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Hot Deformation and Recrystallization of Austenitic Stainless Steel: Part I. Dynamic Recrystallization

A. DEGHAN-MANSHADI, M.R. BARNETT, and P.D. HODGSON

The hot deformation behavior of a 304 austenitic stainless steel was investigated to characterize the evolution of the dynamically recrystallized structure as a starting point for studies of the postdeformation recrystallization behavior. The effect of different deformation parameters such as strain, strain rate, and temperature were investigated. The flow curves showed typical signs of dynamic recrystallization (DRX) over a wide range of temperatures and strain rates (*i.e.*, different Zener–Hollomon (Z) values). However, under very high or very low Z values, the flow curves' shapes changed toward those of the dynamic recovery and multiple peaks, respectively. The results showed that while DRX starts at a strain as low as 60 pct of the peak strain, a fully DRX microstructure needs a high strain of almost 4.5 times the initiation strain. The DRX average grain size showed power-law functions with both the Zener–Hollomon parameter and the peak stress, although power-law breakdown was observed at high Z values.

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I. INTRODUCTION

DYNAMIC recrystallization (DRX) is the most important restoration mechanism during hot deformation of austenite; affecting the final microstructure and, therefore, the mechanical properties of the deformed material. Extensive research has been performed on the mechanisms and the variables involved in this phenomenon.^[1–5] These mechanisms are believed to vary depending on the steel composition and deformation conditions. However, in commercial low-carbon steels, transformation usually restricts the study of DRX of austenite at lower temperatures and, hence, limits the study of this phenomenon over a wide temperature range. Austenitic stainless steels, which do not undergo phase transformation over a wide temperature range, can be used as model alloys to investigate recrystallization.

During deformation, the pre-existing grain boundaries elongate along the deformation direction, grain boundary serration appears, and then new DRX grains nucleate at the serrated grain boundaries. The general descriptive model for DRX is that nucleation occurs at the serrated pre-existing grain boundaries and increases until a layer of DRX grains covers these boundaries. Then the recrystallization reaction proceeds *via* nucleation at the interfaces between the recrystallized and nonrecrystallized material, until new grains consume the structure. This type of DRX structure has been known as a necklace structure. Once the necklace structure is

completed, a steady state is reached; continuing nucleation and growth maintain the structure (with equiaxed grains) at a constant stress.

Despite this simple explanation for DRX phenomenon (*i.e.*, necklacing), its evolution during hot deformation of different materials is not simple and different parameters, such as strain rate, temperature, and initial grain size can affect the necklace structure.^[6–10] Although this structure is observed during hot deformation of many materials and under different deformation conditions,^[11,12] there are still many issues regarding the progress of the DRX microstructure based on this phenomenon and especially the contribution from other DRX mechanisms, such as grain boundary sliding and continuous DRX.

In this study, the effect of thermomechanical parameters including strain, strain rate, and temperature on both the mechanical and microstructural aspects have been investigated. Also, constitutive equations necessary to predict the flow behavior of this steel during hot deformation have been derived. The main focus has been on the evolution of the DRX microstructure as a function of strain under different conditions. The results in this research will be used as the basis for studies on the postdeformation recrystallization in Part II of this work.

II. EXPERIMENTS

The thermomechanical deformation in this research was performed through hot torsion. An AISI 304 austenitic stainless steel with a chemical composition of Fe-0.02 pct C-1.6 pct Mn-8.2 pct Ni-18.5 pct Cr-0.8 pct Cu was used. Torsion samples with a gage length of 20 mm and a diameter of 6.7 mm were machined from the rolled bars. The details of the hot torsion

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equipments and experiment methods have been previously described elsewhere.^[13] For hot (or warm) torsion tests, the samples were heated to 1200 °C at 5 °Cs⁻¹ and held for 3 minutes. A roughing process using a strain of 0.5, at a strain rate of 1 s⁻¹ followed by 120 seconds holding time was applied, and this resulted in a homogeneous microstructure with an average grain size of ~35 μm. The samples were then cooled at 1 °Cs⁻¹ to various deformation temperatures in the range of 600 °C to 1100 °C, held for 2 minutes, and deformed to different strains at different strain rates. To preserve the DRX microstructure, the samples were immediately (< 0.5 seconds) quenched after deformation. Common metallographic preparation was carried out on the samples and their microstructure was analyzed by electron backscattered diffraction (EBSD). The EBSD studies were performed under an accelerating voltage of 20 kV, a working distance of 25 mm from the gun, and an aperture size of 60 μm. The EBSD maps were analyzed using HKL technology channel 5 software (HKL TECHNOLOGY INC., Denmark). The DRX fraction, DRX grain size, and twin boundary fractions were measured from the EBSD maps.

The recrystallized grain size was measured using the linear intercept method,^[14] but to compare the results with the literature values, the mean linear intercept values are multiplied by 1.68 (assuming a tetrakaidecahedral shape for every grain)^[15] to determine the grain diameter. Twin boundaries were not counted as grain boundaries. The linear intercept distribution was plotted based on the geometric scale, a spacing of $\sqrt[10]{10}$, and a total of 15 size intervals.^[16]

III. RESULTS

A. Hot Deformation Flow Curves

A number of stress-strain (σ - ϵ) curves derived from deformation at different temperatures (T) and strain rates ($\dot{\epsilon}$) are presented in Figures 1 and 2. The samples deformed at high temperatures (*i.e.*, $T \geq 850$ °C) and a constant strain rate of 0.01 s⁻¹ (Figure 1) showed typical DRX flow curves, including a sharp peak followed by a long steady state. The steady state then continued until the fracture strain. At temperatures lower than 900 °C, the peak broadened and there was no clear steady state evident in the range of strains examined. However, the dominance of DRX as a restoration process is still evident from the earlier part of flow curve shape.

At low deformation temperatures (*i.e.*, $T \leq 800$ °C), the flow stress peak was not so distinct. While there appeared to be a steady state in the flow curve at $T = 700$ °C, the stress increased continuously with strain until fracture at $T = 600$ °C (Figure 1(a)). The former condition is typically considered as a sign of dynamic recovery (DRV), while the latter implies the continuous work hardening during hot deformation.

An almost similar trend was observed for deformation at a constant temperature (*i.e.*, 900 °C) with different strain rates (Figure 2). At low strain rates (*i.e.*, $\dot{\epsilon} \leq 0.03$ s⁻¹), the typical sequence of a peak stress followed by a steady state in the flow curves implied

Fig. 1—Flow curves of samples deformed at different deformation temperature and a constant strain rate of 0.01 s⁻¹.

Fig. 2—Flow curves of samples deformed at different strain rates and a similar temperature of 900 °C.

that DRX has been the dominant restoration process; although, again it was not a true steady state as some softening continued. As the strain rate increased above 1 s⁻¹, the flow curves showed a broader peak followed by no, or limited, steady state. The deformation heating

generated in these conditions (*i.e.*, high strain rates) is the most likely cause for the shape of these flow curves. This caused a fall in the flow stress of samples deformed at higher strain rates (*i.e.*, 30 s^{-1}), causing them to approach the flow curves at lower strain rates.

B. Work Hardening Curves

The DRX starts at a critical strain that can be determined either by flow curve analysis or through microstructure observations. Flow curve analysis is a simpler and quicker method compared to the microstructural analysis, as the latter needs a large quantity of samples to be examined and it can be difficult to precisely ascertain the new grains. The flow curve analysis has been used in the current work to define the critical strain ε_c for the start of DRX. In some cases, though, the results were confirmed by microstructure observations. The basis of the flow curve analysis method is that DRX (as a restoration process) affects the flow curve shape by changing the rate of work hardening through the introduction of new strain free grains. A mathematical technique, firstly proposed by Kocks and Mecking^[17] and then continued by McQueen and Ryan,^[18,19] can be used to highlight this point in the flow curves. In this method, the onset of DRX corresponds to a deviation in the work hardening curves, *i.e.*, the differentiation of stress with respect to the strain ($\theta = \partial\sigma/\partial\varepsilon$).^[20] It should be noted that the critical stress and strain in this method corresponds to a point where a certain fraction (at least 2 pct) of DRX has occurred in the microstructure to make a noticeable change in the flow curve.

The work hardening rate $\theta = \partial\sigma/\partial\varepsilon$ was calculated from the flow curve and plotted as a function of stress (Figure 3). These work hardening curves consisted of two stages. In the first stage, the work hardening rate decreased rapidly with increasing stress, possibly due to the dynamic recovery, until the start of the second stage. At this point, a change occurred in the slope of the curve and DRX was considered to have started.

Fig. 3—Work hardening curves of samples deformed at different deformation conditions.

The extrapolation of the second stage in the work hardening curve (where the fluctuations related to DRX have started), to $\theta = 0$ establishes the saturation stress σ_s^* , which would correspond to the softening due to DRV alone.^[18] The difference between this saturation stress and the flow stress obtained from the flow curves will determine the softening due to DRX.

The slope change in the work hardening curves is used to identify the critical stress σ_c and strain ε_c for initiation of DRX. It is apparent in Figure 3 that the critical stress derived from the work hardening curves (the slope change point) lies on a specific line at different deformation conditions. Poliak and Jonas^[20] showed that for a material that experiences DRX over a certain range of strain rate and temperature, the onset of DRX may be related to a constant value of the rate of strain hardening rate (*i.e.*, $\Gamma_c = -\partial\theta_c/\partial\sigma$). In the present test conditions, a value of 0.28 was observed for Γ_c , which is very close to the value of $\Gamma_c = 0.27$ reported by Ryan and McQueen^[18] for hot torsion of a 304 austenitic stainless steel. However, a different value of $\Gamma_c = 0.6$ was reported by Poliak and Jonas^[20] for a similar steel tested in hot compression. This difference could be attributed to the effect of deformation mode on the DRX flow curve.^[9]

It is difficult in this method to obtain the exact value for the inflection point in the work hardening curves, and an analytical method was also used to detect this point. In this approach, used by Poliak and Jonas,^[20] the second numerical derivative of flow curves as a function of stress (*i.e.*, $-\partial^2\theta/\partial\sigma^2$ vs σ) has a minimum at the inflection point of work hardening curve. This unique minimum point gives a good estimate of the critical stress (Figure 4). The corresponding strain ε can be extracted from the original flow curve. It is clear that this approach gives a much more easily defined value for σ_c , but does not give an estimate of σ_s^* . Therefore, both methods are required for determination of the two important stress values.

The peak and critical strain, found from the previously described method, can now be represented in the

Fig. 4—Factor of $-\partial^2\theta/\partial\sigma^2$ as a function of flow stress at different deformation conditions.

form of constitutive equations. These equations are usually shown as a function of Zener–Hollomon (Z) parameter. The activation energy of deformation Q can be estimated by fitting a hyperbolic sine function to the stress corresponding to the peak, as first proposed by Sellars and Tegart:^[21]

$$Z = A[\sinh(\alpha\sigma)]^n = \dot{\epsilon} \exp\left(\frac{Q}{RT}\right) \quad [1]$$

where $\dot{\epsilon}$ is the strain rate (s^{-1}); A , α , and n are constants independent of temperature; σ is the stress (MPa); Q is the hot deformation activation energy (J/mol); R is the gas constant; and T is the absolute temperature (K) of deformation. The Zener–Hollomon parameter (Z) encompasses two of the most important hot deformation variables (T and $\dot{\epsilon}$). This relates the equivalence of temperature and strain rate to the microstructure development during hot working. A value of 400 kJ/mol was obtained for this steel, which is in good agreement for other 304 austenitic stainless steels.^[18,22,23]

At a constant initial grain size, the constitutive equations of peak and critical strains are usually shown as power-law functions of Z in the form of^[24]

$$\epsilon = AZ^n \quad [2]$$

where A and n are constants.

The peak and critical strains derived from the flow curves were plotted on a logarithmic scale with respect to Z (Figure 5). With estimation of the values of A and n , similar type equations were developed for both peak (ϵ_p) and critical (ϵ_c) strains:

$$\epsilon_p = 3.6 \times 10^{-3} Z^{0.15} \quad [3]$$

$$\epsilon_c = 2.2 \times 10^{-3} Z^{0.15} \quad [4]$$

These equations indicate that the critical strain is a fraction of the peak strain ($\epsilon_c \cong 0.6 \epsilon_p$). A nearly similar constant ratio between critical and peak strains has been

reported for stainless steel,^[18] copper,^[25] C-Mn steels, and Fe-Ni-C alloys.^[26]

The peak flow stress (σ_p) also showed a power-law function with Zener–Hollomon parameter in the form of

$$\sigma_p = 1.31 Z^{0.13} \quad [5]$$

The results showed good agreement between the measured and calculated (based on Eqs. [3] through [5]) values. However, this equation usually overestimated the stress and strains at high Z values. The most likely reason for such over estimation is the breakdown of the power-law equation at high Z values. As Eq. [5] is based on the creep behavior, slope change (power-law breakdown) at intermediate (or high) stresses is expected due to a change in the deformation mechanism (probably from the diffusion control climb to thermally-activated dislocation glide).^[27,28] However, the role of deformation heating at high Z values can also be important in this slope change.

C. DRX Microstructures

The evolution of microstructure during hot deformation from critical to steady state (strain) was studied by quenching of samples at the end of deformation. Comprehensive microstructure analysis, including crystallographic orientations, twin orientation, subgrain boundaries, grain size distribution, etc., was carried out using the EBSD technique.

The microstructure after the roughing process and cooling to the deformation temperature (considered as the initial microstructure (before hot deformation)) consisted of equiaxed grains with a large quantity (more than 50 pct of high-angle grain boundaries) of annealing twin boundaries (Figure 6). These equiaxed grains were the result of full static recrystallization, and probably some grain growth after the roughing pass and during the following hold and cooling to the deformation temperature. The presence of this quantity of twin boundaries in the microstructure demonstrates that

Fig. 5—Critical and peak strains as functions of Zener–Hollomon parameter.

Fig. 6—EBSD map of initial microstructure after roughing process. High-angle ($\theta > 15$ deg) and twin boundaries shown by black and white lines, respectively.

formation of annealing twins is an important phenomenon during recrystallization of this steel (as expected for a low stacking fault energy (SFE) material). The grain size measurements revealed a grain size (linear intercept) of about 35 μm after roughing and cooling to 900 °C followed by rapid water quenching, with no major effect of cooling temperature observed on the grain size. Although a relatively homogeneous distribution of grain size was observed in the initial microstructure, occasional abnormal grain growth along with fine stable grains were also observed (Figure 6). This occasional difference in grain size has been ascribed to the local segregation of submicron precipitates such as oxides or sulfides.^[29]

Hot deformation at different strain rates and temperatures caused elongation of grains, serration of boundaries, formation of substructure, and finally the initiation of DRX after a critical strain. As deformation temperature and strain rate have significant effects on the deformed microstructure, hot deformation tests were performed at different temperatures and strain rates (*i.e.*, different *Z* conditions) and the quenched microstructures were analyzed by EBSD. The subsequent sections present the results of the EBSD analysis. The medium *Z* condition that showed the most comprehensive DRX microstructure evolution has been presented first, followed by some high and low *Z* deformation conditions.

D. Deformation Structure at 900 °C and 0.01 s⁻¹

At a deformation temperature of 900 °C, the initial grains started to elongate at a low strain of 0.4 (Figure 7(a)). At this strain, which is less than the critical strain to initiate DRX (as measured from the flow curves), grain boundaries were finely serrated and most of the initial twin boundaries have lost their specific twin orientation and changed to normal high-angle grain boundaries. In fact, the frequency of twin boundaries dropped from 50 pct of high-angle grain boundaries in the initial microstructure to less than 15 pct in this microstructure (Figure 7(a)). The bulging of grain boundaries was frequently observed at this strain. Also, a limited number of new DRX grains were formed, mostly at triple junctions (arrows in Figure 7(a)). This indicated that DRX had already started by formation of new small grains, despite the strain being lower than the critical strain predicted from the flow curve analysis (*i.e.*, ε_c predicted from Eq. [4]).

The elongation of grains and serration of boundaries increased with further deformation to the critical strain and a cell substructure formed in the heavily deformed grains; especially in the vicinity of grain boundaries. A considerable number of new DRX grains, much smaller than the initial grains, were formed on the serrated parts of prior grain boundaries (Figure 7(b)). At the peak strain ε_p or a strain just beyond the peak, the new DRX grains decorated a large fraction of prior grain boundaries, and the first layer of a necklace structure started to form around the original grain boundaries (Figure 7(c)). However, some undecorated grain boundaries were still present in the microstructure, while in some areas new DRX grains had formed inside the original grains.

At this deformation condition, the frequency of DRX grains inside the original grains is very low. The first layer of necklace structure was completed at a strain of 1.5, which is beyond the peak strain (Figure 7(d)).

With further deformation, more DRX grains formed at the recrystallization front (the interface of already recrystallized grains and nonrecrystallized area) and subsequent layers of the necklace structure formed (Figure 7(e)). At the steady state strain (Figure 7(f)), most of the initial grains were replaced by the small equiaxed DRX grains. However, some areas of original grains remained unrecrystallized, even after fracture. The final microstructure at the steady-state strain consisted of equiaxed DRX grains with an average linear intercept size of $\sim 7.5 \mu\text{m}$.

E. Deformation Structure at 900 °C and 1 s⁻¹

Increasing the strain rate to 1 s⁻¹ caused a major change in the deformed microstructure compared to the lower strain rate (*e.g.*, 0.01 s⁻¹). At a strain of 0.5, the original grains were slightly elongated in the deformation direction and serrated (Figure 8(a)). Also, most of the twin boundaries lost their twin orientation. Some deformation features also developed inside the grains (arrows in Figure 8(a)). At the peak strain (Figure 8(b)), very fine grains were formed in the vicinity of the serrated grain boundaries, as well as inside the deformed grains. With a further increase in the strain, more new DRX grains were formed on serrated prior grain boundaries, inside the deformed structure, and also on the recrystallization front. The recrystallized microstructure developed with increasing the strain (Figure 8(c)), and a nearly full DRX microstructure was obtained at high strains (Figure 8(d)). This final microstructure has a homogeneous distribution of DRX grains as small as $\sim 3.7 \mu\text{m}$.

It should be noted that in this condition, where the deformation time is very short due to the high strain rate, some portion of the fine grains, especially for lower strains (Figure 8(b)), could be due to the static recrystallization in the delay time of quenching. However, more accurate examinations are necessary to validate this opinion.

F. Deformation Structure at 1100 °C and 0.01 s⁻¹

Samples deformed at this deformation condition, where the deformation temperature is considerably higher (and therefore *Z* is lower) than the previous conditions, developed different recrystallized microstructures. The serration of the grain boundaries at this strain is more visible than at the lower deformation temperature, but some nonserrated grain boundaries were also observed at strains beyond the peak (Figure 9(b)), indicating an inhomogeneity in the deformed microstructure. Large quantities of new grains, much larger than the previous conditions, were formed at a low strain of ~ 0.3 (Figure 9(a)). The fraction of DRX grains increased with increasing strain to the peak. Also, a nearly full DRX microstructure was obtained at strains not very far beyond the peak strain (Figure 9(b)). The steady-state linear intercept reached a value of

Fig. 7—EBSD maps of samples deformed at 900 °C and a strain rate of 0.01 s⁻¹. High-angle ($\theta > 15$ deg) and twin boundaries shown by black and white lines, respectively.

~25 μm , which is much larger than the previous conditions, at a strain of 2 (Figure 9(c)). Unlike the previous deformation conditions, the full DRX microstructure was not homogeneous and very large grains appeared with other areas of small grains (Figure 9(d)).

G. Deformation Structure at Very Low Temperature or High Strain Rates (High Z)

As observed in Figure 1(a), the flow curve of the sample deformed at a very low temperature (*e.g.*, at 650 °C) showed evidence of DRV (*i.e.*, increasing the stress to a steady-state plateau without any peak).

In contrast, the microstructure showed the formation of a large fraction of very fine new grains (Figure 10), indicating DRX. The deformed but unrecrystallized matrix contained a very dense substructure formed through DRV. Therefore, the microstructure at the fracture strain consisted of pancaked grains accompanied by limited serration of pre-existing grain boundaries, a large quantity of small DRX grains, and a dense substructure in the deformed matrix (Figure 10). This suggests that in some cases, the flow curve analysis cannot be relied upon to determine the softening mechanism during deformation and that microstructural observation is also important.

Fig. 8—EBSD maps of samples deformed at 900 °C and a strain rate of 1 s⁻¹. High-angle ($\theta > 15$ deg) and twin boundaries shown by black and white lines, respectively.

H. DRX Kinetics

The DRX volume fraction (X_{DRX}) as a function of applied strain, measured from EBSD maps, is presented in Figure 11. As DRX is a thermally-activated process, X_{DRX} increased with an increase in temperature for a given strain and strain rate. After its initiation (*i.e.*, at $\varepsilon > \varepsilon_c$), the DRX volume fraction increased rapidly with increasing strain until a considerable fraction of the microstructure was recrystallized. The rate of DRX then decreased with further strain. This means that the DRX kinetics exhibited an Avrami type behavior that can be modeled by^[15,30]

$$X_{\text{DRX}} = 1 - \exp(-kt^n) \quad [6]$$

where k is a grain-size-dependent constant and t is the time for X pct recrystallization.

For a constant strain rate, the time from the start of DRX can be replaced by the strain after the initiation of DRX (*i.e.*, $\varepsilon - \varepsilon_c$) to form the following modified Avrami equation:

$$X_{\text{DRX}} = 1 - \exp(-k(\varepsilon - \varepsilon_c)^n) \quad [7]$$

To estimate the Avrami exponent (n), the $\ln(1/(1 - X_{\text{DRX}}))$ vs $\ln(\varepsilon - \varepsilon_c)$ was plotted and an average value of 1.3 obtained, with a negligible dependency to the deformation conditions. However, the n value obtained

under deformation at 1100 °C showed the biggest deviation from the mean value. The most likely reason for this deviation is the difficulty to discriminate (due to very similar sizes) between the recrystallized and unrecrystallized grains during the recrystallization fraction measurements. However, the n derived here is in the range of the values reported in the literature (0.7 to 1.6).^[15,18,31]

I. DRX Grain Size

It is well known^[30] that for conventional discontinuous DRX, the dynamically recrystallized grain size (d_{DRX}) does not change during the progress of DRX. Also, d_{DRX} is normally a power-law function of Zener–Hollomon parameter. The present values of DRX grain size along with some values taken from the literature are given in Figure 12. Here, though, a deviation from linear relationship is observed at very high Z values (dashed line), possibly due to power-law breakdown effects at these Z values. Below this deviation, the relationship between the d_{DRX} and Z can be expressed by

$$d_{\text{DRX}} = 5.2 \times 10^3 Z^{-0.17} \quad [8]$$

Once corrected to the mean grain diameter values (the linear intercept values (L) multiplied by 1.68,

Fig. 9—EBSD maps of samples deformed at 1100 °C and a strain rate of 0.01 s^{-1} . High-angle ($\theta > 15 \text{ deg}$) and twin boundaries shown by black and white lines, respectively.

Fig. 10—EBSD maps of samples deformed at 650 °C and a strain rate of 1 s^{-1} . High-angle ($\theta > 15 \text{ deg}$) and low-angle ($\theta < 15 \text{ deg}$) boundaries shown by thick and thin lines, respectively.

$d = 1.68 \times L$),^[15,18] there is a relatively good agreement between the present data and those previously reported in other work,^[12,18,32–34] which included a variety of initial grain sizes and deformation modes. The remaining differences can be ascribed to the differences in metallography and measurement methods. Importantly, the values cited in the literature were obtained using optical metallography, whereas the present data were

Fig. 11—DRX volume fraction as a function of strain at different deformation conditions.

obtained using EBSD maps. The difficulty in etching the boundaries in stainless steels would suggest that the optical metallography will overestimate the grain size.

Similarly, a power-law relationship between the peak stress and DRX grain size is suggested for various materials. In the current work this relationship could be described as

Fig. 12—DRX grain size as a function of Zener–Hollomon parameter.

$$\sigma_p = 1 \times 10^3 d_{\text{DRX}}^{-0.75} \quad [9]$$

The exponent of 0.75 is similar to previously reported values (0.72,^[22] 0.52,^[34] and 0.81^[18]) for 304 stainless steel. Interestingly, Eq. [9] shows a better fit between the DRX grain size and stress, compared to the Zener–Hollomon parameter in Eq. [8]. As both peak stress and DRX grain size showed power-law breakdown at high Z values, this improved fit between stress and grain size is expected.

IV. DISCUSSION

A. Flow Curve Analysis

In most of the deformation conditions tested here, the dominance of DRX as the main hot deformation restoration mechanism is obvious from the flow curve shapes. The slow kinetics of dynamic recovery, due to the low SFE of austenite in this steel, increased the possibility of DRX as the dominant restoration mechanism during hot, or warm, deformation. Single peak behavior was observed for a wide range of temperatures and strain rates. However, increasing the deformation temperature to 1200 °C (at a strain rate of 0.01 s^{−1}) caused a faint second peak to appear in the flow curve. In other work, the more obvious multiple peaks were observed by decreased the strain rate to 0.001 s^{−1}.^[35]

In contrast to single and multiple peak flow curves, some deformation conditions caused either a continuous increase in the flow stress, as in deformation at 600 °C and 0.01 s^{−1}, or a steady state without any peak, as in deformation at 700 °C and 0.01 s^{−1}. The flow curve in the former condition shows typical work hardening behavior during hot deformation, while in the latter condition, the flow curve suggests DRV as the dominant restoration mechanism. However, the microstructure at fracture in the previous conditions consisted of a large quantity of small grains (Figure 10); evidence of DRX. This implies that DRX was activated as a restoration process, although at a very high strain. Nevertheless, the

Fig. 13—DRX map of 304 austenitic stainless steel at a constant initial grain size of 35 μm. Values on dotted lines show the DRX grain size at given Z .

rate of work hardening (at least for the strain range tested here) is higher than the restoration rate (DRV or DRX), and a balance between work hardening and softening (steady state) is not established before fracture. The presence of a high fraction of low-angle grain boundaries (more than 60 pct in Figure 10) indicates that DRV is a very dominant restoration mechanism in these samples before fracture.

The previous discussion indicates that an exact analysis of the flow curve can provide insights in the mechanism investigation of hot working and, at least, minimize the microstructural examination. On the other hand, the result of this analysis in conjunction with the microstructural observations can be used to predict the different features of a deformed and recrystallized microstructure, and summarized in a DRX map (Figure 13).

The lower limit of this map is the critical strain for the initiation of DRX (ϵ_c) calculated from the flow curve analysis. However, the upper limit (fully DRX microstructure) in this map was determined from the microstructural observations. The vertical dashed lines show the steady-state grain size d_{DRX} at constant Z values. In this map, a critical value of Z is identifiable, above which the initiation of DRX is not possible before the fracture strain. In the present case, this critical Z was about 10²¹.

The present DRX map is plotted for an initial grain size of 35 μm. As the initial microstructure has a pronounced effect on the critical strain and DRX kinetics, this should be taken into consideration when plotting DRX maps. This DRX map indicates that the smallest grain size obtained in this work is about 2.5 μm, attained by deformation at a strain rate of 0.01 s^{−1} and at 750 °C. This implies that it is possible to make a very fine grain microstructure just by considerably reducing the deformation temperature, and the application of a high strain rate is not necessary. However, formation of a fully DRX microstructure with such a small grain size needs a strain higher than the strain achievable before fracture in a torsion sample.

The strain required for a fully (or nearly full) DRX microstructure (ε_f) is a power-law function of Z :

$$\varepsilon_f = 9.1 \times 10^{-3} Z^{0.15} \quad [10]$$

The exponent of Z in Eq. [10] is very close to the Z exponent in Eq. [4]. This implies that there is a linear relationship between the critical and final strains in the form of

$$\varepsilon_f \cong 4.15 \varepsilon_c \quad Z < 10^{17} \quad [11]$$

The value of $Z = 10^{17}$ is another critical value for Z , below which a fully DRX microstructure is achievable just by hot (or warm) deformation.

A combination of Eqs. [9] and [10] gives a new equation of the form

$$\varepsilon_f = 17.2 d_{\text{DRX}}^{-0.88} \quad [12]$$

This equation gives the smallest strain required to achieve a fully DRX microstructure starting with a particular grain size. This also shows that to obtain a full DRX microstructure with grain size less than $1 \mu\text{m}$, a strain of at least 17 is necessary, which is unlikely to be achieved during the industrial hot deformation processes.

However, it is important to note that these equations apply for $d_0 = 35 \mu\text{m}$. A decrease in d_0 would decrease the critical strain (*i.e.*, $\varepsilon_c \propto d_0^{1/2}$).^[6] Therefore, according to Eq. [12], the strain required to obtain a given final DRX grain size will decrease by decreasing the initial grain size. For example, by decreasing the initial grain size to $8 \mu\text{m}$, the strain required to achieve a final grain size of $1 \mu\text{m}$ is about 8. This new strain is much lower than the previous one for an initial grain size of $35 \mu\text{m}$, although still large from an industrial perspective.

B. Microstructural Evolution

There is strong evidence to link the recrystallized microstructure to the deformation conditions of temperature and strain rate (Z value). The necklace structure and its evolution during deformation vary with the deformation conditions, and this type of structure is not clearly visible for all deformation conditions. The clearest necklace structure and subsequent evolution of the DRX microstructure occurred at moderate Z values (*e.g.*, deformation at 900°C and 0.01 s^{-1} in Figure 7). This development is summarized in Figure 14 for a small part of microstructure. After a small strain (Figure 14(b)), the straight prior grain boundaries were distorted and developed an irregular shape in the form of serrations and bulges, where the first dynamically recrystallized grain were located with further deformation to the critical strain (Figure 14(c)).

With an increase in the strain, the bulges started to separate from the original boundaries and formed new grains. At the peak strain (Figure 14(d)), most of these original boundaries were decorated by DRX grains. At this stage, the first layer of the necklace structure has

formed. It has been shown^[11] that the orientation of these new grains, which formed through bulging of original grain boundaries, is very similar to the orientation of the initial grains. The formation of new DRX grains continued with increasing strain and the second layer of necklace structure was formed at a strain close to the start of the steady state (Figure 14(f)). This necklace structure was completed at a strain far beyond the start of the steady-state strain and replaced all of the original deformed microstructure.

At low Z values, where the DRX kinetics is very high, the formation of the necklace structure is not as obvious as for the moderate Z conditions (Figure 9). In this case, the nucleation of DRX grains started at very low strains. It appears that for these conditions, where the mobility of the grain boundary is very high, new grains can separate from the serrated section of the pre-existing grain boundary and grow very quickly. Therefore, a large number of serrated boundaries cannot be maintained for a long period of time. This indicates that the bulging mechanism is very effective at low Z values. However, the formation of a large quantity of new grains at triple junctions and nonequilibrium grain boundary junctions (quadruple junctions) avoids the formation of a necklace structure. The formation of DRX grain at these junctions may indicate the activation of the grain boundary sliding as an important DRX mechanism at such low Z values.^[36–38]

One of the most obvious differences between the low and moderate (or high) Z conditions is the grain size distribution in the final microstructure (Figure 15). Unlike the moderate or high Z values, in the low Z conditions the grain size distribution is not homogeneous and some very large grains were mixed among very small grains. This sort of grain size distribution can be related to the grain growth and grain impingement during deformation. In the low Z condition, the recrystallized grains can grow very fast and the concurrent work hardening does not prevent their growth as much as in high or moderate Z conditions. Therefore, some DRX grains that formed early will grow very quickly and then prevent the growth of the adjacent grains by impingement. But in the case of high Z values, the delay in the initiation of DRX process can cause a high level of work hardening. Therefore, when the work hardening reaches a critical value, a large number of new DRX grains will nucleate. The simultaneous growth of such high fraction of new grains, in addition to concurrent work hardening, can prevent those grains growing to very large sizes due to impingement. However, the role of low temperature or high strain rate (*i.e.*, high Z) in decreasing the grain boundary migration is also very important for the formation of small grains.

V. CONCLUSIONS

The hot deformation behavior of a 304 austenitic stainless steel was investigated to characterize the evolution of the DRX structure. The effect of different deformation parameters such as strain, strain rate, and

Fig. 14—Progress of DRX based on the necklace structure after deformation at 900 °C and 0.01 s⁻¹: (a) initial microstructure, (b) serration of grain boundaries at low strain, (c) nucleation of first DRX grains, (d) decorating of serrated grain boundaries by first layer of necklace structure, (e) formation of second layer of necklace structure on interface of first layer and deformed structure, (f) formation of next layers of necklace, and (g) completion of necklace structure.

Fig. 15—Grain size (linear intercept) distribution under different deformation conditions at a constant strain of 3.

temperature were investigated. The most important results in this study are as follows.

1. Microstructural analysis showed that DRX was operative over a wide range of deformation temperatures and strain rates; even when the curves showed no steady state but continuous work hardening.
2. The peak and critical strains (derived from flow curve analysis) showed power-law functions with the Zener–Hollomon (Z) parameter and the ratio of critical strain to peak strain was ~ 0.6 .
3. A deformation mechanism map was developed to show the different features of the deformed and recrystallized microstructures as a function of temperature and strain.
4. The relationship between the DRX grain size and the Zener–Hollomon parameter deviated from the typical power-law function at high Z values ($Z > 10^{18}$).

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