Evolution of Fine Grained Microstructure and Superplasticity in Warm-Worked 7075 Aluminum Alloy[†]

Xuyue Yang^{††}, Hiromi Miura and Taku Sakai

Department of Mechanical and Control Engineering, The University of Electro-Communications, Chofu, Tokyo 182, Japan

Hot deformation of warm pre-worked 7075 aluminum alloy was studied by means of 2-step tensile testing at 798 K and at various strain rates from around 10^{-5} to 10^{-1} s^{-1} , and metallographic observations. The samples contained unrecrystallized matrices of high density dislocations tangled at fine precipitates, which existed quite stably at 798 K. Rapid flow softening after a stress peak can result from fine grained structure evolved due to continuous dynamic recrystallization and subsequent gradual flow hardening from grain coarsening occurring at high strains. The average size of new grains can be expressed by a unique function of flow stress during deformation. Strain rate dependence of flow stresses as well as flow curves changes clearly in the three regions of strain rate, $\dot{\epsilon}$; i.e. region I (lower $\dot{\epsilon}$), II (medium $\dot{\epsilon}$) and III (higher $\dot{\epsilon}$). Typical superplasticity taking place in region II leads to the total elongation to fracture of over 750% and the stress exponent of 1.7. It is concluded that hot deformation of 7075 aluminum alloy, affected sensitively by the change of microstructures, can be controlled mainly by grain boundary sliding in region II.

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I. Introduction

The pseudo-single-phase aluminum alloys can be superplastic, when fine equiaxial grains less than $10 \mu\text{m}$ are developed either by static or dynamic recrystallization through appropriate thermomechanical treatment⁽¹⁾⁻⁽³⁾. Under dynamic condition, fine recrystallized grain structures can be developed by a continuous reaction during the early stages of hot deformation for warm or cold-worked materials. This seems not only to be an economical route, but also to impart materials with somewhat superior superplasticity and a higher optimum strain rate⁽¹⁾⁻⁽³⁾. After the work of Watts *et al.*⁽⁴⁾, many investigators have reported that several aluminum alloys can be thermomechanically processed to develop fine grained structures thereby exhibiting superplasticity⁽⁴⁾⁻⁽¹⁸⁾. However, the evolution process of fine grains as well as the characteristics and mechanism of superplasticity has not been fully resolved⁽¹⁾⁻⁽³⁾. Sakai and Nakai⁽¹¹⁾ have found that superplasticity takes place in cold-rolled 7075 aluminum alloy which is previously warm-deformed just before hot deformation. The authors⁽¹⁸⁾ have clarified using the same alloy that the microstructures evolved under warm-deformation to over a critical strain can not be recrystallized for a long period of time (e.g. 10^6 s), leading to the appearance of superplasticity under hot deformation. The dynamic evolution of fine grained structure in unrecrystallized matrix, the details of superplastic be-

haviors and also the mechanisms have been remained unclear due to limited observations.

The present paper is concerned with hot deformation behaviors and microstructural changes of cold-rolled 7075 aluminum alloy, which is pre-deformed under appropriate warm conditions⁽¹⁸⁾. The samples were tested in tension mainly at 798 K and at different strain rates, followed by rapid quenching for microstructural observations. The processes of microstructural changes and their relationship with superplasticity are discussed in detail.

II. Experimental Procedure

The commercial 7075 Al alloy used had the following chemical composition; Zn 5.62, Mg 2.63, Cu 1.58, Cr 0.2, Fe 0.16 and balance Al (all in mass %). The hot-rolled 3 mm thickness plates were homogenized at 763 K for 10.8 ks in a molten salt bath and quenched in water, leading to the evolution of equiaxed grains with an average size of about $17 \mu\text{m}$. Then the plates were cold rolled to 1 mm thickness. The specimens with a gauge section of 12 mm length and 5 mm width were machined parallel to the rolling direction. Tensile tests were conducted in vacuum at the temperatures from 623 K to 798 K using an Instron-type testing machine, which was equipped with a hydrogen gas quenching apparatus. The specimens were deformed to a strain of 40% at 623 K and at $4 \times 10^{-3} \text{ s}^{-1}$ (i.e. this is called prior warm-deformation) and then reheated mainly to 798 K and kept for 1 ks, followed by tensile testing at various strain rates from $2.9 \times 10^{-5} \text{ s}^{-1}$ to $2.1 \times 10^{-1} \text{ s}^{-1}$.

The specimens for microstructural observations were quenched by hydrogen gas at an average cooling rate of

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^{††} Graduate Student, The University of Electro-Communications.

over 2000 K/s immediately after warm or hot deformation is ceased. Microstructural observation was carried out using an optical and transmission electron microscopes. The rolling plane and the longitudinal section were electropolished in a solution of 23% perchloric acid and 77% acetic anhydride and then etched using an Dicks-Keller mixed etchant for optical observation. Grain sizes in three directions, i.e. longitudinal (parallel to the rolling direction), short and long transverse directions, were measured by using linear line intercept method and here designated d_L , d_{ST} and d_{LT} , respectively. The average grain size (d) was obtained by the following equation.

$$d = 3 / (1/d_L + 1/d_{LT} + 1/d_{ST}) \quad (1)$$

A part of the specimens was polished by twinjet method using a solution of 80% ethanol and 20% perchloric acid for transmission electron microscopy (TEM). The thin foils were examined at 200 kV by using a JEM-2000FX microscope.

III. Experimental Results

1. High temperature flow behaviors

A series of true stress-normal strain curves at 798 K is shown in Fig. 1. Flow stresses as well as the shape of flow curves are sensitively dependent on strain rate. The flow curves exhibit a sharp stress peak just after yielding, followed by flow softening and then flow hardening at high strains. The stress drop during flow softening decreases with decreasing strain rate and becomes zero at $2.9 \times 10^{-4} \text{ s}^{-1}$. The flow curves at lower strain rates show only a flow hardening till fracture. The elongations to fracture are over 200% at every strain rate except $2.9 \times 10^{-2} \text{ s}^{-1}$.

The strain rate dependence of yield stress (σ_y), peak stress (σ_p) and flow stress minimum after flow softening

Fig. 1 Typical true stress-normal strain curves at 798 K and at various strain rates from $2.9 \times 10^{-5} \text{ s}^{-1}$ to $2.9 \times 10^{-2} \text{ s}^{-1}$ for warm-worked 7075 Al alloy, which was pre-deformed to a strain of 40% at 598 K and at $4 \times 10^{-3} \text{ s}^{-1}$.

($\sigma_{\min}$), and also of the strains associated with σ_p and $\sigma_{\min}$ (ε_p and $\varepsilon_{\min}$) and the total elongation to fracture (ε_T) is shown in Fig. 2[†]. It is evident in Fig. 2 that the strain rate dependence of σ_y and σ_p is almost the same and changes clearly in the three regions of strain rate, $\dot{\varepsilon}$; i.e. region I (lower $\dot{\varepsilon}$), II (medium $\dot{\varepsilon}$) and III (higher $\dot{\varepsilon}$). The total elongation of over 750% and stress exponent of 1.7 appearing in region II suggest typical superplasticity to take place⁽¹⁾⁻⁽³⁾⁽¹⁹⁾⁽²⁰⁾. In contrast, the strain rate dependence of $\sigma_{\min}$, which is smaller than that of σ_p (or σ_y) in region II, is almost identical in region I and II, and so the relationship between $\sigma_{\min}$ and $\dot{\varepsilon}$ can be approximated by a power function with a slope of 2.4. A stress drop from σ_p to σ_s starts to increase with strain rate increase in region II. On the other hand, the exponent for σ_p is 3 and total elongations rapidly decreases to below 200% in region III. This suggests that typical superplasticity disappears in region III.

Fig. 2 Strain rate dependence of (a) yield stress (σ_y), peak stress (σ_p) and the stress at flow curve dip ($\sigma_{\min}$) and (b) the peak strain (ε_p), the strain associated with $\sigma_{\min}$ ($\varepsilon_{\min}$) and the total elongation to fracture (ε_T) at 798 K for warm-worked 7075 Al alloy.

[†] In the case of no stress peak and no stress dip in flow curves, σ_p and $\sigma_{\min}$ were determined from the extrapolated values of ε_p and $\varepsilon_{\min}$ to lower strain rates.

It should be examined whether hot deformation behavior above mentioned can be controlled by different structural changes taking place in unrecrystallized matrix or different deformation mechanisms operate in each region. Microstructural changes during hot deformation will be described in the following sections.

2. Warm-worked microstructures at 798 K

A TEM microstructure just before hot deformation is depicted in Fig. 3. The sample was deformed to 40% at 623 K and at $4 \times 10^{-3} \text{ s}^{-1}$ and then reheated to and kept at 798 K for 1 ks, followed by H_2 gas quenching. It can be seen in Fig. 3 that the microstructure remains unrecrystallized but recovered, leading to the evolution of equiaxial subgrains with an average diameter of $1.3 \mu\text{m}$. High density dislocations and fine precipitated particles developed in each subgrain can exist stably for a long period of time due to their strong interaction⁽¹⁸⁾ (see Fig. 9(a)). Such warm-worked microstructures as well as their changes taking place during static annealing is described in detail elsewhere⁽¹⁸⁾. One of the results is that the microstructural features as in Fig. 3 as well as the room-temperature hardness for warm-worked matrix were scarcely changed by holding at 798 K until around $3.6 \times 10^4 \text{ s}$.

3. Microstructural changes during hot deformation

Optical microstructures of warm-worked 7075 Al alloy, deformed to various strains at 798 K and at

$2.9 \times 10^{-3} \text{ s}^{-1}$, are displayed in Fig. 4. These microstructures were observed on the rolling plane (L-LT). It can be seen in Fig. 4(b) that a fine grained structure is evolved by a deformation to 40% in unrecrystallized matrix as in Fig. 4(a), although their grain boundaries are still unclear. The grain boundaries become more clear with increase in strain. At a strain of 160% ($\approx \epsilon_{\text{min}}$), where flow stress goes to minimum, the evolution of new fine grained structure with high angle boundaries seems to be almost completed. The average grain size of $6 \mu\text{m}$ gradually increases with further straining to over ϵ_{min} . It is evident in Fig. 4(d) that the grains are coarsened and elongated to the tensile direction and also cavitations are formed at some grain boundaries at a strain of 600%.

Changes in TEM micrographs with deformation at 798 K and at $2.9 \times 10^{-3} \text{ s}^{-1}$, which correspond to the microstructures in Fig. 4, are shown in Fig. 5. It is noted in Fig. 5(a) (and Fig. 3) that the recovered substructures with high density dislocations and fine precipitated particles are scarcely changed by deformation to a strain of 5%, although the subgrain size varied from 1.3 to $2 \mu\text{m}$. With further straining (from Fig. 5(b) to (d)) dislocations in subgrains progressively disappear and fine precipitates are coarsened accompanied with subgrain growth. It seems, however, that the feature of high density dislocations pinned by precipitates can still exist stably at high strains.

To examine the change in misorientation between sub-

Fig. 3 TEM microstructure of warm-worked 7075 Al alloy after annealing at 798 K for 1 ks, followed by H_2 gas quenching.

Fig. 4 Optical microstructures of warm-worked 7075 Al alloy deformed to various strains at 798 K and at $2.9 \times 10^{-3} \text{ s}^{-1}$, followed by H_2 gas quenching. (a) $\varepsilon=0\%$, (b) $\varepsilon=40\%$, (c) $\varepsilon=160\%$, (d) $\varepsilon=600\%$.

Fig. 5 TEM microstructures of warm-worked 7075 Al alloy deformed to various strains at 798 K and at $2.9 \times 10^{-3} \text{ s}^{-1}$, followed by H_2 gas quenching. (a) $\varepsilon=5\%$, (b) $\varepsilon=40\%$, (c) $\varepsilon=160\%$, (d) $\varepsilon=300\%$.

grains with deformation, selected area electron diffraction as well as TEM observation was carried out on the substructures above mentioned. Some of the results are depicted in Fig. 6. It can be seen in Fig. 6(a) that diffraction patterns from the subgrains evolved at a strain of 40% are almost similar, and so their misorientation between neighboring subgrains can be within a few degrees. In contrast, the subgrains evolved at a strain of 160% (Fig. 6(b)) are suggested to have high angle boundaries, because the selected area diffraction patterns are clearly different from each other⁽¹⁰⁾⁽²¹⁾. This result reveals that an initial unrecrystallized matrix is completely converted to a fine grained structure with high angle grain boundaries at around $\epsilon_{\min}$, where flow stress reaches its minimum. It is concluded, therefore, that flow softening occurring after stress peak can be attributable to subgrain growth accompanied with increase in the boundary misorientation and also superplasticity in region II is closely related to such a dynamic evolution of fine grained structures.

The changes in average grain or subgrain size (d) with hot deformation were measured by using linear line intercept method against the above optical and transmission electron micrographs (OM and TEM). The average grain sizes in OM, obtained by substituting 3 kinds of grain sizes (i.e. d_L , d_{LT} and d_{ST}) into eq. (1), and in TEM are plotted against strain in Fig. 7. The grain sizes in TEM increase rapidly just after yielding and then gradually with increasing strain, and finally their growth rate become constant at high strains. In contrast, the grain sizes in OM decrease rapidly and approach to a minimum at a strain of about 100%, and then increase conversely with increasing strain. It is interesting to note that the rate of grain growth in OM is almost the same as that in TEM at

high strains. It is reasonable to consider that the difference between them i.e. about $1\ \mu\text{m}$, may arise from the different methods of OM and TEM, and so both grain sizes can be almost the same within experimental scatter⁽²²⁾, because the fine grains evolved at about 160% have high angle grain boundaries (see Fig. 6(b)).

It is concluded from the above analysis that subgrains evolved in warm-worked matrix are grown accompanied with increase in grain boundary misorientation during flow softening, leading to the complete evolution of a new fine grained structure with high angle boundaries at around $\epsilon_{\min}$ ($=160\%$). On the other hand, the new grains start to grow at high strains of over $\epsilon_{\min}$, where flow hardening takes place. It is quite reasonable that flow softening and subsequent hardening can result from grain refinement and coarsening, respectively.

IV. Discussion

1. Microstructural evolution during hot deformation

It has been clarified elsewhere⁽¹⁸⁾ that, when a cold-rolled 7075 aluminum alloy is warm-worked under appropriate conditions, the substructures containing high density dislocations and η -phase precipitates are homogeneously developed in heterogeneous cold-rolling matrices, and so static recovery and recrystallization taking place can be extensively retarded because of the strong interaction between dislocations and precipitates. It is evident in Fig. 5, however, that such stable matrices start to change just after hot deformation is applied. The subgrain sizes measured by TEM (Fig. 7) are replotted against deformation time in Fig. 8, where the ones stati-

Fig. 6 Changes in TEM microstructures and the corresponding selected area diffraction patterns with straining at 798 K and at $2.9 \times 10^{-3}\ \text{s}^{-1}$ for warm-worked 7075 Al alloy. (a) $\epsilon=40\%$, (b) $\epsilon=160\%$.

Fig. 7 Changes in average grain or subgrain size (d) with deformation at 798 K and at $2.9 \times 10^{-3} \text{ s}^{-1}$. d was measured by using linear line intercept method on optical and transmission electron microscopic (OM and TEM) structures.

Fig. 8 Changes in grain or subgrain size for warm-worked 7075 Al alloy with deformation time at a strain rate of $2.9 \times 10^{-3} \text{ s}^{-1}$ or with statically holding time at 798 K.

cally evolved in warm-worked matrix at 798 K are plotted against annealing time for comparison⁽¹⁸⁾. It is evident in Fig. 8 that subgrains do not change statically so much before or even after recrystallization takes place, while they rapidly grow under hot deformation. It is proved in Fig. 6 that such subgrain growth accompanied with increase in boundary misorientation leads to the evolution of fine grained structure with high angle boundaries at high strains of over $\epsilon_{\min}$.

The above phenomenon is termed continuous dynamic recrystallization (DRX), the details of which have not been fully resolved⁽¹⁾⁻⁽³⁾⁽¹⁴⁾. Recently some works on superplasticity accompanied by fine grain evolution have been carried out for several pseudo single phase aluminum alloys including several Zr bearing alloys⁽⁴⁾⁻⁽¹¹⁾⁽¹⁴⁾⁽²¹⁾⁽²³⁾⁽²⁴⁾. Nes⁽⁶⁾ explained that subgrain boundary sliding causes

the boundary misorientations to increase and subgrain growth takes place due to the dissolution of Al_3Zr , leading to the occurrence of continuous DRX. Recently this type of superplasticity has been studied using Al-Mg alloys by Hales and McNelly⁽²¹⁾ and Al-Li-Cu alloys by Liu *et al.*⁽¹⁶⁾. They explained that dislocations produced by hot deformation can be absorbed into subboundaries, resulting in an increase in their misorientations. The mechanism proposed by them can possibly operate during dynamic evolution of fine grains in the present material. Namely, high density dislocations pinned by precipitates become movable due to their dissolution and are absorbed into subboundaries and also subgrain rotation can take place due to grain boundary sliding. These lead to the increase in grain boundary misorientations and finally an evolution of fine-grained structures. This type of evolution process of new grains is clearly different from that of discontinuous DRX, in which nucleation and growth of new grains take place concurrently and so average grain sizes become constant during deformation⁽²⁵⁾⁽²⁶⁾. In the present case, fine grains start to be coarsened gradually due to further dissolution of precipitates just after the completion of new grained structure, leading to an appearance of flow hardening at high strains. These will be discussed in more detail in the next section.

The higher magnification TEM microstructures of Figs. 3 and 5(b) are shown in Fig. 9. It can be seen clearly

Fig. 9 Typical TEM microstructures of warm-worked 7075 Al alloy deformed at 798 K and at $2.9 \times 10^{-3} \text{ s}^{-1}$ to a strain of (a) 0% or (b) 300%, followed by H_2 gas quenching.

in Fig. 9 that some of high density dislocations tangled at precipitated particles are annihilated and the particles are significantly coarsened in near grain boundaries during hot deformation. It is also noted that dislocations released from precipitates seem to move through them and pile up at grain boundaries. There is generally only a few dislocations developed during superplasticity, because grain boundary sliding mainly operate⁽¹⁰⁾⁽¹⁴⁾. It is suggested from the present observations, therefore, that dislocation motion as well as grain boundary sliding takes place frequently during superplasticity of the present 7075 aluminum alloy and so both the mechanisms can operate in connection with each other.

2. Superplasticity and continuous DRX

When hot deformation can be controlled by grain boundary sliding, the deformation equation is given as follows⁽¹⁾⁻⁽³⁾,

$$\dot{\epsilon} = K_1 d^{-p} \sigma^n \exp(-Q/RT) \quad (2)$$

where K_1 , p , n and Q are experimental constants and the others have usual meanings. The strain rate dependence of peak stress (σ_p) is almost the same as that of yield stress (σ_y) (Fig. 2), and so σ_p is considered to be controlled by an initial microstructure before hot deformation. The strain rate dependence of σ_p changes clearly in the three regions of strain rate, i.e. region I (lower $\dot{\epsilon}$), II (medium $\dot{\epsilon}$) and III (higher $\dot{\epsilon}$). This is similar to the results of typical superplasticity previously reported⁽¹⁾⁻⁽³⁾⁽¹⁹⁾. In region II a maximum elongation of over 750% appears and the stress exponent is 1.7, which is almost the same as those in previous results (e.g. $n = 1.4 \sim 2$)⁽¹⁾⁻⁽³⁾⁽²⁷⁾. In contrast, the strain rate dependence of the stress minimum (σ_{min}) is almost the same in region I and II and so can be approximated by a power function with n of 2.4. σ_{min} is associated with ϵ_{min} where a fine grained structure is formed completely, and so its strain rate dependence

can be controlled more frequently by grain boundary sliding. It is something strange that the stress exponent for σ_{min} ($=2.4$) is larger than that for σ_p ($=1.7$) associated with unrecrystallized matrix. The reason will be discussed as follows.

The changes in grain structures developed at strains of ϵ_{min} and 300% with strain rates are shown in Fig. 10. It can be seen that the grain boundaries are delineated more clearly and the average grain sizes increase with increasing strain or decreasing strain rate. The relationship between the stress at ϵ_{min} (σ_{min}) and the dynamic grain size (d) evolved at ϵ_{min} at various temperatures and strain rates is displayed in Fig. 11. The d is expressed by a unique function of σ_{min} irrespective of temperature and strain rate used and so its relationship can be approximated by the following equation,

Fig. 11 Relationship between dynamic grain size (d) evolved at around ϵ_{min} and the stress at ϵ_{min} (σ_{min}) for warm-worked 7075 Al alloy at various temperatures and strain rates.

Fig. 10 Optical microstructures for warm-worked 7075 Al alloy. The samples were deformed at 798 K and at various strain rates to [(a) to (c)] the strains associated with σ_{min} (ϵ_{min}) and [(d) to (f)] a strain of 300%, followed by H_2 gas quenching. (a) and (d) $2.9 \times 10^{-3} s^{-1}$, (b) and (e) $2.9 \times 10^{-4} s^{-1}$, and (c) and (f) $2.9 \times 10^{-5} s^{-1}$.

$$\sigma_{\min} = K_2 d^{-N} \quad (3)$$

where K_2 and N are constants. $N=5$ is derived from Fig. 11. For many discontinuous DRX materials, DRX grain sizes evolved at high strains are uniquely related to flow stress and their relationships are also represented by eq. (3), although N ranges between 0.75 and 1⁽²⁵⁾⁽²⁶⁾. The high value of N in the present alloy is partly due to the difference of continuous and discontinuous DRX, and also grain growth is effectively retarded by the pinning effect of precipitated particles⁽²⁸⁾.

By the way, as the strain rate dependence of σ_p and $\sigma_{\min}$ can be controlled mainly by grain boundary sliding, the following equations at a temperature are obtained from eq. (2).

$$\dot{\epsilon} = K_3 \sigma_p^n \quad (4)$$

$$\dot{\epsilon} = K_4 \sigma_{\min}^{n'} \quad (5)$$

where K_3 and K_4 are constants. n and n' are 1.7 and 2.4 in Fig. 2. It is evident from the discussions previously argued that σ_p and n are controlled by an initial microstructure before deformation and, on the other hand, $\sigma_{\min}$ and n' by new fine grained structure evolved at $\epsilon_{\min}$. Therefore, n' for $\sigma_{\min}$ can be expressed by the following equation which is derived from eqs. (2) and (3).

$$n' = n + p/N \quad (6)$$

If $n=1.7$, $N=5$ and $p=2\sim3$ in the previous reports⁽¹⁾⁻⁽³⁾ are given in eq. (6), n' takes a value of 2.1~2.3. This is nearly equal to the experimental value of 2.4 for $\sigma_{\min}$. It is concluded, therefore, that n' for $\sigma_{\min}$ becomes larger due to the variation of grain size accompanied by continuous DRX and the strain rate or temperature dependence of σ_p and $\sigma_{\min}$ can be expressed essentially by the same type of deformation equation.

On the other hand, if T and $\dot{\epsilon}$ in eq. (2) are constant, σ is expressed as follows,

$$\sigma = K_5 d^{p/n} \quad (7)$$

where K_5 is a constant. It is understood from eq. (7) that flow stress should decrease accompanied with grain refinement taking place due to continuous DRX. The period of flow softening at $2.9 \times 10^{-3} \text{ s}^{-1}$ corresponds roughly to that of grain refinement measured by OM (see Figs. 1 and 7). Flow hardening accompanied by grain coarsening taking place at high strains also can be explained by using eq. (7). In contrast, σ - ϵ curves appearing in region I show only flow hardening during deformation. The grain size evolved at $\epsilon_{\min}$ is about $9 \mu\text{m}$ at $2.9 \times 10^{-5} \text{ s}^{-1}$ (see Fig. 11), which is already larger than the initial size of $8 \mu\text{m}$. Namely, flow hardening associated with grain coarsening in region I is understood by using eq. (7). It is concluded, therefore, that deformation in region I also can be controlled by grain boundary sliding.

V. Conclusions

The superplastic behavior and fine grained structure

evolved in unrecrystallized matrices of cold- and warm-worked 7075 aluminum alloy was investigated by means of hot tensile tests and microstructural observations. The main results obtained can be summarized as follows.

(1) The strain rate dependence of flow stresses can be changed clearly in the three regions of strain rate. A maximum elongation of over 750% and a stress exponent of 1.7 appearing in region II suggest typical superplasticity to take place.

(2) The strain rate dependence of peak stress (σ_p) and yield stress (σ_y) for an initial unrecrystallized matrix is almost the same, but clearly larger than that of flow stress minimum ($\sigma_{\min}$) in region II, where new fine-grained structures and developed.

(3) The unrecrystallized structures containing high density dislocations tangled at fine precipitates are changed to fine grained structures with high angle boundaries; i.e. continuous dynamic recrystallization takes place assisted by grain boundary sliding. The relationship between flow stress and dynamic grain size is expressed by a following equation, which does not depend on temperature or strain rate.

$$\sigma = Kd^{-5}$$

(4) Flow softening taking place after a stress peak and subsequent flow hardening at high strains can result from grain refinement and coarsening, respectively.

(5) The deformation equation for $\sigma_{\min}$ is essentially the same as those for σ_p and σ_y , although it is apparently changed by the variation of grain size evolved by continuous dynamic recrystallization. It is concluded that deformation in region I and II can be controlled mainly by grain boundary sliding.

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