Dynamic Recovery and Static Recrystallization of 1.8 % Al Steel in Hot Deformation*

By Chiaki OUCHI** and Tomoyoshi OKITA**

Synopsis

Dynamic recovery and the following static recrystallization process of 1.8 % Al steel were investigated by a hot compression machine equipped with a precise control system for quenching after hot deformation. The temperatures and strain rates of deformations were varied in the ranges from 800 to 1100 °C, and from $5 \times 10^{-4}$ to 10.3 $s^{-1}$, respectively. The subgrain structures developed by dynamic recovery process were observed by interference microscope and transmission electron microscope, and these were correlated with steady state flow stress. The substructural changes in static recovery and recrystallization after dynamic recovery were investigated to understand the nucleation mechanism of static recrystallization after dynamic recovery. Subgrain coarsening at a high rate took place in the recovery stage, and subgrain coalescence along the grain boundary appeared to be the nucleation mechanism of recrystallization. The effects of hot deformation conditions on static recrystallization kinetics were also studied and analyzed based on grain boundary migration rate.

I. Introduction

The microstructures developed during hot working are influenced primarily by processes of dynamic and static restoration. A lot of basic research of recovery and recrystallization of steel have been carried out on the cold working and annealing process where the effect of dynamic recovery was negligible. On the other hand, the microstructural change after hot working is always associated with the progress of dynamic recovery or recrystallization in a hot-deformed state, and therefore many investigations of dynamic recovery process have been conducted on various metals and alloys during the last two decades. These studies resulted in classification of metals into dynamic recovery type and dynamic recrystallization type. Past investigations on iron and steels or iron-base alloys have shown that the austenite, including steels deformed in the austenitic region as well as the austenitic steels, exhibited basically dynamic recrystallization, and that most ferrite behaved as dynamic recovery type.1-7)

To understand a true microstructural nature of dynamic recovery or recrystallization, it is quite important to quench the specimens after deformation rapidly enough to suppress the progress of static recovery and recrystallization, the kinetics of which generally accelerate with the increase of strain rate and strain of deformation. Most studies of dynamic restoration processes have been carried out with a delay time between the finish of deformation and quenching of less than one to a few seconds. This delay time may be too long in some cases for complete elimination of the static effect. The present authors developed the hot compression machine equipped with a more precise control system for quenching after hot deformation, and reported the microstructural features of dynamic recrystallization of austenite in HSLA steels and 18-8 stainless steel.7) The objectives of the present studies, which were carried out as part of a series of investigations of dynamic restoration process in steels, were to grasp the microstructural nature of dynamic recovery and to investigate its influence on static recovery and recrystallization. To meet these objectives ferritic steels such as Fe-Si, Fe-Cr or Fe-Mo must be selected. A 1.8 % Al steel, of which the dynamic and static restoration process had been rarely investigated, was used as one of those ferritic steels expected to cause the dynamic recovery behavior. The hot deformation conditions were varied similarly to those in the previous studies of austenite,7) covering strain rates from $5 \times 10^{-4}$ to 10.3 $s^{-1}$ and deformation temperatures from 800 to 1100 °C. The interference microscope technique was specially used to observe the substructures developed by the dynamic recovery process and particular focus was put on observation of the substructural changes in static recovery and recrystallization after dynamic recovery. The microstructural natures, and dynamic and static restoration characteristics of 1.8 % Al steel are discussed in comparison with those in austenite of the HSLA steels and 18-8 stainless steel or in ferrite of pure iron.

II. Experimental

The chemical composition of a 1.8 % Al steel is shown in Table 1. The steel was melted in a 50 kg vacuum induction heating furnace. Slabs of 80 mm thickness were hot rolled to 20 mm thickness with the finish-rolling temperature of 1100 °C. Ferrite grain size in this plate was 750 μm. A part of the slabs was hot rolled at a finish-rolling temperature of 950 °C that brought about the finer grain size of 190 μm. The cylindrical specimens with diameter of 8 mm and length of 12 mm were machined from the mid thickness of the plate. The hot compression testing equipment used in this study was described in detail in a previous paper.7) This equipment has a high frequency induction coil in vacuum; strain rate

<table>
<thead>
<tr>
<th>C</th>
<th>Si</th>
<th>Mn</th>
<th>P</th>
<th>S</th>
<th>sol. Al</th>
<th>T. N</th>
</tr>
</thead>
<tbody>
<tr>
<td>0.014</td>
<td>0.10</td>
<td>0.01</td>
<td>0.008</td>
<td>0.005</td>
<td>1.81</td>
<td>0.0032</td>
</tr>
</tbody>
</table>

* Presented to the 93rd ISIJ Meeting, April 1977, at The University of Tokyo in Tokyo. Manuscript received December 25, 1981.
© 1983 ISIJ
** Steel Products Laboratory, Technical Research Center, Nippon Kokan K.K., Minamiwatarida-cho, Kawasaki-ku, Kawasaki 210.
can be varied from $5 \times 10^{-4}$ to $10 \, \text{s}^{-1}$. The very elaborate cooling system in this machine enables to quench the specimen by He gas with a delay time of less than 0.1 s after deformation and with cooling rate of around 500 °C/s.

The reheating temperatures for hot compression testing were selected low enough to avoid coarsening of the initial grain size. The specimen with grain size of 750 μm was reheated at 1 100 °C for 5 min and then deformed in the temperature range from 1 100 to 800 °C. The specimen with grain size of 190 μm was reheated at 900 °C for 5 min, and then deformed at 900 °C. The deformation was performed at the various strain rates as described above, and the compressive strain was 0.7, if not otherwise stated. The true stress-strain (S-S) curves were displayed on a plotter through processing by an in-line microcomputer. For investigation of the microstructures developed in the dynamic restoration process, the deformed specimens were quenched immediately after deformation under the condition described above. The static recovery and recrystallization were studied by holding the deformed specimens for the various lengths of time at the deformation temperature, followed by He gas quenching. In this experiment the steel with initial grain size of 750 μm was used. The cooling system in the present equipment enabled to investigate the progress of static recovery and recrystallization for quite short holding times, for example much less than one second.

The deformed and quenched specimens were cut into half parallel with the direction of deformation. The microstructures of the center part of specimens were observed by an optical microscope. Interference microscope was also used; this was particularly useful to observe the substructure in the scale of optical microscope. Microstructure was revealed by the etchant of 3 % nital. A transmission electron microscope was also used to observe the finer substructure of the deformed specimens. The thin foil was prepared by jet polisher, using the solution of 5 % perchloric and 95 % acetic acid. The recrystallized fraction in the study of static recrystallization kinetics after hot deformation was obtained by point counting approximately 200 points under optical microscope.

### III. Results

1. The Effect of Temperature and Strain Rate on Dynamic Recovery

The examples of the S-S curves for the various combinations of temperature and strain rate are shown in Figs. 1 and 2. The S-S curve behaviors of 1.8 % Al steel were always the same in all the deformation conditions investigated here. That is, work hardening was observed in the early stage of strain, then followed by the steady state flow behavior at higher strain. This type of S-S curve has been commonly observed in dynamic recovery type of metals such as Al, ferritic iron and ferritic alloy steels. The increase of strain rate or the decrease of deformation temperature increased the steady state flow stress ($\sigma_s$) and the start strain of steady flow ($\varepsilon_c$). As shown in Fig. 3, a linear correlation between $\sigma_s$ and $\varepsilon_c$ was obtained. Figure 2 includes the S-S curves for steels with different initial grain size. The change of grain size from 750 to 190 μm resulted in a very small variation in the whole S-S curve behavior. This contrasts with dynamic recrystallization behavior of austenite,\(^6\) where a change in the grain size caused a relatively large difference in flow stress and S-S curve behavior.

When the hot deformation process is controlled by a thermally activated process, the relation among

![Fig. 1. True stress-strain curves of 1.8 % Al steel.](image)

![Fig. 2. True stress-strain curves of 1.8 % Al steel.](image)

![Fig. 3. Relation of steady-state flow stress and start strain of steady-state flow.](image)
temperature, strain rate and flow stress is expressed by:

\[ \dot{\varepsilon} = A \sigma^n \exp \left( -\frac{Q}{RT} \right) \] ...........................(1)

or \[ \dot{\varepsilon} = A' \sinh \left( \alpha \sigma_s \right)^n \exp \left( -\frac{Q}{RT} \right) \] ..........(2)

where, \( A, A', \alpha, m, n \): constants
\( Q \): an activation energy
\( R \): the gas constant
\( T \): the absolute temperature.

The values of \( \alpha \) and \( Q \) obtained at the steady state flow stress condition in 1.8 % Al steel were 0.15 (kgf/mm²)⁻¹ and 76 kcal/mol, respectively. This value of \( Q \) was lower than those of austenite (around 96 kcal/mol) in HSLA steels and 18-8 stainless steel investigated under the same deformation conditions with those in the present study. The Zener–Hollomon parameter, \( Z = \exp \left( \frac{Q}{RT} \right) \), is shown against \( \sinh \left( \alpha \sigma_s \right) \) in Fig. 4. A correlation of Eq. (2) held in a wide range of \( Z \) value from \( 4 \times 10^9 \) to \( 7 \times 10^{14} \) s⁻¹, and the relevant \( n \) value was obtained as 4.3 from this figure.

For the case of Eq. (1), a linear relation between \( \ln \sigma \) and \( \ln Z \) was observed in the stress range below 7 kgf/mm². The corresponding \( m \) value was 4.6 in the range of \( \sigma \) from 0.8 to 7 kgf/mm². The linear correlation between \( \ln \sigma \) and \( \ln Z \) deviated upwards with the increase of \( \sigma \) in the stress region above 7 kgf/mm². The \( Q \) value obtained from Eq. (1) is similar to the value of 2.8 % Si iron (80 kcal/mol) obtained by Uvira and Jonas.

The examples of the microstructures in the specimens quenched immediately after hot deformation are shown in Photos. 1 and 2, which were observed by the conventional optical microscope and interference microscope, respectively. It is clear from comparison of the photomicrographs that the subgrain boundaries as well as the original grain boundaries are more definitely observed by an interference microscope. Photograph 1 includes the microstructures with the different initial grain sizes. The subgrain size is fairly large because of the deformation condition with low \( Z \) value, but the shape and size of the subgrains were almost the same for both initial grain sizes. That is, the microstructure of dynamic recovery is influenced very little by the initial grain size. Photograph 2 shows the microstructural changes with the deformation conditions, covering the \( Z \) value from \( 2.2 \times 10^{10} \) to \( 1.2 \times 10^{14} \) s⁻¹. For the deformation with high strain rate, the original grain boundaries which were deformed in a pancake shape, were clearly distinguished, but these became more scattering and less clear with the decrease of strain rate. The subgrains which were observed in Photos. 2 (b) to (d) coarsened continuously with the decrease of \( Z \) value, and this was accompanied with an increase of equi-axiality of subgrains. The substructure developed by hot deformation with high \( Z \) value as shown in Photo. 2 (a) was too fine to be observed by the optical microscope. Therefore, observation by transmission electron microscope was conducted on the specimen deformed at high strain rate or at lower temperatures. These examples are shown in Photo. 3, in which photographs, (a) and (b), revealed the substructure of microstructure shown in Photo. 2 (a). It is evident that well-defined subgrains with high equi-axiality are still formed. Hexagonal network dislocations as well as straight and parallel dislocations were observed in the subgrain matrix (Photo. 3 (b)).
Photo. 2. Changes of dynamically recovered microstructures with $Z$ values. Interference microscope.

Photo. 3. Transmission electron micrographs of dynamically recovered microstructures. Arrow mark indicates dislocation network.
and some of them appeared to be just on the stage of subboundary formation or annihilation of dislocations. All the substructural features developed by the dynamic recovery process are brought about from concurrent process of formation and annihilation of dislocations, and the former process becomes more predominant than the latter with the increase of Z value. These examples are shown in Photos. 3 (c) and (d). The increase of the Z value increased the dislocation density continuously, decreased subgrain size, and tended to decrease the equi-axiality of subgrains. These changes caused subgrain formation to approach to more like cell-wall formation as seen in Photo. 3 (d).

Optical measurement of the subgrain size ($d_\varepsilon$) in the dynamic recovery structure was carried out, and thin foil observation by electron microscope was also used for the finer subgrain size less than 3 μm. The results were shown in Fig. 5, where the relation between $d_\varepsilon$ and $\sigma_s$ was expressed by

$$\sigma_s = K_s \cdot d_\varepsilon^N$$

with $K_s = 17.3$ and $N = 0.74$ ...

(3)

It is found that the subgrain size is controlled primarily by flow stress ($\sigma_s$) determined by the deformation condition, but not by the initial grain size. This is consistent with the result that the initial grain size influenced flow stress very little. From Fig. 5, the $N$ value of 0.74 was obtained. This value was relatively small as compared with the values obtained for dynamic recovery of other metals such as pure iron or Al, where the $N$ values of around 1.0 to 1.2 were obtained.

All results described above were obtained under a given compressive strain of 0.70. The progress of dynamic recovery with strain was investigated by the interruption of deformation at strains below 0.70. Figure 6 shows the changes of subgrain size with strain, and the arrow marks in this figure indicate the start strain of steady state flow behavior. The well defined subgrain was formed in the early stage of strain just below $\varepsilon_0$ and the subgrain size did not change with the increase of strain, although it depended on flow stress in each deformation condition. This result indicates that balancing the rates of generation and annihilation of dislocations during deformation gives rise to the substructure with a given subgrain size, as well as a constant flow stress in the steady state flow region. The grain boundary bulging which was observed in the HSLA steels and 18-8 stainless steel was never observed in the present steel.

2. Static Recovery and Recrystallization after Dynamic Recovery

The effects of the deformed microstructure and the temperature on the static restoration process were investigated. Specimens with the coarse initial grain size of 750 μm were deformed to a strain of 0.70, isothermally held for various lengths of time at the deformation temperature, and then quenched. Figure 7 shows the progress of static recrystallization with holding time at 1000 °C after deformation at three different strain rates. The increase of strain rate evidently accelerates recrystallization.

This effect of strain rate has been observed in other studies of hot deformation. The stored energy in the deformed material is the driving force of static recovery and recrystallization, and this is primarily determined by dislocation density in the deformed structure. For the case of hot deformation, as described above, the substructure of dynamic recovery was widely varied by strain rate under given strain and temperature, and variation of substructures was related to flow stress rather than strain. That is, the prominent effect of strain rate on static recrystallization kinetics, shown in Fig. 7, is brought about for the different substructures developed by dynamic recovery.
Figure 8 shows the effects of the temperatures on static recrystallization for two cases. For the case of Fig. 8 (a), the deformation temperatures which were same with the holding temperatures after deformation were varied from 1 050 to 900 °C under a fixed strain rate of 10.3 s⁻¹. This change of the deformation temperatures resulted in variation of flow stress from 6.3 to 11.6 kgf/mm². This means that higher stored energy is involved in the deformed state with the decrease of the deformation temperatures. Therefore, retardation effect of static recrystallization with the reduction of temperatures shown in Fig. 8 (a) is evaluated in a relatively smaller degree, compared with the case of the fixed initial deformed microstructure. On the other hand, Fig. 8 (b) shows the temperature dependence of recrystallization kinetics obtained under a constant deformed state. That is, the strain rates at 1 000 and 950 °C were adjusted to attain a given flow stress of 6.3 kgf/mm², which was the flow stress obtained at 1 050 °C with a strain rate of 10.3 s⁻¹. The strain of 0.7 was kept constant in all the cases. The decrease of the temperature from 1 050 to 950 °C in this case obviously caused a larger extent of retardation of recrystallization compared with the result of Fig. 8 (a).

Examples of the microstructural changes with static recovery and recrystallization are exhibited in Photos. 4 and 5. Both were taken by interference microscope, particularly focusing on the stage from recovery to onset of recrystallization. A series of photographs in Photo. 4 are the cases of holding time of 0.1, 1 and 5 s at 1 000 °C after a strain of 0.7 at 1 000 °C with a strain rate of 5.0 × 10⁻² s⁻¹. It is clearly seen by an interferential contrast that the progress of static recovery caused a very rough appearance of substructures and this seems to be brought about by subgrain coarsening in some areas of the substructure (Photos. 4 (a) and (b)). Photograph 4 (b) seems to be a microstructure just before formation of recrystallization nuclei. The area consisting of coarse subgrains appears to change to a recrystallized new grain during further extended holding (Photo. 4 (c)). The new grain observed there has a very smooth interface and contains no substructure inside the grain, while the surrounding of a new grain still keeps the well-defined subgrain structure.
Photograph 5 shows the progress of static recrystallization under the same conditions as the case of Photo. 4, except that the strain rate in hot deformation was raised to 10.3 s⁻¹. Holding times at 1000 °C are (a) 0.04 s, (b) 0.1 s and (c) 1 s.

Photograph 5 shows the progress of static recrystallization under the same conditions as the case of Photo. 4, except that the strain rate in hot deformation was raised to 10.3 s⁻¹. Holding times at 1000 °C are (a) 0.04 s, (b) 0.1 s and (c) 1 s. Most of the new grains straddled the boundaries. The longer holding time up to one second increased the fraction of recrystallization to around 30 % (Ref. Fig. 7), and this was caused by increases of the numbers of recrystallized grains and also by growth of each new grain.

The progress of recovery and recrystallization was always accompanied with the increase of the average subgrain size in substructures; an evident example is seen in Photo. 5 (c). The change of the average subgrain size with holding time was quantitatively measured as shown in Fig. 9. The arrow marks shown in this figure indicate the holding time when the first recrystallized grain was formed (Ref. Fig. 7). The subgrain size with holding time continuously increased even after the first new grain was formed, and then it tended to reach a given subgrain size determined by the deformation conditions and the holding temperatures.

**IV. Discussion**

The stress–strain curve behavior and the microstructural features observed in the specimens quenched immediately after deformation certainly indicated that the present steel behaved as a typical dynamic recovery type of metal. The well-defined subgrain structures formed by most deformation conditions investigated here were very similar to the dynamically recovered microstructures in Al or its alloys investigated by McQueen and Hockett. In a previous report by the present authors, the substructures of dynamic recrystallization of austenite were investigated under the same range of hot deformation conditions as in the present study. In this case the well-defined subgrain structure was not observed. From these comparisons, the stacking fault energy of 1.8 % Al steel is presumed to be as high as that of Al, and this may facilitate cross slip or climb of dislocations to form the well-defined subgrain structure.

This high stacking fault energy of the present steel, at least compared with that of austenite in HSLA steel or 18-8 stainless steel, may be associated with its b.c.c. structure, because most of the ferritic steels or alloys investigated by others also exhibited dynamic recovery behavior. Only one exception for this was high purity iron studied by Glover and Sellars, which was reported to give rise to dynamic recrystallization in the ferritic region. Although their conclusion was based on both the S–S curve behavior.
and the microstructural observation, it should be noted that the microstructure demonstrated as a dynamically recrystallized grain in the high purity iron was often very similar to the microstructure developed by static recovery during a very short holding time after deformation in the present steel. Photograph 4 (b) is one of these examples, where the boundary between the region of the preferentially coarsened subgrain and its surrounding substructure appears to correspond to the dynamically recrystallized grain with evidence of migration of the boundary pointed out by Glover and Sellars. If this correspondence is true, occurrence of dynamic recrystallization in high purity iron appears to be still questionable, as the dynamic restoration process of metals cannot be classified based on only the S-S curve behavior.

Several mechanisms of nucleation of static recrystallization have been proposed such as subgrain coalescence, boundary migration, or bulge nucleation. Bulge nucleation was observed in both dynamic and static recrystallization in austenite, but it was not observed in static recrystallization after dynamic recovery in 1.8 % Al steel. Kozasu and Shimizu investigated precisely the microstructural changes from recovery to recrystallization after hot rolling in 18-8 stainless steel and found very slow progress of static recovery. On the contrary, the present steel exhibited continuous growth of subgrain in recovery stage. This difference of static recovery kinetics in these steels seems to be also due to a large difference of stacking fault energy between the two steels. Although it was not confirmed whether the subgrain growth took place by the mechanism of coalescence or boundary migration, the overall picture observed in the nucleation stage here was very similar to the nucleation model proposed recently by Jones et al., where subgrain coalescence along the grain boundary was proposed.

The activation energy of static recrystallization was evaluated based on the method by English and Backofen. The average migration rate of recrystallizing front, \( G \) is expressed by the following equation,

\[
G \cdot A = \frac{dX_v}{dt} \quad \text{.................(4)}
\]

with

\[
A = 2 \cdot m, \quad \text{.................(5)}
\]

where, \( A \): the total interfacial area per unit volume of recrystallizing front

\( X_v \): recrystallized volume fraction

\( m \): the number of recrystallizing fronts intersected by unit length.

For the case of \( X_v=0.30 \), \( G \) values at each temperature were determined and an Arrhenius type plot for this is given in Fig. 10. Corresponding to the results shown in Fig. 8, activation energies for both cases of constant strain rate and constant flow stress were obtained. Although the former case gave an apparent activation energy, the \( Q \) value of 86 kcal/mol obtained in this case was almost same with that of 18-8 stainless steel obtained in the hot rolling studies by Kozasu and Shimizu. On the other hand, the case of constant flow stress which resulted in a constant initial deformed structure at each holding temperature gave rise to much higher \( Q \) value of 125 kcal/mol. The grain boundary migration rates of 1.8 % Al steel obtained by both methods were one or two order higher than that of 18-8 stainless steel. There is no reported activation energy of recrystallization or of self diffusion for 1.8 % Al steel. Glover and Sellars investigated static recrystallization of pure iron in the temperature range from 500 to 900 °C and obtained the \( Q \) value ranged from 67 to 79 kcal/mol depending on purity of iron. When the values of \( G \) obtained in pure iron are extrapolated to the higher temperature range investigated here, they become much higher than migration rates obtained in 1.8 % Al steel. From these comparisons, it appears that static recrystallization kinetics of 1.8 % Al steel may be influenced to a large extent by Al atoms in Fe-Al system. Al has a large size-misfit parameter among the various solid solution elements of the iron-base binary alloys, and it may maintain this large misfit at elevated temperature. Thus a large Al addition into iron may be able to retard the average boundary migration rate or the static recrystallization kinetics through the size-misfit effect of Al atom.

The average grain boundary migration rate is influenced by the microstructure or stored energy in the deformed state besides the temperature and alloying elements. In the hot deformation, the stored energy is related to flow stress or \( Z \) value rather than strain. Therefore, the effect of flow stress on \( G \) values was investigated as shown in Fig. 11. Flow stress was varied by changing strain rate at 1000 °C. \( G \) values increased with the increase of flow stress, and the relation of \( \sigma \) and \( G \) was expressed by the following equation taking 2.6 kgf/mm² as the reference point.

\[
G = 4.1 \times 10^{-4} \exp [0.84(\sigma-2.6)] \quad \text{(mm/min)} \quad \text{(6)}
\]
V. Conclusions

Dynamic recovery and the following static restoration process of 1.8% Al steel were investigated by hot compression equipment. The temperatures and strain rates of deformation were varied from 800 to 1100 °C, and from 5×10⁻⁴ to 10.3 s⁻¹, respectively. The following results were obtained.

1) The stress-strain curves of 1.8% Al steel behaved as a typical dynamic recovery type of metal in a wide range of deformation conditions. An activation energy of dynamic recovery process in this steel was obtained as 76 kcal/mol.

2) The substructures developed by the dynamic recovery process, which were clearly observed by interference microscope, were well-defined subgrain structures in most of the deformation conditions investigated here. This subgrain structure started to be formed at a strain somewhat below the onset of steady state flow, and the subgrain size was constant in the steady state flow region. The subgrain size depended primarily on the Z value or steady state flow stress, but not on the initial grain size.

3) The original grain boundaries were deformed into a pancaked shape at a high strain rate, but they became more scattering and less clear with the decrease of strain rate. Grain boundary bulging was not observed during straining.

4) The subgrains developed in dynamic recovery process coarsened in the following static recovery stage, and at longer holding times tended to reach a fixed subgrain size determined by the initial subgrain size and the temperature. A statically recrystallized new grain started to be formed during the stage of subgrain coarsening. The original grain boundaries, particularly triple points of boundaries, were the preferential nucleation sites for recrystallized grains.

5) A static recrystallization nucleus appeared to be formed by the subgrain coalescence mechanism along the grain boundary, and no grain boundary bulging was observed.

6) An activation energy (Q) of grain boundary migration obtained by a constant flow stress method was 125 kcal/mol. The relatively lower grain migration rate and the higher Q value obtained in 1.8% Al steel compared with pure iron seemed to be associated with the effects of solid solution of Al in iron and its large size-misfit.

Fig. 11. Effect of flow stress on the migration rate of recrystallizing front.

REFERENCES

7) G. Ouchi and T. Okita: Trans. ISIJ, 22 (1982), 543.