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Effect of Cu and Ni on the mechanical properties and fracture behavior of Al–Si–Mg cast alloys

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The effect of addition of Cu or Ni on the microstructure and mechanical properties of an Al-6Si-0.5Mg alloy was investigated. The alloys were cast into pre-heated metal moulds, homogenized (24 h at 500°C), solution-treated (2 h at 540°C) and aged for 1 h at temperatures of up to 300°C; the age-hardening stages were monitored by hardness measurements. Tensile properties of the aged alloys were evaluated at nominal strain rates of 10^{-3} s^{-1}. The yield strength and fracture strength of the alloys were found to increase with increasing ageing temperature and to be maximum in the peak-aged condition (1 h at 225°C), whereas the ductility and impact toughness decreased with increasing ageing temperature. The addition of 2 wt.% Cu to the Al-6Si-0.5Mg alloy resulted in an increase of the hardness and tensile strength and substantial reduction of the ductility and fracture toughness. The fracture toughness of the Al-6Si-0.5Mg-2Ni alloy, containing 2 wt.% Ni, was lower for all ageing treatments and lowest at the peak-aged condition. SEM fractographs of broken specimens showed both ductile and ductile-brittle fracture behavior.

Al-6Si-0.5Mg alloy / Microstructure / Tensile strength / Fractograph

Introduction

Age-hardened Al-alloys are widely used in engineering applications, due to the significant improvement in yield strength and hardness obtained by controlled thermo-mechanical treatments. Unstable fast fracture, even if the alloy is ductile, becomes frequent because the strengthening lowers the level of toughness, and this becomes a problem for large scale structures. Since fracture of many engineering components is promoted under dynamic conditions, there is a need to understand the fracture behavior of materials under dynamic loads. Moreover, knowledge about fracture characteristics under impact load will probably become more important, because the applications in transportation vehicles will increase [1].

Al–Mg–Si alloys have been widely used in transportation systems owing to their fair strength, weldability and corrosion resistance. The precipitation sequence of solution-treated Al–Mg–Si ternary alloys during artificial aging has been reported to be: α supersaturated solid solution (SSS) → GP-I zones → metastable needle-like β″ precipitates (or so-called GP-II zones; formed through the transformation of GP-I as nuclei) → metastable rod-like (or lath-like) β′ precipitates → stable β phase [2].

The mechanical properties of cast Al–Si alloys largely depend on grain size and morphology, dendrite arm spacing (DAS), amount and distribution of secondary phases [3]. Structural refinement leads to improvement of the mechanical properties. Addition of small amounts of Cu, Mg or Ni strengthens Al–Si alloys, and also the presence of Si provides good casting properties [4]. Addition of Cu to Al–Si alloys leads to the formation of the Al_2Cu phase and other intermetallic compounds, which influence the strength and ductility [3-7]. Among the elements added to Al–Si–Mg alloys for increasing strength and grain-size control, copper has attracted considerable attention. Cu additions reduce the natural aging rate of Al–Mg–Si alloys, but generally increase the kinetics of precipitation during artificial aging [8]. In addition to the phases that precipitate in the ternary alloys, the equilibrium precipitate in high-Cu Al–Mg–Si/Cu alloys, Q, was identified as $Al_5Cu_2Mg_8Si_6$ or $Al_4CuMg_5Si_4$ [9,10]. Additions of Ni lead to the formation of Al_3Ni in the aluminum matrix. Due to the high values of Young's modulus and tensile strength, the compound Al_3Ni has been applied as reinforcement for aluminum-matrix composites. [11,12]. A report regarding the properties of the Al–Zn–Mg/Al_3Ni system indicates that an increased amount of Al_3Ni not only raises the yield strength but also reduces the time for peak-aging [13].

Considering that Cu and Ni can strengthen aluminum alloys through precipitation and dispersion hardening, respectively, the objective of this study was to explore their influence on the microstructure characteristics and tensile properties of an Al-6Si-0.5Mg alloy.

Experimental

Melting was carried out in a natural gas heating pot furnace under a suitable flux cover (degasser, borax, *etc.*). In the process of preparation of the alloys, an Al–Si master alloy and pure aluminum (99.7 % purity) were melted first. Cu (99.98 % pure), in the form of sheet, was added by plunging to Alloy-2, while Ni chips were charged into the bottom of the crucible for Alloy-3. Finally, magnesium ribbon (99.7 % purity) was added into the liquid metal solution. The final temperature of the melt was always maintained at 910±15°C. The melt was degassed with solid hexachloroethane (C_2Cl_6) and homogenized by stirring at 700°C before casting. Casting was done into metal moulds, 15 mm×150 mm×300 mm, preheated to 200°C. All the alloys were analyzed by wet chemical and spectroscopic methods. Table 1 shows the compositions of the alloys.

The cast samples were first cleaned properly to remove the oxide layer from the surface. All the alloys were homogenized at 500°C for 24 h. The specimens for hardness, tensile and impact measurements were prepared from the homogenized alloys according to ASTM standards. The homogenized samples were solution-treated at 540°C for 2 h and quenched into a solution of salt in iced water. For hardness and resistivity measurements the solutionized samples were aged for 1 h at different temperatures up to 400°C.

The hardness was measured by a Rockwell hardness testing machine (F scale) and the average of seven consistent readings was accepted as the representative hardness value of the alloy.

For mechanical properties measurement, the tensile and impact samples were aged for 1 h at temperatures up to 300°C. Tensile testing was carried out at a strain rate of 10-3 s-1. The averages of three consistent test results were accepted as the tensile test values for the corresponding samples. Charpy impact test samples, 10 mm × 10 mm × 55 mm, with a 45° V-notch (2 mm depth and 0.25 mm root radius) were hit by a pendulum at the opposite end of the notch. The absorbed energy required to produce two fresh fracture surfaces was recorded in Joule. Selected samples were observed under a Scanning Electron Microscope and also under an optical microscope.

Results and discussion

Microstructure

Fig. 1, Fig. 2 and Fig. 3 show optical micrographs of as solution-treated alloys Al-6Si-0.5Mg, Al-6Si-0.5Mg-2Cu, and Al-6Si-0.5Mg-2Ni. The α(Al) face-centered-cubic solid solution is the predominant phase (light grey) in the microstructure of these alloys. The silicon phase, which is soluble in aluminum, and the other alloying elements, form a binary eutectic with α(Al). In the micrographs, the Si eutectic and primary particles are dark grey. The Al_2Cu phase is not well seen; it appears in the form of small particles, slightly darker than the white aluminum matrix. The change of morphology of the eutectic Si after solution treatment is obvious – the plate-like eutectic Si has broken into small particles. The fragmentation process was accelerated by the homogenizing (24 h at 500ºC) and solutionizing (2 h at 540°C), as can be seen in Fig. 1, Fig. 2 and Fig. 3. During solutionizing, the Si particles underwent coarsening.

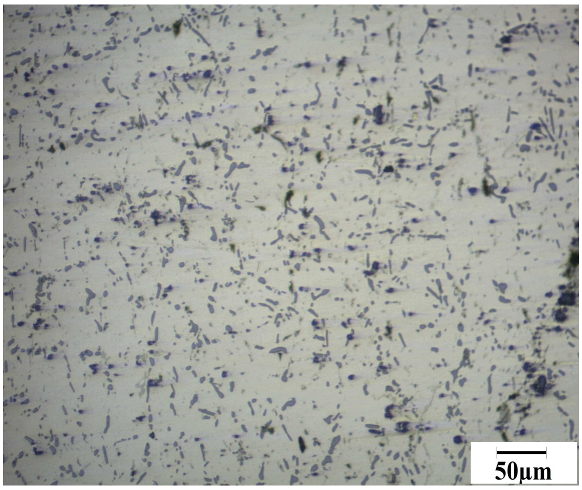


Fig. 1 Microstructure of a solution-treated Al-6Si-0.5Mg cast alloy (Alloy-1).

Table 1 Chemical composition of the alloys used in this work (wt.%).

Alloy	Si	Mg	Cu	Ni	Fe	Mn	Ti	Al
Al-6Si-0.5Mg (Alloy-1)	5.802	0.441	0.006	0.006	0.146	0.002	0.099	Ballast
Al-6Si-0.5Mg-2Cu (Alloy-2)	5.801	0.497	1.980	0.003	0.300	0.004	0.094	Ballast
Al-6Si-0.5Mg-2Ni (Alloy-3)	5.935	0.440	0.007	2.220	0.141	0.003	0.088	Ballast

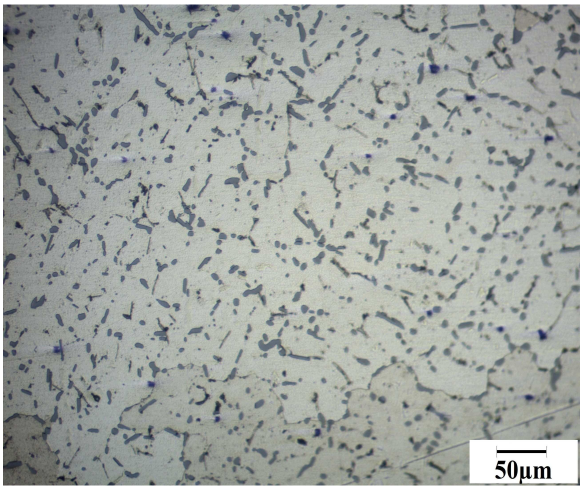


Fig. 2 Microstructure of a solution-treated Al-6Si-0.5Mg-2Cu cast alloy (Alloy-2).

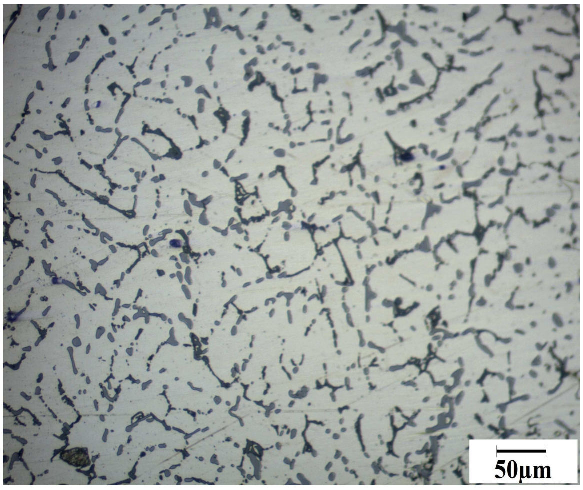


Fig. 3 Microstructure of a solution-treated Al-6Si-0.5Mg-2Ni cast alloy (Alloy-3).

When the alloys were kept at the temperatures of homogenizing and solutionizing for a longer period of time, shape perturbations in the Si particles began to arise, until the particles were broken into a series of spherical crystals. This process happened due to the instability of the interfaces between two different phases and was driven by a reduction of the total interfacial energy.

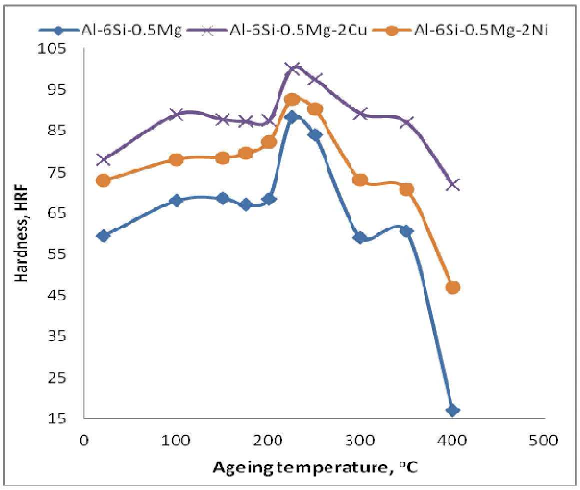
The Al-6Si-0.5Mg-2Cu alloy shows eutectic acicular silicon and very few coarse primary silicon particles, embedded in the dendritic aluminum matrix. The Al_2Cu particles are rather coarse, mainly elongated along the grain boundaries, but also forming small pockets (Fig 2).

The Al-6Si-0.5Mg-2Ni alloy reveals the formation of Al_3Ni in the aluminum matrix through eutectic reaction during solidification. Ni stabilizes the contiguity of the eutectic network by increasing the volume fraction of rigid phases (Si + Al_3Ni) in the eutectic (Fig. 3).

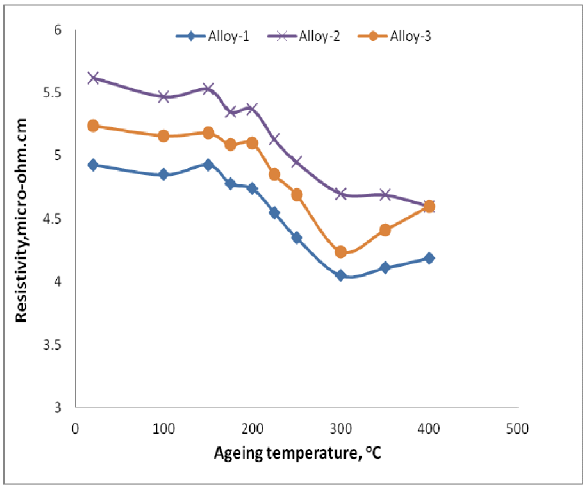
Properties

Effect of Cu or Ni on age-hardening of Al–Si–Mg cast alloys

Fig. 4 shows hardness and resistivity against ageing temperature after ageing for 1 h. When Cu or Ni is added to Alloy-1, the hardness increases strongly in both the as-quenched condition and in the peak-aged condition (maximum hardness), but the electrical resistivity decreases. For all of the alloys the peak-aged condition was achieved by ageing for 1 h at ~225°C. The hardness values of Alloy-2 (2 wt.% Cu) and Alloy-3 (2 wt.% Ni) are greater than the hardness of the base alloy (Alloy-1). The hardness of Alloy-2 is greater than that of Alloy-3 at all ageing conditions. Thus the effect of Cu on Al-6Si-0.5Mg precipitation hardening is greater than that of Ni.



a



b

Fig. 4 Variation in hardness (a) and resistivity (b) with ageing temperature for alloys aged for 1 h (Table 2).

The hardness decreased with increasing aging temperature beyond 250°C. However, the Cu-bearing alloy exhibited stronger resistance to softening, in comparison with the Ni-bearing (Alloy-3) and base Al-6Si-0.5Mg alloy (Alloy-1).

Effect on fracture strength

Fig. 5 shows the effect of ageing temperature on the ultimate tensile strength (fracture strength) of the alloys. The fracture strength of the Cu-containing alloy (Alloy-2) was higher than for the Ni-containing alloy (Alloy-3) and the base alloy (Alloy-1, containing neither Cu nor Ni) at all ageing conditions. In the solution-treated condition the increase in fracture strength was 59.09 % for the Cu-containing alloy (Alloy-2) and 34.55 % for the Ni-containing alloy (Alloy-3). In the peak-aged condition, the fracture strength of the alloy containing Cu (Alloy-2) was 21.14 % higher than for the same alloy in the solution-treated condition and 48.25 % higher than for the peak-aged Al-6Si-0.5Mg alloy. The alloy containing Ni (Alloy-3) showed 20.95 % higher fracture strength at the peak-aged condition than the same alloy in the solution-treated condition, and 25.17 % higher fracture strength than the peak-aged Al-6Si-0.5Mg alloy. Increasing the ageing temperature beyond 225°C caused a drop in the fracture strength.

Effect on yield strength

Fig. 6 illustrates the influence of Cu or Ni additions on the yield strength (0.2 % proof strength) of the Al-6Si-0.5Mg cast alloy at various ageing temperatures. The yield strength of the alloys increased with increasing ageing temperature, the maximum being attained at the peak-aged (1 h at 225°C) condition. In the as-quenched condition, the alloy containing Cu (Alloy-2) showed 63.63 % and the alloy containing Ni 47.73 % higher yield strength than the base alloy (Alloy-1) containing neither Cu nor Ni. In the peak-aged condition the values were 62.39 % and 27.35 % higher, respectively, than for Alloy-1. The intermetallic particles may contribute as a reinforcement agent in the Al-alloy matrix. The higher strength may be attributed to the effect of precipitation hardening.

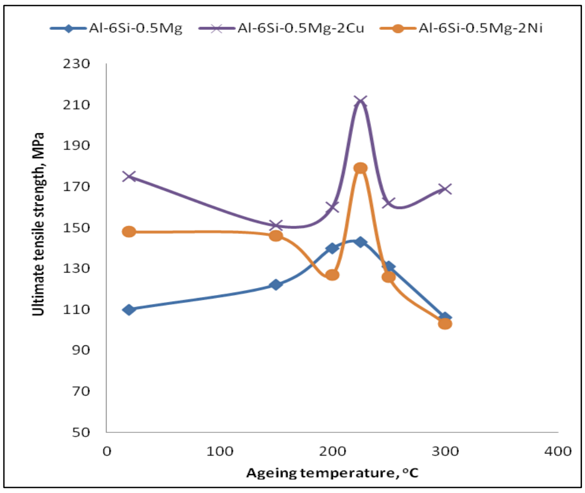


Fig. 5 Variation of fracture strength with ageing temperature for alloys aged for 1 h (Table 3).

Effect on ductility

Fig. 7 shows the influence of Cu or Ni on the ductility (% elongation) of the Al-6Si-0.5Mg alloy for different ageing temperatures. In the peak-aged condition (1 h and 250°C), the alloy containing Ni (Alloy-3) showed the lowest ductility. The change in ductility in the alloy containing Cu (Alloy-2) was not significant in the peak-aged condition. At temperatures beyond 250°C, the ductility of the alloys increased significantly.

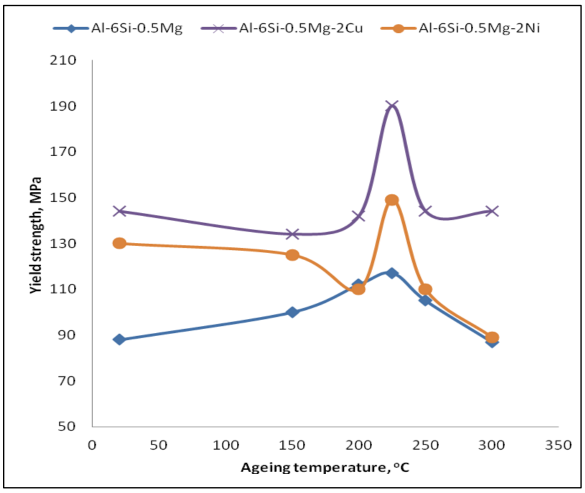


Fig. 6 Variation of yield strength with ageing temperature for alloys aged for 1 h (Table 3).

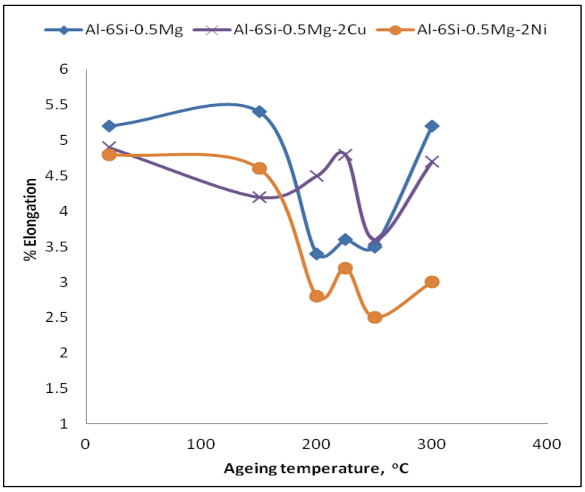


Fig. 7 Variation of elongation with ageing temperature for alloys aged for 1 h (Table 3).

Effects on impact energy

Fig. 8 shows the absorbed energy of the alloys as a function of the artificial ageing temperature. The heat treatment, i.e. solutionizing and ageing, greatly influenced the capacity of absorbing energy. The composites showed higher toughness in the as solution-treated condition, than in the peak-aged (1 h at 225°C) condition. The toughness of the Al-6Si-0.5Mg alloy (Alloy-1) is higher than that of the Ni-bearing Alloy-3. The Ni-containing alloy shows the lowest fracture toughness at the peak-aged condition and low values also for the other ageing conditions. The change in fracture toughness of the Cu-containing alloy (Alloy-2) shows a similar trend as that of the Ni-containing alloy (Alloy-3).

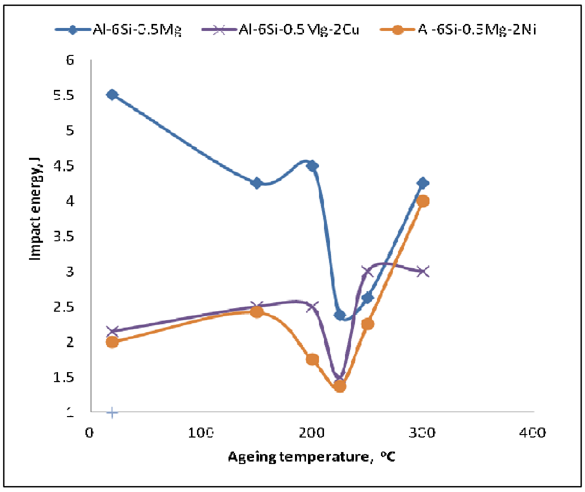


Fig. 8 Variation of impact energy with ageing temperature for alloys aged for 1 h (Table 3).

Effect on fracture behavior

Fig. 9, Fig. 10 and Fig. 11 show the fractured surfaces (after tensile test) of the alloys at the peak-aged condition. The surfaces appear rough and normal to the axis of loading. On a microscopic scale, the fracture surfaces seem to contain many microvoids in the matrix. The dimples are neither uniform nor circular in shape. Precipitate particle fracture, interface debonding and matrix crack are the main failure modes. Matrix-intermetallic particles decohesion is also observed for these alloys. The fracture mechanism is ductile, involving nucleation, growth, and coalescence of voids in the matrix around the intermetallic particles. The voids grow under the applied load and the influence of local plastic constrain, until a coalescence mechanism is activated, followed by the total failure of the alloys.

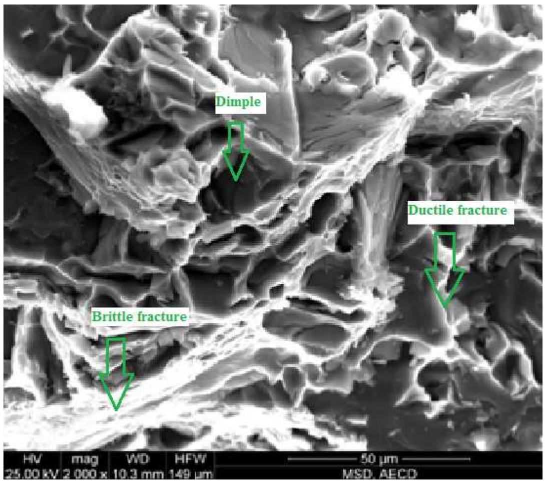


Fig. 9 SEM fractograph of Alloy-1, aged at 225°C for 1 h and tensile tested at a strain rate of 10-3 s-1.

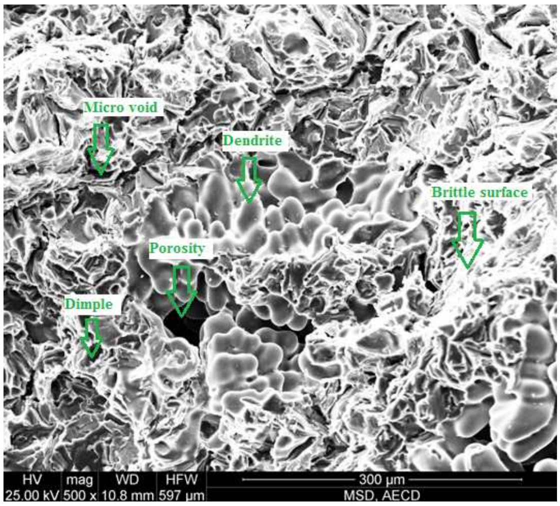


Fig. 10 SEM fractograph of Alloy-2, aged at 225°C for 1 h and tensile tested at a strain rate of 10^{-3} s^{-1}.

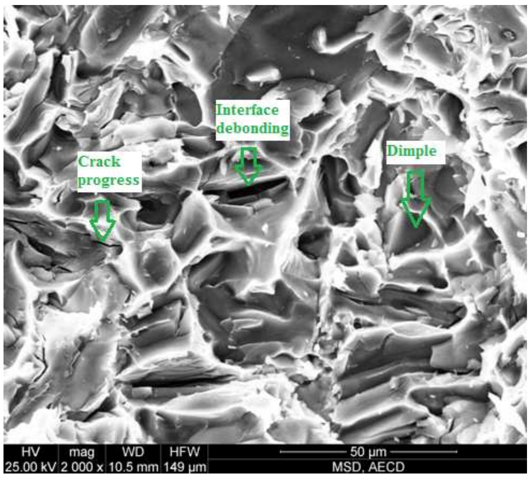


Fig. 11 SEM fractograph of Alloy-3, aged at 225°C for 1 h and tensile tested at a strain rate of 10^{-3} s^{-1}.

Table 2 Hardness (HRF) and resistivity (in μΩ·cm) after ageing for 1 h at different temperatures.

Alloy	Property	20°C	100°C	150°C	175°C	200°C	225°C	250°C	300°C	350°C	400°C
Alloy-1	Hardness	59.5	68.1	68.6	67	68.5	88.3	84	59	60.6	17
	Resistivity	4.93	4.85	4.93	4.78	4.74	4.55	4.35	4.05	4.11	4.19
Alloy-2	Hardness	78	89	87.8	87.3	87.5	100.1	97.5	89.2	87.1	72
	Resistivity	5.62	5.47	5.53	5.35	5.37	5.13	4.95	4.7	4.69	4.6
Alloy-3	Hardness	73	78.1	78.5	79.5	82.4	92.6	90.4	73.1	70.7	47
	Resistivity	5.24	5.16	5.18	5.09	5.1	4.85	4.69	4.24	4.41	4.6

Table 3 Mechanical properties after different ageing treatments.

Alloy	Ageing treatment	Fracture strength, MPa	Yield strength, MPa	Elongation, %	Impact energy, J
Al-6Si-0.5Mg (Alloy-1)	None	110	88	5.2	5.5
	1 h at 150°C	122	100	5.4	4.25
	1 h at 200°C	140	112	3.4	4.5
	1 h at 225°C	143	117	3.6	2.38
	1 h at 250°C	131	105	3.5	2.63
	1 h at 300°C	106	87	5.2	4.25
Al-6Si-0.5Mg-2Cu (Alloy-2)	None	175	144	4.9	2.15
	1 h at 150°C	151	134	4.2	2.5
	1 h at 200°C	160	142	4.5	2.5
	1 h at 225°C	212	190	4.8	1.5
	1 h at 250°C	162	144	3.6	3
	1 h at 300°C	169	144	4.7	3
Al-6Si-0.5Mg-2Ni (Alloy-3)	None	148	130	4.8	2
	1 h at 150°C	146	125	4.6	2.42
	1 h at 200°C	127	110	2.8	1.75
	1 h at 225°C	179	149	3.2	1.37
	1 h at 250°C	126	110	2.5	2.25
	1 h at 300°C	103	89	3	4

Conclusions

Addition of Cu or Ni to Al-6Si-0.5Mg cast alloys resulted in improved strength, but reduced ductility and impact fracture toughness. The effect of Cu addition to the Al-6Si-0.5Mg cast alloy was more effective than Ni addition in improving the properties. Maximum tensile strength was found after ageing at ~225°C for 1 h. 2 wt.% Ni addition to the Al–Si–Mg cast alloy reduced the ductility maximum, as compared to same wt.% Cu addition. Consequently, at a higher value of tensile strength, Cu has better effect on ductility than Ni. The investigated alloys showed ductile or ductile-brittle mixed fracture behavior during tensile testing.

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