Simulation and experimental study on microstructure evolution of 5CrNiMoV steel during multi-directional non-isothermal forging

Zhiqiang Hu, Kaikun Wang and Li Yang

1 University of Science and Technology Beijing, Beijing, People’s Republic of China
2 China Iron and Steel Research Institute Group, Beijing, People’s Republic of China

E-mail: kkwang@mater.ustb.edu.cn

Keywords: microstructure evolution, prior parent austenite grains, textures, non-isothermal forging, FEM simulations

Abstract

Microstructure evolution during the hot forming shows a significant impact on material’s mechanical properties. To explore the deformation characteristics of 5CrNiMoV steel, numerical simulation and microscopic phase-field simulation of the multi-direction forgings were carried out. The strain distribution at each pass was investigated and the evolution of temperature, effective strain, effective strain rate, and grain size was acquired. The hot forging trials were carried out and three typical regions of forgings were taken to study the microstructure evolution. Detailed microstructure characterizations showed that the constructed parent austenite grain size of the forging in typical regions was slightly larger than the simulation results due to the grain coarsening during the air cooling. There were large amounts of high angle grain boundaries (HAGBs) for the occurrence of complete dynamic recrystallization and many bulging grain boundaries showed that discontinuous dynamic recrystallization (DRX) could be the governing mechanism of nucleation and growth of dynamic recrystallization (DRX). Besides, the hot deformation texture changed significantly during the non-isothermal forging and the texture component differed remarkably at different regions of the forging. The main hot deformation texture components were Cube {001}<100> and Goss {011}<100>.

1. Introduction

Microstructure evolution of steels during hot deformation shows a significant influence on the final mechanical properties. Many deformation mechanism events, such as grain boundary sliding and bulging, dislocation slip and twinning, dynamic recrystallization (DRX), meta-dynamic recrystallization (MDRX) and dynamic recovery (DRV), may occur during deformation [1–4]. These events will have a great impact on the behavior of flow stress, the microstructure evolution and the final mechanical properties. Usually, the flow stress curves obtained from the isothermal hot compression tests can be also used to illustrate the situation of metallurgical events that occurred during the deformation. The hot deformation behaviors of steel and its microstructure evolution of austenite for temperatures ranging between 800 °C and 1100 °C with strain rate up to 2 s⁻¹, were studied based on the hot compression tests [5, 6]. Other similar works were carried out to study the impact of the hot deformation conditions on the deformation behavior and microstructure evolution [7–10].

Although there are many published papers concerning the hot deformation behaviors and the microstructure evolution, most of the research usually focuses on the samples compressed at relatively low strain rates by Gleeble simulator [7, 11, 12]. Limited research has been carried out to study the effect of non-isothermal conditions on the austenite microstructure evolution. However, most forgings need to be acquired under non-isothermal conditions on account of the requirement of low cost and high efficiency, especially for the large shaft parts and forging dies. The hot forging process usually causes the inhomogeneous distribution of strain, stress and temperature, which results in the inhomogeneity of microstructure and mechanical properties [13, 14]. To investigate the microstructure of Inconel 718, pancake forging was performed using a hydraulic press and samples at different locations of the pancakes were investigated through the experiment and FEM analysis [15]. To study the impact of
strain magnitude on the microstructure evolution during hot forging, three cylindrical samples were forged at room temperature and typical regions of the samples were analyzed by electron back-scattered diffraction (EBSD) [16].

For non-austenitic steel, the microstructure after hot forging is mainly composed of martensite and bainite. However, the prior austenite microstructure evolution during the non-isothermal forging has a significant impact on subsequent microstructure properties. Much research has focused on the microstructure evolution of austenitic steels [17, 18], while few studies have been done on the non-austenitic steels. Based on the models for the DRX volume fraction and grain size of 33Cr23Ni8Mn3N austenitic steel, the hot compression was simulated and the corresponding experiment was carried out to analyze the microstructure evolution mechanism by Ji et al [19]. Besides, Cellular Automata (CA), Cellular Automata (CA) and Phase Field (PF) are commonly used to calculate the hardening and microstructure evolution by introducing a uniform strain field and dislocation model [20–22]. This work is to study the high-temperature austenite microstructure evolution of non-austenitic steel 5CrNiMoV during the non-isothermal hot forging. Dependent on the initial microstructure and chemical composition of the steel and the external hot deformation conditions, the main mechanism of microstructure evolution probably was dynamic recrystallization (DRX), grain rotation accommodating (GBs), the dislocation glide, and so on [23, 24]. To elaborate the deformation behaviors and microstructure evolution, the Finite Element Method (FEM) coupling simulation with hot deformation and microstructure was used to acquire the distribution of temperature, strain, strain rate and grain size. The typical region microstructure of the forgings was characterized and analyzed by electron back-scatter diffraction (EBSD).

2. Materials and methods

2.1. Materials

5CrNiMoV steel used in the research was sampled from a large forged die with a chemical composition (wt. %) of 0.54C - 0.25Si - 0.72Mn - 0.012P - 0.003S - 0.96Cr - 1.58Ni - 0.36Mo - 0.074V - (bal.) Fe. The heat treatment for the large forged die consists into heating up at 880 °C for 10h and cooling in air to 300 °C, heating up at 670 °C for 40h and cooling in the furnace to 150 °C, and finally cooling in the air to room temperature. From the large forged die, cylinder specimens of φ60 × 90 mm dimensions were machined.

2.2. Simulation

Multi-direction forgings have a significant impact on the microstructural evolution of 5CrNiMoV forged die. To study the deformation characteristics of 5CrNiMoV steel during non-isothermal hot forging, a coupling simulation with heat transfer and microstructure was carried out in DEFORM software. The constitutive equation and equations for the volume fraction of dynamic recrystallization (DRX) used in the simulation were acquired from the raw data in the article [25].

\[
\dot{\varepsilon} = 1.17 \times 10^{14} \left( \sinh (0.0128 \sigma) \right) ^{4.927} \times \exp \left( -390238 \right) \times \exp \left( -8.314 T \right)
\]

\[
X_{\text{DRX}} = 1 - \exp \left( -0.693 \left( \frac{\varepsilon - \varepsilon_0}{\varepsilon_{0.5}} \right) ^2 \right)
\]

\[
\varepsilon_{0.5} = 0.058 \times 10 ^ {0.1219} \exp \left( 26561.57 \times 8.314 T \right)
\]

\[
\varepsilon_p = 0.002335 \times 10 ^ {0.14057} \exp \left( 54855.76 \times 8.314 T \right)
\]

\[
\varepsilon_e = 0.727 \varepsilon_p
\]

\[
D_{\text{DRX}} = \left[ 3.0367 \times 10 ^ {25} \right] ^ {1/4.414} \times \exp \left( -444339 \right) \times \exp \left( -8.314 T \right)
\]

Other parameters, such as Young’s Modulus, Poisson’s ratio, thermal expansion, thermal conductivity and heat capacity, were calculated by JMatPro and input into the software. They were shown in table 1.

As shown in figure 1, the non-isothermal hot forging process consisted of one axial upsetting (Stage I) and two radial forgings (Stage II & III), and the reduction of three stages were respectively 50%, 33% and 40%. The coupled thermo-mechanical tetrahedral element was used to discretize the workpiece. The element number of the workpiece was 124974 and the size ratio of the maximum and minimum element was 2.5. The workpiece was modeled as a plastic body. Both the upper and lower dies were regarded as rigid bodies at a temperature of 300 °C. The workpiece, with an initial grain size 300 μm, was forged at a temperature of 1250 °C and the upper die moved at a 10 mm s⁻¹ velocity. Detailed simulation parameter settings were shown in table 2. Besides, to compare with the FEM simulation, the software MICRESS was used to simulate the austenite microstructure evolution of three typical regions (Center of the forging, 1/4 section of the forging and 1/8 section of the forging)
during the non-isothermal forging. The main input parameters of the phase-field model were the recrystallization energy, the recrystallization temperature range, the recrystallization nucleation position (grain boundary or in the grain), the maximum nucleation number and the grain boundary mobility. Based on the initial microstructure, the microstructure evolution of recrystallization can be simulated by inputting the corresponding temperature, strain and strain rate. The average grain size of the initial microstructure was set as 300 μm and the microstructure at the first pass deformation can be obtained by inputting the temperature, effective strain and effective strain rate at the point, which were acquired from the FEM simulation. To obtain the microstructure at the next pass deformation, the previous microstructure was firstly taken as the initial microstructure, and then the temperature, effective strain and effective strain rate at the corresponding point were input.

2.3. Experiment
To compare with the simulation results, two specimens were forged. The non-isothermal hot forging experiment was implemented in the hydraulic press with die temperatures of 300 °C. The specimens were preheated at 1250 °C for one hour before forging in a resistance furnace attached to the hydraulic press. Two cylindrical samples were with identical dimensions of 90 mm in height and 60 mm in diameter. For one sample, a height reduction of 50% was obtained in one single blow at a rate of 10 mm s\(^{-1}\). For the other, three forging passes of 50% axial upsetting, 33% radial forging, and then 40% radial forging after turning 90 degrees, were carried out at the same forging rate, respectively. After the hot forging process, the specimens were cooled in the air. As shown in figure 2, two specimens were sectioned in half, parallel to the upsetting direction (UD) and three

| Table 1. Thermal and materials properties of 5CrNiMoV steel. |
|-----------------|-----------------|
| Item            | Value           |
| Young’s modulus/(Mpa) | 90726           |
| Poisson’s ratio  | 0.36            |
| Thermal expansion/(1/°C) | 163994         |
| Thermal conductivity/(N/sec/°C) | 30.60        |
| Heat capacity/(N/mm²/°C) | 4.90           |
| Emissivity      | 0.70            |

| Table 2. Parameter settings. |
|-----------------------------|
| Parameters                  | Value     |
| Initial temperature of the workpiece/°C | 1250      |
| Initial temperature of the upper die/°C | 300       |
| Initial temperature of the bottom die/°C | 300       |
| Initial grain size/μm | 300       |
| Environment temperature/°C  | 30         |
| Convection coefficient with the environment/N/(s.mm.°C) | 0.02     |
| Convection coefficient with the die/N/(s.mm.°C) | 11.0      |
| Radiation heat transfer coefficient of the workpiece | 0.9       |
| Friction coefficient        | 0.3        |
| Number of elements of the workpiece | 124974   |
| Number of simulation steps  | 228        |
| Speed of the upper die/mm.s\(^{-1}\) | 10         |
| Step increment control/mm/step | 0.3       |
typical regions (center of the forging, 1/4 section of the forging and 1/8 section of the forging) were taken to analyze the microstructure evolution. Microstructure of as-received and as-forged specimens was compared through EBSD analyses. The samples were firstly mechanically polished, then electrochemically polished in 10% nitric acid alcohol solution at room temperature for 10s with a 30V applied potential. The EBSD analyses of typical regions were carried out in a ZeissAuriga field emission gun Scanning Electron Microscope (SEM) equipped with Oxford Instruments Nordlys Nano EBSD system. The acquired data were further analyzed by Channel 5 software.

3. Results and discussion

3.1. Analysis of the simulation results

Figure 3 shows the strain distribution across the specimen section in the course of the upsetting process (stage I), the first radial forging process (Stage II) and the second radial forging process (stage III). As shown in figure 3, the distribution of effective strain was not uniform, showing the characteristic gradient pattern (‘X’). The highest value of effective strain was always discovered in the center of the forgings. As the deformation passes increased, the strain gradient increased gradually. The dead zones of the effective strain were not only for the friction with the dies, but also for the cooling effect of the dies. The regions near the contact surfaces showed large areas with a significant drop in strain. The inhomogeneous deformation was generally developed due to the non-isothermal conditions on the hydraulic press.

Besides, as shown in figure 4(a), three typical points (the center of the section, 1/4 of the section and 1/8 of the section) were chosen to analyze the changes of temperature, effective strain, effective strain rate, dynamic recrystallization fraction and average grain size during the forging process. Figure 4(b) depicted the thermal evolution at the tracking points. A small temperature increase was calculated during the upsetting stage, with the 1250 °C reaching up to about 1270 °C, due to the relatively high strain accumulated in the workpiece. As the
forging progressed, the temperature gradually decreased, but it was always higher than the required forging-ending temperatures. It can be seen in figure 4(c) that the effective strain of the three typical points gradually increased during the multi-pass forging process and the effective strain of the three typical points was 3.49, 2.97 and 2.19 at the end of the forging. As shown in figure 4(d), during the upsetting stage, the effective strain rate of the center point was the largest, and the other two points were relatively small. During the radial forging process, the evolutions of the effective strain rate in the three points were complicated. From figures 4(e) and 4(f), the complete dynamic recrystallization occurred in the three points after the upsetting, and the austenite grains were remarkably refined to about 65 μm. During the radial forging process, three typical points also underwent dynamic recrystallization, and the average austenite grain size was 26.55, 26.43 and 32.68 μm, respectively.

As shown in figure 5, the simulated microstructure of the three typical regions during the non-isothermal forging is acquired and the average grain size can be directly read in MICRESS. The average grain size of typical
region P1 at three stages was about 66 μm, 72 μm and 26 μm, the average grain size of typical region P2 at three stages was about 66 μm, 63 μm and 27 μm and the average grain size of typical region P3 at three stages was about 67 μm, 62 μm and 32 μm. Compared the austenite grains at stage I, the grains at stage III were significantly refined with the deformation increasing. However, the austenite grains at stage II, especially the center of the forging (P1), coarsened mainly due to small strain in the vertical direction. Overall, the grain evolution law was in accordance with the FEM simulation.

3.2. Microstructure evolution

3.2.1. Analysis of the initial microstructure
Software Channel 5 is used to process the acquired EBSD data. The initial microstructure of as-received steel obtained by the EBSD technique is shown in figure 6. According to the phase map (figure 6a), the initial microstructure consisted of a single α-Fe phase. The microstructure was mainly composed of equiaxed grains (average aspect ratio was 1.8). The average grain size of as-received steel 5CrNiMoV was 5.01 μm and the fraction of the large grains (the mean diameter greater than 10 μm) was 12.6%. Grain boundaries were usually distinguished by setting the boundaries misorientation value in Channel 5. Boundaries with misorientation angles between 2° and 5° were defined as low angle boundaries (LAGBs). Boundaries with misorientation greater than 15° were high angle boundaries (HAGBs) and others were medium grain boundaries (MAGBs). As shown in figure 6(b), the fractions of HAGBs and MAGBs were 0.517 and 0.138, respectively. It suggested that there were still some defects forming for 5CrNiMoV steel after heat treatment. The fraction of LAGBs was only 0.345, which was mainly caused by the movement of dislocation. The inverse pole figure (IPF) map (figure 6(c)) included orientation information and different colors represented different crystal orientations. The misorientation distribution was shown in figure 6(d) and the average correlated misorientation angles were calculated as 13.3°. Usually, 'Correlated' means misorientation calculated according to the neighbouring points; 'Uncorrelated' implies the misorientation calculated based on the random points; and 'Random' means the misorientation calculated based on the random misorientation distribution [26]. As shown in figure 6(d), the ‘Correlated’ grain boundaries took the largest fraction in a misorientation angle around 2° (LAGBs) and a peak during the transition from LAGBs to HAGBs appeared at about 55°. Moreover, the ‘Uncorrected’ and the ‘Random’ curves were almost coincident, indicating that there were no preferred orientations or formation of textures in the base steel.
3.2.2. Evolution of parent austenite grain morphology and size

To understand the impact of deformation parameters (temperature, strain and strain rate) on microstructure evolution of 5CrNiMoV steel during the non-isothermal forging process, two specimens were forged and 3 typical regions (figure 2) distributed across the thickness were considered for microstructural analysis.

The forging process of 5CrNiMoV steel is conducted in the austenite temperature region, and the prior austenite microstructure evolution has a great influence on the subsequent microstructure properties. However, the air-cooled microstructure after hot forging is generally martensite due to its favourable hardenability. Lots of techniques have been used to illustrate the parent grain microstructure during the hot forming or after the transformation. The research on the reconstructed parent grains acquired by the orientation of the daughter phase has shown major developments recently [27]. In the study, Nishiyama-Wassermann (NW) was employed as the crystallographic orientation relationship (OR) that described the transformation. Figures 7 and 8 show the reconstructions of the prior austenite grain structure in ARPGE. As shown in figure 7, the parent austenite grains remained coarse at the forging stage I and the average grain size of the forging at the half (P1), the quarter (P2) and the eighth (P3) was found to be 77.7, 78.4 and 104.5 μm. With the further straining in figure 8, the parent austenite grains were remarkably refined at the forging stage III. The equiaxed grains were developed and the grains were significantly refined, especially in the center of the forging. The average parent austenite grain size of the forging at the half (P1'), the quarter (P2') and the eighth (P3') at the forging stage III was respectively calculated as 43.8, 44.7 and 51.3 μm. It was found that the average grain sizes measured by the experiment were slightly larger than the simulation results. It was probably that meta-dynamic recrystallization (MDRX) and static recrystallization (SRX) were not taken into account during the simulation. Moreover, the grains would coarsen during the air cooling after hot forging in the experiment. Besides, compared with stage I, the distribution of grain size and grain shape in stage III was much more uniform and equiaxed (figure 8), which may attribute to the multiple dynamic recrystallization and growth of the recrystallization grains at room temperature after hot forging.

To visualize the orientation relationship between the reconstructed parent austenitic phase and the martensitic phase obtained at room temperature, the crystallographic orientation of the martensite block with reconstructed austenite grain boundaries in bold black line was shown in figures 9(a), 9(b). It was suggested that the martensitic packets and blocks were coarsening with increasing austenitizing temperature and the increasing
prior austenite grains. A parent austenite grain was usually divided into several packets with the same habit plane. These packets were separated by grain boundaries with misorientation angles higher than 15°, while martensitic laths are usually separated by low angle grain boundaries [28]. Figure 9(c) showed the misorientation profile along with line A in a parent austenite grain, which indicated that several paralleled blocks were arranged in the parent austenite grain and several paralleled blocks formed a packet.

3.2.3. Evolution of parent austenite grain boundary

Figure 10 shows the grain boundary (GBs) maps of typical regions after the whole forging. In these GBs maps, the black, green, blue and red lines represent HAGBs, LAGBs, MAGBs and CSL Σ3 boundaries, respectively. It was observed from figure 10 that the volume fraction of the HAGBs of the three typical regions was 0.954, 0.923 and 0.877, which gradually decreased with the decrease of the strain (P1’-P3’). It suggested that after the whole forging there were few defects, especially in the center (P1’) of the forging, due to the happen of complete DRX. At the eighth (P3’) of the forging, the grains were filled with some low angle grain boundaries, which probably resulted from the movement of the accumulated dislocations during the non-isothermal forging. The fraction of 10°–15° misorientations decreased from 0.138 of the as-received steel to about 0.07 of hot-forgings, which indicated that continuous dynamic recrystallization (CDRX) had little impact on microstructure evolution. Moreover, as shown in figure 10, it was observed that there were many bulging grain boundaries with HAGBs. It was induced by deformation strain and brought about grain boundary sliding. Besides, the serrated HAGBs were the potential nucleation sites of the dynamic recrystallization (DRX) by grain boundary bulging [29], which implied that the effect of discontinuous dynamic recrystallization (DDRX) had a progressive enhancement. Therefore, during the non-isothermal forging, DDRX was concluded to be the main nucleation and grain growth mechanism of DRX [26].

Besides, a relatively large fraction (about 0.159) of Σ3 CSL boundaries (60° around <111> axis) could be observed in red lines in figure 10(c). It is widely known that Σ3 boundaries usually show relatively lower energy and possess better functional properties, such as thermal stability and corrosion resistance [30, 31], than the common HAGBs. It can significantly put down to hardening, and become the effective barriers to restrain dislocation slip [30]. In this case, the long distance between the <112> partial dislocations on {111} slip planes restricted the screw dislocations slipping and edge dislocations climbing, which lead to the formation Σ3 CSL boundaries. Besides, it was found that Σ3 CSL boundaries were mainly distributed around the {011} <001>
oriented grains, which verified that Σ3 CSL boundary was a migration grain boundary easy to form Goss texture and it was favorable for forming Goss texture, as shown in Figure 13(d).

Figures 11(a)–11(c) shows the misorientation-angle distribution of the typical regions (P1’, P2’ and P3’) of 5CrNiMoV steel after the whole forging. The misorientation distribution was more uniform from the eighth to the center (P3’–P1’) of the forging. Compared with the as-received steel, the highest peak corresponded to much more HAGBs induced by hot deformation and recrystallization. At the eighth (P3’), a peak concentrated on misorientation angle 60° with the main rotation axis of <111>, which might be due to the deformation twins. At the center (P1’), there was a relatively smooth transition from LAGBs to HAGBs for the misorientation distribution and a distinct peak around 45° was found. There were smaller differences between the ‘Uncorrected’ and the ‘Random’ curves (figure 11(a)), which may be for that larger strain and strain rate made the parent austenite grains more uniform under dynamic recrystallization. Relatively large differences between the ‘Uncorrected’ and the ‘Random’ curves were observed in figure 11(c), which suggested that a preferred orientation or textures formed during the multi-directional forging. Further analysis of the texture type would be made through pole figures. Besides, the average misorientation angles from the center to the eighth (P1’–P3’), was calculated as 38.9°, 38.4° and 40.7°, respectively, and increased rapidly from 13.3° obtained from the initial microstructure, which was probably ascribed to the occurrence of DRX [29, 31].

3.2.4. Analysis of Schmid factor during forging
The grain Schmid factor distribution at different regions (P1’, P2’, and P3’) is presented in Figure 12. The value of the Smith factor corresponding to different colors referred to the color gradient diagram below. It can be observed that the grain of the three images is basically covered by red, red gradient, yellow and the minimum value was 0.29. It indicated that the grain orientation of crystal grains at different regions was favorable for slippage, which was for the fact that the austenite grains were face-centered cubic structures, theoretically having 12 slip systems. Areas with smaller Schmid factor were not favorable for slipping and were often accompanied by LAGBs. It can be found that the Schmid factors in the center (P1’) of the forging were higher and more uniform than that at the eighth (P3’) of the forging and the fractions of Schmid factor above 0.4 was 0.962, which contributed to grain boundary movement and grains rotation. Besides, Schmid factor of small grains was relatively higher, which indicated that the ability of small grains to coordinate deformation was stronger.

Figure 11. Misorientation angle distributions of 5CrNiMoV steel: (a), (b) and (c) respectively for typical region P1’, P2’ and P3’ at the last radial forging process (stage III).

Figure 12. Schmid factor map of 5CrNiMoV steel: (a), (b) and (c) respectively for typical region P1’, P2’ and P3’ after the non-isothermal forging.
3.2.5. Texture evolution during the forging

Many studies concerning the texture evolution of bcc material have been performed during the cold forming. However, the deformation texture during hot formation has not yet been clarified, especially for the non-austenitic steels [32–34]. As shown in figure 13(a), the maximum texture intensity of the as-received steel was 4.03 times the random distribution in [100] pole figure, and a weak texture with the main texture components Brass [011]<211> and S [123]<634> was obtained. It was observed in figure 13 that the texture changed significantly after the non-isothermal forging and the texture type differed remarkably at different regions of the forging. As shown in figure 13(b), the maximum texture intensity was 5.77 and obvious recrystallization texture was observed in the center (P1’) of the forging. From the [100] pole figure in figure 13(b), a Wulff net was used to cover the pole figure and it was obtained that the maximum of random distribution (MAD) distributed at 0° and 90° in the normal direction (ND), which indicated a preferred orientation of (200) face. In the rolling direction (RD), the

![Figure 13. Pole figures of the as-received steel and post-forged specimens in typical regions: (a) the as-received steel; (b), (c), (d) respectively for the typical region P1’, P2’ and P3’ after the non-isothermal forging.](image-url)
The simulation results showed that the inhomogeneous deformation was generally developed for the non-isothermal conditions of the hydraulic press. The distribution of effective strain in the specimen showed the characteristic gradient pattern (X') and the strain gradient increased gradually with the strain. Analysis of the three typical tracking points showed that the temperature of the typical points slowly dropped and was higher than the required forging-ending temperatures, and there was always a temperature gradient. The effective strain of the three typical points gradually increased during the hot forging, while the evolution of the strain rate was relatively complicated for the multiple forging. Dynamic recrystallization occurred in all three typical points, and the austenite grains were significantly refined.

2. The parent austenite grains were more uniform and equiaxed with the deformation increased, which may attribute to the multiple dynamic recrystallization and growth of the recrystallization grains. The average parent austenite grain size obtained in the experiment was larger than the simulation results mainly due to the grain coarsening during the air cooling after hot forging. The distribution of grain boundaries showed that there were few defects due to the occurrence of complete DRX. Small amounts of MAGBs indicated that continuous dynamic recrystallization (CDRX) had little impact on microstructure evolution and many bulging grain boundaries showed that DDRX could become the main nucleation and grain growth mechanism.

### Table 3. Volume fraction of main texture components of the steel.

| Sample                  | Cube/\%  | Goss/\% | Brass/\% | Copper/\% | S/\%   |
|-------------------------|----------|---------|----------|------------|--------|
| As-received steel       | 2.38     | 4.21    | 16.1     | 5.34       | 24.9   |
| Center of the forging(P1) | 17.1     | 2.55    | 3.92     | 4.56       | 5.1    |
| Quarter of the forging(P2) | 32.9     | 4.82    | 4.23     | 4.25       | 11.3   |
| Eighth of the forging(P3)   | 11.3     | 16.3    | 15.2     | 8.99       | 14.3   |

MAD also distributed at 0° and 90°, which implied the [200] preferred orientation. Thus, the texture component in the hot-forged microstructure of 5CrNiMoV steel was expressed as Cube(001) <100>, which probably resulted from the recrystallization and grain growth [35, 36]. The acquired texture components of typical regions P1', P2' and P3', such as Cube(001) <100>, Goss(011) <100>, Brass(011) <211> and S(123) <634>, were shown in figure 13. For better comparison, Texture Component of the software Channel5 was employed to quantitatively calculate the content of various texture components and the volume fraction of main texture components was listed in table 3.

As is observed in figure 13 and table 3, the as-received steel depicted a texture with the main texture components of Cube(2.38%), Goss(4.21%), Brass(16.1%), Copper(5.34%), and S(24.9%). Compared with the texture of the as-received steel, the texture of the forging changed remarkably during the multi-directional non-isothermal forging and the Cube texture component was strengthened, especially in the center(P1') and quarter (P2') of the forging. Besides, deformation texture components, such as S, Brass and Copper, still existed due to relatively small strain and strain rate at the eighth(P3') of the forging. Moreover, comparing the texture components of the three typical regions P1', P2' and P3', it was known that the center of the forging(P1') showed typical recrystallization texture without obvious deformation texture component. However, in the quarter(P2') and the eighth(P3') of the forging, there were typical deformation texture components, such as S and Brass component, in addition to the recrystallization texture.

Considering the deformation characteristics, the final texture components in the different regions of the forging can reveal the evolution of texture in the forging during hot deformation. As is known that the texture component S (123) <634> in the deformed forging is favourable for the rapid development of Cube texture component. However, the first disappearing texture component was the Brass and Copper, and then it was the S texture component, which was consistent with the findings in many other papers [37, 38]. Besides, the Goss texture component, found in the eighth(P3') of the forging, was attributed to the grain growth and secondary recrystallization.

### 4. Conclusions

In the present work, the deformation characteristics of 5CrNiMoV steel forged under non-isothermal conditions was analyzed through FEM simulations. The hot forging experiments were carried out in a hydraulic press and the microstructural evolutions of typical regions were studied by scanning electron microscope equipped with an EBSD camera. The main results are as following:

1. The simulation results showed that the inhomogeneous deformation was generally developed for the non-isothermal conditions of the hydraulic press. The distribution of effective strain in the specimen showed the characteristic gradient pattern (X') and the strain gradient increased gradually with the strain. Analysis of the three typical tracking points showed that the temperature of the typical points slowly dropped and was higher than the required forging-ending temperatures, and there was always a temperature gradient. The effective strain of the three typical points gradually increased during the hot forging, while the evolution of the strain rate was relatively complicated for the multiple forging. Dynamic recrystallization occurred in all three typical points, and the austenite grains were significantly refined.

2. The parent austenite grains were more uniform and equiaxed with the deformation increased, which may attribute to the multiple dynamic recrystallization and growth of the recrystallization grains. The average parent austenite grain size obtained in the experiment was larger than the simulation results mainly due to the grain coarsening during the air cooling after hot forging. The distribution of grain boundaries showed that there were few defects due to the occurrence of complete DRX. Small amounts of MAGBs indicated that continuous dynamic recrystallization (CDRX) had little impact on microstructure evolution and many bulging grain boundaries showed that DDRX could become the main nucleation and grain growth mechanism.
3. The study of Schmid factor distribution showed that the grain orientation of crystal grains at different regions was favorable for slippage, which was for the fact that the austenite grains were face-centered cubic structures. The hot deformation texture changed significantly after the non-isothermal forging and the texture component differed remarkably at different regions of the forging. The main texture component in the hot-forged microstructure of 5CrNiMoV steel was Cube{001} < 100 >, which were resulted from the recrystallization and grain growth. A small amount of deformation texture, such as S, Brass and Copper, remained at the eighth (P3’) due to relatively small strain and strain rate during hot forging.

Acknowledgments

This work was financially supported by the National Key Research and Development Program of China (2017YFB0701803, 2017YFB0701801).

ORCID iDs

Zhiqiang Hu @ https://orcid.org/0000-0002-9619-8183

References

[1] Lou Z, Liu H, Yang G et al 2019 Microstructural characteristics and mechanical properties in the laser beam welded joints of high-strength microalloyed steel J. Mater. Eng. Perform. 28 3724–36
[2] Batista M N, Marinelli M C and Alvarez-Armas I 2019 Effect of initial microstructure on surface relief and fatigue crack initiation in AISI 410 ferritic-maritensitic steel Fatigue Fract. Eng. Mater. Struct. 42 61–8
[3] Rahimi S, Konkova T N, Violatos I et al 2019 Evolution of microstructure and crystallographic texture during dissimilar friction stir welding of duplex stainless steel to low carbon-manganese structural steel Metallurgical And Materials Transactions A-Physical Metallurgy And Materials Science 50A 684–87
[4] Ramesh R, Dinaharan I, Kumar R et al 2019 Microstructure and mechanical characterization of friction-stir-welded 316L austenitic stainless steels J. Mater. Eng. Perform. 28 498–511
[5] Eskandari M, Mohtadi-Bonab M A, Zarei-Hanzaki A et al 2018 Effect of hot deformation on texture and microstructure in Fe-Mn austenitic steel during compression loading J. Mater. Eng. Perform. 27 1555–69
[6] Mandal G K, Rajinkanth V, Kumar S et al 2017 Microstructure evolution during hot deformation of a micro-alloyed steel Trans. Indian Inst. Met. 70 1019–33
[7] Ahmadabadi R M, Naderi M, Mohandesl i et al 2018 Dynamic recrystallization behavior of aisi 422 stainless steel during hot deformation processes J. Mater. Eng. Perform. 27 560–71
[8] Huo Y, Wang B, Lin J et al 2017 Hot compression deformation behavior and microstructure evolution rule of a high-speed railway axle steel Indian J. Eng. Mater. Sci. 24 447–54
[9] Li J, Zhao G, Chen H et al 2018 Hot deformation behavior and microstructural evolution of as-cast 304L antibacterial austenitic stainless steel Mater. Res. Express 5 0254052
[10] Dong J, Li C, Liu C et al 2017 Hot deformation behavior and microstructural evolution of Nb-V-Ti microalloyed ultra-high strength steel J. Mater. Res. 32 3777–87
[11] Luo R, Zheng Q, Zhu J et al 2017 Dynamic recrystallization behavior of Fe-20Cr-30Ni-0.6Nb-2Al-Mo alloy Rare Met. 38 181–8
[12] Yang Q, Li C and Zhu M 2019 Modeling of the dynamic recrystallization kinetics of a continuous casting slab under heavy reduction Metallurgical And Materials Transactions A-Physical Metallurgy And Materials Science 50A 357–76
[13] Perez M 2018 Microstructural evolution of Nimonic 80a during hot forging under non-isothermal conditions of screw press J. Mater. Process. Technol. 252 45–57
[14] Chen K, Wu J, Shi H et al 2015 Transition of deformation behavior and its related microstructure evolution in Nimonic 80A under hot-to-warm working Mater. Charact. 106 179–84
[15] Krishna S, Korthick N K, Avula A K et al 2019 Microstructural evolution of inconel 718 during pancake forging Trans. Indian Inst. Met. 72 1485–8
[16] Konkova T, Rahimi S, Mironov S et al 2018 Effect of strain level on the evolution of microstructure in a recently developed AD730 nickel based superalloy during hot forging Mater. Charact. 139 437–45
[17] Eskandari M, Mohtadi-Bonab M A, Zarei-Hanzaki A et al 2018 Effect of hot deformation on texture and microstructure in Fe-Mn austenitic steel during compression loading J. Mater. Eng. Perform. 27 1555–69
[18] Nayan N, Gurao N P, Mutyu S V S N et al 2015 Microstructure and micro-texture evolution during large strain deformation of inconel alloy IN718 Mater. Charact. 110 236–41
[19] Ji H, Cai Z, Pei W et al 2020 DRX behavior and microstructure evolution of 33Cr23Ni8Mn3N: Experiment and finite element simulation Journal Of Materials Research And Technology-JMR&T 9 4340–55
[20] Zhou G, Li Z, Li D et al 2017 A polycrystal plasticity based discontinuous dynamic recrystallization simulation method and its application to copper Int. J. Plast. 91 98–76
[21] Li H, Sun X and Yang H 2016 A three-dimensional cellular automata-crystal plasticity finite element model for predicting the multiscale interaction among heterogeneous deformation, DRX microstructural evolution and mechanical responses in titanium alloys Int. J. Plast. 87 154–80
[22] Yoshimoto C and Takaki T 2014 Multiscale hot-working simulations using multi-phase-field and finite element dynamic recrystallization model JSJ Int. 54 452–9
[23] Yang Y, Wu S Q, Li G P et al 2010 Evolution of deformation mechanisms of Ti-22.4Nb-0.73Ta-1.34O alloy during straining Acta Mater. 58 2778–87
[24] Zhang W J, Song X Y, Hui S X et al. 2014 Tensile behavior at 700 degrees C in Ti-Al-Sn-Zr-Mo-Nb-W-Si alloy with a bi-modal microstructure. Materials Science And Engineering A-Structural Materials Properties Microstructure And Processing 595 159–64

[25] Hu Z, Wang K and Yang Y. 2020 Deformation behavior and evolution of austenite microstructure and texture during hot compression of 5CrNiMoV Steel. Metallography, Microstructure, and Analysis 9 576–87

[26] Yan Z, Wang D, He X et al. 2018 Deformation behaviors and cyclic strength assessment of AZ31B magnesium alloy based on steady ratcheting effect. Materials Science And Engineering A-Structural Materials Properties Microstructure And Processing 723 212–20

[27] Back J G and Engberg G. 2017 Investigation of parent austenite grains from martensite structure using EBSD in a wear resistant steel. Materials 10 453

[28] Zheng Y, Wang F, Li C et al. 2018 Effect of microstructure and precipitates on mechanical properties of Cr-Mo-V alloy steel with different austenitizing temperatures. ISIJ Int. 58 1126–35

[29] Liu Z, Li P, Xiong L et al. 2017 High-temperature tensile deformation behavior and microstructure evolution of Ti55 titanium alloy. Materials Science And Engineering A-Structural Materials Properties Microstructure And Processing 680 259–69

[30] Huang W, Yuan S, Chai L et al. 2019 Development of grain boundary character distribution in medium-strained 316L stainless steel during annealing. Met. Mater. Int. 25 364–71

[31] Sharma N K and Shekhar S. 2017 Microstructure and property evolution for hot-rolled and cold-rolled austenitic stainless steel 316L. Trans. Indian Inst. Met. 70 1277–84

[32] Rout M, Biswas S, Ranjan R et al. 2018 Deformation behavior and evolution of microstructure and texture during hot compression of AISI 304LN stainless steel. Metallurgical And Materials Transactions A-Physical Metallurgy And Materials Science 49A 864–80

[33] Peng C, Callaghan M D and Li H. 2015 Post-deformation microstructure and texture characterization of Fe-18Mn-0.6C-1.5Al TWIP Steel. Steel Res. Int. 86 1461–8

[34] Barbier D, Gey N, Allain S et al. 2009 Analysis of the tensile behavior of a TWIP steel based on the texture and microstructure evolutions. Materials Science And Engineering A-Structural Materials Properties Microstructure And Processing 500 196–206

[35] Omale J I, Ohaeri E G, Szpunar J A et al. 2019 Microstructure and texture evolution in warm rolled API 5L X70 pipeline steel for sour service application. Mater. Character. 147 453–63

[36] Lu X, Fang F, Zhang Y et al. 2018 On goss orientation in strip cast grain-oriented silicon steel. Steel Res. Int. 89 1700403

[37] Liu J, Sha Y, Hu K et al. 2014 Formation of cube and goss texture after primary recrystallization in electrical steels. Metallurgical and Materials Transactions A-Physical Metallurgy AND Materials Science 45A 134–8

[38] Betanda Y A, Helbert A, Brisset F et al. 2014 Influence of$ N Fe48%Ni alloys tapes. Adv. Eng. Mater. 16 933–9