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Effects of Mg and Cu Additions on Superplastic Behavior in MA Aluminum Alloys

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Abstract MA Al alloys are examined to determine the effects of alloying of Mg and Cu and rolling on tensile deformation behavior at 748 K over a wide strain rate range(10^{-4}-10^3/s). A powder metallurgy aluminum alloy produced from mechanically alloyed pure Al powder exhibits only a small elongation-to-failure(ε_f < ~50%) in high temperature(748 K) tensile deformation at high strain rates($\dot{\varepsilon}$ = 1-10^2/s). ε_f in MA Al-0.5~4.0Mg alloys increases slightly with Mg content(ε_f = ~140% at 4 mass%). Combined addition of Mg and Cu(MA Al-1.5%Mg-4.0%Cu) is very effective for the occurrence of superplasticity(ε_f > 500%). Warm-rolling(at 393-492 K) tends to raise ε_f. Lowering the rolling-temperature is effective for increasing the ductility. The effect is rather weak in MA pure Al and MA Al-Mg alloys, but much larger in the MA Al-1.5%Mg-4.0%Cu alloy. Additions of Mg and Cu and warm-rolling of the alloy cause a remarkable reduction in the logarithm of the peak flow stress at low strain rates ($\dot{\varepsilon}$ < ~1/s) and sharpening of microstructure and smoothening of grain boundaries. Additions of Mg and Cu make the strain rate sensitivity(the m value) larger at high strain rates, and the warm-rolling may make the grain boundary sliding easier with less cavitation. Grain boundary facets are observed on the fracture surface when ε_f is large, indicating the operation of grain boundary sliding to a large extent during superplastic deformation.

Key words tensile deformation, strain rate, superplasticity, cavitation, grain boundary sliding.

1. Introduction

Powder-metallurgy aluminum(Al) alloys produced from mechanically alloyed(MA) Al powders consist of fine (submicron in diameter) grains(or subgrains).[1,2] Further, very fine(a few 10 nm in diameter) particles are generally distributed by a several vo1ume% homogeneously in the matrix. They may by Al_2O_3 and Al_4C_3 formed by the reaction of Al with ethanol which was incorporated into the Al powder during MA process.[3-5] Therefore, superplastic deformation due to fine grained structure is expected to occur at elevated temperatures under high strain rates. In fact, a large elongation-to-failure has been reported in tension of some MA Al alloys at extremely high strain rates above 10^{-1}/s.[6-8]

Previous works have shown that, without exception, the maximum value of tensile elongation is smaller in a MA Al-Mg alloy(IN9052) than in MA Al-Mg-Cu alloys (IN9021 and IN90211) despite their similar microstructure (uniform dispersion of very fine carbides and oxides in a fine grained matrix).[6-9] Further, it was found in the preliminary experiment that an alloy produced from MA pure Al powder, that is also similar in microstructure to the above alloys, exhibited only a small tensile elongation (a few 10% in maximum). These facts propose that some sort of alloying elements, which may be disso1ved in the matrix at the deformation temperature, also affects the tensile elongation, probably through some modification of microstructure besides particle dispersion and grain size. Moreover, it has been found that thermomechanical processing like warm-rolling to deformation influences the amount of tensile elongation in an ingot-metallurgy Al-Mg alloys.[10)]

The purpose of the present paper is to examine the effects of alloy additions of Mg and Cu and warm-rolling on the high-temperature superplastic behavior in MA Al alloys, and to know the microstructure responsible for high strain rate superplasticity.

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2. Experimental Procedure

Chemical compositions of MA Al alloys examined in the present work are, by mass%, Al-1.7%O-1.1%C, Al-0.5%Mg-1.7%O-1.1%C, MA Al-2.0%Mg-1.7%O-1.1%C, Al-4.0%Mg-0.8%O-1.1%C and Al-1.5%Mg-4.0%Cu-0.8% O-1.1%C. The last two are IN9052 and IN9021, respectively, produced by Incomap Co. The alloys will be designated MA pure Al, MA Al-0.5Mg, MA Al-2.0Mg, MA Al-4.0Mg and MA Al-1.5Mg-4.0Cu in this paper. The difference in volume% of fine particles due to the difference in oxygen content was ~2%. Plates(4 mm thick) were machined from extruded rods of these alloys and rolled to 1 mm thickness(reduction in thickness: ~75%), parallel to the extruding direction, at three different temperatures: 393, 423 and 493 K. Specimens for tension tests(length 10 mm, width 5 mm, thickness 1 mm) were machined from the rolled sheets, parallel to the rolling direction. Specimens were finally annealed at 823 K for 1 h and then quenched in iced- water. After heating quickly to 753 K(deformation temperature) and holding at the temperature for l0 min, specimens were deformed at constant nominal strain rates($1 \times 10^{-4} - 1 \times 10^{3}$/s) with hydraulic testing machines. The matrices of the alloys are supposed to be in a state of solid solution at the deformation temperature; Al_3Mg_2 and $CuMgAl_2$ phases in MA Al-Mg alloys and MA Al-1.5Mg-4.0Cu, respectively, may have been mostly dissolved.

Thin foils for TEM observation were prepared from specimens by spark-erosion machining and standard twin-jet electro polishing, and examined in a JEM-3010 electron microscope operated at 300 kV. Fracture surfaces were inspected by SEM(JSM-890S).

3. Results and Discussion

3.1 Stress vs. Strain Behavior

Examples of nominal stress, σ_n, vs. nominal strain, ε_n, curves obtained at 753 K under extremely high and low strain rates(5 and 1×10^{-4}/s) are given in Fig. l for MA pure Al, MA Al-4.0Mg and MA Al-1.5Mg-4.0Cu. σ_n vs. ε_n curves for MA Al-0.5Mg and MA Al-2.0Mg were located between those for MA pure Al and MA Al-4.0Mg. The following can be seen from the figure: (1) Both maximum flow stress and elongation-to-failure are larger at higher strain rate; (2) In contrast to the deformation at room temperature, the maximum flow stress decreases with the alloy additions of Mg and Cu; (3) Elongation-to-failure increases with the alloying. It may be noteworthy that the tensile elongation of MA pure Al is very small (<1%) at low strain rate.

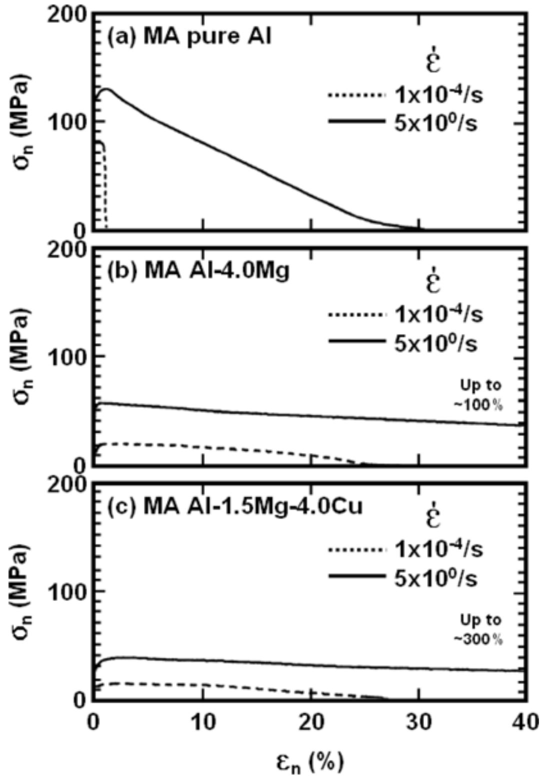

Fig. l. Nominal stress, σ_n vs. nominal strain, ε_n, curves at 748 K for (a) MA pure Al, (b) MA Al-4.0Mg and (c) MA Al-1.5Mg-4.0Cu rolled at 423 K and finally annealed at 773 K.

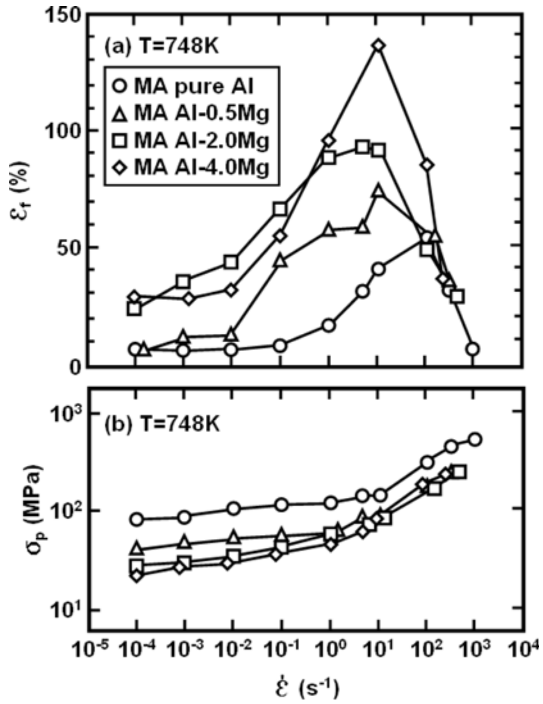

Fig. 2. Effect of Mg addition on (a) elongation-to-failure, ε_f vs. nominal strain rate, $\dot{\varepsilon}$, and (b) peak flow stress, σ_p vs. $\dot{\varepsilon}$ relations in MA Al-Mg alloys rolled at 423 K and finally annealed at 773 K.

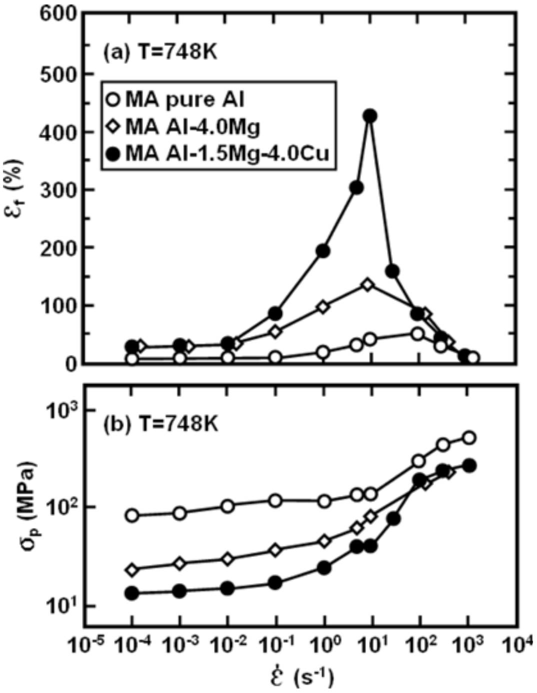

Fig. 3. Effect of combined addition of Mg and Cu on (a) elongation-to-failure, ε_f vs. nominal strain rate, $\dot{\varepsilon}$, and (b) peak flow stress, σ_p vs. $\dot{\varepsilon}$ relations in Al-1.5Mg-4.0Cu rolled at 423 K and finally annealed at 773 K. Results on MA pure Al and MA Al-4.0Mg are also shown.

3.2 Tensile Elongation and Strain Rate Sensitivity

3.2.1 Effect of addition of alloying elements, Mg and Cu

Figure 2 shows changes in the elongation-to-failure, ε_f, and the peak flow stress, σ_p, as a function of the nominal strain rate, $\dot{\varepsilon}$, in some MA Al-Mg alloys. The value of ε_f reaches a maximum in the neighborhood of $\dot{\varepsilon} = 10/s$. The strain rate sensitivity, $m = d(\ln \sigma_p)/d(\ln \dot{\varepsilon})$, a parameter governing the rate of neck development, correspondingly takes a large value(0.3-0.4) in high strain rates($\dot{\varepsilon} > \sim 1/s$), while it is less than 0.1 at lower ones($\dot{\varepsilon} < \sim 1/s$). The maximum values both in ε_f and m increase with the increase in the amount of alloying element, Mg.

Figure 3 shows ε_f vs. $\dot{\varepsilon}$ and σ_p vs. $\dot{\varepsilon}$ relations in MA Al-1.5Mg-4.0Cu along with the results on MA pure Al and MA Al-4.0Mg for comparison. The tendency observed in the addition of Mg becomes more pronounced by the combined addition of Mg and Cu. Variation in the deformation behavior with the alloying can be characterized by the features: an increase in the maximum of ε_f at high strain rates(around $\dot{\varepsilon} = \sim 10/s$), a remarkable decrease in $\ln \sigma_p$ at low strain rates($\dot{\varepsilon} < \sim 1/s$), and a slight decrease in the strain rate at which ε_f takes the maximum.

At least from a phenomenological viewpoint, one can say that the decrease in $\ln \sigma_p$ at low strain rates leads to the increase in the m value and consequently to the increase in ε_f at high strain rates. The third feature may be related to the increase in the grain size former to deformation, as will be shown later.

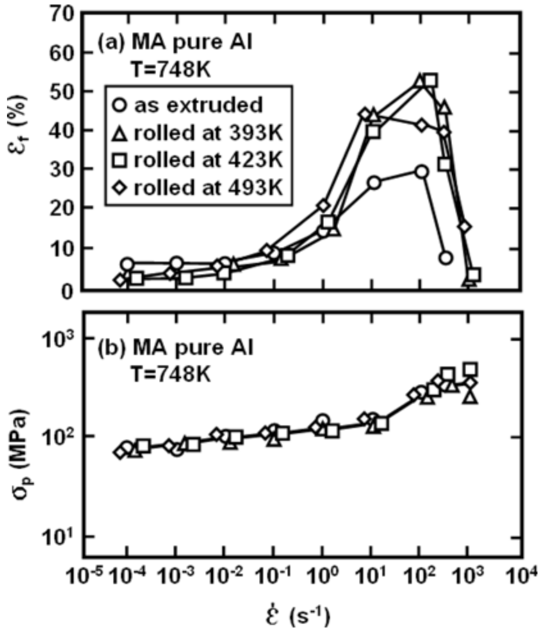

Fig. 4. Effect of Warm-rolling on (a) elongation-to-failure, ε_f vs. nominal strain rate, $\dot{\varepsilon}$, and (b) peak flow stress, σ_p vs. $\dot{\varepsilon}$ relations in MA pure Al finally annealed at 773 K.

3.2.2 Effect of rolling

Figure 4 shows changes in ε_f vs. $\dot{\varepsilon}$ and σ_p vs. $\dot{\varepsilon}$ relations with warm-rolling former to deformation in MA pure Al that exhibited the most poor ductility among the alloys examined. The maximum value of ε_f observed around $\dot{\varepsilon} = \sim 50/s$ becomes larger by the rolling, particularly at low temperature(Fig. 4(a)). However, the amount of the increase in ε_f is considerably small(~25%), reflecting no notable change in the σ_p vs. $\dot{\varepsilon}$ relation(Fig. 4(b)). In MA Al-4.0Mg, too, the effect of warm rolling was found to be rather small.

In MA Al-1.5Mg-4.0Cu that exhibited the largest elongation, on the other hand, the tensile ductility can be largely improved by the rolling; the increase in the peak value of ε_f reaches ~300% after rolling at the lowest temperature(393 K) (Fig. 5(a)). This may be resulted from the increase in the m value caused by the reduction in $\ln \sigma_p$ at low strain rates below 1/s(Fig. 5(b)).

3.3 Microstructure

Let us consider here, from a viewpoint of microstructure, the reason for the increase in tensile ductility by the alloy addition and the prior warm-rolling.

Examples of TEM micrographs of MA pure Al and MA Al-1.5Mg-4.0Cu after rolling at 423 K and subsequent

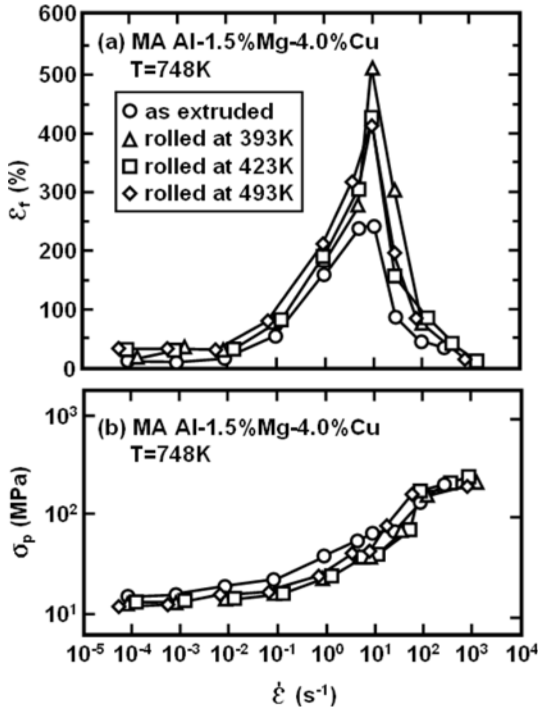

Fig. 5. Effect of Warm-rolling on (a) elongation-to-failure, ε_f vs. nominal strain rate, $\dot{\varepsilon}$, and (b) peak flow stress, σ_p vs. $\dot{\varepsilon}$ relations in MA Al-1.5Mg-4.0Cu finally annealed at 773 K.

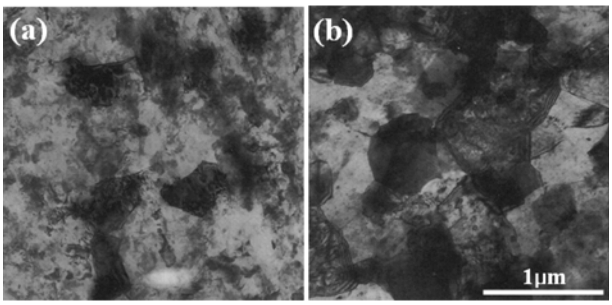

Fig. 6. TEM micrographs of (a) MA pure Al and (b) MA Al-1.5Mg-4.0Cu rolled at 423 K and finally annealed at 773 K.

annealing at 823 K are given in Fig. 6. Even MA pure Al, which exhibited very poor ductility, consists of fine grains(or subgrains) indeed. However, the internal structure of this alloy has not yet been recovered before deformation (after the above thermomechanical treatment); the crystal lattice is highly strained and the grain boundaries are irregular as indicated by complex bend contours in the few thickness fringes and grain interiors at grain boundaries, respectively. On the other hand, MA Al-1.5Mg-4.0Cu, which exhibited large elongation, has been almost fully recovered; the grain size is larger compared to that of MA pure Al, the grain interiors are rather free from dislocations and clear thickness fringes can be seen at grain boundaries. The internal structure of MA Al-4.0Mg was found to be in an incompletely recovered state.

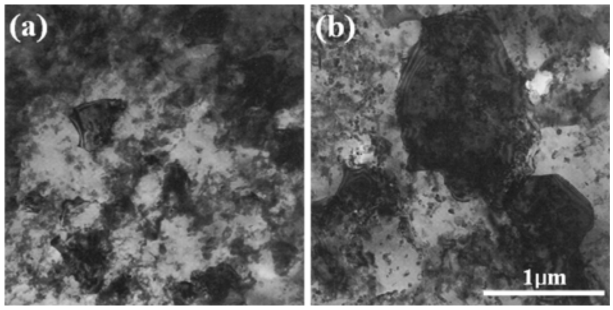

Fig. 7. TEM micrographs of (a) MA pure Al and (b) MA Al-1.5Mg-4.0Cu after high-temperature extrusion.

These observations may suggest that the characteristic changes in deformation behavior with the alloying of Mg and Cu, namely the decrease in ln σ_p at low strain rates below l/s and the increase in ε_f at higher strain rates around 10/s, are ascribed to the smoothening of grain boundaries; more smooth boundaries would be able to slide under lower stresses with less cavitation. Further, the strain rate giving the peak of ε_f showed a tendency to decrease with the alloying(Fig. 3). This might be related to the increase in grain size by the alloying(compare Figs. 6(a) and 6(b)). At present, we do not know the exact reason for the development of more well-defined grain(or subgrain) structure with the alloying. It might be due to the increase in the self diffusion of Al by the solute atoms, Mg and/or Cu, as suggested by Bieler *et al.*.[9]

Figure 7 shows TEM micrographs taken from the as-excluded MA pure Al and MA Al-1.5Mg-4.0Cu. One can see the effect of warm rolling on the evolution of microstructure by comparing micrographs in Fig. 6 with those in this figure. In MA pure Al, the effect is small; the internal structure is highly strained in both the extruded and the subsequently rolled states. In MA Al-1.5Mg-4.0Cu, on the other hand, the effect is considerably large; the partially recovered structure consisting of rather coarse grains in the as-extruded state can be made more sharp(more well-defined) in nature and finer in scale by the rolling. Taking into account that the sharpening of structure involves an increase in the boundary misorientation due to the incorporation of excess dislocations, which have been induced during rolling, into pre-existing boundaries, the enhancement of superplastic deformation by the rolling observed in MA Al-1.5Mg-4.0Cu is reasonably considered to be a reflection of the sharpening and refinement of microstructure.[10]

SEM micrographs of fracture surface in MA pure Al and MA Al-1.5Mg-4.0Cu deformed at T = 753 K and $\dot{\varepsilon}$ = 10/s are given in Fig. 8. In the former(ε_f = ~40%), the surface is wavy and many dimples are formed, indicating that the fracture is essentially transgranular. In the latter(ε_f = ~400%), grain boundary facets can be seen

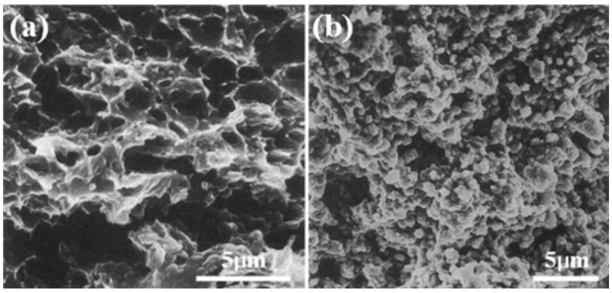

Fig. 8. SEM micrographs of fracture surface in (a) MA pure Al and (b) MA Al-1.5Mg-4.0Cu rolled at 423 K and finally annealed at 773 K. T = 748 K, $\dot{\varepsilon}$ = 10/s.

and the fracture is intergranular. These facts may suggest that the large elongation due to high strain rate superplastic deformation in MA Al alloys is mostly caused by the grain boundary sliding, as reported previously.[7)]

4. Conclusions

(1) Powder metallurgy aluminum(Al) alloy produced from mechanically alloyed pure Al powder(MA pure Al) exhibits only a small elongation-to-failure(ε_f < ~50%) in high temperature(748 K) tensile deformation at high strain rates($\dot{\varepsilon}$ = 1-10^2/s). ε_f in MA Al-0.5~4.0Mg alloys increases slightly with Mg content(ε_f = ~140% at 4 mass%). Combined addition of Mg and Cu(MA Al-1.5%Mg-4.0%Cu, IN9021) is very effective for the occurrence of superplasticity(ε_f > 500%).

(2) Warm-rolling(at 393-492 K) tends to raise ε_f. Lowering rolling-temperature is effective for the ductility increase. The effect is rather weak in MA pure Al and MA Al-Mg alloys, but much larger in MA Al-1.5%Mg-4.0%Cu alloy.

(3) Alloy additions of Mg and Cu and warm-rolling cause a remarkable reduction in the logarithm of peak flow stress at low strain rates($\dot{\varepsilon}$ < ~1/s), and sharpening of microstructure and smoothening of grain boundaries. The former makes the strain rate sensitivity(the m value) larger at high strain rates, and the latter may make the grain boundary sliding easier with less cavitation.

(4) Grain boundary facets can be observed on the fracture surface when ε_f is large, indicating the operation of grain boundary sliding to a large extent during superplastic deformation.

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