EBSD study of a hot deformed austenitic stainless steel

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The microstructural evolution of a 304 H austenitic stainless steel subjected to hot compression was studied by the electron backscattered diffraction (EBSD) technique. Detailed data about the boundaries, coincidence site lattice (CSL) relationships and grain size were acquired from the orientation imaging microscopy (OIM) maps. It was found that twins play an important role in the nucleation and growth of dynamic recrystallization (DRX) during hot deformation. Moreover, the conventional discontinuous DRX (DDRX) was found to be in charge of grain refinement reached under the testing conditions studied. Furthermore, the recrystallized fraction (X) was determined from the grain average misorientation (GAM) distribution based on the threshold value of 1.55°. The frequency of high angle boundaries showed a direct relationship with X. A time exponent of 1.11 was determined from Avrami analysis, which was related to the observed single-peak behavior in the stress–strain flow curves.

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1. Introduction

Hot deformation processing plays an important role in the industry for production of steels with required mechanical properties while maintaining the production costs as low as possible. The main idea is enhancing the properties of the material by processing not by alloying. Therefore, the parameters of the forming process must be carefully controlled to produce a fine microstructure with desired shape, size, distribution and orientation of phases. Different strategies can be applied to refine the microstructure during and after hot deformation of steels, which are based on phenomena such as dynamic recrystallization (DRX) [1], metadynamic recrystallization (MDRX) [2], static recrystallization (SRX) [3], austenite pancaking [4], precipitation and phase transformation [5].

During the course of hot deformation, grain refinement can be achieved by dynamic recrystallization [4,6]. In some materials such as aluminum, dynamic recovery (DRV) can balance work hardening, and a plateau in the flow curve is then achieved. However, in many materials such as austenite phase in steels, the kinetics of DRV is low and the work hardening cannot be balanced only by DRV. As a result, the dislocation density increases gradually by strain and eventually the recrystallization occurs during deformation. It has been recognized that DRX is one of the most important softening mechanisms in hot deformation processing and has profound effects on the microstructure, grain size and flow stress [7]. It was shown by Sellars and McTegart [8] and others [9,10] that hot working can be considered as a thermally activated process and it can be described by strain rate equations similar to those employed in creep studies. Due to the great impact of DRX on hot working flow stress and its effect on the microstructure and properties of the material after processing, the prediction of DRX kinetics is of considerable importance in the modeling of industrial hot working processes.

The critical strain for initiation of DRX could be determined by metallography. However, this technique requires extensive sampling before and after the critical strain. Furthermore, phase changes during cooling from hot working temperature alter the deformed structure, which in turn render difficulties for metallographic analysis. Ryan and McQueen [11] observed that the presence of a stress peak at a constant strain rate flow curve leads to an inflection in plots of strain hardening rate (θ) versus stress (σ). Moreover, the points of inflection in θ–σ plots where the experimental curves separate from the extrapolated lower linear segments give critical conditions for initiation of DRX. Later, Poliak and Jonas [12,13] have shown that this inflection point corresponds to the appearance of an additional thermodynamic degree of freedom in the system due to the initiation of DRX.

The characterization of grain structure and the substructure inside grains is a prerequisite for the study of high-temperature
phenomena such as the nucleation and growth of DRX grains. Traditionally, optical microscopy (OM) and transmission electron microscopy (TEM) have been used to study the hot deformed samples. The former have limitations to characterize the substructure and grain boundary character, while the latter is limited to small parts of the specimen. Electron backscattered diffraction (EBSD) in the scanning electron microscope (SEM) is a powerful tool to study hot deformed samples. EBSD enables highly detailed crystallographic information to be acquired automatically from large areas of a polished specimen within a relatively short time by stepping a SEM beam across the sample and characterizing each point in order to obtain information about individual grains and to establish direct neighborhood relationships between these grains [14].

The present work is focused on the study of the microstructural evolution and DRX in a 304 H stainless steel subjected to hot compression. Detailed data about misorientations, grain boundary character, grain size, recrystallization mechanism and recrystallized fraction were acquired by EBSD.

2. Experimental materials and procedures

A 304 H austenitic stainless steel with chemical composition of 0.064 wt% C–18.35 wt% Cr–8.96 wt% Ni–1.52 wt% Mn–0.45 wt% Mo–0.83 wt% Si was used in this work. Cylindrical specimens were prepared, which were 11.4 mm in height and 7.6 mm in diameter for hot compression test.

2.1. Hot compression test

An Instron 4507 universal deformation machine was used for uniaxial hot compression tests. Tantalum foils and boron nitride solution were used to reduce friction between the anvils and the specimen. The specimen was heated up to a deformation temperature of 1050 °C and held there for 5 min for homogenizing purposes and then deformed at true strain rate of 0.1 s⁻¹ under different true strains ranging from 0.21 to 0.8. After deformation, samples were immediately quenched in water.

2.2. Microstructural studies

The hot deformed samples were sectioned parallel to the compression direction. The electrolytic etch in 60% HNO₃ was used for optical metallography. For EBSD analyses, the samples were ground with SiC papers, mechanically polished with diamond slurries and then were finely polished with 0.04 μm colloidal silica solution for one hour. A Zeiss UltraPlus analytical field emission gun scanning electron microscope (FEG-SEM) equipped with an EBSD detector, provided by HKL Technology, was used for EBSD studies. Step sizes of 0.3, 0.4 and 0.08 μm were used to scan the max deformed portion available at the cut surface. The Channel 5 software was used to analyze and display the data. In the generated OIM (orientation imaging microscopy) maps, high angle boundaries (misorientations greater than 15°) and low angle boundaries (misorientations less than 15°) were shown as black and white lines, respectively. Different colors were used for high angle boundaries with a coincidence site lattice (CSL) relationship. As a result, only ordinary high angle boundaries were shown with black lines. Σ3 twin boundaries (misorientation of 60° about a (1 1 1) axis), Σ5, Σ7, Σ9, Σ11 boundaries were respectively shown as red, green, navy blue, pink and yellow lines.

3. Results and discussion

3.1. Flow behavior

Flow curves obtained at different strains are shown in Fig. 1a. All curves show a similar behavior, which is as indicative of the appropriateness of the deformed samples to represent the same deformation condition of 1050 °C–0.1 s⁻¹. The complete flow curve exhibits typical DRX behavior with a single peak stress followed by a gradual fall towards the steady state stress. The work hardening rate-stress (dσ/dε) curve analysis was performed to reveal if DRX occurred [12,13]. The dσ/ε curve for this deformation condition is shown in Fig. 1b. This figure shows a clear inflection point, which is related to the occurrence of DRX. The appearance of this inflection point is identical to appearance of a minimum in −dθ/dσ curve, which is shown in Fig. 1c.

Fig. 2a shows the microstructure of the sample for strain of 0.28. It can be seen that the initial grain boundaries are corrugated and
serrated and some bulges and new fine grain can be observed, which are indicators of the occurrence of DRX. Fig. 2b shows the flow curve and some superimposed OIM band contrast microstructures at different strains. It can be seen that the initial annealed material and the dynamically recrystallized material consist of equiaxed grains with many annealing twins, while the deformed material shows a high fraction of low angle grain boundaries. Owing to its very low dislocation density, the quality of the band contrast map for the annealed sample is higher than the recrystallized and deformed samples, which have higher dislocation densities due to the operation of concurrent deformation.

3.2. Characteristic points of the flow curve

The analyses based on work hardening rate ($\dot{\varepsilon}$) were performed to reveal if dynamic recrystallization (DRX) occurred and also to determine the characteristic points of the flow curves [13,15]. The critical stress ($\sigma_C$) and strain ($\varepsilon_C$) for initiation of DRX were detected from the inflection in plot of $\dot{\varepsilon}$ versus $\varepsilon$ (as shown in Fig. 1c) and Inflection in plot of $\ln \dot{\varepsilon}$ versus $\sigma$, respectively. Moreover, the peak stress ($\sigma_P$) and peak strain ($\varepsilon_P$) were respectively detected from the first occurrence of $\dot{\varepsilon} = 0$ in the $\sigma$–$\varepsilon$ and $\sigma$–$\sigma$ curves. The values of $\sigma_P$, $\sigma_C$, $\varepsilon_P$ and $\varepsilon_C$ were determined as 129.65 MPa, 121.28 MPa, 0.4 and 0.2, respectively. Therefore, the normalized critical stress and strain can be expressed as $\sigma_C/\sigma_P = 0.93$ and $\sigma_C/\sigma_P = 0.5$, respectively. The value of normalized critical strain is within the range reported for steels, which is between 0.3 and 0.9 [13,15–19] but is lower than a previous report on 304 stainless steel [20]. The values of studied strains in this work are 0.21, 0.28, 0.4, 0.63, 0.72 and 0.8, which correspond respectively to normalized strains ($\varepsilon/\varepsilon_P$) of 0.5, 0.7, 1, 1.57, 1.8 and 2.

3.3. Hot deformed microstructures

The OIM band contrast microstructures at different values of strain and their associated misorientation distribution histograms are shown in Fig. 3. As illustrated in Fig. 3a, in the annealed sample, most of boundaries are of high angle type, many of them have $\Sigma 3$ twin relationship, and the fraction of low angle boundaries is very low. Some of these twin boundaries are coherent (straight) and some others are incoherent (curved) [21]. However, a large fraction of boundaries in the sample deformed to strain of 0.21 (Fig. 3b), are sub-boundaries, which is a result of dislocation generation and dislocation boundary formation (e.g. cell walls and geometrically necessary boundaries) during deformation and dynamic recovery (DRV). These changes can be better realized from Fig. 3g. It can be seen that the frequency of twins and other high angle boundaries has decreased significantly and a very strong peak in the low angle part of the histogram has been developed. Besides the formation of low angle boundaries, some of the twin boundaries have lost their twin characteristics (unique angle and axis relationship) due to effects of concurrent deformation and grain rotation [20–22]. Conversely, at higher strains (Fig. 3c–f), the new DRX grains start to consume the deformed microstructure which results to decreasing the fraction of low angle boundaries and increasing the fraction of high angle ones. Interestingly, a large fraction of newly generated boundaries after the initiation of DRX have $\Sigma 3$ twin relationship (Fig. 3g). This point will be future considered in the following section.

3.4. The role of annealing twins

Fig. 4 shows the misorientation distribution of some samples. It can be seen that at strain of 0.21, the intensity of the twin peak decreased and a little broadened, which is an indication of the deviation from the ideal orientation [23]. However, after a little progress of DRX at strain of 0.28, the peak intensity increased and the deviation decreased, which shows that there are newly nucleated twins. It has been reported that the formation of twins plays an important role in the nucleation and growth of DRX in materials with low stacking fault energy [21,22,24–26]. The grain boundary migration is frequently accompanied by twinning [27,28]. During high temperature or low strain rate deformation, bulging can easily take place and depending to the deformation conditions, the bulged part of a sertared grain boundary can lead to the formation of a new DRX grain through the evolution of strain-induced dislocation sub-boundaries or twinning [22,29,30]. The role of twinning in separation of bulged part and nucleation of DRX can be seen in Fig. 5. This figure reveals that the other CSL boundaries can also take part in creation of new boundaries for separating grains.

In some studies it was found that the majority of twins formed during the growth of DRX grain, because the dislocation sub-boundaries behind the bulged parts were free of twins at incipient stages of DRX [22]. There is evidence that during the formation of further necklace layers and also during steady state condition, in which bulging requires a very high boundary curvature on the existing small recrystallized grains [21], the new DRX grains form trough the repeated nucleation and growth of annealing twins on the migrating interface [22]. The twin boundaries evolved near the interface of DRX and parent grains might either glide away laterally to the migrating boundary [31], or by the dissociation of initial grain boundary into a coherent twin part and a mobile one.
with moderate to high misorientation [29]. Some evidence of this effect can be seen in Fig. 5.

The propagation of slip dislocations across grain boundaries could produce additional \{111\} steps on the boundaries. The movement of these steps could result in additional twins [31,32]. Twinning can play a significant role in lowering the boundary energy of a growing grain and also increasing the mobility of its moving boundary, which can help the progress of recrystallization [33]. However, twinning sometimes results in immobile boundaries [34].

Annealing twins are mostly found in statically recrystallized microstructures (Fig. 3a). However, the present study and other relevant works [21,22,24–26] show that twinning is also typical for DRX. In some cases, some of the twin boundaries in the DRX
3.5. Discontinuous DRX

Recrystallization may occur in two ways: (1) heterogeneously throughout the material, that is, it may be formally described in terms of nucleation and growth stages (discontinuous), or (2) uniformly, i.e., the microstructures evolve gradually with no identifiable nucleation and growth stages (continuous). The discontinuous dynamic recrystallization (DDRX), which commonly takes place in metals with low and medium stacking fault energy, involves grain boundary migration. Conversely, continuous dynamic recrystallization (CDRX), which takes place in alloys in which dislocation motion is inhibited, involves continuous fragmentation of substructure to form crystallites bounded with high angle boundaries with little or no grain boundary migration [35]. Progressive lattice rotation is a common type of CDRX in hot working and due to its nature, the presence of boundaries with 10–15° misorientation is necessary for its nucleation in the substructure [35]. Fig. 6a shows the fraction of boundaries versus deformation strain. Since each point was obtained from a single EBSD scan, there are no error bars shown on the data. It can be seen that the fraction of low angle boundaries (<10°) increases abruptly at the beginning of the deformation but decreases continuously during the progress of DRX. The converse effect can be seen for high angle grain boundaries and Σ3 twin boundaries. However, the fraction of low angle boundaries with misorientation in the range of 10–15° remains very low (around 1% of all boundaries) during the hot deformation. Fig. 6b shows the change in the fraction of different misorientation angle scopes during deformation. It can be seen that the fraction of high angle boundaries increases with strain after the initiation of DRX but the fraction of boundaries with misorientation in the range of 10–15° does not change significantly. Moreover, there is no evidence of loose high angle boundaries inside the grains (Fig. 3). Loose high angle boundaries, which are incomplete boundaries inside grains, can be produced by the operation of progressive lattice rotation mechanism. Therefore, it is concluded that the conventional DDRX is the only significant operating mechanism of DRX in the present conditions [35,36].

3.6. Evolution of grain size

The change in average grain size versus strain is shown in Fig. 7a. As can be seen, the average grain size decreases at increasing strain. This effect is also evident from the OIM maps of Fig. 3. The initial grain size (D0) is 19.7 μm, while grain size at strain of 0.8 (D) is 5 μm. Although this is not the steady-state grain size but is close to it. The grain size at strain of 0.8 along with some reported data of steady-state grain size for a hot compressed 304 stainless steel are shown in Fig. 7b. As can be seen, the value of 5 μm is in well agreement with

![Fig. 4.](image)

**Fig. 4.** The right part of the misorientation distribution histograms of some samples.

grains lose their twin characteristics due to effects of concurrent deformation and grain rotation and as a result, twinning may be concealed [21].

![Fig. 5.](image)

**Fig. 5.** OIM band contrast map for the sample deformed to strain of 0.28.

![Fig. 6.](image)

**Fig. 6.** The fraction of different boundaries: (a) the fraction of boundaries versus strain and (b) the change in the fraction of different misorientation angle scopes during deformation.
the general trend in literature data. The value of $D_0/D$ is about 4. It means that this deformation condition (deformation temperature of 1050 °C under strain rate of 0.1 s$^{-1}$) resulted to grain refinement and the observed single peak behavior in the stress–strain curve is justified [37].

By progressing DRX, the grains not only have become finer, but also have developed high fraction of $\Sigma 3$ twin boundaries (Fig. 6a). Fig. 8 shows the fraction of different CSL boundaries versus deformation strain. It can be seen that the fraction of $\Sigma 3$ and $\Sigma 9$ boundaries at strain of 0.8 are 53.3% and 2.1%, respectively, while the corresponding values for the annealed sample are 54.8% and 4.2%, respectively. The fraction of other $\Sigma$ twin boundaries was very low in all samples and is not shown in this figure.

3.7. Recrystallization fraction

Sometimes, the recrystallized grains by necklace DRX are easily distinguishable [20,22]. However, in some cases such as the present study, the division of recrystallized grains is difficult. Therefore, a quantitative method is required for this purpose. Several approaches based on EBSD data have been proposed to separate recrystallized grains from the deformed ones and to find the recrystallized fraction, which are based on the grain internal statistics. The dislocation density in the deformed grains is high and these dislocations are arranged in dislocation structures, which result in local misorientations of several degrees within grains. However, recrystallized grains have much lower dislocation densities. Therefore, the methods based on grain internal statistics can be used for dividing the recrystallized grain form the deformed matrix.

One method to characterize the distribution of the low-angle misorientations of the grains is to determine the overall spread in orientation within the grains, in which the grain orientation spread (GOS) is calculated by averaging the deviation between the

**Fig. 7.** DRX grain size: (a) the change in average grain size versus strain by consideration of all high angle boundaries and (b) the grain size at strain of 0.8 along with some reported data of steady-state grain size for a hot compressed 304 stainless steel.

**Fig. 8.** The fraction of different CSL boundaries versus strain.

**Fig. 9.** The distribution of GAMS for different samples.
orientation of each point in a given grain and the average orientation of that grain [14,26,38,39]. The GOS values for recrystallized grains are lower than those of deformed grains. Another useful method is the grain average misorientation (GAM), which is the average misorientation between all neighboring pairs of points in a grain [14,40]. GAM is based on point-to-point misorientations inside a grain. The GAM values for recrystallized grains are also lower than those of deformed grains. However, for the methods based on GAM and GOS, a threshold value of these parameters is required to distinguish a recrystallized grain from a deformed one.

In the present work, the GAM approach was used to find the recrystallized fraction (X). The distribution of GAM is shown in Fig. 9. As displayed in Fig. 9a, at a GAM value of 1.55°, there is a sharp change in the GAM distribution in all deformation strains, especially for strains after the peak point of the flow curve. Fig. 9b, which is the same figure on double-logarithmic scale, better shows that there is multimodal distribution of GAM for each sample. However, the distribution on the left hand side is related to the low GAM values and it seems that a threshold value of 1.55° is suitable for all samples. Therefore, grains with GAM less than 1.55° were considered as recrystallized grains. Based on this technique, the recrystallized grains in the OIM map of the annealed sample are shown with green color in Fig. 10. As can be seen, almost all grains were labeled as recrystallized. However, the exact recrystallized fraction is 0.9, while this is a completely (statistically) recrystallized sample. Therefore, a deviation of ±0.1 was considered for values of recrystallized fraction determined by this method.

![Fig. 10. The OIM map of the annealed sample showing recrystallized grains with green color. The high angle, low angle and Σ3 twin boundaries are shown as black, white and red lines, respectively. (For interpretation of the references to color in this figure, the reader is referred to the web version of the article.)](image)

![Fig. 11. The OIM map of the samples deformed to strain of (a) 0.21, (b) 0.28, (c) 0.4, (d) 0.63, (e) 0.72 and (f) 0.8. The recrystallized grains are shown with green color. The high angle, low angle and Σ3 twin boundaries are shown as black, white and red lines, respectively. (For interpretation of the references to color in this figure, the reader is referred to the web version of the article.)](image)
The OIM maps showing the recrystallized grains with green color for samples at different strains are shown in Fig. 11. It can be seen that the recrystallized fraction increases with increasing strain. However, there are some recrystallized grains at strain of 0.21, which is close to the onset of DRX. The recrystallized fraction at this strain is about 4%. This implies that the calculated critical strain for the onset of DRX does not correspond to the strain at which the very first DRX grains form. At the calculated critical strain, the progress of DRX is high enough to be detectable on the flow curve [20,41].

Fig. 12. The recrystallization curves: (a) the recrystallized fraction versus true strain, (b) the recrystallized fraction versus recrystallization time, and (c) plot used to determine the Avrami constants.

The relation of the frequency of high angle boundaries and recrystallized fraction: (a) the fraction of different boundaries versus recrystallized fraction and (b) the dependency of recrystallized fraction and the fraction of high angle boundaries to strain.

3.8. Avrami analysis

Fig. 12a shows the recrystallized fraction versus true strain. The recrystallization curve of Fig. 12a has the well-known sigmoid shape, but without incubation time. Note that the critical point is assigned to the start of recrystallization \( t=0 \). Fig. 12b shows the recrystallized fraction versus recrystallization time. The values of recrystallization time \( t \) were calculated according to strain rate \( (\frac{\varepsilon}{S} = \frac{1 - \varepsilon}{\varepsilon - \varepsilon_c}(t - 0)) \). Now, it is possible to apply the Johnson–Mehl–Avrami–Kolmogorov (JMAK) equation as shown below:

\[
X = 1 - \exp(-kt^n)
\]

where \( n \) and \( k \) are terms associated with the nucleation mode and the nucleation and growth rates, respectively [42]. It follows from Eq. (1) that the plot of \( \log \ln[1/(1 - X)] \) versus \( \log t \) can be used to determine the values of \( n \) and \( k \) as shown in Fig. 12c. The value of \( n \) was determined as 1.11, which is consistent with the value of 1.3 for a 304 stainless steel based on torsion test [20] and a Ti-modified austenitic stainless steel [43].

A low Avrami exponent (\( <1 \)) is an indication of nucleation in the interfacial surface of grain and twin boundaries, while a high coefficient (\( >2 \)) indicates that nucleation takes place on grain and twin edges [44]. In the latter case, dynamic recrystallization is controlled by nucleation due to the relatively small number of nucleation sites that results to a small number of nuclei that grow uninhibitedly, which results to a larger final grain size than the initial one during
cyclic DRX. However, at low Avrami exponents, dynamic recrystallization is growth controlled. Therefore, a large number of growing nuclei results in the single peak behavior and grain refinement \[42\]. The time exponent of 1.11 in the present work indicates that nucleation takes place at the surface of grain and twin boundaries.

3.9. The relation of recrystallized fraction with other factors

As can be seen in Fig. 13a, the fraction of low angle boundaries (<15°) decreases as the recrystallized fraction increases. The converse effect can be seen for high angle grain boundaries. It seems that these trends can be related to the recrystallization fraction. Fig. 13b shows the dependency of recrystallized fraction and also the fraction of high angle boundaries to deformation strain. The good agreement between these two curves signifies that the fraction of high angle boundaries is also a good measure of recrystallization progress.

As indicated before, the grain average misorientation (GAM) by consideration of a threshold value of 1.55° was used to distinguish between recrystallized and deformed grains. A relevant concept based on GAM is the average GAM (AGAM) for each sample. Based on the distribution of grain average misorientation, the weighted average of GAM was calculated and presented as the AGAM for that distribution. Fig. 14a shows the dependence of AGAM on strain. It is evident that the value of AGAM first increases due to the development of deformation substructures and then decreases due to the progress of DRX.

The variation of AGAM with recrystallized fraction is shown in Fig. 14b. There is a linear relationship between recrystallized fraction and AGAM. The value of AGAM for the annealed sample is 0.68°. By assigning the recrystallized fraction of 1 to reciprocal of this value, the values of recrystallized fraction for other samples were calculated. The results are shown in Fig. 14c. Except the case of strain of 0.21, the calculated recrystallized fractions by AGAM are in a good agreement with the calculated values using the threshold value of 1.55°. The large error in the case of strain of 0.21 is related to the high frequency of grains with relatively low GAM (and therefore incorrectly low weighted average) in this sample.

4. Conclusions

(1) The flow curve of the 304 H stainless steel under deformation temperature of 1050 °C with strain rate of 0.1 s⁻¹ exhibited typical DRX behavior with a single peak stress followed by a gradual fall towards the steady state stress.

(2) In the annealed sample, most of boundaries were of high angle type and many of them had Σ3 twin relationship. However, a large fraction of boundaries in the sample deformed to strain of 0.21 were sub-boundaries, which was a result of dislocation generation and dislocation boundary formation during deformation and dynamic recovery (DRV). Conversely, at higher strains, the new DRX grains started to consume the deformed microstructure which resulted to decreasing the fraction of low angle boundaries and increasing the fraction of high angle ones. Interestingly, a large fraction of newly generated boundaries after the initiation of DRX had twin relationship and the annealing twins played an important role in the nucleation and growth of dynamic recrystallization during hot deformation.

(3) The fraction of low angle boundaries with misorientation in the range of 10–15° remained very low (around 1% of all boundaries) during the hot deformation. Moreover, there was no evidence of loose high angle boundaries (as a result of progressive lattice rotation) inside the grains. Therefore, the progressive lattice rotation was not confirmed as a supplementary DRX mechanism and the conventional DDRX was the sole operating mechanism of DRX and was responsible for the observed grain refinement.

(4) The recrystallized fraction (X) was determined from the grain average misorientation (GAM) distribution based on the threshold value of 1.55°. This threshold value was determined from the multimodal distribution of GAM. Moreover, the frequency of high angle boundaries and the reciprocal of average GAM (AGAM) showed a direct relationship with X. The time exponent of 1.11 was determined from Avrami analysis, which was related to the observed single-peak behavior in the stress–strain flow curve.
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