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Effect of Scandium Additions on the Tensile Properties of Cast Al-6Mg alloys

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Effect of aging on the mechanical properties of Al-6Mg alloy doped with varying concentration of scandium ranging from 0.2 to 0.6 wt.% is analyzed. As-cast samples were aged isochronally for 60 min at different temperatures ranging from 100 to 500 °C. Evaluation of mechanical properties of the aged Al-6Mg (Sc) alloys was done by employing an Instron testing machine. Various strain rate of testing were used to find out the values of strain-rate sensitivity of the experimental alloys. The influence of scandium is much pronounced on yield strength than on the tensile strength. Alloys with higher scandium content have shown higher yield strength and the values of strain-rate sensitivity '*m*' at peak-aged condition have been found to be comparatively high at higher scandium concentration. The fracture of the experimental alloys occurs through microvoid coalescence.

Keywords age hardening, Al-Mg alloys, grain-refinement, microvoid, precipitates, strain-hardening exponent, strain-rate sensitivity

1. Introduction

Extensive research has been conducted on Al-Mg alloys containing scandium (Ref 1-6). The addition of scandium in Al-Mg alloy system is intended to take advantage of the unique precipitation hardening characteristics of scandium in aluminum. Scandium forms a LI_2 phase, Al_3Sc, which precipitates coherently. Magnesium aluminide does not enter the precipitate structure and hence the strengthening effect of Al_3Sc is additive to the solid solution strengthening due to magnesium. Reportedly, Al_3Sc is the most potent strengthener on equal atomic fraction basis. These precipitates are also effective on stabilizing substructure, thus allowing the use of strain hardening and stabilization treatments to improve the strength properties quite considerably. Though previous investigators have examined the microstructure and properties of Al-Mg-Sc alloys, a detail study on the mechanical properties of Al-Mg-Sc alloy systems, on consideration of the compositional and process variables, seems to be needed. It is the purpose of the current investigation to advance this work forward and systematically explore the Al-Mg-Sc alloy system. This work is intended to determine if alloys of this type could be made competitive with other low-density systems for high-performance applications. The ability of Al_3Sc precipitates to stabilize substructure envisages the use of strain hardening for enhancement of mechanical properties of the alloy. Since strain-hardening behavior of alloys is sensitive to strain rate of testing, the study on the tensile properties of the alloys under various thermal and mechanical treatments has been carried out for various strain rates of testing. Due to fine distribution of Al_3Sc precipitates, superplastic effect in Al-Mg-Sc alloy has been studied by previous workers (Ref 7, 8). The microstructure and mechanical properties of a cold-rolled Al-Mg-Sc alloy have been characterized. The alloy exhibited superplasticity at relatively high strain rates ($\sim 10^{-2}$ s^{-1}). At a strain rate of 10^{-2} s^{-1} there exists a wide temperature range (475–520 °C) within which the tensile elongation is over 1000%. There also exists a wide strain rate range (10^{-3} to 10^{-1} s^{-1}) within which the tensile elongation is over 500%. The presence of scandium in the alloy results in a uniform distribution of fine coherent Al_3Sc precipitates which effectively pin grain and subgrain boundaries during static and continuous recrystallization. As a result, the alloy retains its fine grain size (~7 μm), even after extensive superplastic deformation (>1000%) (Ref 8). Superplasticity in metals and alloys are characterized by strain-rate sensitivity values.

The present authors have already reported microstructural variations of Al-6Mg alloy due to addition of scandium (Ref 9), and also its effect on the fracture toughness of the alloy (Ref 10). The present work focuses on the effect of age hardening of Sc-added Al-6Mg alloy on the tensile properties.

2. Experimental

Melting of the alloys was carried out in a resistance heating pot furnace under the suitable flux cover (degasser, borax, etc.). Commercially pure aluminum (99.5% purity) and aluminum-scandium master alloy (2 wt.% Sc) were melted in a clay-graphite crucible. Magnesium ribbon (99.7% purity) was then added into the melt. The final temperature of the melt was maintained at 780 ± 15 °C. Variation of the scandium percentage was brought about by varying the quantity of addition of master alloy. Casting was done in 12.5 mm × 50 mm × 200 mm

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Table 1 Chemical composition of the experimental alloys (wt.%)

Alloy	Mg	Sc	Zr	Ti	Cu	Fe	Mn	Ni	Si	Zn	Cr	Sn	Al
1	6.10	0.000	0.000	0.001	0.081	0.382	0.155	0.003	0.380	0.136	0.002	0.002	Bal
2	5.90	0.200	0.001	0.002	0.081	0.345	0.132	0.003	0.360	0.174	0.002	0.002	Bal
3	5.97	0.400	0.000	0.002	0.071	0.314	0.107	0.002	0.335	0.124	0.002	0.002	Bal
4	6.02	0.600	0.001	0.003	0.061	0.293	0.086	0.002	0.320	0.126	0.003	0.002	Bal

cast iron molds preheated to 200 °C. The chemical compositions of the alloys were determined both chemically and spectroscopically (Table 1).

The alloys were aged at different temperatures between 100 and 500 °C, with an increment of 100 °C for 1 h. Hardness of different alloys aged at different temperatures was measured in Vickers hardness testing machine at 5 kg load to assess the age-hardening effect of the alloys. For each samples, average of nine observations with standard deviation around 5 MPa is reported. Tensile testing was carried out in an Instron testing machine of model no. 4204, using different cross head speed to maintain the strain rate of 10^{-2}/s, 10^{-3}/s, and 10^{-4}/s. The averages of five consistent test results were accepted as the tensile test values for the corresponding samples. Fractographic observations of the surfaces fractured by tensile testing were carried out in a Jeol Scanning Electron Microscope.

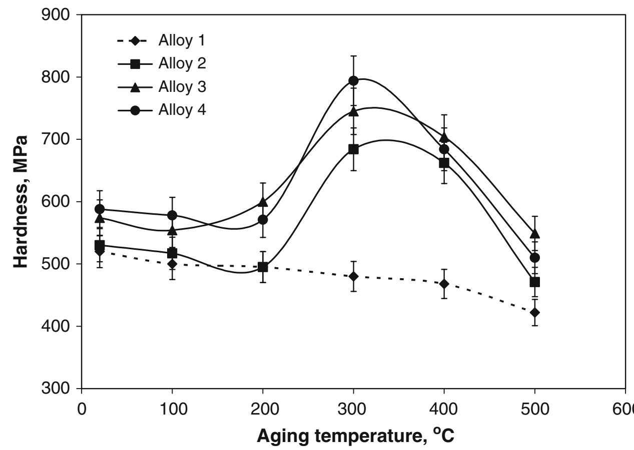

Fig. 1 Isochronal aging curve of the cast alloys. Aged for 1 h

3. Results

The results of isochronal aging of all the four alloys are shown in Fig. 1. It is seen that all the alloys except the binary alloy (alloy 1) have shown appreciable aging response. Alloy 1 has, however, shown a continuous softening at increasing aging temperatures, with a steeper hardness drop beyond 400 °C. The extent of age hardening of scandium-bearing alloys seems to increase marginally with increasing scandium content in the alloys (~15% for 0.2 wt.% increment at peak aged condition). The peak hardness of the 0.6 wt.% Sc alloy is appreciably higher (~30%) than that of 0.2 wt.% Sc alloy. The peak hardness has occurred at 300 °C for all the alloys under investigation. Beyond peak hardness values, the softening is due to usual overaging.

The results of tensile tests of the alloys under various processing conditions are tabulated in Table 2. The variation of ultimate tensile strength, yield strength, and percentage elongation under various aging conditions of the alloys 1-4 are shown in Fig. 2-4. The test values obtained at a strain rate of testing 10^{-3} s^{-1} are used to plot the graphs in Fig. 2-4. It is seen from Fig. 2 that the scandium-doped alloys experience extra strengthening due to age-hardening effect and the maximum in tensile strength value is achieved at an aging temperature of 300 °C. It is further observed that increasing scandium content up to 0.4 wt.% Sc increases the tensile strengths of the alloys. Increase in scandium content beyond this has marginal drop in the tensile strength value under all aging conditions.

From the nature of variation in yield strength with aging temperature of the cast alloys (Fig. 3), it appears that the yield strength of the scandium-added alloys increases to peak value at an aging temperature of 300 °C. Beyond 300 °C further aging has lowered the yield strength of the alloys. The base alloy (Al-Mg alloy) does not show any variation in yield strength with aging temperature. It is to be noted that yield strength of the alloys before aging does not depend appreciably upon the scandium content of the alloys; however, when the alloys are aged at 300 °C or 350 °C the yield strength increases with increasing scandium up to a level of 0.4 wt.% Sc. Beyond this level of scandium addition a marginal fall in yield strength is observed. The sensitivity of yield strength toward scandium concentration is found to be most pronounced when the alloy is aged at 300 °C, the temperature at which the peak hardness occurs during isochronal aging. In alloys aged at 300 °C, increasing scandium to 0.4 wt.% leads to an increase in yield strength by about 175 MPa over the binary Al-6 wt.%Mg alloy. Figure 4 demonstrates the variation of elongation percent with aging temperatures of alloys 1-4. It is observed that at the aging temperature for which age hardening is maximum (300 °C), the ductility values of the alloys pass through minima. The ductility value of the aged alloy 4 is found to be less than all the other alloys for all aging conditions. Alloys 2 and 3 record their maximum ductility after aging at 200 °C. When the ductility variation against scandium content is plotted for all aging conditions of the alloys, it is observed that ductility maxima occur at 0.4 wt.% Sc. However, maximum ductility is obtained when aging of 0.4 wt.% Sc alloy is carried out at 200 °C.

From Table 2 it appears that increasing strain rate has increased the tensile and yield strength of the experimental alloys. It is further observed that the change in strength properties is found to be less (~20 MPa) in case of alloys 2-4 aged at 300 °C. It is seen that the strain-hardening exponent '*n*' of the experimental alloys range from 0.342 to 0.597 after aging at 300 °C. For alloys with minor additions (alloys 2-4), the strain-hardening exponent values lie within 0.342-0.460. The variation of strain-hardening exponent, with the strain rate of testing for the experimental alloys under various thermal and mechanical treatments, is shown in Fig. 5. It shows that strain-hardening

Table 2 Tensile properties of the cast alloys

Alloy no.	History	Strain rate, s^{-1}	Tensile properties: UTS, MPa	YS, MPa	Elongation, %	*n* value	*m* value
1	As cast	10^{-3}	197	137	8.5		
	As cast/aged at 200 °C	10^{-3}	205	125	11.5		
	As cast/aged at 300 °C	10^{-2}	180	130	8.8	0.597	0.049
		10^{-3}	145	110	8.8	0.514	
		10^{-4}	127	100	9.2	0.475	
	As cast/aged at 350 °C	10^{-3}	180	100	10		
2	As cast	10^{-3}	200	134	8.2		
	As cast/aged at 200 °C	10^{-3}	220	145	11.5		
	As cast/aged at 300 °C	10^{-2}	265	245	8.1	0.460	0.053
		10^{-3}	250	230	8.3	0.420	
		10^{-4}	248	230	9.5	0.400	
	As cast/aged at 350 °C	10^{-3}	230	175	8.4		
3	As cast	10^{-3}	253	165	11.2		
	As cast/aged at 200 °C	10^{-3}	277	171	14.0		
	As cast/aged at 300 °C	10^{-2}	338	325	8.3	0.350	0.058
		10^{-3}	334	315	8.6	0.344	
		10^{-4}	329	303	10.0	0.342	
	As cast/aged at 350 °C	10^{-3}	336	301	9.00		
4	As cast	10^{-3}	219	156	7.8		
	As cast/aged at 200 °C	10^{-3}	200	144	8.5		
	As cast/aged at 300 °C	10^{-2}	317	285	7.1	0.455	0.069
		10^{-3}	300	273	7.0	0.454	
		10^{-4}	300	267	7.2	0.440	
	As cast/aged at 350 °C	10^{-3}	289	250	7.5		

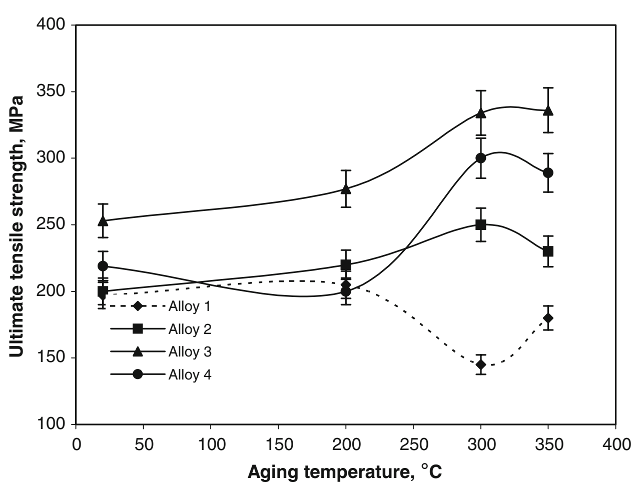

Fig. 2 Variation of ultimate tensile strength (10^{-3} s^{-1}) with aging temperature of cast alloys isochronally aged for 1 h

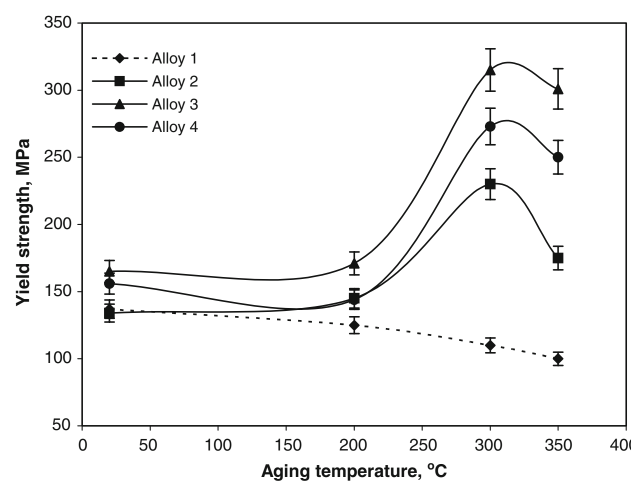

Fig. 3 Variation of yield strength (10^{-3} s^{-1}) with aging temperature of cast alloys isochronally aged for 1 h

exponent of alloys 3 and 4 are almost independent of strain rate of testing within the range employed in the present experiment. For other alloys, a certain increase in '*n*' value with increasing strain rate is noticed. It appears from Table 2 that the strain-rate sensitivity (*m*) of the experimental alloys lies within 0.049-0.069.

Fractography of alloy 3, cast and annealed at 300 °C, shows that the microvoid coalescence is the chief mode of failure. A big crack along the grain boundary is indicative of delamination at high angle boundary. Numerous microvoids are observed in Fig. 6. The characteristic dimples are observed in Fig. 7. Some kinds of particles are observed to be present at the base of the dimples. Fracture might have been initiated by these particles. Testing at higher strain rate does not exhibit much difference in the character of fracture. Though dimples are visible in Fig. 8, a crack along grain boundary (shown in the figure) appears to be the chief cause of failure. Dimples are clearly delineated in Fig. 9. This implies a ductile fracture. Particles are noticed at the dimple base andthat these particles are involved in creating microvoids is also noticed in the same fractograph (Fig. 9).

4. Discussions

The results of the present experiment clearly indicate that the strengthening of the alloys due to aging is purely due to addition of scandium. When added in small concentrations scandium is known to refine the grain structure of cast metal

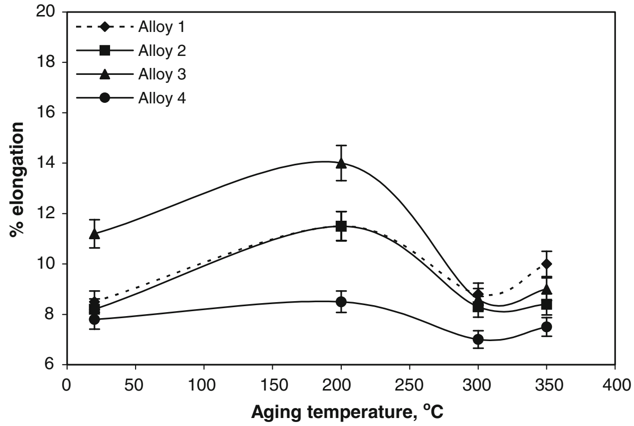

Fig. 4 Variation of percent elongation (10^{-3} s^{-1}) with aging temperature of cast alloys isochronally aged for 1 h

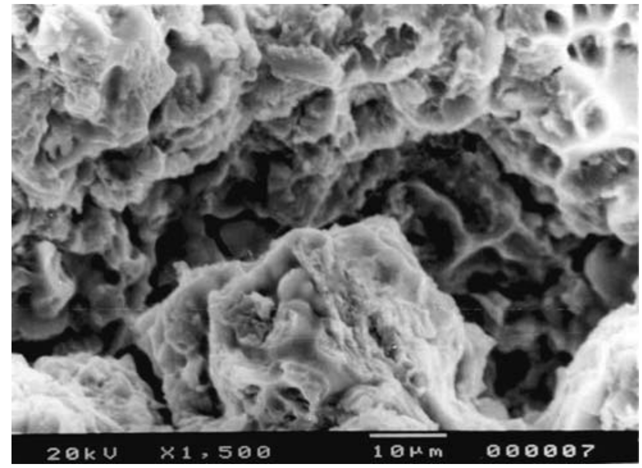

Fig. 7 SEM fractograph of cast alloy 3, aged at 300 °C for 1 h and tensile tested at strain rate of 10^{-3} s^{-1}

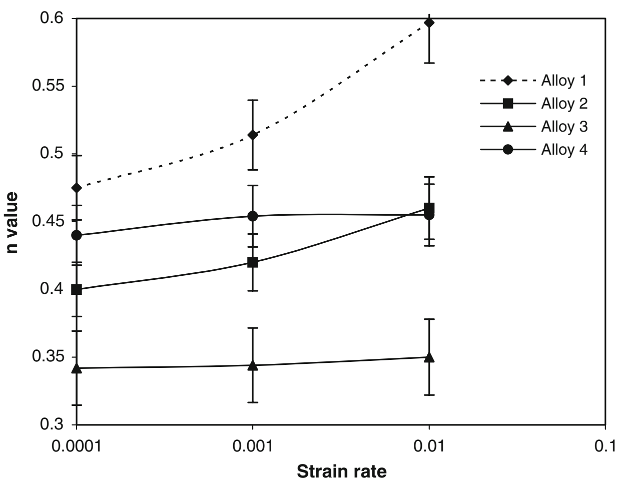

Fig. 5 Variation of strain-hardening exponent with strain rate of testing of cast alloys aged at 300 °C

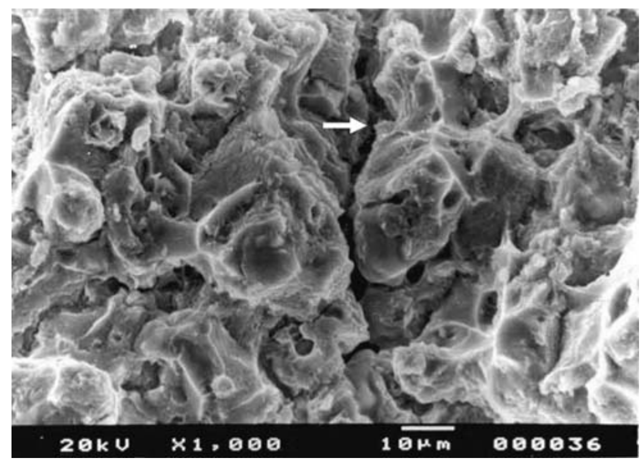

Fig. 8 SEM fractograph of cast alloy 3, aged at 300 °C for 1 h and tensile tested at strain rate of 10^{-2} s^{-1}

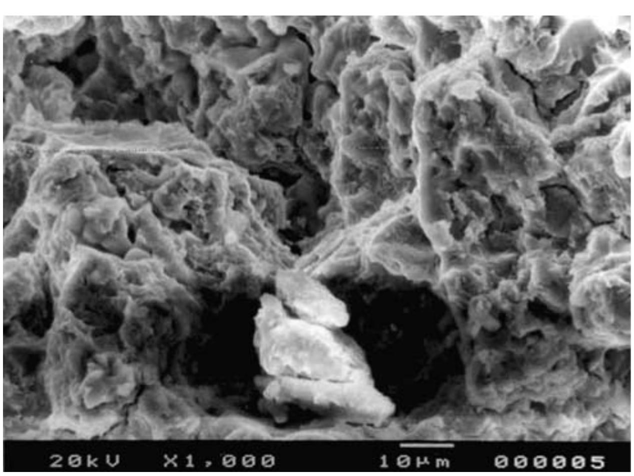

Fig. 6 SEM fractograph of cast alloy 3, aged at 300 °C for 1 h and tensile tested at strain rate of 10^{-3} s^{-1}

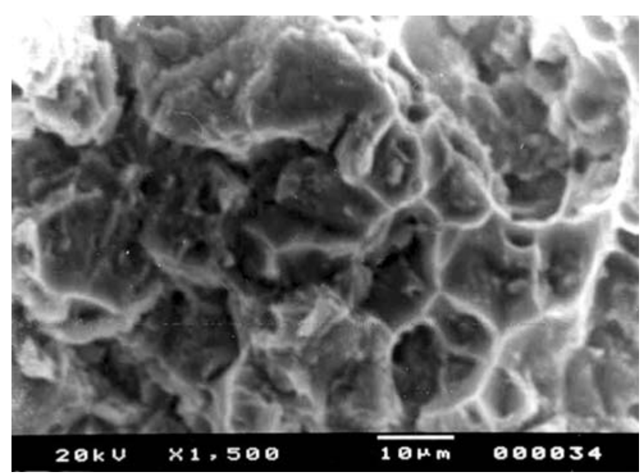

Fig. 9 SEM fractograph of cast alloy 3, aged at 300 °C for 1 h and tensile tested at strain rate of 10^{-2} s^{-1}

and to form a supersaturated solid solution upon solidification (Ref 6, 7). Formation of supersaturated solid solution assures a high precipitation hardening effect upon decomposition of this solid solution with the formation of fine coherent equilibrium Al_3Sc precipitates (Ref 11-13). No aging response is, however, visible for the base alloy. In the cast condition, the β-phase being already present in the microstructure of the matrix of alloy 1, 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 (Ref 14).

From the isochronal aging curves it is seen that Al_3Sc precipitates form most rapidly at around 300 °C, where the peak-aging 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 aging curve (Fig. 1) is thought to be due to internal stress relieving of the rapidly solidified castings. From the results of Fig. 2 it is clear that strengthening due to aging takes place to the maximum extent at the aging temperature of 300 °C. Incidentally, this is the temperature at which maximum aging hardness has occurred. Thus, improvement in strength during aging is due to the formation of Al_3Sc precipitates, which are known to have LI_2 crystal structure. Figure 2 clearly delineates that improvement in yield strength of the alloys due to aging is more than that for ultimate strength. Thus with the progress of aging yield tensile ratio increases. Material toughness is related to this ratio. It is known that yield strength is a structure-sensitive property of the material and hence formation of fine precipitates is more responsive to yield strength. It is quite clear that beneficial effect of scandium in improving the strength properties of the Al-Mg alloys is maximized at 0.4 wt.% (Fig. 2, 3). It, therefore, appears that the optimum scandium addition can be limited to 0.4 wt.% to get the maximum benefits in strength. Figure 3 showing the effect of scandium on the yield strength of the aged alloys gives a clear evidence that the precipitates of Al_3Sc is mainly responsible for variation of strength of the alloys with aging. From the same figure one may observe that significant influence on yield strength is obtainable only for aging temperatures 300 °C or 350 °C where an appreciable amount of Al_3Sc is formed. Till the time, the amount of Al_3Sc precipitates has not been perceptible in the microstructures; no variation in yield strength is practically noticed (200 °C aging). Again, the influence of scandium is much pronounced on yield strength than on the tensile strength due to the already stated reasons that fine coherent precipitates of Al_3Sc are responsive much more to the yield behavior of the alloys. Hence higher volume fraction of the precipitates in alloys with higher scandium content exerts greater influence on the yield strength. Such scandium dependence of yield strength of aluminum alloys has been observed by earlier workers too (Ref 15). Calculations have shown that combining the contributions from coherency and order hardening gives an overall increase in strength due to Al_3Sc of a magnitude close to what could be observed experimentally (Ref 16).

The occurrence of ductility minima at the peak-aged condition is easily understandable since the inhomogeneous deformation due to cutting mechanism is operative during tensile loading and it always leads 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. This is reflected in the form of minimum percent elongation of the alloys aged at 300 °C whence the density of fine precipitates is most. This conjecture further explains why alloy 4 with 0.6 wt.% Sc shows the least ductility. Higher scandium content implies higher volume fraction of precipitates and hence lower ductility. Scandium is a very good grain refiner in Al-Mg alloy. Grain refinement is known to have a beneficial effect on ductility. Therefore the trend of curves in Fig. 4 is a mere reflection of grain-refining ability of scandium in Al-6Mg alloy. Here again one would notice that no extra benefit might be obtained on increasing the addition of scandium beyond 0.4 wt.% as the grain refinement of the experimental alloy system has maximized at 0.4 wt.% Sc.

It is known that tensile strength of alloys increases with increasing strain rate. It is also known that yield stress and flow stress at low plastic strains are considerably dependent on strain rate of testing. In the present experiments it is recorded that both tensile and yield strength have increased with increasing strain rate. Thus the experimental results commensurate with the existing knowledge in this regard. In the present case, the plastic strain is, in general, small and it assumes slightly higher value only after overaging. Therefore the results of the present experiments do not reveal any difference in the responsiveness of yield and tensile strength toward the strain rate of testing. The lower values of strain-hardening exponent of scandium-doped alloys over that of the base alloy is attributed to the formation of Al_3Sc, which increases the yield stress. Thus extra hardening during postyielding plastic deformation is limited by its tensile strength, which is not far greater than the yield strength as substantiated by the high yield/tensile ratio of the alloys.

Velocity of mobile dislocations is proportional to strain rate. Again, velocity of dislocations is strongly dependent on flow stress ($v = A\sigma^{m'}$, with usual notations). It may be deduced that for a specific strain increment an increase in the velocity of mobile dislocations would lead to an increase in '*n*' values as it increases the flow stress. Increasing strain rate leads to an increase in the velocity of dislocations. This is why the '*n*' values of the experimental alloys are seen to increase with increasing strain rate of testing. However, for alloys 3 and 4 the microstructure after aging at 300 °C contains maximum amount of fine precipitates of Al_3Sc. Within the range of variation in strain rate of testing, the velocity of mobile dislocations cannot appreciably increase due to the inhibitions stated above. Therefore '*n*' value remains essentially independent of strain rate of testing (Fig. 5).

It is known that strain-rate sensitivity of metals and alloys are rather low (<0.1) at room temperature although it increases significantly above half the melting point. 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 temperatures. In fact, superplastic behavior is already reported in similar alloys (Ref 8, 17, 18) where '*m*' values were found to range from 0.33 to 0.50 within a temperature span of 350-475 °C. In the present case also the room temperature values of '*m*' are comparatively high.

5. Conclusions

1. The age-hardening effect shown by the alloys are due to addition of scandium.
2. Improvement in strength during aging is due to the formation of Al_3Sc precipitates. The improvement in

yield strength of the alloys due to aging is more than that for ultimate strength.

3. The influence of scandium is much pronounced on yield strength than on the tensile strength, as fine coherent precipitates of Al_3Sc are much more responsive to the yield behavior of the alloys. Scandium improves yield strength quite considerably.
4. '*n*' value remains essentially independent of strain rate of testing.
5. The fracture of the experimental alloys occurs through microvoid coalescence.

References

1. L.S. Toropova, D.G. Eskin, M.L. Kharakterova, and T.V. Dobatkina, *Advanced Aluminum Alloys Containing Scandium, Structure and Properties*. Baikov Institute of Metallurgy, Moscow, Russia, 1998
2. L.I. Kaygorodova and V.P. Domashnikov, Investigation of the Influence of Scandium on the Structure and Properties of an Aluminium-Magnesium Alloy During Natural Ageing, *Phys. Met. Metall*, 1990, **68**(4), p 160–164
3. Yu.M. Vainblat, S.S. Khayurov, and L.B. Ber, Aluminium Alloys Containing Scandium, *Tekhnol Legk Spl.*, 1996, **3**, 18–22
4. N.I. Turkina and V.I. Kuz'mina, Phase Relations in the Al-Mg-Sc System, *Izv. Akad Nauk SSSR*, 1976, Met. No.4, p 208–212
5. L.M. Dougherty, I.M. Robertson, and J.S. Vetrano, Direct Observation of the Behavior of Grain Boundaries During Continuous Dynamic Recrystallization in an Al-4Mg-0.3Sc Alloy, *Acta Mater.*, 2003, **51**(15), p 4367–4378
6. T. Aiura, N. Sugawara, and Y. Miura, The Effect of Scandium on the As-Homogenized Microstructure of 5083 Alloy for Extrusion, *Mater. Sci. Eng.*, 2000, **280**, p 139–145
7. Yu.A. Bazin and B.A. Baum, About Mechanism of Alloy Modifying with Soluble Additions, *Tsvetnye Metally.*, 1994, **35**, p 130–136
8. T.G. Nieh, R. Kaibyshev, L.M. Hsiung, N. Nguyen, and J. Wadsworth, Subgrain Formation and Evolution During the Deformation of an Al-Mg-Sc Alloy at Elevated Temperatures, *Scripta Mater.*, 1997, **36**(9), p 1011–1016
9. M.S. Kaiser, S. Datta, A. Roychowdhury, and M.K. Banerjee, Age Hardening Behaviour of Wrought Al-Mg-Sc Alloy, *Mater. Manuf. Process.*, 2008, **23**(1), p 74–81
10. M.S. Kaiser, S. Datta, A. Roychowdhury, and M.K. Banerjee, Ageing Effect of Ternary and Quaternary Additions on the Fracture Toughness Behaviour of Cast Al-Mg Alloys, *J. Inst. Eng. (I) MM Div.*, 2007, **88**, p 3–9
11. M.E. Drits, L.S. Toropova, Yu.G. Bykov, et al., Metastable Ai-Sc Phase Diagram in the Aluminum-Rich Region, *L-'v. Akad. Nauk SSSR. Metally.* 1983, **1**, p 179–182
12. M.E. Drits, J. Dutkiewicz, L.S. Toropova, and J. Salawa, The Effect of Solution Treatment on the Ageing Processes of Al-Sc Alloys, *Cryst. Res. Technol.*, 1984, **19**, p 1325–1331
13. M.E. Drits, L.B. Ber, Yu.G. Bykov, L.S. Toropova, and G.K. Anastaseva, Ageing of Alloy Al-0.3 at.% Sc, *Phys. Met. Metall*, 1984, **57**, p 118–126
14. I.J. Polmear, *Light Alloys, Metallurgy of the Light Metals*. Edward Arnold (Publishers) Ltd., London, 1981
15. T. Torma, E. Kovacs-Csetenyi, L. Vitalis, J. Stepanov, and M. Butova, The Effect of Scandium Addition on the Mechanical Properties of Pure Aluminium and of an AlMg6 Alloy, *Mater. Sci. Forum*, 1987, **13/14**, p 497–503
16. B.A. Parker, Z.F. Zhou, and P. Nolle, The Effect of Small Additions of Scandium on the Properties of Aluminium Alloys, *J. Mater. Sci.*, 1995, **30**, p 452–458
17. W.F. Smith and N.J. Grant, Effects of Chromium and Copper Additions on Precipitation in Al-Zn-Mg Alloys, *Metall. Trans. A*, 1971, **2**, p 1333–1338
18. A.J. Ardell, Precipitation Hardening, *Metall. Trans. A*, 1985, **16**(8), p 2131–2136