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Properties of Al-6Mg-*x*Sc (*x* = 0 to 0.6 wt.%) Alloy Subjected to Thermal Treatment: a Review

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Abstract. Aluminum–magnesium alloys are regularly used to manufacture different types of sheets, automotive trims, and architectural components, which are very intricate in shape. Additionally it is important due to their excellent properties of high-strength to weight ratio, corrosion resistance and weldability. Magnesium increases the strength of the alloys but there is a tendency to form β-phase Al_8Mg_5 compound, usually denoted as Al_3Mg_2 precipitates along grain boundaries to facilitate intergranular fracture. Numerous studies have been conducted on these alloys to make their use potential as different places. The use of scandium in Al-Mg alloys is meant for taking the advantage of grain refinement along with the unique precipitation strengthening behavior through the formation of Al_3Sc precipitates with aluminum, a stable LI_2 phase coherent with the matrix. The purpose of this paper is to review and discuss recent developments on Al-6Mg alloy through scandium addition at different levels under thermal treatment.

1. INTRODUCTION

Aluminum-magnesium alloys belong to the 5xxx series alloys which were first developed in the 1930s [1]. In view of the areas of application, the properties of interest of these alloys are said to be sheet materials with high strength, high formability, high toughness, good corrosion resistance, superior machinability and weldability [2-4]. Al-Mg alloys also have a good application as cast alloys. These alloys are often used to fabricate different types of automotive trims and architectural components, which are very complex in shape [5,6]. Incidentally, this particular alloy system is not age hardenable and hence is made to derive its strength through work hardening. Increasing magnesium content in Al-Mg alloy is able to increase its work hardening rate [7,8]. However this leads to a higher tendency for intercrystalline failure presumably due to the formation of β-phase Al_8Mg_5 or Al_3Mg_2 precipitates at the slip band and grain boundaries. Upon cold working this particular problem is aggravated. Again the heavily cold worked alloy suffers from the problem of age softening due to local recovery. So the stability of the alloy in respect of time remains to be adequate unless the structure is stabilized by low temperature annealing. Unfortunately this reduces the strength of the alloy [9-11].

Properties of aluminum and its alloys can be extensively improved through the accumulation of transition metals like Ni, Fe, Cr, Ti, V, Er, Zr, and Sc etc. [12-16]. A good amount of works regarding the use of these elements in Al-Mg alloys has been reported [17-20]. It was found that the modifying action of scandium on cast aluminum is non-uniform. The structure is substantially refined with increase in the concentration of scandium [20,21]. The use of scandium in Al-Mg alloys is meant for taking the advantage of the unique precipitation hardening behavior of scandium. The Al-Sc solid solution decomposes to form a fine dispersion of the homogeneously nucleated equilibrium LI_2 Al_3Sc precipitates, which can produce a significant ageing response. A small addition of scandium to aluminum-magnesium or other aluminum alloys significantly increases the hardness and tensile strength during ageing [22-24]. Scandium does not form any second phase intermetallic compounds with other alloying elements such as iron, man-

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ganese and chromium. These precipitates are also effective on stabilising substructure, thus allowing the use of strain hardening and stabilisation treatments to improve the strength properties quite considerably [1,25].

There is a limitation for the alloys with a high Mg content. From the available 5xxx series alloys, this contains about 6 wt.% magnesium considered as the strongest non-heat-treatable aluminum alloy. Despite this, it is considered as the most demanding material in different uses due to a good combination of properties [26-28]. This paper is intended to review all the experimental results along with the analysis to study the above mentioned issues. These are related to the age hardening, microstrucral modification, tensile, fractural, DSC and texture behavior of Al-6Mg alloy containing varying amounts of scandium under thermal treatment.

2. INVESTIGATED MATERIALS AND ROUTINE

Investigated aluminum-magnesium alloys containing scandium at various levels were prepared by melting of commercially pure aluminum, magnesium ribbon and aluminum-scandium master alloy. In this purpose a resistance heating pot furnace was used. Chemical compositions of the following four alloys showing the contents of additional minor impurity elements of Cu, Fi, Mn, Si, Zn etc. are stated elsewhere [29,30].

Alloy 1	Al-6 wt.% Mg
Alloy 2	Al-6 wt.% Mg-0.2 wt.% Sc
Alloy 3	Al-6 wt.% Mg-0.4 wt.% Sc
Alloy 4	Al-6 wt.% Mg-0.6 wt.% Sc

The cast and cold deformed alloys were aged at different temperatures for one hour. Vickers hardness of different aged alloys was measured at 5 kg load to evaluate the age hardening effect of the alloys. Electric resistivity was calculated from conductivity data which was measured with an Electric Conductivity Meter, type 979. Tensile test was carried out according to ASTM specification. Optical and Scanning Electron Microscope investigation were carried out in usual way [29]. The DSC scan of the alloys was conducted in a Du Pont 900 instruments under inert N_2 gas atmosphere over a temperature range from 50 °C to 650 °C at heating rate of 10 °C/min. The activation energy of transformations in different conditions was calculated by using the method of Nagasaki-Maesono analysis [31]. The X-ray diffraction analyses of the alloys were carried out by using a PHILIPS PW1830 diffractometer with Cu-K_α radiation.

3. REVIEW ON PROPERTIES

Cold rolling behaviour.

Kaiser [29] conducted a test on cold rolling behavior of Al-6Mg alloy with different amount of Sc ranging from 0 to 0.6 wt.% Sc. The results in providing the performance of the hardness of the alloys increase with the deformation because of the strain hardening effects. The increases of rolling reduction means more plastically deformed of the metal, as a result additional dislocations density are generated [32]. The supplementary dislocations inside a material makes more strengthening of the material. Higher hardness values were obtained for higher Sc added alloys due to the grain refining effects as well as fine precipitates of Al_3Sc throughout casting [10,33]. Fig. 1 shows the variation of hardness of different Sc added Al-Mg alloys with the percentages of cold deformation.

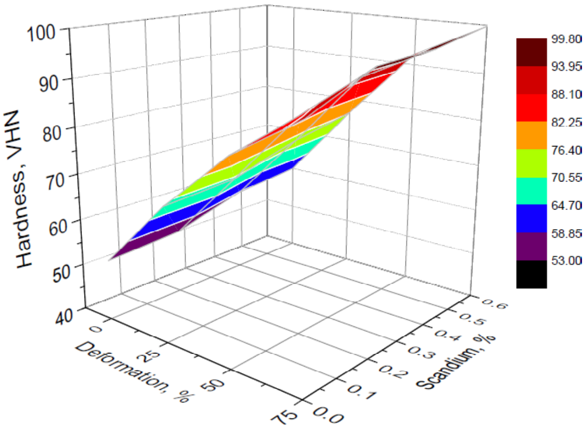

Fig. 1. The influence of scandium content and degree of deformation on the hardness of Al-6Mg alloys.

Age-hardening behaviour.

Kaiser et al. [30] investigated the effect of the Sc content on the age-hardening behavior of cast Al-6Mg alloys. Fig. 2 shows the results of isochronal ageing at different temperature for one hour of the alloys with different Sc content. Addition of scandium has shown appreciable ageing response. Without scandium the alloy do not exposed any ageing response however shown a continuous softening at increasing ageing tempera-

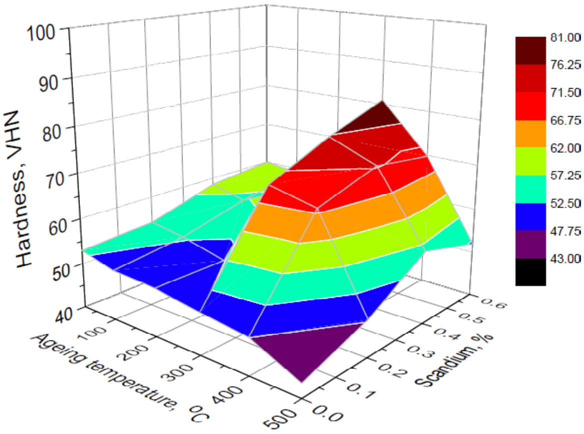

Fig. 2. The influence of scandium content and ageing temperature on the hardness of cast Al-6Mg alloys.

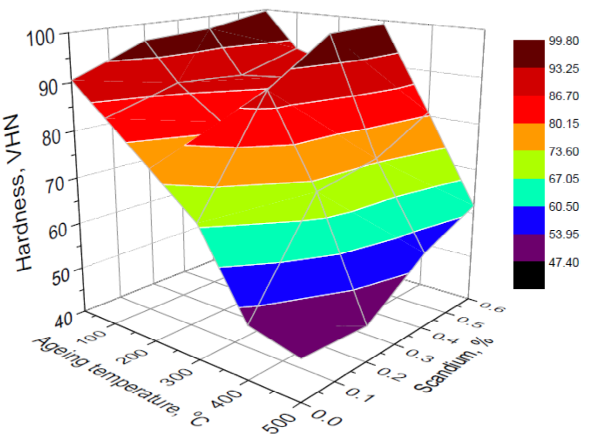

Fig. 3. The influence of scandium content and ageing temperature on the hardness of 75% cold rolled Al-6Mg alloys.

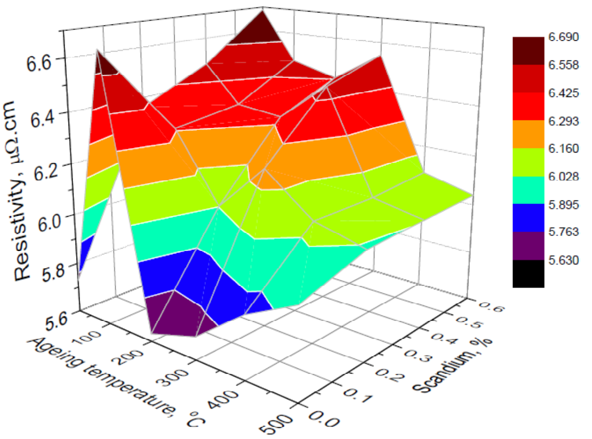

Fig. 4. The influence of scandium content and ageing temperature on the resistivity of 75% cold rolled Al-6Mg alloys.

tures, with a steeper hardness drop beyond 400 °C. In the cast condition the b-phase being already present in the microstructure of the matrix of Al-Mg alloy, precipitation hardening due to the formation of aluminides of magnesium is not envisaged. Moreover Al-Mg alloys are known to be incapable of producing significant age hardening even though the binary phase diagram contains a sloping solvus [1]. The extent of age hardening of scandium bearing alloys increases marginally with increasing scandium content in the alloys. The results of the present experiments clearly indicate that the age hardening effect shown by the alloys are purely due to addition of scandium. Scandium when added, the alloy forms a supersaturated solid solution upon solidification [34,35]. From the isochronal ageing curves it is seen that Al_3Sc precipitates form most rapidly at around 300 °C, where the peak-ageing hardness values are obtained. The softening of the alloys at higher temperature may be due to particle coarsening effect. The initial softening shown in the isochronal ageing curve is thought to be due to internal stress relieving of the rapidly solidified castings.

Kaiser et al. [36] has also made the similar age-hardening study related to cold worked by 75% of the above discussed alloys. The three dimensional graph is presented in Fig. 3 regarding the influence of scandium and ageing temperature on the hardness of cold rolled alloys. They reported that the scandium free alloy shows a continuous softening due to stress relieving, recovery and recrystallization of the strained grains [37]. All other alloys demonstrate age hardening response with peak hardness value at 300 °C. The age hardening of the alloys containing scandium is attributable to the formation of Al_3Sc precipitates [38]. The maximum attainable hardness due to ageing the cold worked alloy has not exceeded the hardness values obtained due to cold working alone. This implies that the precipitation of Al_3Sc is not dislocation induced [20]. At the initial stage of ageing the alloys become softer and subsequent to ageing enable the maximum ageing hardness to reach a magnitude which is comparable to the hardness of cold working alloys. When the alloys are aged at higher temperature a sharp decrease in hardness is observed for all the alloys beyond 300 °C. The nucleation of Al_3Sc is facilitated indirectly by the presence of higher dislocation density. Moreover extensive cold working also generates large number of vacancies, which form vacancy-scandium atom complexes of high binding energy. The substitutional atoms, including Sc atoms in aluminum, diffuse by the vacancy mechanism. Thus, their diffusion mobility can only increase with an increase in the concentration of vacancies. Beyond peak hardness, over ageing effect due to coarsening of the precipitates is seen to have taken place. At higher ageing temperature there is ample scope for dislocation annihilation and this softens the material [39].

Electrical resistivity

From the same study [36] the resistivity curves of the cold rolled alloys as shown in Fig. 4 that the resistivity fall at a steeper rate during the initial period of ageing. Following this, resistivity peaks are noted in the ageing curves. The initial drop in resistivity during isochronal ageing of the alloys is indicative of dislocation rearrangement within the cold worked alloys [40]. The escalate in resistivity in the Sc free alloy is accredited to the formation of magnesium rich precipitate viz. GP zones at 100 °C and β/-metastable phase at 200 °C onwards. Fine zones spread out the free electrons incomprehensibly, result the increases in resistivity until particle coarsening becomes so prominent as to diminish the incoherent scattering of electrons. Scandium doped alloys, when aged at 300 °C the precipitation of Al_3Sc be liable for resistivity peaks. The higher volume fractions for upper Sc content also demonstrate the superior value of resis-

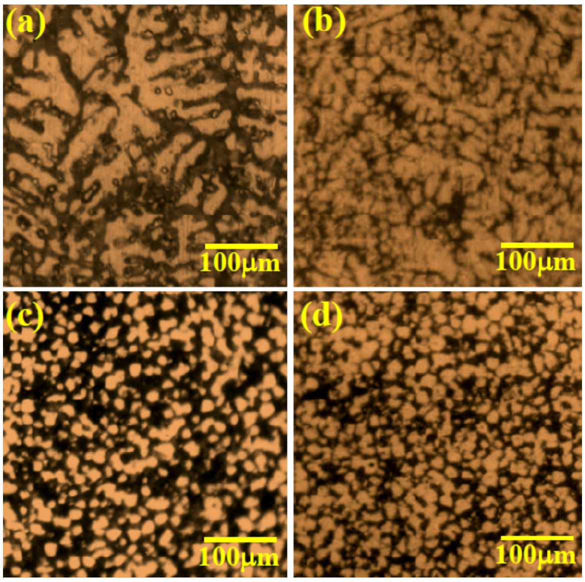

Fig. 5. Optical micrograph of cast Al-6Mg alloys (a) 0.0Sc, (b) 0.2Sc, (c) 0.4Sc, and (d) 0.6Sc.

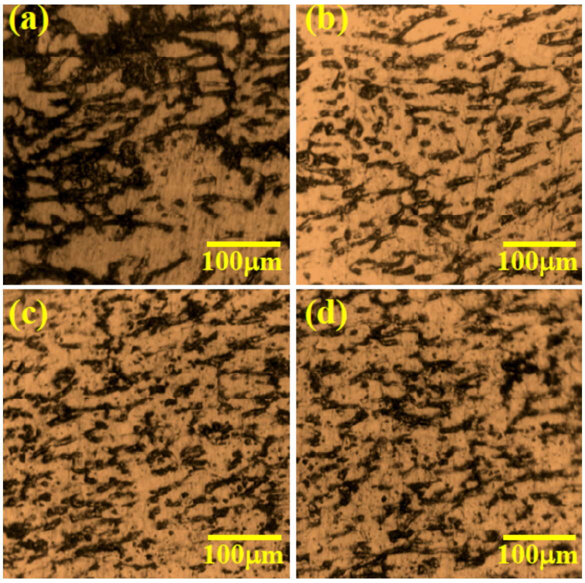

Fig. 6. Optical micrograph of 75% cold rolled Al-6Mg alloys (a) 0.0Sc, (b) 0.2Sc, (c) 0.4Sc, and (d) 0.6Sc.

tivity. The dislocation movement is obstructed by the precipitates so lead to resistivity humps. The final steady decrease in resistivity at higher ageing temperature stems from particle coarsening, recovery and recrystallization which reduces the number of scattering centers [41].

Optical Microscopy

In the previous work [30] the influence of Sc content ranging from 0-0.6% on the microstructure of cast Al-6Mg alloys was analyzed. As presented in Fig. 5, the optical microstructure of Sc free alloy shows dendrites with black second phase particles inside the dendrites spaces. It is well establish that this type of microstructure contain α+β eutectic within the primary dendrites of β aluminium rich solid solution. Where â is composed of intermetallics, primarily Al_8Mg_5 or Al_3Mg_2 along with aluminides of trace iron, chromium, zirconium, manganese, etc. [1,42]. The figure in case of 0.2% Sc added alloy indicates considerably refined dendrites of the alloy. This level of Sc does not able to refine the grains large extend, but capable to consequent shrinking of dendrite arm spacing. It was measured that the arm spacing lie between 20-40 μm against a value of around 45 μm in Sc free alloy [43,44]. Addition of Sc formed the $L1_2$ Al_3Sc primary phases during solidification which also act as heterogenous nucleation sites, consequence increase in solidification speed and the gap between liquids and solidus becomes thin. Higher level of Sc direct to solidification speed as evidenced in the optical microstructures where more the dendrite refinement along with fraction and number density of second phase constituents are observed [45]. More than 0.4Sc addition that is 0.6Sc the grain size remains practically unchanged. From the statement of Al-Sc alloy phase diagram the eutectic temperature varies between 655 and 659 °C and the solid solubility of Sc in Al is approximately 0.35 wt.% [46].

According to Ref. [10] the cold worked structures as presented in Fig. 6 are comprised of elongated grains along with the rolling direction. The overall appearance is columnar grains with second phase particles remaining aligned along the grain boundaries. In case of 0.2Sc alloy the cold worked structure is typified by elongated grains. However crystallite size is found to be less than that in the cold worked Sc free base alloy. Fragmented dendrites, elongated along the direction of rolling, are observed in the microstructure of 0.4Sc and 0.6Sc cold worked alloys.

The microstructure of 75% cold rolling scandium free alloy ageing at 300°C for one hour shows partially recrystallized grains with sufficient second phase particles at the grain boundaries (Fig. 7). In case of 0.2Sc added alloy there is no evidence of recrystallization after ageing at 300 °C for one hour. Recrystallization could not be effected also in 0.4 and 0.6Sc added alloys after that ageing condition. It may be noted that acicularity of grains is maximum in 0.2Sc added alloy, which contains large elongated grains. On the contrary higher Sc bearing alloys display elongated grains of smaller size. Higher volume fraction of precipitates dominates the recrystallization start temperature. However, at 300 °C, the finely dispersed Al_3Sc in 0.2Sc alloy is sufficient to restrain recrystallisation fully by hindering the movement of sub-boundaries and grain boundaries [47]. On increasing the temperature to 400 °C, the second phase constituent is almost dissolved in Sc free alloy and there is nothing to hinder dislocation movement. As a result

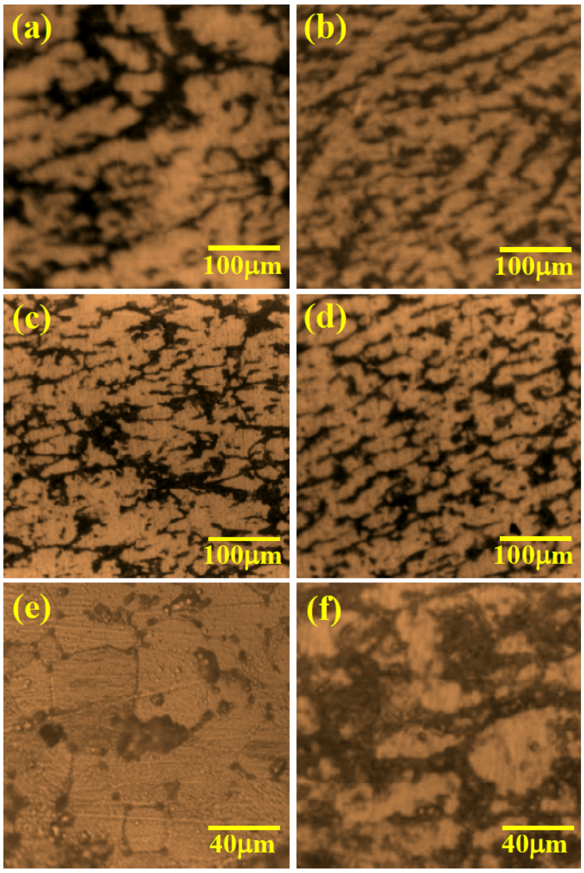

Fig. 7. Optical micrograph of 75% cold rolled Al-6Mg alloys aged at 300 °C for one hour (a) 0.0Sc , (b) 0.2Sc , (c) 0.4Sc, (d) 0.6Sc and aged at 400 °C for one hour (e) 0.0Sc , (f) 0.6Sc.

recrystallisation becomes complete. In 0.6 wt.% Sc containing alloy the supersaturated solution decomposes to form Al_3Sc at around 300°C. These precipitates are known to be resistant to coarsening. There are reports saying that increasing the annealing temperature of Al-Mg-Sc alloy from 300 °C to 400 °C increases the size of Al_3Sc precipitates from 4 nm to 13 nm. The precipitates of Al_3Sc remain coherent with the matrix at higher temperature even when their size increases to 100 nm [48]. In this case however the precipitate size is around 15 nm when aged at 400 °C. Therefore dislocation pinning force is very large. As a result, recrystallization is not possible.

Scanning Electron Microscopy

Scanning electron microstructures of the cast alloys substantiate the observations made in the study of optical microstructures. The Al-6Mg base alloy shows coarse dendrites with high quantity of second phase constituents. The second phase is found to be contained in the inter-dendritic space [1]. In scandium added alloy by 0.6%, not only there happens refinement of dendrites but also the second phase constituent has been reduced in amount. When the same Al-6Mg alloy

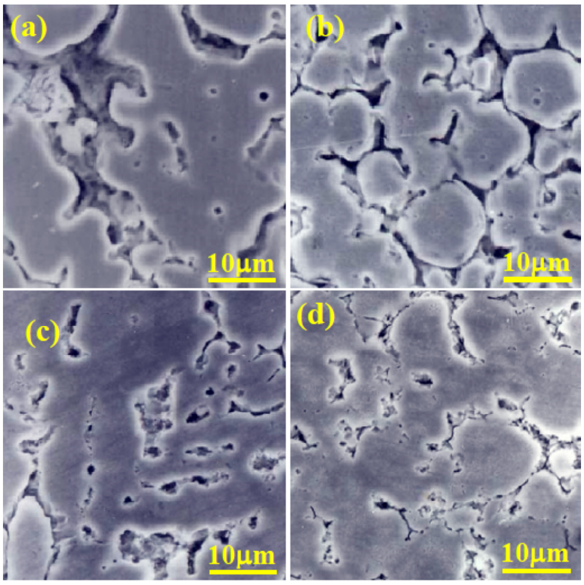

Fig. 8. SEM micrograph of cast Al-6Mg alloy (a) 0.0Sc (b) 0.6Sc and aged at 300 °C for one hour (c) 0.0Sc (d) 0.6Sc.

is aged at 300 °C for one hour the amount of second phase decreases in quantity seeing that partially recrystallized. However, dendritic pattern is not removed fully from the microstructure. The second phase constituents are seen to be segregated mostly at the grain boundaries. In case of 0.6Sc added alloy, the grain refinement has taken place with minimal amount of second phase at the grain boundaries as well as retains its fine grains as shown in Fig. 8. It was due to grain refining effect as well as inhibited recrystallization properties of Sc as discussed elaborately earlier in the Ref. [20].

Tensile properties

The results of tensile tests of different scandium added Al-6Mg alloys under various processing conditions are graphically represented in Figs. 9-12 [49]. When the cold rolled alloys are aged for one hour, it is found that the tensile strength of the base alloy decrease with increasing ageing temperature (Fig. 9). After the onset of ageing of Sc added alloys the tensile strength increases slowly and then reaches a peak at 300 °C. The enhancement of the strength by about 45-50 MPa takes place due to precipitation strengthening through the formation of appreciable amount precipitates of Al_3Sc. When aged at high temperature beyond 300 °C, all the alloys attained the sharp decrease in strength. This is due to precipitation coarsening as well as recovery and recrystallization mechanisms occur in deformed microstructure. The rate of strength decreeing of Sc added alloys are always low. The dispersedly distributed Al_3Sc particles decreases softening rate as well inhibit recrystallization during the ageing. The coherent bound-

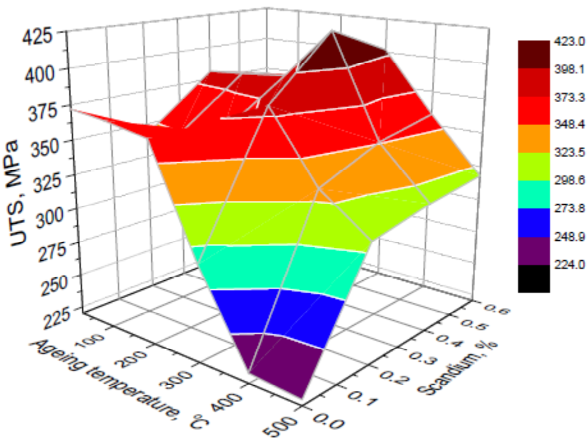

Fig. 9. The influence of scandium content and ageing temperature on the ultimate tensile strength of the alloys.

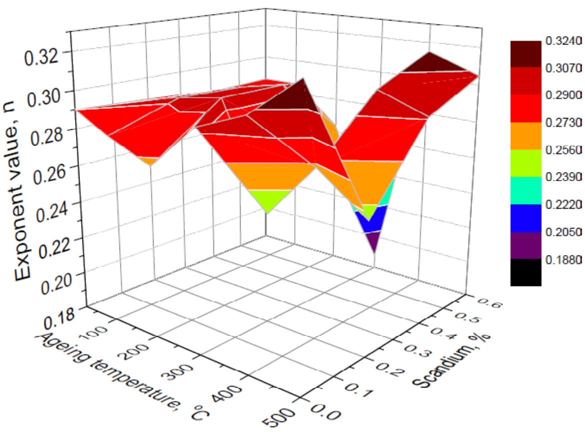

Fig. 11. The strain hardening exponent of the alloys as a function of scandium content and ageing temperature.

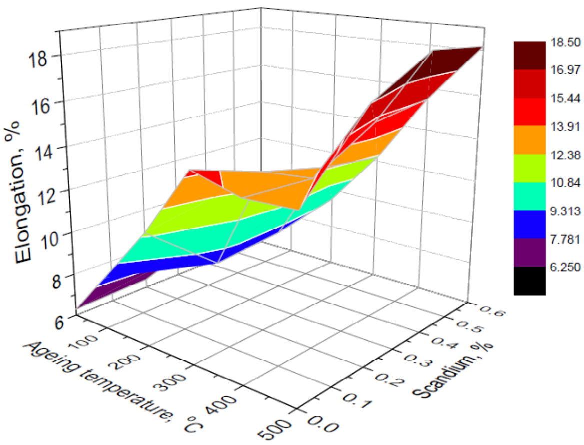

Fig. 10. The influence of scandium content and ageing temperature on the elongation of the alloys.

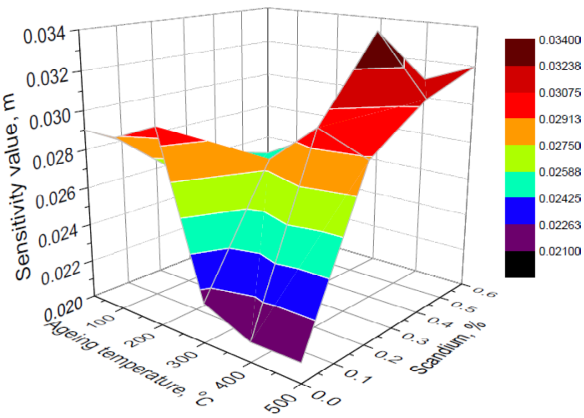

Fig. 12. The strain rate sensitivity of the alloys as a function of scandium content and ageing temperature.

ary energy between the particles and matrix is low since particles are dispersed in the matrix. Thus, these coherent boundaries are stable. As a result, Al_3Sc particles can impede dislocation slip and effectively pin the subboundary and grain boundary. It is difficult for grains to grow even when ageing at high temperatures. From the results of these experiments, it appears that the optimum scandium addition can be limited to 0.4 wt.% to get the maximum benefits in strength. More than 0.4 wt.% Sc exhibits lower tensile properties due to larger particle size and inter-spacing, which result from the higher supersaturation degree. The higher supersaturation degree leads to larger nucleation rate and accelerates the coarsening of Al_3Sc particles significantly, which a phenomenon called Ostwald is ripening. It is generally accepted that dislocations can only bypass hard Al_3Sc particles during deformation, resulting in an increase in the strength due to the Orowan strengthening [50,51].

According to Ref. [49], variation of ductility of the cold rolled alloys as a function of scandium content at various ageing temperature are three-dimensionally plotted in Fig. 10. With increasing ageing temperature, the elongation of the alloys showed a rising trend. The character of this elongation with ageing temperature can be separated into three phases: the first phase of ageing up to 200 °C, the elongation increased sharply, the second phase as the ageing temperature at 200 °C-400 °C, the change of elongation of the alloy tends to be slow and the final stage of ageing beyond at 400 °C, the change of elongation is high. But the Sc free alloy shows the different with a trend of gradually increased with ageing temperature. Cold rolling of the alloys results a large number of deformation energy storage, during ageing recovery and recrystallization can be occurred which increase the elongation through offset the deformation. Grain refinement is known to have a beneficial effect on ductility. Therefore the trend of the curves is a mere reflection of grain refining ability of scandium in Al-6Mg alloy. The occurrence of ductility minima at the peak aged condition is easily understandable since the inhomogeneous deformation due to cutting mechanism being operative during tensile loading would always lead to a lowering of toughness. In fact the fine precipitates of Al_3Sc act as the early nucleation sites for microvoids. Therefore, fracture resistance of the material decreases.

An elongation value of about 18% is obtained in more than 0.4Sc added alloys after ageing at 500 °C with a benefit in elongation by about 6% from the peak ageing temperature at 300 °C. From the results it is perceived that no extra benefit obtained beyond 0.4 wt.% scandium addition as the grain refinement of the experimental alloy system has maximized at 0.4 wt.% Sc. It may be further explained that with 0.6 wt.% Sc shows the least ductility for their higher volume fraction of precipitates into the alloy and hence lower ductility [51,52].

The *'n'* value that is strain hardening exponent of the materials corresponds to homogeneous deformation ability, where the lower value indicates that the deformed material has better plasticity. Three dimensional graph in Fig. 11 clearly indicates that there is no remarkable variation of n value up to ageing at 200 °C for all the level of scandium addition alloys [49]. It is because no major precipitations are generated at this temperature. The '*n*' value for scandium added alloys drastically changed at the temperature around 300 °C where the Al_3Sc precipitate as well as precipitation hardening becomes maximum as already confirmed by the ageing curves [43,53]. Thus further hardening to a large extent due to straining is improbable. When the alloys are aged, partial recovery and precipitate coarsening take place. These create extra provision for dislocation-dislocation interaction during straining. This is reflected by an increase in the '*n*' values of the scandium alloys aged beyond at 300 °C.

Strain rate sensitivity '*m*' value depends on the applied stress and corresponding strain rate which is an indicator of superplastic potential of a material [54]. The higher value is a sign of the material more stable against the local strain rate. Kaiser et al. [49] analyzed the relationship between strain rate sensitivity and the ageing temperature for different Sc added Al-6Mg alloys. Experimental studies confirmed that the strain rate sensitivity of the alloys depend on both the ageing temperature and Sc contents. It is apparent from the figure that Sc added alloys, the strain rate sensitivity increases with the ageing temperature. Without Sc, the '*m*' value of the alloy is seen to decrease with increase in ageing temperature. Scandium refines the grain size of Al-6Mg alloy and also leads to the formation of coherent precipitates of Al_3Sc. As a result, scandium doped Al-6Mg alloy is microstructurally conducive to superplasticity at elevated temperature. In fact superplastic behaviour is already reported in similar alloys where 'm' values were found range from 0.33-0.50 within a temperature span of 350 °C-475 °C [51,55]. In the present case, the room temperature values of '*m*' are comparatively low and hint upon its inherent superplastic behaviour at elevated temperatures. At room temperature although grain size is fine enough, so far as the deformation behaviour is concerned grains are primarily the subgrains. However the sub-boundaries are generally immobile with respect to grain boundary sliding. So deformation is primarily controlled by dislocation slide through the lattice. The increasing 'm' values of the cold rolled alloys with ageing temperatures are consistent with the corresponding enhancement of tensile ductility as evidence in Fig. 12. This is presumed to be due to the advent of power law break down as enumerated earlier [56].

Fracture behavior

Kaiser et al. [49] evaluated the influence of Sc on the fracture behavior of cold rolled Al-6Mg alloys at the peak aged condition that is aged at 300 °C for 1 hour. After tensile tested at strain rate of $10^{-3}s^{-1}$, the obtained fracture surface shows microvoid coalescence (Fig. 13). The fracture is seen to be transgranular shear type. A mixed mode of fracture is present. In some cases delamination tearing is also visible. Formation of coarse dendrites has been observed at the failure zone of the scandium free base alloy and the failure was of brittle nature [57]. The fractograph of Sc added alloys consist of higher quantity of dimples, which is indication that the tensile specimen failed in relatively a ductile mode under the accomplishment of tensile loading. The number of fine dimples causes the grains refinement by addition of Sc. The alloys also show the elongated manner of dimples due to cold deformation of the alloys. Absent

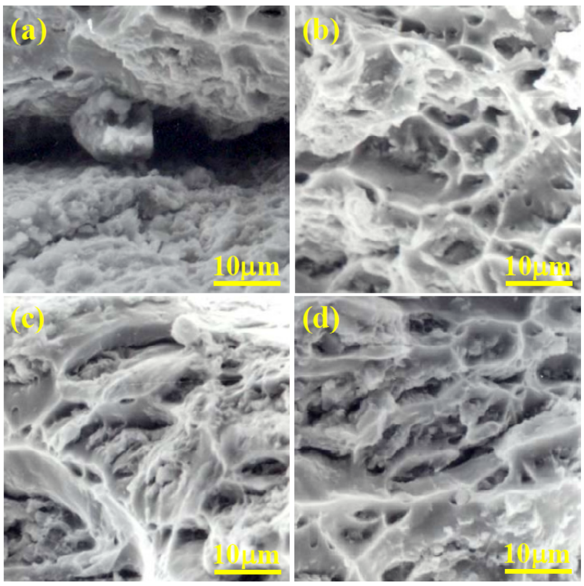

Fig. 13. SEM fractograph of 75% cold deformed Al-6Mg alloys aged at 300 °C for 1 hour (a) 0.0Sc, (b) 0.2Sc, (c) 0.4Sc, and (d) 0.6Sc after tensile tested at strain rate of $10^{-3}s^{-1}$.

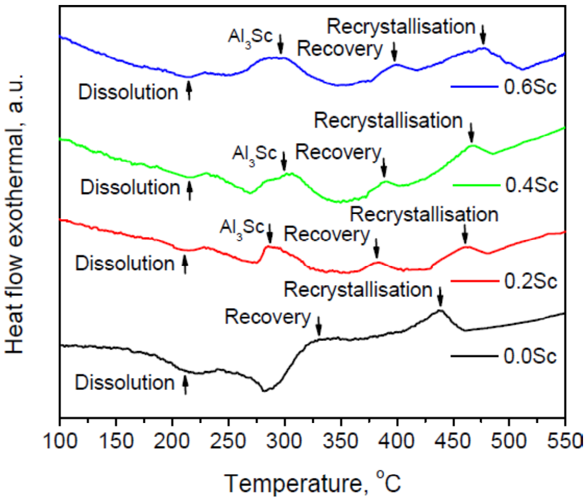

Fig. 14. DSC heating curve of Al-6Mg alloys with different scandium contents.

of such elongated grains of the base alloy due to partially recrystallized at ageing 300 °C for one hour.

DSC study

Kaiser used the differential scanning calorimetry study to measure enthalpy changes due to changes in the physical and chemical properties of different scandium added Al-6Mg alloys as a function of temperature [58]. The experimental alloys when cold rolled by 50% has given rise to the DSC heating curve as shown in Fig. 14. Al-6Mg alloy always have some metastable β' phase as stated earlier which gives way to the formation of b-phase [1]. Two separate DSC peaks noted for the alloy is suggestive of the probable dissolution of β' phase for subsequent formation of β-phase. Therefore the endothermic noted at 215 °C in the base alloy seems to correspond to the dissolution of β'-phase. The activation energy of the process, 133 KJ/mol, is close to the activation energy for diffusion of magnesium in aluminium as earlier study has been reported to be 130 KJ/mol [59]. Thus the dissolution kinetics is dependent upon the diffusion of magnesium in the alloy. The broad exothermic at 325 °C gives activation energy 118 KJ/mol. This value is quite close to the activation energy of 120 KJ/mol for recovery in Al-Mg alloy as reported elsewhere [60]. So the peak is thought to be concerned with the recovery process taking place in the deformed sample. The following exothermic is relatively sharp and occurs at 435 °C. Recrystallisation temperature is found similar to what has been reported earlier [60]. However its activation energy is found to be somewhat higher (201 KJ/mol) than the previously reported value of 190 KJ/mol. This may be due to hindrance to dislocation movement by certain iron-aluminides which might have been present due to relatively higher iron content of the alloy (~0.38 wt.%)

The DSC heating curve of different Sc added alloys endothermic noted at 215 °C seems to correspond to the dissolution of b'-phase with the activation energy of 134, 136, and 137 KJ/mol for 0.2, 0.4, and 0.6Sc alloys, respectively. The peak temperature for dissolution is more or less same. Because there was no major precipitation of Al_3Sc taken place within these temperatures which hinder this behavior. Sc added alloys showed exothermic at 290 °C correspond to the precipitation of Al_3Sc taking place during the DSC heating run. The calculated activation energy for this precipitation reactions are 180 KJ/mol, 177 KJ/mol, and 176 KJ/mol, respectively. It may be mentioned that activation energy for diffusion scandium in aluminium is reported to be 173 KJ/mol [61]. Following this a broad exothermic peak is seen to appear at 380 °C, 385 °C, and 390 °C, respectively, in the heating curve. This signifies the recovery process taking place at a higher temperature. The activation energy for these recovery processes is found to be 117 KJ/mol, 119 KJ/mol,and 123 KJ/mol in that order. All the 0.2, 0.4, and 0.6Sc added alloys appeared a further exothermic peak at 453 °C, 455 °C, and 457 °C associated with activation energy of 207 KJ/mol, 210 KJ/mol, and 211 KJ/mol, respectively. This exothermic peak is representative of recrystallization. Recrystallization of the base alloy takes place at 435 °C, while the Sc added alloys attained the higher temperature viz. 455 °C depends on higher level of Sc. Also, it is noticed that in this particular case the activation energy for recrystallization is also rather high. Thus kinetics of recovery and recrystallization is greatly delayed in Sc added alloys. This is because Al_3Sc precipitates have high coherency strains which strictly impede the migration of dislocations.

XRD study

Kaiser [62] evaluated the influences of scandium on the development of texture in cast Al-6Mg alloy. The variations of relative intensities of different reflecting planes with scandium content of the alloys are shown in Fig. 15. The investigational alloys show that (111) planes give rise to highest intensity at all scandium levels. In case of (200) reflection, the increase in scandium content beyond 0.2 wt.% has led to appreciable increase intensity. (220) plane records decrease in intensity for all level of scandium addition. While the (311) reflecting planes show a significant rise in its intensity during the same composition region. The intensities of (200) plane does not influence by scandium up to 0.2 wt.% to any significant extent. It is evocative of the information that at this level of scandium content is not able to control the orientation of nucleation at vast amount. The atomic scattering factor of scandium is higher than aluminum and magnesium and Sc atoms are randomly distributed

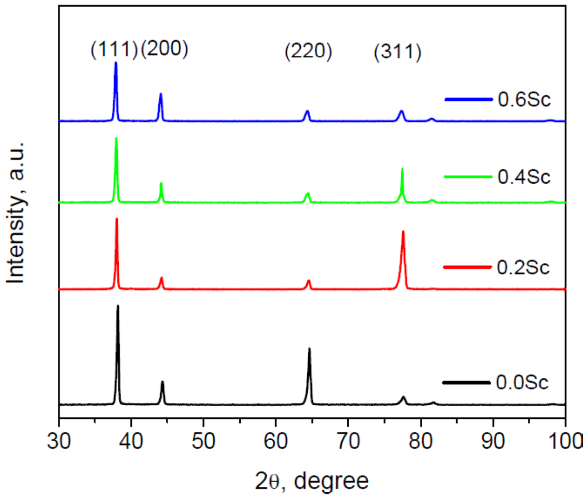

Fig. 15. XRD pattern for Al-6Mg alloys with different scandium contents.

in the alloy matrix. There are some positive changes in structure factor due to the average atomic scattering factor which shall have to be computed for statistically averaged atoms. The multiplicity factor of (220) is higher than (200) and (111) set of planes, whereas it is least in (200). Hence the relative enhancement of integrated intensity due to structure factor effect will be greatest in (220) planes. The relatively high magnitude of increase in the intensity of (311) reflection is ascribed to its high multiplicity factor thereby making it more responsive to the increase in structure factor in scandium treated alloy [63,64].

4. SUMMARY AND CONCLUSIONS

From this literature review related to lightweight Al-6Mg alloy the following important points may be peaked out.

The review clearly indicates that the age hardening effects are exposed by Al-6Mg alloys simply due to addition of scandium. It is seen that Al_3Sc precipitates form most speedily at around 300 °C, where the peak-ageing hardness values are obtained. The extent of age hardening increases with increasing scandium content into the alloys up to 0.4 wt.% Sc but marginally increases beyond this level.

Small concentrations of Sc refine the dendrites of the cast Al-6Mg alloys but no extra benefit obtained beyond 0.4 wt.% Sc addition as the grain refinement of the experimental alloy system.

From the reviewed results, it becomes visible that the optimum scandium addition can be limited to 0.4 wt.% to get the maximum benefits in tensile strength. Additional Sc performs the inferior tensile properties due to higher degree of supersaturation. The strain hardening exponent ‘*n*’ value for scandium added alloys drastically changed at the ageing temperature around 300 °C where the precipitate becomes maximum. On the other hand ‘*m*’ value of Sc added alloys show general trend of increasing with ageing temperature since the higher value is a sign of the material more stable against strain rate.

The fracture mechanism of the cold worked alloys is a mixed mode of fracture. The fractograph consist of higher quantity of dimples in case of Sc added alloys as failed in relatively a ductile mode but Sc free alloy failed in brittle nature with lower quantity of dimples created for its coarse dendrites structure.

DSC heating run gives an idea about the precipitation of Al_3Sc into the Sc added alloys which taking place around at 290 °C. The precipitates delay the recrystallization process of the Al-6Mg alloys and higher volume fraction keep up the recrystallization temperature.

The addition of scandium in Al-6Mg alloy influences development of texture. The intensities of (200) planes influence by scandium addition around to 0.4 wt.%. Less than this level the alloys are not able to control the orientation of nucleation at vast amount. (220) plane records decrease in intensity for all level of scandium addition while the (311) reflecting planes show a significant rise in its intensity during the lower composition region.

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