Mechanical Properties and Blow Forming of Rolled AZ31 Mg Alloy Sheet

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Rolling was conducted at 373–673 K for AZ31 Mg alloy; mechanical properties of the rolled Mg alloy were investigated by tensile and blow forming tests. The grain sizes of all the rolled specimens were smaller than that of the specimen prior to rolling. At tensile temperatures under 373 K, the rolled specimens showed much higher 0.2% proof stresses than the non-rolled specimens due to their fine-grained microstructure. However, the strength of the rolled specimens decreased significantly at 473 K. Superplastic behavior was obtained at 573–723 K for the specimens rolled at 498 K. Blow forming tests demonstrated that specimens rolled at 498 K exhibited a high degree of formability at 723 K.

1. Introduction

In magnesium at room temperature, the critical resolved shear stresses for the non-basal slips of the prismatic and pyramidal slips are much greater than those that for basal slips,[1,2] and hence non-basal slips are difficult to operate at room temperatures. Five independent slip systems are necessary for a polycrystalline material to be able to undergo a general homogeneous deformation without producing cracks. However, the number of independent modes for basal slips is only 2.[3] This gives rise to poor formability at room temperature of polycrystalline Mg and its alloys. Because the difference in critical resolved shear stress between the basal slip and the non-basal slips decreases with increasing temperature, the formability increases with increasing temperature. In addition, grain refinement leads to significant improvement of formability at elevated temperatures because superplasticity is attained for fine-grained alloys.[4–7]

It is well known that a fine-grained microstructure is required to attain superplasticity.[8] It has been reported that fine-grained microstructures can be obtained by hot rolling, hot extrusion, hot forging, *etc.*, of Mg alloys.[7,9–11] Grain refinement by hot deformation processing is attributed to dynamic recrystallization.[12] Recently, Mukai *et al.*[6] have demonstrated superplastic forming of a rolled AZ61 sheet. However, there are few systematic studies on the relationship between mechanical properties and processing conditions for rolled Mg alloys. In the present paper, rolling was conducted over a wide temperature range of 373–673 K for AZ31 Mg alloy and the mechanical properties of the rolled Mg alloy were investigated using tensile tests at between room temperature and 723 K. Furthermore, blow forming tests were carried out at 723 K to investigate the superplastic formability of the rolled Mg alloy.

2. Experimental Procedure

An AZ31 (Mg–3 mass%Al–1 mass%Zn–0.15 mass%Mn) alloy ingot was prepared and annealed at 691 K for
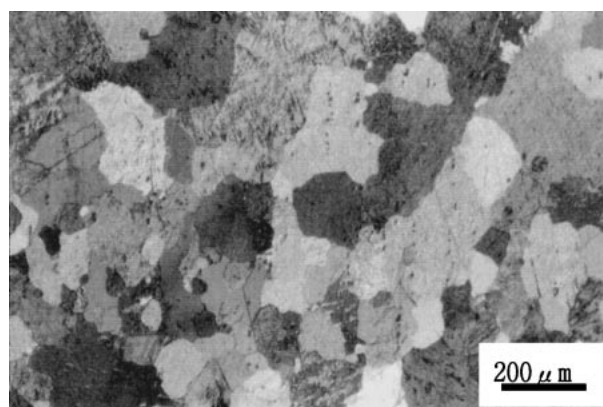

Fig. 1   Microstructure of the annealed AZ31 Mg alloy ingot.

2.6×10^5 s. The microstructure of the annealed ingot is shown in Fig. 1. Its grain size was 156 µm. Blocks 7 mm in thickness and 50 mm in breadth were cut from the annealed ingot and rolled at 373, 423, 473, 498, 523, 573 and 673 K using a rotational speed of 500 mm/s under a roll diameter of 100 mm. Reduction per pass was 0.4 mm down to a thickness of 2.2 mm, followed by 0.2 mm per pass down to a thickness of 1.0 mm. Rolling reduction was 85.7%.

The microstructures of the RD-TD plane, the ND-RD plane and the TD-ND plane of the rolled sheets were observed by optical microscope, where the RD is the rolling direction, the TD is the transverse direction and the ND is the normal direction. From the measurement of the apparent grain size of each plane using the intercept method, the grain size was determined by

$$d = 1.73 \times (d_{RT} \times d_{NR} \times d_{TN})^{1/3} \tag{1}$$

where d is the grain size, d_{RT} is the grain size in the RD-TD plane, d_{NR} is the grain size in the ND-RD plane and d_{TN} is the grain size in the TD-ND plane. The (0002) pole figure of the RD-TD plane in the specimen rolled at 498 K was measured at a position of about 0.5 mm under the surface.

Tensile specimens 10 mm in gage length, 5 mm in gage breadth and 1 mm in gage thickness were machined from the

annealed ingot and the rolled sheets. Tensile tests were carried out at room temperature, 373, 473 and 573 K at an initial strain rate of $1.7 \times 10^{-3}\,\mathrm{s}^{-1}$ for the annealed ingot and the specimens rolled at 498, 523, 573 and 673 K, with the angle between the tensile direction and the rolling direction at 0 degrees. Furthermore, additional tensile tests, where the angle between the tensile direction and the rolling direction was 45 and 90 degrees, were conducted at room temperature and 723 K for the specimen rolled at 498 K to investigate the anisotropy of their tensile properties. Step strain rate tests were also carried out at 573, 673 and 723 K with 1.5×10^{-5}–$1.5 \times 10^{-1}\,\mathrm{s}^{-1}$ for the specimen rolled at 498 K to investigate its superplastic properties. In addition, elongation to failure was measured for the specimen rolled at 498 K by tensile tests at 573, 673 and 723 K at 10^{-5}–$10^{-1}\,\mathrm{s}^{-1}$.

Discs 70 mm in diameter were machined from the sheet rolled at 498 K. The specimens were clamped between two hollow dies and bulged at 723 K using N_2 gas pressure. Domes and cups to heights up to 20 mm were formed in the die cavity, which was 40 mm in diameter.

3.　Results and Discussion

Rolling from 7 to 1 mm in thickness was tried at 373, 423, 473, 498, 523, 573 and 673 K. At rolling temperatures of more than 498 K, the specimens could be rolled to a thickness of 1 mm in without cracking. In this case, the rolling reduction was 85.7%. However, when the rolling temperature was lower than 473 K, the specimens could not be rolled to a thickness of 1 mm without cracking. The relationship between the rolling reduction and the rolling temperature is shown in Fig. 2. The rolling reduction decreased with decreasing rolling temperature with falling temperatures from 473 K. This is because the activity of non-basal slips drops with decreasing temperature.

The microstructures of the rolled specimens are shown in Fig. 3, where the RD-TD plane is observed. The rolling temperature is (a) 373 K, (b) 498 K and (c) 673 K, respectively. The grains of the rolled specimens were almost equiaxed and the grain sizes of the rolled specimens were smaller than that of the specimen prior to rolling, indicating that dynamic recrystallization was caused by rolling in the
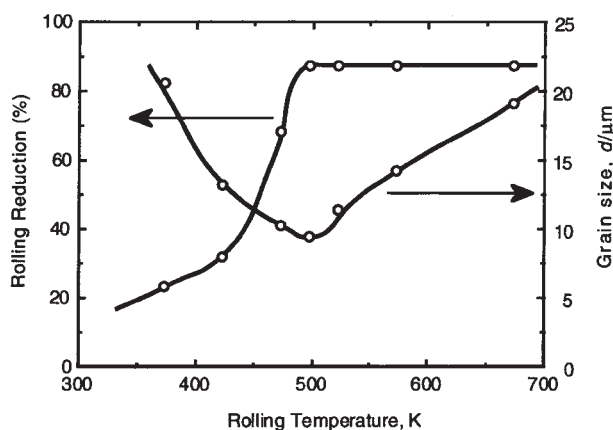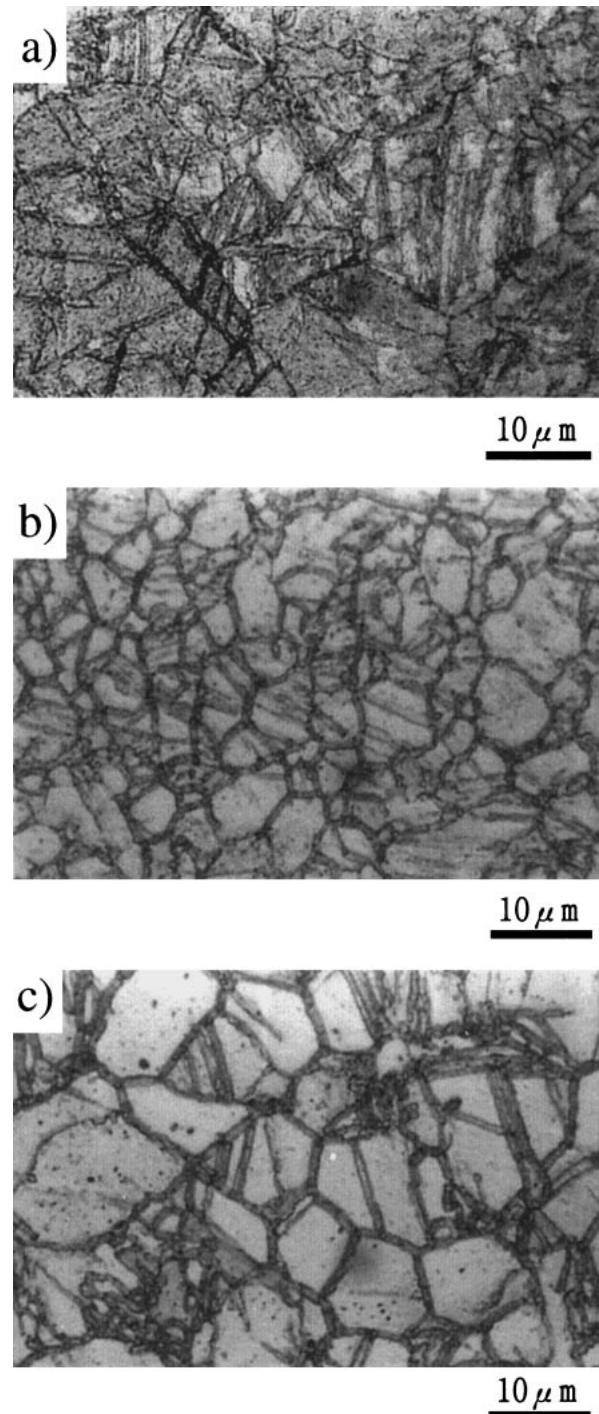

Fig. 3　Microstructures of the rolled specimens, where the RD-TD plane is observed. The rolling temperature is (a) 373 K, (b) 498 K and (c) 673 K, respectively.

temperature range investigated. Also, twinning was observed in all the rolled specimens. It is known that twin nucleation is not thermally activated: it occurs at locations of high stress concentration.[13] Hence the fact that twinning was observed in all the rolled specimens suggests that stress concentration occurred during rolling, irrespective of the rolling temperature, indicating deformation twinning.

The variation in grain size of the rolled specimens as a function of rolling temperature is shown in Fig. 2. The minimum grain size of 9.4 µm was obtained at the rolling

Fig. 2　The relationship between rolling reduction, grain size and rolling temperature.

temperature of 498 K. The grain size increased with decreasing rolling temperature from 498 K. This is because the rolling reduction decreased with decreasing rolling temperature from 498 K.

Tensile tests were carried out at room temperature, 373, 473 and 573 K for the annealed ingot and the specimens rolled at 498, 523, 573 and 673 K, where the angle between the tensile direction and the rolling direction was 0 degrees. The variations in 0.2% proof stress and elongation to failure as a function of tensile temperature are shown in Fig. 4 and in Fig. 5, respectively. It is noted that the rolled specimens showed much higher 0.2% proof stress than the annealed specimen in a tensile temperature range under 373 K. Magnesium and its alloys show a large Taylor factor due to a lack of slip systems, and hence they exhibit a grain size strongly dependence on stress, allowing high strength to be attained by grain refinement.[14,15] For this reason, the high strength of the rolled specimens can be attributed to their small grain size. However, the strength of the rolled specimens decreased significantly at 473 K. In particular,

the 0.2% proof stresses at 573 K for the rolled specimens were slightly lower than those for the annealed ingot.

Chapman and Wilson[16] showed that the elongation to failure at room temperature increases with decreasing grain size for Mg and its alloys. In the present investigation, however, the elongation of the specimen rolled at 673 K was larger than those of the others, and the trend of increased elongation with decreasing grain size was not obtained at room temperature. This may be related to the presence of twins in the rolled specimens. On the other hand, the elongation at 573 K increased with decreasing grain size. In particular, the specimen rolled at 498 K with a small grain size of 9.4 μm showed large elongation of 256%.

To investigate the anisotropy of tensile properties for the specimen rolled at 498 K, tensile tests were conducted at the angles between the tensile direction and the rolling direction of 45 and 90 degrees. The nominal stress-nominal strain curves at the angles of 0, 45, and 90 degrees are shown in Fig. 6, where the tensile temperatures are (a) room temperature and (b) at 723 K. It should be noted that there was little anisotropy in tensile properties at room temperature. The stress during deformation and elongation are also showed little anisotropy at 723 K, although there is a difference in the initial stage. This trend is the same as the results obtained by Kaneko *et al.*[17] The Lankford values at the angles of 0, 45, and 90 degrees are listed in Table 1. It can be seen that there
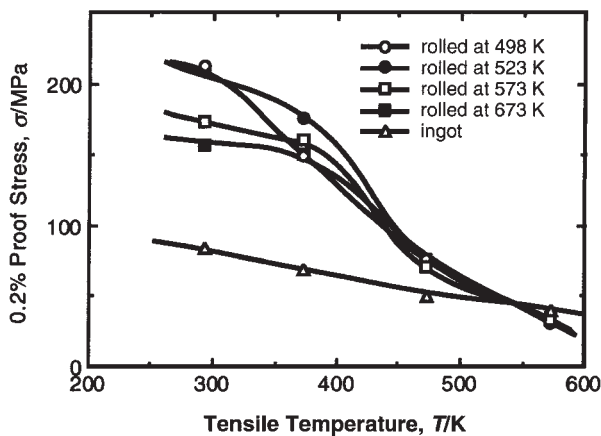

Fig. 4 The variation in 0.2% proof stress as a function of tensile temperature for the annealed ingot and the specimens rolled at 498, 523, 573 and 673 K, where the angle between the tensile direction and the rolling direction is 0 degrees.
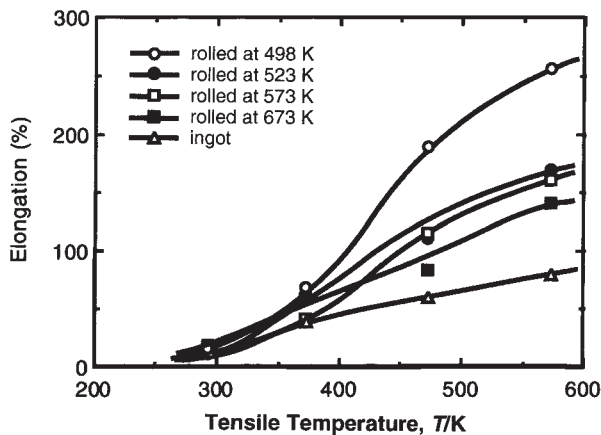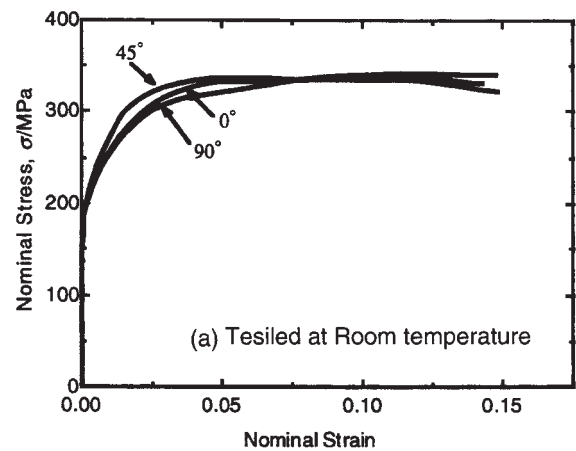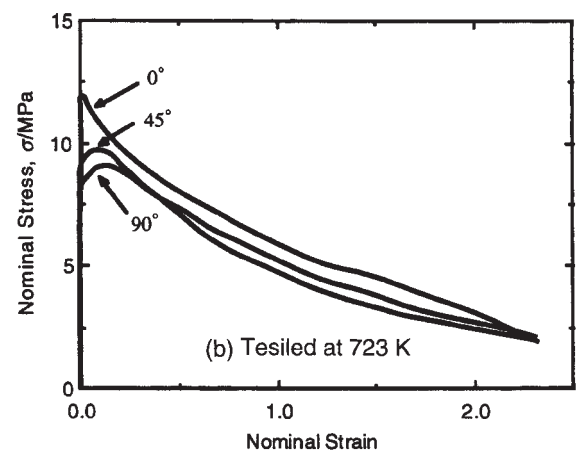

Fig. 5 The variation in elongation to failure as a function of tensile temperature for the annealed ingot and the specimens rolled at 498, 523, 573 and 673 K, where the angle between the tensile direction and the rolling direction is 0 degrees.

Fig. 6 The nominal stress-nominal strain curves at the angles of 0, 45, and 90 degrees for the specimen rolled at 498 K, where the tensile temperatures are (a) room temperature and (b) 723 K.

Table 1 The Lankford values at the angles of 0, 45, and 90 degree for the specimen rolled at 498 K.

	0°*	45°*	90°*
Room temperature	3.4	3.7	3.7
723 K	1.2	1.3	1.1

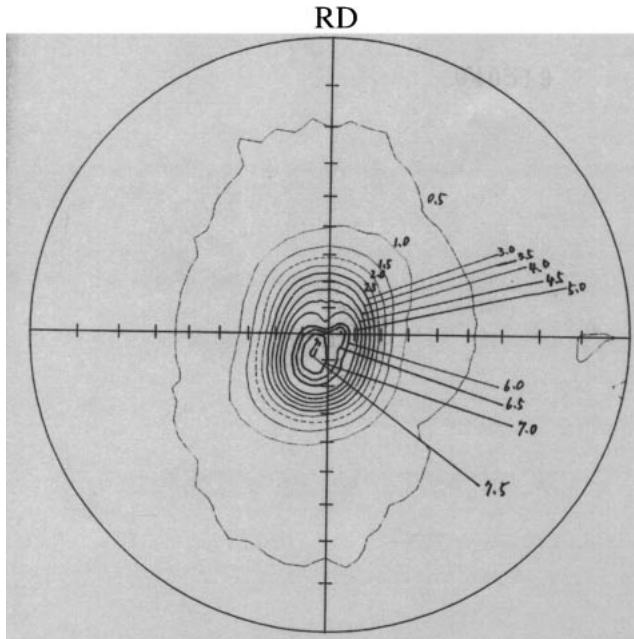*The angle is the one between the tensile direction and the rolling direction.

Fig. 7 The (0002) pole figure of the RD-TD plane in the specimen rolled at 498 K.
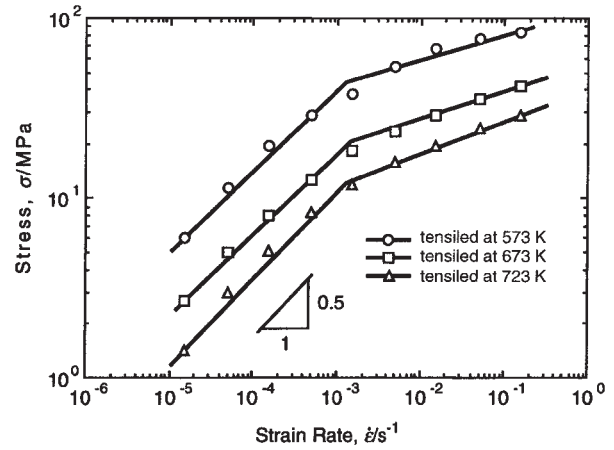

Fig. 8 The variation in stress as a function of strain rate for the specimen rolled at 498 K, where stress is the true stress at the true strain of 0.1.
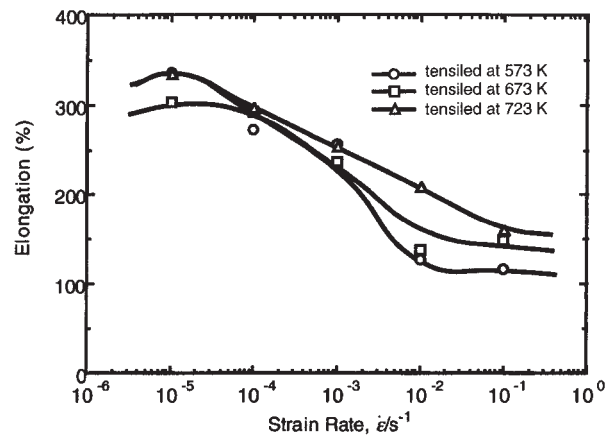

Fig. 9 The variation in elongation to failure as a function of strain rate for the specimen rolled at 498 K.

was little anisotropy in the Lankford value of the rolled Mg alloy.

The (0002) pole figure of the RD-TD plane in the specimens rolled at 498 K is shown in Fig. 7. The basal plane tended to be parallel to the RD-TD plane. The strong texture parallel to the RD-TD plane is likely to be responsible for the low anisotropy seen in the tensile properties of the rolled Mg alloys.

The variation in stress at 573–723 K as a function of strain rate for the specimen rolled at 498 K is shown in Fig. 8. It can be seen that a high strain rate sensitivity of about 0.5 was obtained in the low strain rate range at less than $10^{-3}\,\mathrm{s}^{-1}$. Clearly, superplastic behavior occurred in the strain rate range at less than $10^{-3}\,\mathrm{s}^{-1}$. In the high strain rate range at more than $10^{-3}\,\mathrm{s}^{-1}$, however, the strain rate sensitivity was about 0.2, and no superplastic behavior occurred.

The variation in elongation to failure at 573–723 K as a function of strain rate for the specimen rolled at 498 K is shown in Fig. 9. A large elongation of more than 200% was attained in the strain rate range of under $10^{-3}\,\mathrm{s}^{-1}$, which was in agreement with the strain rate range in which the high strain rate sensitivity of 0.5 was obtained. In this range, elongation was independent of the tensile temperature. However, in the high strain rate range showing a low strain rate sensitivity of 0.2, elongation increased with increasing

tensile temperature. It should be noted that a large elongation of 209% was obtained at 723 K with a high strain rate of $10^{-2}\,\mathrm{s}^{-1}$.

Blow forming tests were carried out at 723 K for the specimen rolled at 498 K. The blow conditions were pressure of 0.68 MPa with forming time of 1800 seconds for Condition A; pressure of 1.23 MPa with forming time of 130 seconds for Condition B; and pressure of 1.84 MPa with forming time of 10 seconds for Condition C. The stress at the dome apex can be given by[18]

$$\sigma_{\mathrm{c}} = \sigma_{\mathrm{t}} = \frac{PB}{4S}\left(H + \frac{1}{H}\right) \qquad (2)$$

where σ_{c} is the stress in the circumferential direction, σ_{t} is the stress in the thickness direction, P is the forming pressure, B is the die radius, S is thickness and H is the relative dome height. From eq. (2) and the results seen in Fig. 8, the strain rate is in the order of $10^{-4}\,\mathrm{s}^{-1}$ for Condition A, $10^{-3}\,\mathrm{s}^{-1}$ for Condition B and $10^{-2}\,\mathrm{s}^{-1}$ for Condition C. The optimum results of blow forming tests for each condition are shown in Fig. 10. The cup shape was successfully formed for Condition A. The high degree of formability for Condition A results from superplastic deformation. In this case, however, a long

Fig. 11 Tensile properties at room temperature of the specimen formed under Condition A and the non-formed specimen.
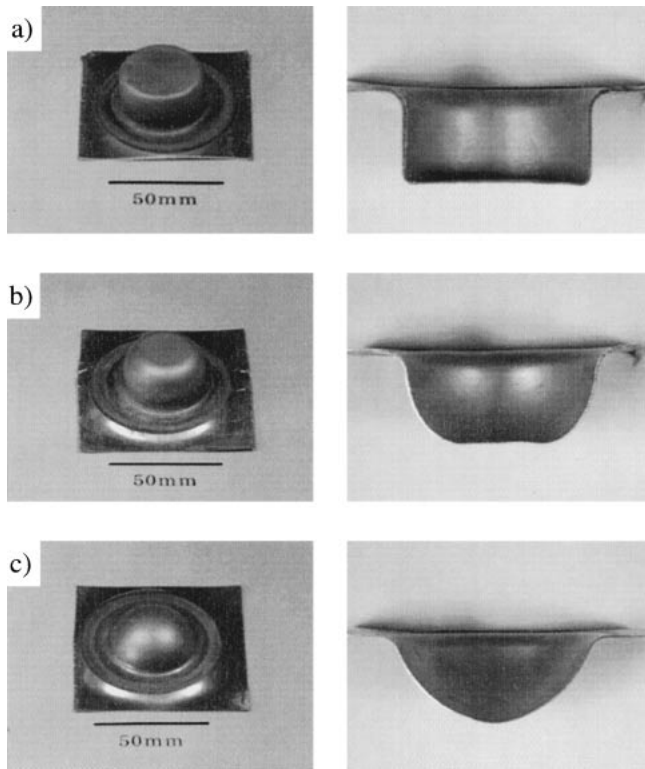
Fig. 10 The optimum results without cracking of the blow forming tests at 723 K for the specimen rolled at 498 K. (a) Condition A with pressure of 0.68 MPa and forming time of 1800 seconds, (b) Condition B with pressure of 1.23 MPa and forming time of 130 seconds, and (c) Condition C with pressure of 1.84 MPa and forming time of 10 seconds.

Fig. 12 Microstructure of the specimen formed under Condition A.

forming time of 1800 s was needed. On the other hand, the dome shape was formed in Condition C in a very short forming time (= 10 s). The strain rate for Condition C was beyond the superplastic region, exhibiting a high strain rate sensitivity of 0.5. However, good formability was attained for Condition C. This is likely to be attributed to the minimal anisotropy of mechanical properties and large elongation of about 200%.

Superplastic forming often causes a reduction in service properties.[19,20] Hence, it is worthwhile comparing mechanical properties before and after superplastic forming. Tensile properties at room temperature for the specimen formed under Condition A and the non-formed specimen are illustrated in Fig. 11. The formed specimen is fabricated from the bottom of the cup, which is confirmed from the difference between initial thickness and final thickness of this specimen, where the strain is about 1.2 in the range of 7.5 mm from bottom center. The non-formed specimen is given the same thermal history as the formed specimen. It can be seen that elongation significantly decreased as a result of forming. The 0.2% proof stress and the tensile strength were also slightly decreased.

The microstructure of the specimen formed under Condition A is shown in Fig. 12. The grain size was 22.5 μm, indicating that grain growth occurred during forming. Cavities were also observed. Therefore, it is likely that the reduction in service properties caused by forming can be attributed to grain growth and cavitation.
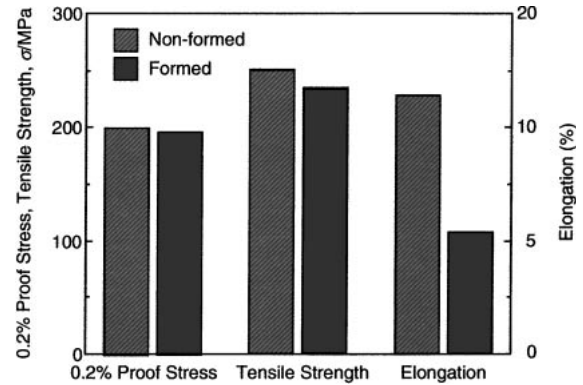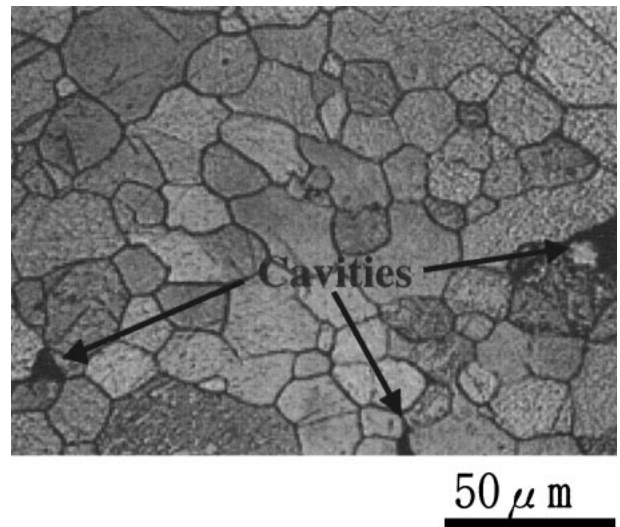
4. Conclusions

Rolling was conducted at 373–673 K for AZ31 Mg alloy, after which the mechanical properties of the rolled Mg alloy were investigated by means of tensile tests. Blow forming tests were also carried out at 723 K. The results were as follows.

(1) When the rolling temperature was higher than 498 K, the specimens could be rolled from 7 to 1 mm thickness. However, when the rolling temperature was under 473 K, this was not possible.

(2) The grains of the rolled specimens were almost equiaxed, and the grain sizes of the rolled specimens were smaller than that of non-rolled specimens. The minimum grain size of 9.4 μm was obtained at a rolling temperature of 498 K.

(3) In a tensile temperature range under 373 K, the specimens rolled at 498–673 K showed much higher 0.2% proof stress than the non-rolled specimens. However, the strength of the rolled specimens decreased significantly at 473 K. In particular, the 0.2% proof stresses at 573 K for the rolled specimens were slightly lower than those of the non-rolled specimens.

(4) The specimens rolled at 498 K exhibited little anisotropy in tensile properties at room temperature. The stress during deformation and elongation resulted in minimal anisotropy at 723 K, although there were some differences in the initial stage.

(5) The specimen rolled at 498 K showed superplastic behavior, namely, high elongation above 200% and a high strain rate sensitivity of 0.5 at 573–723 K at under $10^{-3}\,\mathrm{s}^{-1}$. The specimen also showed large elongation of about 200% at 723 K with a high strain rate of $10^{-2}\,\mathrm{s}^{-1}$, despite the low strain rate sensitivity of 0.2.

(6) The cup was made of a sheet rolled at 498 K by blow forming with a long forming time of 1800 s at 723 K. The dome was made by blow forming over a short forming time of 10 s.

(7) Blow forming had a harmful influence, attributable to grain growth and cavitation, on the mechanical properties of the rolled alloy.

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