardening during heat treatment [10]. Furthermore, the precipitation of Al_9FeNi and Al_3CuNi phases in high volume fractions (Table 3) would facilitate

cracking, since they would act as stress raisers, causing instability in the flow strain, and hence the ductility of this alloy will reduce accordingly.

In general, the room-temperature tensile properties in the T6 heat-treated conditions are better than those obtained with the T5 treatment for the alloys studied. The use of permanent mold casting, which is the technique used for producing the test bars, would preserve copper and magnesium in considerable amounts in solid solution due to the high solidification rate. Thus, by applying artificial aging directly after solidification (i.e., T5 temper), large proportions of the dissolved Cu and Mg in the solid solution will form strengthening dispersoids. The T6 heat treatment, on the other hand, involves solutionizing the as-cast structure at a sufficiently high temperature in order to dissolve higher amounts of Cu and Mg in solid solution in order to form a supersaturated solid solution upon quenching and to achieve further modification of the eutectic Si particles. Thus, the artificial aging in the case of T6 treatment will precipitate fine strengthening particles in larger proportions than in the case of T5 treatment, which will enhance the alloy strength to a greater extent, however, at the expense of ductility.

For the alloys studied, the ambient-temperature tensile testing data are listed in Table 5 along with the quality index values, which were calculated using Equations (1) and (2) according to the models developed by Drouzy et al. [47] and Cáceres [48], respectively:

$$Q = S_{UTS} + d \log e_f, \quad (1)$$

$$Q_c = K [(qn)^n \exp^{-qn} + 0.4 \log(100qn)], \quad (2)$$

where Q and Q_c (MPa) are the quality indices according to Drouzy and Cáceres models, S_{UTS} refers to the ultimate tensile strength (MPa), e_f refers to the percentage elongation to fracture, d is a material constant equal to 150 MPa, K is the strength coefficient (MPa), relative quality index (q), and n is the strain-hardening exponent.

Figure 4 shows the quality chart for the alloys studied based on Drouzy [47] model and depicts variations in the quality index values based on the chemical composition, and the applied heat treatment. Equations (1) and (3) were used to develop the “iso- Q ” and “iso- YS ” lines in the chart:

$$S_{p(ys)} = aS_{UTS} - b \log e_f + C, \quad (3)$$

where coefficients a , b , and c were quantified as 1, 60 MPa, and -13 MPa, respectively.

The quality index values Q and Q_c listed in Table 5 show the same trend in variations, however, with different values. The difference between Q and Q_c increases in the T5 and T6 heat-treated conditions, which can be accredited to the fact that the alloy quality is affected by the net amount by which the increase in strength is balanced by the reduction in ductility. As can be seen from Table 5 and Figure 4, the best quality values for the alloys are obtained after solution heat treatment, attributed to the microstructural changes that take place during the solution treatment including the dissolution of strengthening elements, the homogenization

of the segregated as-cast structure, and the spheroidization of the eutectic silicon particles. These changes will significantly enhance the alloy ductility, in addition to a limited enhancement in UTS values. Consequently, the quality index values of the as-quenched (or solution heat-treated) alloys are remarkably higher than those in the as-cast condition.

While the T6 heat treatment improves the UTS values considerably, in comparison to the UTS values obtained after solution treatment, it does so at the expense of the ductility. This trade-off between UTS and ductility values will certainly affect the quality index values and not necessarily in a positive way. Likewise, a similar behavior is noted for the T5-treated alloys, as well, as can be inferred from Figure 4.

It is evident that the superior ductility of the base alloy M1S is the main reason for the improved quality indices of this alloy in the majority of the conditions studied. On the other hand, the mutual enhancement in the strength and ductility values of alloys M2S and M3S compared to those of alloys M4S and M5S is responsible for the higher quality indices of the former compared to those of the latter.

The addition of Mn in alloy M3S results in transforming the needles of the β -iron phase into the less detrimental α -iron phase; this favorable morphological change is believed to improve the ductility and strength values of alloy M3S. Whereas the structure of alloy M2S contains Ni-bearing phases with acicular morphologies and β -iron needles that negatively affect the mechanical properties. Accordingly, the quality index values of alloy M3S are higher than those of the 2 wt.% Ni-containing alloy M2S in the as-cast, SHT, and T5 conditions. In contrast, the quality index values of the T6-treated alloys M2S and M3S show a reversed behavior according to the marginal variations in their tensile properties and hence the quality index values.

From Table 5, it is observable that the quality index values calculated using Drouzy's approach are higher than those obtained by the model developed by Cáceres for all treatment conditions per each alloy, except for solution-treated conditions. This can be attributed to the improved ductility of as-quenched conditions that will allow more accurate determination of the material parameters (K and n). Thus, it would be advisable to calculate the quality index values of solution-treated conditions using Cáceres' model, especially for materials with low ductility at room temperature.

3.2.2. Elevated-Temperature Tensile Testing. Figure 5 reveals the elevated-temperature tensile properties obtained at 250°C for the alloys studied. By tensile testing at 250°C, all the investigated alloys endure some softening owing to the possible coarsening of the strengthening precipitates that exist during tensile testing at room temperature (Figure 3). Figures 5(a) and 5(b) demonstrate that additions of Ni and Mn in different amounts and combinations to the base alloy, i.e., alloys M2S through M5S, slightly improve the strength values of the base alloy in the range of 5–15 MPa for both as-cast and T5-treated conditions. The tight enhancement in

TABLE 5: Variation in average UTS, YS, % El, Qc, and Q values of the alloys studied in as-cast, SHT, and T5- and T6-treated conditions obtained at ambient temperature.

Alloy	Condition	UTS (MPa)	YS (MPa)	Total strain (%)	Plastic strain (%)	E (GPa)	n	K	q	Q _c (Equation (2))	Q (Equation (1))	Difference (Q - Q _c) (MPa)
M1S	As-cast	260.86	194.39	1.58	1.16	62.66	0.18	580.56	0.07	277.88	290.50	12.62
	SHT	300.34	202.51	6.22	5.74	62.50	0.17	579.26	0.34	512.36	419.43	-92.93
	T5	289.50	258.75	0.94	0.48	63.06	0.13	591.69	0.04	217.06	285.43	68.38
	T6	341.95	324.99	1.1075	0.56	62.45	0.10	576.91	0.06	281.53	348.60	67.07
M2S	As-cast	320.35	235.85	1.63	1.16	67.58	0.19	763.78	0.06	342.14	352.19	10.05
	SHT	327.74	212.61	3.20	2.70	66.23	0.17	759.32	0.16	531.09	403.43	-127.66
	T5	299.97	277.19	0.80	0.34	65.64	0.15	701.96	0.02	167.10	285.22	118.12
	T6	362.33	351.57	0.80	0.24	64.71	0.11	707.79	0.02	196.04	348.16	152.12
M3S	As-cast	325.78	245.72	1.76	1.26	64.73	0.15	648.89	0.08	354.71	362.55	7.84
	SHT	334.75	209.43	3.81	3.30	64.47	0.12	646.47	0.27	549.26	421.97	-127.29
	T5	314.58	279.39	0.98	0.49	64.14	0.13	634.49	0.04	238.21	313.10	74.89
	T6	357.01	355.14	0.75	0.20	64.32	0.08	624.21	0.02	192.02	338.50	146.48
M4S	As-cast	304.40	224.61	1.33	0.87	66.10	0.20	836.77	0.04	293.64	322.87	29.23
	SHT	312.32	207.09	2.22	1.75	66.56	0.18	821.35	0.10	469.59	364.27	-105.32
	T5	276.44	254.14	0.71	0.31	68.43	0.20	857.07	0.02	99.28	254.51	155.24
	T6	315.98	315.98	0.5705	0.11	67.10	0.10	634.74	0.01	77.52	279.41	201.89
M5S	As-cast	297.07	226.59	1.27	0.83	66.91	0.20	792.01	0.04	273.91	312.80	38.89
	SHT	299.79	206.64	1.99	1.54	67.35	0.17	780.35	0.09	437.14	344.62	-92.52
	T5	284.61	277.00	0.66	0.22	64.74	0.15	722.11	0.01	97.20	257.45	160.25
	T6	314.04	314.04	0.56	0.11	67.89	0.11	677.64	0.01	45.43	276.32	230.89

FIGURE 4: Drouzy quality chart representing the relation between the UTS and the percent elongation to fracture values of the alloys studied in the as-cast, SHT, and T5- and T6-treated conditions obtained at ambient temperature.

strength values of the T5-treated conditions can be ascribed to the limited variations in the microstructural features of as-cast structures, as well as to the low proportion of strengthening precipitates in the structure of T5-treated alloys.

In regard to the ductility values in the as-cast condition, Figure 5(c) reveals that the highest ductility value is observed to be associated with the base alloy M1S with a value of

~3.6%, followed by the ductility of the Mn-containing alloys M3S and M5S, and ending up with the lowest ductility values for alloys M2S and M4S, containing 2 and 4 wt.% Ni, respectively. The higher ductility values of the as-cast Mn-containing alloys M3S and M5S can be attributed to the well-refined Si particles, Table 4, and the transformation of a considerable amount of β -Al₅FeSi needles, which may act

FIGURE 5: Variation in average (a) UTS, (b) YS, and (c) % El values of the alloys studied in the as-cast, solution heat-treated (SHT), and T5- and T6-treated conditions obtained at 250°C.

as crack initiators, into the less detrimental α -Al₁₅(Mn, Fe)₃Si₂ phase with script-like and/or sludge morphologies.

The addition of Ni, on the other hand, lowers the ductility values, even for alloy M5S that contains 0.75 wt.% Mn, when compared to alloy M3S, which is Ni-free. The reduction in ductility values of the as-cast Ni-containing alloys can be directly correlated to the presence of acicular Ni-bearing phases with sharp edges, such as Al₃Ni, Al₉FeNi, and Al₃CuNi phases. Variation in chemical compositions has a limited effect on the ductility values obtained at 250°C for the T5-treated alloys, since the maximum absolute difference in the ductility values of alloys M2S through M5S is found to be ~0.44%. Application of T5 and T6 heat treatments reduces the ductility observed in the as-cast condition. The ductility values in the T6-treated condition are generally lower than those obtained with T5 treatment conditions.

The application of the T6 heat treatment enhances the strength values of as-cast conditions regardless the alloy composition, as shown in Figures 5(a) and 5(b). The enhanced strength values of alloy M4S after T6 heat treatment may be attributed to the presence of δ -Al₃CuNi and eutectic Al-Al₃Ni phases that prove to contribute effectively to the elevated-temperature strength of alloy M4S, in spite of a considerable amount of Cu that is consumed in forming the δ -Al₃CuNi phase, which will certainly affect the amount

of fine Al₂Cu dispersoids formed, which is consistent with the findings reported in references [38, 49].

Interestingly, alloys M3S (354 + 0.75 wt.% Mn) and M4S (354 + 4 wt.% Ni) exhibit the highest and almost identical strength values at 250°C for the different conditions examined. Moreover, alloy M3S is considered to be more favorable between the two, since it exhibits higher ductility than that of the Ni-containing alloy M4S. These two alloys exhibit the best strength values (UTS and YS) in the T6-treated conditions among the investigated alloys; whereas, the strength values of the other three alloys, i.e., alloys M1S, M2S, and M5S, are close to each other and lower than the strength values obtained for alloys M3S and M4S by ~36 MPa. This observation would emphasize the advantageous role of adding Mn instead of Ni to enhance the elevated-temperature tensile properties along with its economic impact.

Generally, the closeness of the elevated-temperature strength values of the alloys studied can be credited to the presence of 0.3 wt.% Zr in each alloy, whereby the formation of the fine metastable L1₂-Al₃Zr particles is expected in the microstructures of all the alloys, which, in turn, will improve the alloy strength in a common manner.

For elevated-temperature tensile properties, the concept of the quality index will be discussed according to the concept of Drouzy et al. [47] (Q). Table 6 demonstrates the

TABLE 6: Variation in average UTS, YS, % El, and Q values of the alloys studied in as-cast and T5- and T6-treated conditions obtained at 250°C.

Alloy	Condition	UTS (MPa)	YS (MPa)	% El	Q (Equation (1))
M1S	As-cast	169.95	157.56	3.67	254.65
	T5	172.07	161.51	2.02	217.91
	T6	217.65	213.65	1.92	260.14
M2S	As-cast	186.32	172.84	2.25	239.15
	T5	196.25	178.85	1.60	226.71
	T6	223.72	222.33	2.15	273.65
M3S	As-cast	175.76	162.84	2.85	243.99
	T5	181.07	177.45	1.61	212.16
	T6	248.74	245.62	1.54	276.68
M4S	As-cast	180.53	167.35	1.77	217.57
	T5	181.86	179.13	1.32	199.85
	T6	253.58	253.32	1.06	257.35
M5S	As-cast	176.48	162.32	2.45	234.92
	T5	191.10	184.02	1.76	227.86
	T6	222.09	220.55	1.32	239.96

elevated-temperature tensile data along with the quality index values (Q) of the alloys studied calculated using Equation (1). Figure 6 shows the quality chart obtained based on the calculations of Drouzy et al. [47]. As may be seen, the quality index values obtained at 250°C do not show wide variation in values, as was observed in the case of the ambient-temperature data. This limited variation can be understood in light of the balanced variation in UTS and ductility values obtained at the elevated temperature of 250°C. For example, the base alloy M1S in the as-cast condition exhibits the highest ductility value of 3.67% and a UTS value of 169.95 MPa, while the lowest ductility is experienced by alloy M4S for the T6-treated condition with a value of 1.06% along with a UTS value of 253.58 MPa. By calculating the quality indices of these two conditions, they reveal Q values of 263.95 and 257.35 MPa for as-cast M1S and T6-treated M4S, respectively. Those two extreme conditions show that despite the considerable variation in the UTS values on the one hand, and ductility values on the other, for these two conditions, the quality indices in both cases remain almost unchanged due to the balanced trade-off between the UTS and ductility values. The relatively low UTS and ductility values obtained at elevated temperature (250°C) for the T5-treated condition result in the T5-treated alloys exhibiting minimum Q values among the conditions studied.

Another interesting observation is that the quality index values for alloys M2S and M3S in the T6-treated condition are found to be the maximum for the alloys and conditions studied. This observation highlights the enhanced characteristics of alloy M3S, which contains 0.75 wt.% Mn and emphasizes the positive influence of the high Mn-addition on the elevated-temperature tensile properties, which are found to be more or less comparable to those obtained with the addition of 2 and 4 wt.% Ni to the same base alloy.

The favorable role of the blocky α -Al₁₅(Fe, Mn)₃Si₂ phase, termed as sludge particles, in improving the tensile properties of alloy M3S at 25°C and 250°C can be witnessed

from the fractograph of alloy M3S after testing at 250°C presented in Figure 7. The sludge particles appear to retard the propagation of the cracks developed in the other phases, which will subsequently enhance the mechanical properties of the respective alloy. Yet, another interesting observation made from this figure is that, while many of the intermetallic phase particles appear cracked, as indicated by the solid arrows, the sludge particles, however, are crack-free. Whereas the multiple cracked Ni-rich phases exist on the fracture surface of alloy M2S after testing at 250°C are believed to contribute to the crack initiation process and hence deteriorate the tensile properties of this alloy as witnessed in Figure 8.

3.2.3. Hardness Values. Figure 9 illustrates the variation in the hardness values of the alloys as a function of the applied heat treatment. At first glance, one can observe from Figure 9 that the hardness values of different alloys show insignificant variations for the same conditions. For each alloy, the peak-aged condition exhibits the highest hardness value among all conditions. Also, the tailored alloys, i.e., M2S through M5S, show better hardness values than those obtained for the base alloy in all the conditions studied.

Variations in hardness values of the alloys in the as-cast condition can be attributed to the additions of Ni and/or Mn made to the base alloy M1S. It was seen that additions of Ni and/or Mn in various amounts increased the volume fractions of intermetallic compounds considerably, as listed in Table 3. The variations in hardness values follow the same trend as variations in the percentage volume fraction of intermetallic compounds. Thus, the base alloy M1S exhibits the lowest hardness value in the as-cast condition, having the lowest volume fraction according to Table 3; and the highest hardness value of the same condition is associated with alloy M4S that has the highest volume fraction of intermetallic compounds. This observation highlights the effective role of intermetallic compounds in enhancing the hardness values [19, 50].

The dissolution of the strengthening elements over the course of the solution treatment reduces the hardness values, in spite of the expected improved homogeneity in composition and evolution of the eutectic silicon morphology following solution treatment. This behavior emphasizes the crucial role of intermetallic phases in influencing the mechanical performance of alloys. It is established that the hardness value of a specific alloy corresponds to the combination of the tensile yield strength and work-hardening rate of the alloy [50, 51]. The order of the alloys studied according to the hardness value in the as-quenched condition matches to a large extent their order with respect to their yield strength in the same as-quenched condition (Figure 3).

Direct artificial aging following casting of test bars, i.e., T5-temper treatment, introduces slight improvements in the hardness values with respect to those obtained for the as-cast condition. This can be attributed to the limited changes in the microstructure of the as-cast alloys/bars following direct artificial aging without solution treatment. The slight

FIGURE 6: Drouzy quality chart representing the relation between the UTS and the percent elongation to fracture values of the alloys studied in the as-cast and T5- and T6-treated conditions obtained at 250°C.

(a)

(b)

FIGURE 7: SEM images of T6-treated M3S alloy after testing at 250°C: (a) BSE image showing various cracked intermetallic phases and crack-free sludge particles and (b) EDS spectrum corresponding to the point of interest in (a).

increase in the hardness values in the T5-treated condition emphasizes the positive role of employing a high solidification rate in the casting process. The high solidification rate allows for partial solubility of Cu and Mg in the α -Al matrix, such that subsequent artificial aging will precipitate a limited amount of strengthening precipitates, to produce the marginal increase in hardness values observed.

With respect to the peak-aged condition, the hardness values of alloys M2S through M5S are almost identical approaching ~100 HRF, whereas the hardness value of the base alloy for the same T6-treated condition is ~96 HRF. This variation can be ascribed to the combined effect of the strengthening precipitates and intermetallic compounds in

the four alloys. The improvement in hardness of the base alloy in the T6-treated condition compared to the as-cast case is mainly attributed to the effect of the strengthening precipitates formed after the T6 treatment because of the low volume fraction of intermetallic phases observed in the microstructure of the base alloy as listed in Table 3 (namely, ~2.51% in the as-cast condition and 1.11% in the as-quenched condition), which is too low compared to the other alloys.

3.2.4. Impact Properties. The impact bars used in the present study were not notched based on three considerations: (i) the expected low toughness of 354-type alloys, (ii) increasing the

FIGURE 8: SEM images of T6-treated M2S alloy after testing at 250°C: (a) BSE image showing microcracks associated with Ni-rich phases and (b) EDS spectrum corresponding to the point of interest in (a).

FIGURE 9: Variation in Rockwell hardness value (HRF) as a function of heat treatment conditions for the alloys studied.

FIGURE 10: Variation in total impact energy value as a function of heat-treatment conditions for the alloys studied.

measurement accuracy by excluding uncertainties associated with machining of notches, and (iii) emphasizing the effects of microstructural constituents.

The variation in the toughness values of the alloys studied as a function of the applied heat treatment is displayed in Figure 10. It is evident that values of the total absorbed energy for the alloys studied are relatively low in the as-cast, T5-treated, and T6-treated conditions compared to those obtained in the solution heat-treated conditions.

The morphology of the eutectic silicon particles and the volume fraction of intermetallic compounds [52] present are supposed to determine the impact properties of the as-cast alloys studied in the present investigation. It is worth mentioning that the sphericity and roundness parameters (in percentage) of the eutectic silicon particles in the as-cast condition did not vary substantially with respect to the alloy composition, as shown in Table 4. Therefore, the only

parameter affecting the impact properties of the as-cast alloys is the presence of intermetallic compounds. As can be inferred from Figure 10, the order of alloys according to the absorbed energy during impact testing matches that with respect to the volume fraction of intermetallic compounds shown in Table 3. Thus, it can be deduced that increasing the volume fraction of intermetallic compounds will increase the amount of absorbed energy and hence improve the impact properties.

Thermal-modification of eutectic silicon particles associated with solution heat treatment can contribute positively to the impact properties by (i) producing more rounded edges of the Si particles instead of the relatively sharp edges obtained in as-cast conditions and hence better resistance to cracks initiation and propagation, resulting in higher toughness values, and (ii) producing well-separated silicon particles through the fragmentation of the interconnected

fibrous silicon structure present in the Sr-modified as-cast structures, which will make available greater areas of the ductile α -Al matrix and hence improve the impact properties [50]. In light of the aforementioned points, the impact properties of the alloys studied in the as-cast condition substantially improved by applying solution heat treatment at 495°C for 5 hours. The increase in the total absorbed energy values for each alloy after solution treatment are as follows: (i) 15 J for the base alloy M1S, (ii) 11 J for alloy M2S, (iii) 12 J for alloy M3S, (iv) 9 J for alloy M4S, and (v) 8 J for alloy M5S.

The impact properties of an alloy are directly related to its ductility. Figure 11 illustrates the relationship between the impact properties and ductility values of the alloys studied. The relationship between these two properties shows a linear trend with a high goodness of fitting represented by the high value of R^2 . The order of alloys with respect to their impact energy values (Figure 10) matches with the order of the alloys with respect to the ductility values (Figure 3(c)) obtained from room temperature tensile testing.

The impact properties of the investigated alloys in T5- and T6-treated conditions are close in values and lower than the values obtained in the as-cast and as-quenched conditions, respectively. For the alloy studied, the presence of fine precipitates in T5- and T6-treated conditions promotes the initiation of fine cracks, which will eventually reduce the impact properties [53]. This would explain the reduced impact properties observed in the T5- and T6-treated conditions of the alloys studied.

The Al_2Cu phase particles seem to control the impact properties of the alloys studied rather than the eutectic silicon particles due to the following observations. The impact properties of the investigated alloys have no significant variations with respect to the condition studied, i.e., the total energies absorbed by the five alloys in the as-cast condition are close in their values, the same for the T5- and T6-treated conditions. This may be ascribed to the same copper content in the studied alloys and the existence of Al_2Cu -phase particles in their microstructures. On the other hand, the total absorbed energy values vary widely for the as-quenched alloys. This wide variation can be attributed to the dissolution of Al_2Cu phase particles during the course of solution treatment, so that the impact properties are no longer dependent on the Al_2Cu particles but on other microstructural features reported to have noticeable differences. Similar observations have been previously reported for copper containing alloys by Paray et al. [53].

4. Conclusions

This article discussed the effects of the addition of transition elements and applied heat treatments on the microstructural characteristics of test bars of Zr-containing 354-type Al-Si-Cu-Mg cast alloys including volume fractions of intermetallic compounds formed and the eutectic silicon particle characteristics followed by evaluating the ambient- and elevated-temperature tensile properties, hardness, and impact properties. The most important findings are as follows:

FIGURE 11: Correlation between impact energy and ductility values of the alloys studied in the solution-heat treated condition.

- (1) The proposed additions and heat treatments enhance the overall mechanical performance of the alloys, namely, the ambient- and elevated-temperature tensile properties, hardness, and impact properties
- (2) For the Mn-containing alloys, the improvement in properties results from the formation of polygonal sludge particles in the form of blocky α - $\text{Al}_{15}(\text{Fe}, \text{Mn})_3\text{Si}_2$ alongside the script-like α -iron phase, which resists crack propagations
- (3) Blocky sludge particles prove to be beneficial to the mechanical properties through blocking the propagation of cracks and stand crack-free after the tensile testing till fracture
- (4) Alloys M3S (354 + 0.75 wt.% Mn) and M4S (354 + 4 wt.% Ni) exhibit the highest and almost identical strength values at 250°C for the different conditions examined
- (5) Addition of 0.75 wt.% Mn to the base alloy is considered more favorable than adding 4 wt.% Ni, since it results in higher ductility values (cf. 1.54% in alloy M3S and 1.06% in M1S for T6-treated conditions tested at 250°C); this finding is of considerable economic benefits
- (6) The variations in hardness values and impact properties followed the same trend as variations in the percentage volume fraction of intermetallic compounds
- (7) The impact properties of the alloys are highly influenced by the Al_2Cu phase particles rather than the eutectic silicon particles

Data Availability

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

Conflicts of Interest

The authors declare that they have no conflicts of interest.

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