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Flow stress and microstructural evolution in as rolled AZ91 alloy during hot deformation①

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Abstract: The flow stress behavior and microstructural evolution in as rolled AZ91 alloy were studied by tensile test at temperature ranging from 573 K to 698 K and with initial strain rate of 10^{-3}－$1 s^{-1}$. The results show that the activation energy of the AZ91 alloy is 96 kJ/mol and the stress exponent is 3.5. Grain refinement is observed during hot deformation. It is found that grain refinement is due to dynamic recrystallization, but grain size is decreased with increasing Zener-Hollomon parameter. The deformation mechanism is dislocation cross-slip, dislocation climbs and twin intersections, which are probably suitable to the development of substructures. Twins always exist in matrix whether they are at high temperature or at low temperature. Strain induced precipitation of $Mg_{17}Al_{12}$ occurs in the AZ91 alloy and stabilizes the subgrain boundaries.

Key words: magnesium alloy; grain refinement; recrystallization; strain softening; strain hardening

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1 INTRODUCTION

Mg alloys are the lightest construction materials for many engineering components due to their low density, higher ductility and suitable strength[1－3]. However, Mg alloys have poor formability and limited ductility at room temperature because of its HCP structure[4]. Therefore, it is required for Mg alloys to be deformed at warm temperature(> 498 K)[5], especially at elevated temperature, when prismatic slip is activated so that the formality of Mg alloys can be improved in terms of high ductility and makes it easy to simplify the design and manufacturing processing. Generally speaking, they are related to flow stresses and microstructures, which determine the efficiency of them. However, the strain distribution on various locations of the deformed products during forming processing is not homogenous so that it brings the different microstructures. In this study, as rolled AZ91 alloy is subject to uniaxial tension test at elevated temperature under various strain rates in order to get details of flow stresses and the microstructures and provide some information for sheet forming.

2 EXPERIMENTAL

The AZ91 alloy used in the present study has a chemical composition of Mg-9Al-0.8Zn-0.3Mn(mass fraction, %) in the shape of blocks. The blocks were solution-treated at 673 K for 20 h, and rolled for 11 passes with a reduction of 10%－20% per pass. The final thickness of block reached 1.4 mm with total reduction of 73%. Tensile specimens were machined from the sheets with a gauge of 10 mm × 5 mm. Constant velocity tensile tests were carried out at temperature ranging from 573 K to 698 K with initial strain rate $\dot{\varepsilon}= 10^{-3}$－$1 s^{-1}$. The size of grains was measured by optical microscope ($d= 1.74 L$).

3 RESULTS

3.1 Flow stress behavior

Fig. 1 shows the flow stress behavior of as rolled AZ91 alloy at different deformation conditions. Apparently, the general characteristics of the flow stress behavior are different. Below 623 K, the flow curves display high peak stress at a high corresponding strain and significant strain hardening after the peak. Above 623 K, the peak stress and its corresponding strain are small compared with those below 623 K with a feature of strain softening. However, the flow curves at temperature of 673 K or 698 K exhibit a large steady regime with a feature of complete dynamic recrys-tallization(DRX). When initial strain rate is changed, the peak stress and its corresponding strain follow the same patterns as temperature is decreased

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or increased. Furthermore, the peak stress and its corresponding strain are increased with increasing strain rate, as shown in Fig. 1(b). It is evident that the AZ91 is sensitive to temperature and strain rate.

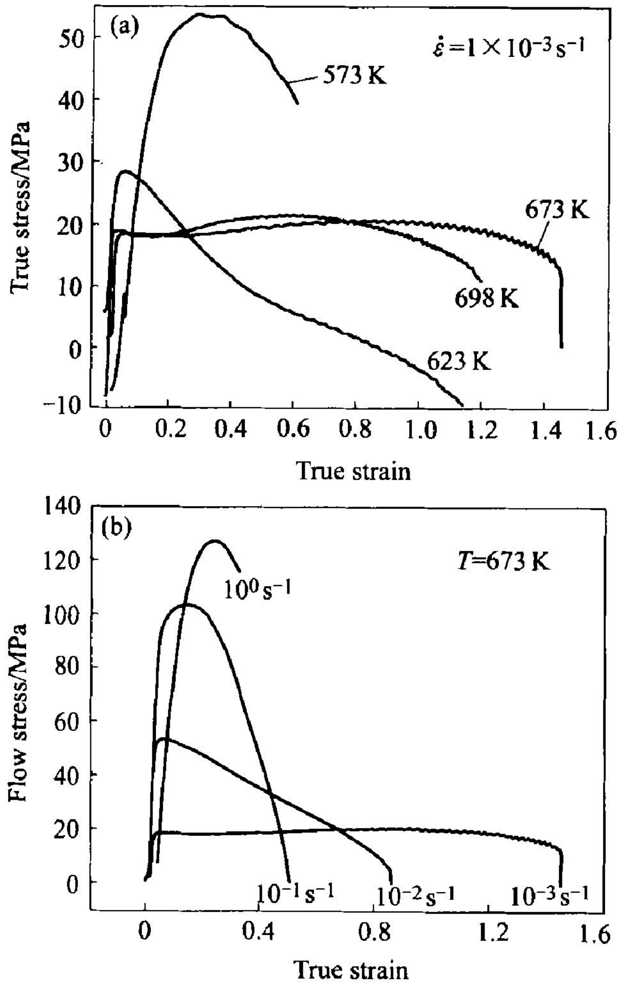

Fig. 1 Flow stress curves of AZ91 alloy obtained under different conditions

(a) —Strain rate of $10^{-3}s^{-1}$; (b) —T = 673 K

Figs. 2 and 3 are the constitute analysis for the AZ91 alloy under various deformation conditions. It is clearly shown that these curves almost follow straight lines in certain temperature range. In fact, these lines are obtained in a logarithmic plot where the behavior is better represented by the following equation:

$$A\sinh(\alpha\sigma)^n = \dot{\varepsilon}\exp(Q/RT) \quad (1)$$

where A, α(= 0.052 MPa^{-1}) are material constants[6], R is gas constant, n is stress exponent, Q is active energy and T is absolute temperature.

Through changing the form of this equation, it is obtained

$$Z = \dot{\varepsilon}\exp(Q/RT) = A\sigma_p^n \quad (2)$$

where σ_p is peak stress, Z is Zener-Hollomon parameter. Therefore,

$$n = (\partial\lg\dot{\varepsilon}/\partial\lg\sigma_p)_{T=\text{constant}} \quad (3)$$

$$Q = 2.3nR[\partial\lg\sigma_p/\partial(1/T)]_{\dot{\varepsilon}=\text{constant}} \quad (4)$$

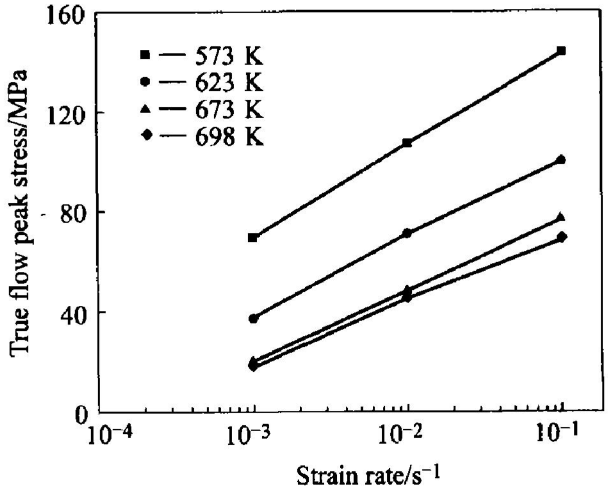

Fig. 2 Variation of normalized stress with strain rate in double logarithmic plot

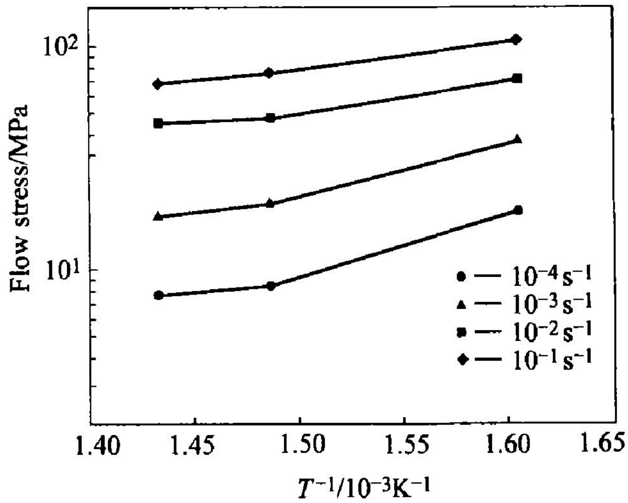

Fig. 3 Normalized peak stress as a function of temperature

According to the slope of these lines and these formulas, Q and n can be calculated respectively, Q = 96 kJ/mol, n = 3.5[7].

3.2 Microstructure changes during hot deformation

Fig. 4(a) shows the optical microstructures of the specimen before tensile testing. It can be seen that the grains are not equiaxial and distribute nonuniformly. Some grains are elongated along rolling direction, and twins exist on matrix. As soon as the specimen is heated to 673 K for 30 min, the grains rapidly grow as a result of static grain growth, and the rolled microstructures transform into coarse-grained structure whose average size is about 30 μm, as shown in Fig. 4(b). When the specimen is imposed to various strains at temperature of 673 K with initial strain rate of $\dot{\varepsilon}$ = $1\times10^{-3}s^{-1}$, the grain size is decreased with increasing strain. For ε = 0.1(> ε_c, critical strain), small amount of fine recrystallized grains appear at the initial grain boundaries, especially at triple junctions. When ε is 0.3, the whole matrix of specimen are taken up by uniform grains indicating that DRX occurs completely. When strain surpasses 0.3, the grains become small and the grain size changes a little with increasing strain, which is in consistant with little change of the flow stress. In addition, second phase $Mg_{17}Al_{12}$ on matrix is uniformly distributed with increasing strain probably indicating dynamic precipitation occurs during hot deformation[8], and its size is not the same at deferent temperatures; whereas the second particles make a different attribution in the efficiency of pining grain boundary at different temperatures[9, 10].

When temperature is decreased, the grain size is decreased, as shown in Fig. 5. This illustrates

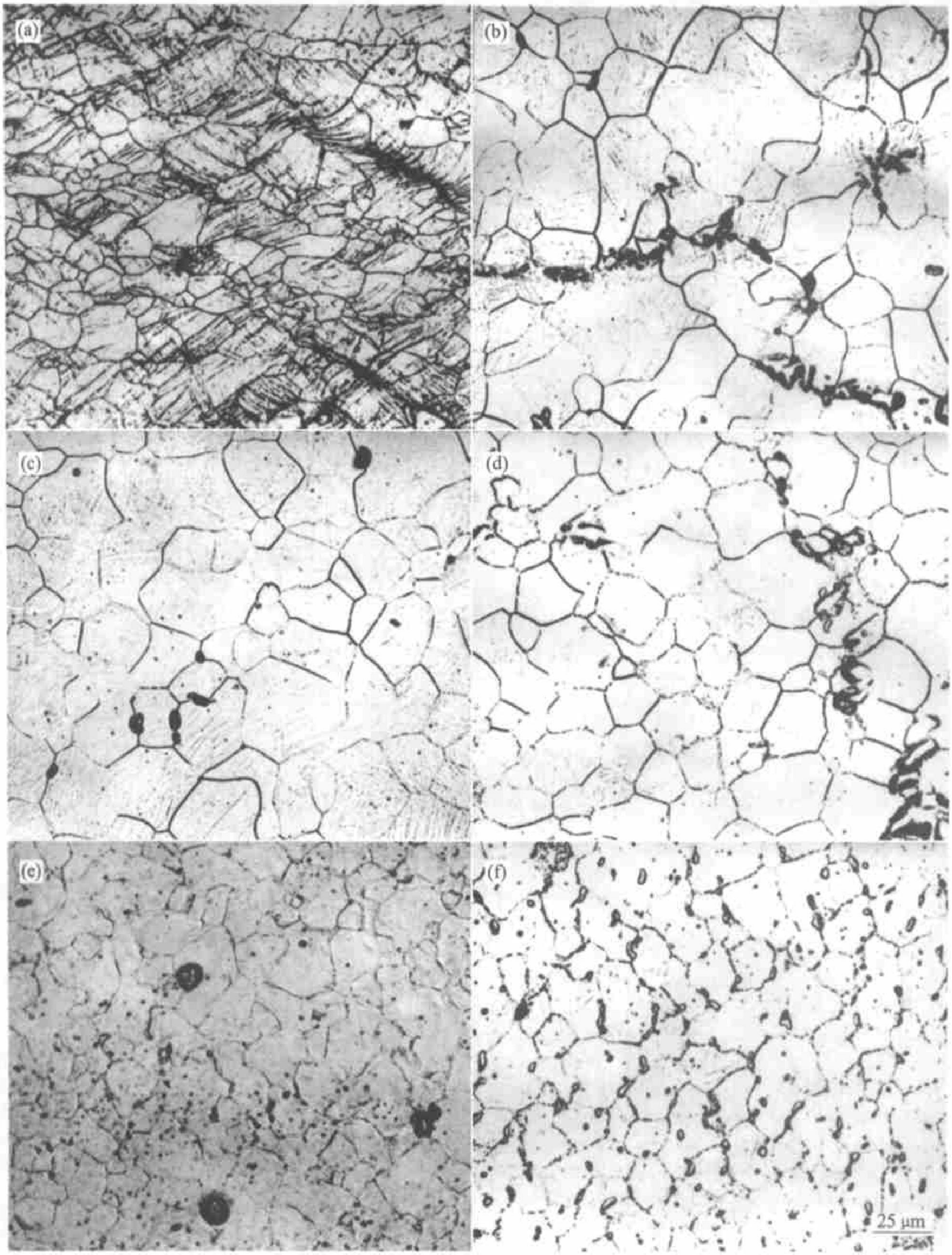

Fig. 4 Optical microstructures of AZ91 alloy at 673 K
(a) —As rolled; (b) —Heated for 30 min; (c) —ε= 0.1; (d) —ε= 0.3; (e) —ε= 0.5; (f) —ε= 1.0

that the recrystallization grain size is controlled by the Zener-Hollomon parameter[8]. Fig. 6 shows the relationship between dynamic recrystallized grain size and Z-parameter in the AZ91 alloy. It is clearly evident that the grain size decreases with increasing Z value.

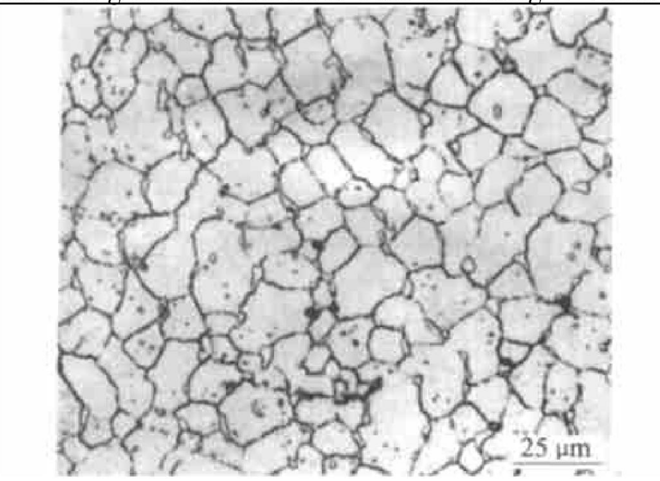

Fig. 5 Optical microstructure at 623 K (ε= 1.3)

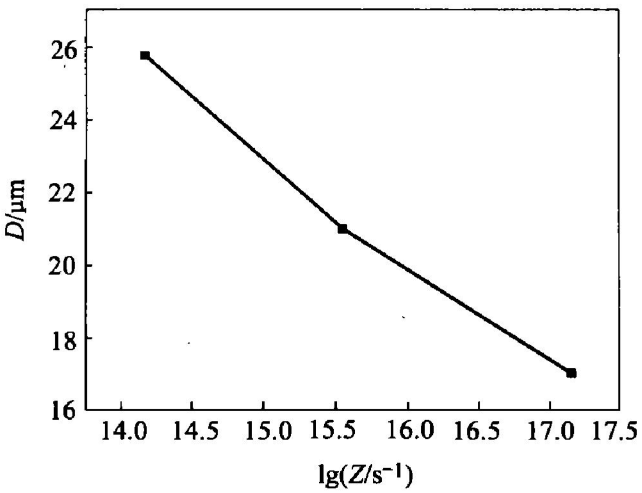

Fig. 6 Relationship between recrystallization grain size and Z-parameter in AZ91

4 DISCUSSION

Strain softening occurs with a feature of DRX in Mg alloys above 573 K[10]. Furthermore, a peak in the flow curve followed by softening to a new steady state regime is indicative of DRX during which strain hardening and softening reach a balance. Fig. 7 shows the relationship between strain hardening rate and strain at various temperatures. The traces of the curves corresponding to higher temperatures show a distinct region where hardening rate remains constant. At low temperature, hardening rate decrease with increasing strain continuously. It is clearly seen that a minimum value of ε is corresponded to the point at which hardening rate begins to remain constant, the storage energy reaches the maximum, and the dissipation rate is at the minimum where DRX begins and approaches to the steady state with strain slowly. However, DRX just starts followed by the fracture of the specimen at 573 K. This illustrates that deformation machanism is different. For instance, twining interrupters DRX (twins are known to increase with lower temperatures and higher strain rate[9]).

At 573 K, basal slip and non-basal slip, twining and cross-slip are mainly deformation machanism. Twins(A), dislocation(B) and substructures(C) reveal within some grains, as shown in Fig. 8. Some slip lines created by cross-slip are found only in near grain boundaries where the stresses are highly concentrated[11]. It is evident that density of twining and dislocations are higher.

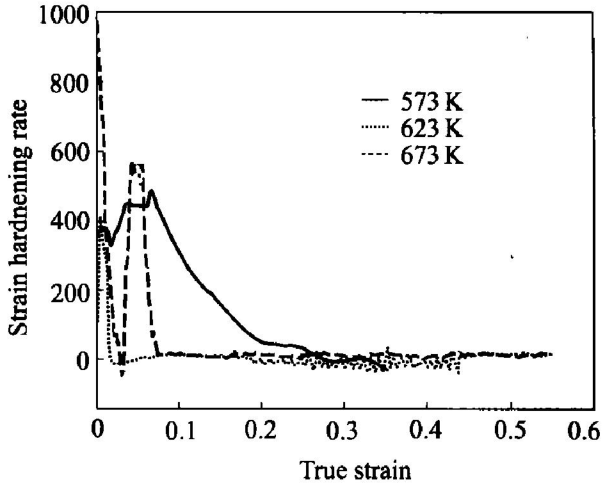

Fig. 7 True strain dependence of strain hardening rate

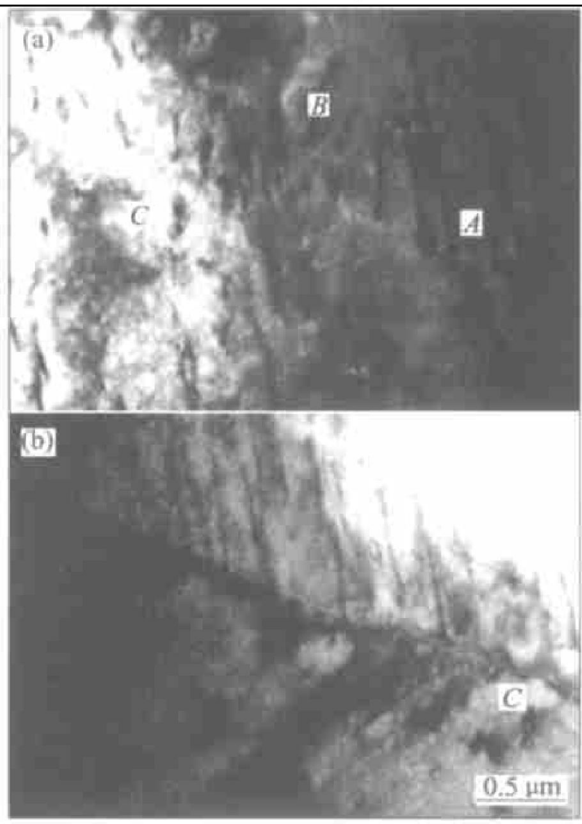

Fig. 8 TEM images of specimen deformed to ε= 0.3 at 573 K

A —Twins; B —Dislocations; C —Substructure

In some other regimes, substructures develop in twin bands. Probably the distorted substructures where twins cross are finally transformed into recrystallized grains[12], and also the twins and dislocations interact each other to form substructures. The result of their interaction may be reflected by the strain-hardening rate, as shown in Fig. 7.

At 673 K, extensive multiple slip or cross-slip and high temperature climb of dislocations are identified as controlling mechanism for strain plastic flow[11]. When strain is little($\varepsilon = 0.1$), some low temperature structures such as sharp twins(*A*) also exist in matrix and develop into a bamboo characteristic as intensive slip polygonizes into disoriented bands, which are likely to develop into substructures during hot deformation, as shown in Fig. 9. It is considered that the formation of substructures is also likely to be aided by additional slip system operation as a result of both higher temperature and more complex stresses near twin intersections. Furthermore, the matrix develops into substructures from both basal dislocation climb and non-basal dislocation activity with straining. The substructure is likely to form at twin intersections. With increasing strain($\varepsilon = 1.0$), these substructures change into DRX grains, and the density of dislocation decreases, as shown in Fig. 9. Therefore, the dislocation cross-slip, dislocation climbs and twin intersections control the rate of plastic deformation.

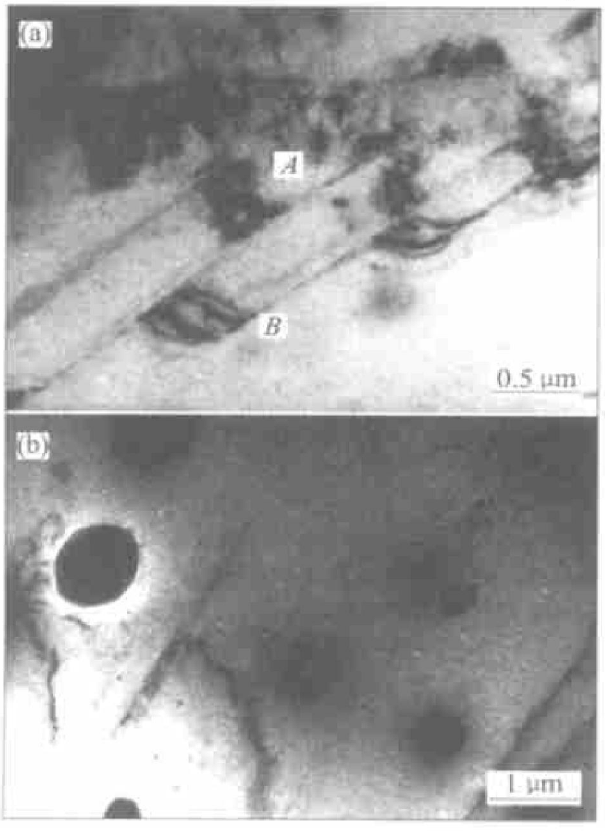

Fig. 9 TEM images of specimen deformed to $\varepsilon = 0.1$(a) and $\varepsilon = 1.0$(b) at 673 K
(*A* Dislocations; *B* —Twins)

5 CONCLUSIONS

1) The hot deformation behavior of AZ91 alloy is interpreted with the activation energy of 96 kJ/mol, and the stress exponent of 3.5.

2) DRX is attributed to grain refinement, and grain size is decreased with increasing *Z*-parameter.

3) At high temperature, dislocation cross-slip, dislocation climbs and twin intersections are probably suitable to the development of substructures. Twins always exist in matrix whether they are at high temperature or at low temperature.

4) Strain induced precipitation of $Mg_{17}Al_{12}$ occurs in the AZ91 alloy and stabilizes subgrain boundaries.

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