Mechanism of crack initiation and propagation of 316LN stainless steel during the high temperature tensile deformation

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Abstract

Due to the difficult deformability and cracking in forging process, the mechanism of crack initiation and propagation of 316LN stainless steel during the high-temperature tensile deformation are investigated. The thermal tension experiment is carried out on the Gleeble-1500D machine to reveal the relation between the crack propagation path and the degree of the dynamic recrystallization. By the microstructure analysis of the scanning electron microscopy (SEM) on the fracture behavior of various strains and strain rates, the dynamic recovery and dynamic recrystallization occur simultaneously, the formation of cracks was accompanied by recrystallization, and the crack propagation show both scale effect and interface effect. The intergranular ductile fracture perpendicular to the principal stress direction is found to be closely relative to the recrystallization microstructure, where the core position for crack initiation is around the intersection of the tricrystal boundaries. It is found that the tricrystal boundary cracking from the inclusion of aluminum oxide ($\text{Al}_2\text{O}_3$) is critical to the decrease of the fracture properties and plasticity, while a certain degree of dynamic recrystallization and the dynamic recovery happened simultaneously from 1000 °C to 1200 °C benefits for the plasticity. Thus, the interface between the austenite matrix and the alumina-based brittle inclusions is easy to form local stress concentration on the tricrystal boundary due to the stacking of dislocations, which becomes the core of crack initiation and thus increases the tendency of forging cracking at high temperature.

1. Introduction

AP1000 is a two-loop pressurized water reactor (PWR) developed by Westinghouse. The AP1000 third-generation main pipeline of 316LN steel in usage of the nuclear power station connects the reaction vessel and the steam generator [1], which is the core equipments of the nuclear station in China thus to be called as the aorta of the nuclear power plant. This kind of pipeline made of 316LN stainless steel is commonly subjected to the atmospheric pressure of 1500 MPa, at the high temperature of 616 K, the alternating stress at the interface at the high flow rate in the corrosion of radioactive medium in working environment [2, 3], thus it puts forward the higher and more strict requirement of the microstructure consistency and structural reliability [4].

316LN austenitic stainless steel composed of the ultra-low carbon and nitrogen containing is commonly used to product the integrally forged AP1000 pipe. The AP1000 nuclear power main pipe has a large diameter, and the two flange joints need to be integrally formed in the forging process. However, this kind of 316LN austenitic stainless steel has a narrow forging temperature range and the strong crack sensitivity [5–7], which can not be strengthened directly by heat treatment [8–11], as there is no phase transformation during heat treatments. The tensile deformation is accompanied with the recrystallization [12, 13], creep [14] and rupture behavior. Previous researches on 316LN stainless steel mainly focus on the microstructure and mechanical properties promoted only by forging due to the poor resistance to intergranular stress corrosion cracking,
through both simulation techniques [15–17] and experimental measurements [18–22]. However, the previous work investigations only concentrate on the effect of strain rate and notch size on fatigue crack and fracture in forming process [23, 24], the cracking initiation mechanism is unrevealed in need of a solution. It is therefore urgent to find the cracking initiation mechanism of 316LN steel to provide useful hot deformation process parameters during the forging process, which is also the key to realize the performance controlling in the practical production of AP1000 pipeline.

In our previous study, the formation of cracks in 316LN including the nucleation, growth, and aggregation of microscopic cavities during the stretching process under high temperature has been studied [11]. It was also found that stress triaxiality plays an important role in the prediction of the destruction of the material.

In this study, thermal tensile experiments were carried out on the Gleeble-1500D so as to reveal the crack initiation mechanism and propagation mechanism of 316LN during high-temperature plastic deformation. The relationship between crack propagation and dynamic recrystallization are studied and analyzed by means of the metallographic optical microscope (OM), scanning electron microscopy (SEM) and energy dispersive spectrum (EDS). The formation mechanism of cracks under external force and the microstructure evolution mechanism were explored by controlling the dependent variables in the hot tensile deformation process. With the help of mesoscopic damage mechanics, this study aims to reveal the mechanism of crack initiation and propagation of 316LN austenitic stainless steel during the high-temperature plastic deformation, to fatherly solve the technical problem of the difficulty to deform and cracking in forging process.

2. Experimental material and methods

The chemical composition of the original cast ingot was tested, shown in table 1.

The plate sampling and the initial microstructure of specimens before tensile process was given in figure 1. From figure 1(a), the thickness of the initial steel plate after the forging and rolling was greater than 50 mm, and the test surface was chosen on the location at a quarter of the width of the steel sheet and on the middle position from the surface to the center. Microstructure of 316LN steel at room temperature distributed a lot of austenite grains and the grain size grade of averaged austenite grain was about 4.3, containing a large number of twins in the microstructure, seen in figure 1(b).

The total of 42 cylinder tensile specimens of Φ10 × 121.5 mm forged 316LN stainless steel were used. Each two samples were tested by the same condition, then the mechanical properties were obtained and averaged. Sample were heated to 1200 °C at the speed of 5 °C/s on the Gleeble-1500D at first, kept for 5 min, and then cooled to the experimental temperature at a cooling rate of 5 °C s⁻¹. The tensile specimen was subjected to a high-temperature tensile test within the region of 900 °C–1200 °C (900 °C, 950 °C, 1000 °C, 1050 °C, 1100 °C, 1150 °C and 1200 °C, separately) at a strain rate of 0.005 s⁻¹, 0.05 s⁻¹ and 0.5 s⁻¹. Besides, the deformed samples at different temperatures of 900 °C, 1000 °C, 1100 °C and 1200 °C were axially split in the direction of stretching.

Table 1. The chemical composition of 316LN (17Cr-12Ni-2Mo-0.1 N) (%).

| Element | C    | Mn   | Si   | P    | S    | Cr   | Ni   | Mo   | Cu   | N    |
|---------|------|------|------|------|------|------|------|------|------|------|
|         | 0.020| 2.00 | 0.7  | 0.025| 0.005| 16.00| 11.00| 2.00 | 0.10 | 0.10 |

(a) Sampling with the thickness over than 50mm  (b) Initial microstructure of specimens

Figure 1. The plate sampling and the initial microstructure of specimens before tensile process.
tensile fracture. After the procedures of polishing and corrosion (the corrosion solution is 4 g CuSO₄ + 20 ml HCl + 20 ml H₂O), the metallographic microstructure near the fracture were observed. Finally, fracture microstructures during the high-temperature tensile process were analyzed by means of the analysis of OM, SEM and EDS.

3. Experimental results and analysis

3.1. Characteristics of experimental stress-strain curves

Figure 2 shows the true stress-strain curves of 316LN specimen at different temperatures (900 °C, 1000 °C, 1050 °C, 1100 °C, 1150 °C, 1200 °C) and different strain rates (0.005 s⁻¹, 0.05 s⁻¹, 0.5 s⁻¹).

As is shown in figure 2, the flow stress and the shape of the true strain-stress curves are sensitive dependent on temperature and strain rate. The rheological stress curves of 316LN steel under high temperatures show the typical characteristics of dynamic work-hardening and dynamic recrystallization. As can be seen from figures 2(a)–(f), for a fixed temperature (within the region of 900 °C–1200 °C), the flow stress generally increases with the increases of strain rate due to the increase of dislocation density and the dislocation multiplication rate. It can be seen that the flow stress firstly increases to the peak stress and then decreases with the increase of strain. Meanwhile, under the same strain, the higher the temperature is, the lower the stress is, seen in figures 2(a)–(f). Parameters of peak stress σ₀ₚ, peak strain ε₀ₚ, and rupture strain εₚ of 316LN steel are shown in table 2.

It can be seen from table 2 that the peak stress σ₀ₚ and the steady state stress σₛₚ decrease with the decrease of strain rate at the same temperature and strain rate. Besides, the σ₀ₚ and σₛₚ are the lowest at 1200 °C, compared with the other temperatures of 900 °C–1150 °C. It should be noticed that the fracture strain decreases with the decrease of strain rate. This is because the decrease of strain rate prolongs the diffusion time and increases the precipitation of carbides and second phase particles that play the role of precipitation and solution strengthening [23]. With the increase of strain rate from 0.005 s⁻¹ to 0.5 s⁻¹, the activation energy decreases, there is not enough time for the matrix to soften and for the internal dislocations to disappear, resulting in the increased odds of microcracks [9, 23].

3.2. Metallographic microstructure analysis of the fracture region

Microstructures near the fracture region at 900 °C, 1050 °C, 1150 °C and 1200 °C at the strain rate of 0.5 s⁻¹ are shown in figures 3–6, respectively.

It can be seen from figure 3(a) that, during the tensile process at 900 °C, the austenite grains are elongated in the stretching direction, where cracks are initiated and expanded along the austenite grain boundaries [22], exhibiting irregular distribution. There exists different crack growth under various tensile directions. From figure 3(b), the microvoids are irregular, and the cracks formed along large grain boundaries propagate in the axial direction of external stress. It should be noticed that the tensile specimen only undergo dynamic recovery during the elastoplastic deformation process till to fracture [16], due to the relative low deformation temperature and the high deformation rate. The strain energy accumulated during the tensile process is relatively lower, without dynamic recrystallization occurred.

As is shown in figure 4(a) at 1050 °C, the nucleation, initiation and aggregation of the cracks are still along the austenite grain boundary without directivity, and the averaged size of the formed voids gradually increases. It is because that, with the increase of the deformation temperature, the ability of atomic diffusion in the grain boundary is enhanced. From figure 4(b), cracks surrounded by grain boundaries is accompanied with the dynamic recrystallization, which disperses the crack propagation path, thus weakening the driving force of crack propagation to some extent [21]. Due to the increase of deformation temperature that is conducive to the occurrence of dynamic recrystallization [13, 25], local dynamic recrystallization occurred in the plastic deformation region near the fracture surface.

During the tensile process at 1150 °C, the dynamic recrystallization occurs sufficiently in the tensile deformation process near the fracture region, and the crack formation and the crack expansion have certain directivity, seen in figure 5(a). The greater the deformation amount nearby the fracture vicinity is, the more dynamic recrystallization core formed is [26], and vice versa. With the plastic deformation continued, the void of the crack is reduced and the direction of the crack propagation gradually tends to be parallel to the axial tensile stress, seen in figure 5(b). The dynamic recrystallization takes place in the plastic deformation area near the fracture surface [13, 25], and cracks are mainly surrounded by the formed dynamically recrystallized grains.

From figure 6(a) at 1200 °C, the fully dynamic recrystallization occurs in the plastic deformation region near the fracture, and the formed grains relatively grow uniformly [27]. The occurrence of dynamic recrystallization provides more small grain boundaries that are beneficial to the plasticity and slipping, which restrains the crack initiation and propagation effectively at the higher deformation temperature [28]. The void and cavities defects in
the central fracture region are furtherly reduced, seen in figure 6(b), and the crack formation and the crack propagation have the obvious directionality under axial stress.

In all, the dynamic recrystallization is closely related to the deformation temperature and deformation amount, and the large amount of dynamic recrystallization during crack initiation and propagation actually has a kind of gradient effect. At a certain deformation temperature, the larger the deformation amount is, the greater the driving force of recrystallization is, as the more deformation energy stored \[13, 25\]. Thus it can be inferred that during the tensile process from 1000 °C to 1200 °C, the crack initiation is also the place where the dislocation accumulates, accompanied by the dynamic recrystallization with the higher energy stored at the larger strain \[12\]. The small degree of dynamic recrystallization provides more paths, which is favorable for crack propagation \[21\]. While the larger dynamic recrystallization disperses the energy of the crack propagation and

Figure 2. The true strain-stress curves of 316LN steel under various deformation conditions.
Table 2. Parameters of peak stress $\sigma_p$, peak strain $\varepsilon_p$ and rupture strain $\varepsilon_r$ of 316LN steel.

| Temperature ($^\circ$C) | Parameters | 0.5 s$^{-1}$ | 0.05 s$^{-1}$ | 0.005 s$^{-1}$ |
|--------------------------|------------|-------------|--------------|--------------|
| 900 $^\circ$C             | $\sigma_p$ | 372.74      | 297.91       | 211.16       |
|                          | $\varepsilon_p$ | 0.43        | 0.31         | 0.23         |
|                          | $\varepsilon_r$ | 0.53        | 0.42         | 0.34         |
| 1000 $^\circ$C            | $\sigma_p$ | 218.41      | 192.67       | 129.41       |
|                          | $\varepsilon_p$ | 0.20        | 0.26         | 0.15         |
|                          | $\varepsilon_r$ | 0.46        | 0.43         | 0.33         |
| 1100 $^\circ$C            | $\sigma_p$ | 158.57      | 125.97       | 75.959       |
|                          | $\varepsilon_p$ | 0.26        | 0.21         | 0.17         |
|                          | $\varepsilon_r$ | 0.50        | 0.41         | 0.29         |
| 1150 $^\circ$C            | $\sigma_p$ | 141.04      | 92.23        | 72.56        |
|                          | $\varepsilon_p$ | 0.28        | 0.17         | 0.18         |
|                          | $\varepsilon_r$ | 0.55        | 0.45         | 0.31         |
| 1200 $^\circ$C            | $\sigma_p$ | 102.48      | 81.52        | 54.07        |
|                          | $\varepsilon_p$ | 0.23        | 0.14         | 0.16         |
|                          | $\varepsilon_r$ | 0.59        | 0.46         | 0.38         |

(a) Microstructure near fracture zone  (b) The local enlarged microstructure near fracture zone

Figure 3. Microstructure near the fracture region at 900 $^\circ$C.

(a) Microstructure near fracture zone  (b) The local enlarged microstructure near fracture zone

Figure 4. Microstructure near the fracture region at 1050 $^\circ$C.
effectively restrains the crack propagation. Cracks occur when the external deformation conditions of the ingot are beyond the limit of plastic deformation during the process of forging.

3.3. Tensile fracture analysis

In order to study the high temperature fracture state of 316LN, microstructures of the tensile fractures at 900 °C, 1050 °C, 1150 °C and 1200 °C at the strain rate of 0.5 s\(^{-1}\) are observed by SEM, seen in figures 7–10.

As shown in figure 7(a), the grain boundary of 316LN steel at 900 °C undergoes the stages of the slip and diffusion, with shallow and deep dimples on the fracture surface. The fracture morphology is intergranular ductile fracture, and the results show that the shapes of dimples varies greatly due to different grain sizes, seen in figure 7(b). When the deformation temperature is lower, the value of the thermo-plasticity index is lower, and the dimple formed along the fracture is relatively shallow. Obviously, the size of the dimple formed is related to the size of the grain size perpendicular to the axial tensile stress, that’s to say, the larger grain size corresponds to the deeper dimple formed along the grain boundary nearby fracture [29], and vice versa.

Figure 8(a) shows the fracture morphology at at 1050 °C, the fracture surface is rugged, the number of the surface dimples is increased, and there exist some clusters of small dimples. With the dimples become deeper, the plasticity increases and dislocation rings tend to pile up around the second phase or inclusion, thus form microvoids at the interface, seen in figure 8(b). The crack propagation finally leads to the intergranular fracture [29, 30]. The dimples become deeper as compared with figures 7(a)–(b), where the obvious second phase can be seen at the bottom of the dimples.

From figure 9(a) at 1150 °C, the majority of the fracture areas experience the dynamic recrystallization with different grain sizes around the initial boundaries, and the dimples formed become deeper due to the decrease of the dislocation density, the decrease of dislocation resistance and the increases of the accumulated strain energy.
From figure 9(b), it can be seen that the grain boundary is clear, the deformed grains under the axial stress has obvious directivity. The inner surface of the dimple shows characteristics of the intergranular ductile fracture, and a second phase inclusion is embedded at the bottom of the deeper dimple.
As shown in figure 10(a), the macro-fracture has a good plastic deformation ability at 1200 °C, the macroscopical fracture shows obvious necking, with the fracture diameter of about 1.8 mm. The fracture is intergranular fracture with the micro-voids formed in the center of fracture, where the oxidation degree of grain boundary is serious [31, 32]. From figure 10(b), the grain boundary has been heavily oxidized during the fracture process, and the interface has become blurred. The fracture dimples formed at 1200 °C become much deeper than the dimples formed at 900 °C. Due to the high resistance to plastic deformation under the large deformation degree at the high deformation temperature, fully dynamic recrystallization occurs near the fracture surface [33].

From the analysis of the metallographic microstructure near the fracture, it can be seen that the crack initiation and propagation mainly proceeds along the original austenite grain boundary. The migration of grain boundaries terminates the diffusion of vacancy along grain boundaries, thus crack initiation and propagation become difficult [34]. It is therefore that the large area dynamic recrystallization can effectively restrain the crack initiation and propagation during the high temperature plastic deformation process [25]. The fracture of 316LN steel at elevated temperature is ductile fracture. With the increase of the temperature and strain rate, the depth of dimples as well as the plasticity increase. The crack propagation state and dynamic recrystallization state exist simultaneously [24, 28].

4. Fracture properties and analysis of 316LN steel

4.1. Analysis of nonmetallic inclusions

During the fracture analysis under the temperature of 1150 °C, clusters of non-metallic inclusions are found in individual fracture surfaces in figure 11(a), and the distribution morphology is shown in figure 11(b).

From the mesoscopic damage mechanics, the source of crack initiation commonly appears nearby the brittle inclusions and second-phase particles in cast ingots during the forging cracking. Clusters of non-metallic inclusions cause microvoids and cracks, seen in figure 11.

In order to further determine its characteristics, the chemical compositions of these points are analyzed qualitatively by EDS, seen in figure 12.

According to SEM analysis and EDS analysis of fracture surface from figure 11 to figure 12, the main inclusions existed in 316LN steel are the brittle non-metallic inclusion of aluminum oxide (Al2O3), with the distribution of the cluster formed. As the melting point of the Al2O3 inclusion is high, the cluster existence in matrix of 316LN makes the plastic deformation ability very poor during the high-temperature forging process. Thus, the contact areas around the aluminum oxide inclusion easily cause the stress concentration, leading to the crack initiation and propagation [5, 20].

As the inclusion of alumina in 316LN is a brittle inclusion, its plasticity is quite different from that of the matrix at high temperature. During the thermal plastic deformation process, dislocations is easily piled up to cause the local stress concentration, which becomes the core of crack initiation [5]. The existence of inclusions destroys the continuity of the matrix, and the contact region between the matrix and the inclusion is easy to cause stress concentration and form the crack under the external stress during the forging process. The second-phase particles or inclusions contained induce the cavity to nucleate, grow, and polymerize until it is connected to the macroscopic cracks, the crack propagation gradually continues and finally the damage is happens.

Figure 10. The fracture morphology at 1200 °C.
4.2. Analysis of crack propagation mechanism

According to the metallographic analysis of the fracture area in figures 3–6, the initiation and propagation of 316LN crack are mainly along the grain boundary during the high-temperature tensile process.

To verify the cracks formation process, the microcrack of 316LN steel on the fracture face is analyzed at the strain rate of 0.5 s$^{-1}$ at 1050 °C, shown in figure 13.

As is shown in figure 13, the orientation of grain is disordered in different regions, and it exist a difference in the direction of the tensile stress. When the non-metallic inclusion or other defects is under the action of axial stress in figure 13(a), the microcracks are gradually formed and enlarged, seen in figure 13(b). The closer to the center region, the larger the axial tensile stress is [35]. These microcracks mainly propagate along the grain boundary, seen in figure 13(c), which has intensified the void increase and the crack propagation along the grain boundary, in figure 14(d). It should be noticed in figure 13 that the propagation of the microcrack is neither vertical nor parallel to the direction of maximum principal stress, with a certain degree with the maximum
principal stress direction instead. With the increase of the stress, the microcrack gradually develops into a large cavity. The closer to the fracture face the micro-crack is, the larger the size of the cavity is.

At the junction of the trigeminal grain perpendicular to the principal axis of stress, the typical crack initiation and propagation at the strain rate of 0.5 s\(^{-1}\) at 1150 °C is shown in figure 14.

As can be seen from figure 14 at 1150 °C, the atoms on the grain boundary slip easily and the vacancies on the grain boundary tend to accumulate across the junction of the tricrystal boundaries, especially. After crack initiation, the direction of crack propagation is related to the grain boundary tendency, and generally presents a certain angle with the axial stress, as shown in figures 14(a)–(b). If the direction of propagation is parallel to the axial tensile stress or the cracking direction is propagated at a small angle, the peak stress at the crack tip will decrease significantly [36]. Therefore, the junction of the tricrystal boundaries is the weakest point of the grains at high temperature, where stress concentration is likely to occur under the action of external axial stress, leading to crack initiation and subsequent propagation along the grain boundaries, seen in figures 14(c)–(d).

As is shown in figure 15, under the strain rate of 0.005 s\(^{-1}\) at 1200 °C, creep occurs at higher temperature, and the microcracks form and develop along the grain boundary. Meanwhile, the grain growth leads to the decrease of grain boundary area, which is beneficial to the void formation and growth. It should be noticed that the trigeminal grain boundary is an important structural unit consisting of three grain boundaries intersected by three adjacent grains [37]. As atoms and vacancies on grain boundaries are more active, the microcracks can be thermally activated and diffused at high temperature. The grain boundary during high temperature tensile process has the characteristics of the high solubility and the high diffusibility. Commonly, the higher the temperature is, the stronger the diffusivity is, which is due to that the atomic diffusion rate on grain boundaries is higher than that in the grain. Thus, the grain boundary is relatively easy to slide under the external force [38].

From the analysis and discussion of figures 13–15, the process of crack initiation, propagation and fracture during high temperature tension can be drawn, through the explanation of intergranular cracks formation and cracking around trigeminal grain boundaries, seen in figure 16.

As is shown in figure 16, when the grain boundary is perpendicular to the axial tensile stress, the cracks first appear during the tensile process, seen in figure 16(a), and then propagates along the grain boundary with the increase of the external stress [11, 39]. When it encounters on the other grains, the crack propagation is hindered and the direction of crack propagation changes with the grain boundary direction, seen in figure 16(b). The joint of the trigeminal grain boundary is easy to cause the stress concentration under the external axial stress, which leads to the crack initiation and propagation along the grain boundaries, seen in figure 16(c). After a long period of propagation, the crack penetrates the interface between brittle inclusions and grain boundaries until the fracture happens, seen in figure 16(d). After the necking and fracture processes during the tensile plastic

Figure 14. Evolution of the intergranular cracks around trigeminal grain boundaries at 1150 °C.

Figure 15. Evolution of the intergranular cracks around trigeminal grain boundaries at 1200 °C.
deformation, the necked portion of the specimen will be only subjected to axial stress. That’s to say, the greater the damage stress intensity factor is, the more favorable the crack initiation is [31, 32].

Overall, the trigeminal grain boundary restrains the grain boundary sliding, and the great disharmony during the deformation process leads to a large stress concentration [40]. If the stress concentration on grain boundaries cannot be relaxed by the sliding or migration on grain boundary during the plastic deformation process, the crack initiation might occur under the external tensile stress [41, 42]. The closer to the central fracture region the microcrack initiation is, the easier the crack propagation is. This is because the three-dimensional axial stress during the necking process is different till to the fracture [43], accompanied with the reduce of the plastic index during the forging process on the tendency of cracking from the boundaries of the brittle inclusions [44, 45].

4.3. Fracture properties and analysis

Figure 17 shows the reduction of area of 316LN stainless steel at different strain rate under various deformation temperatures.

As can be seen from figure 17, when the strain rate is 0.05 s⁻¹ and 0.5 s⁻¹, there is a large amount of dynamic recrystallization in the higher temperature region (greater than 950 °C). With the increase of temperature, the deformation resistance decreases, the plastic properties increase and the reduction of area (or section shrinkage ratio) obviously increases.

It should be noticed that for the relative high temperature region I (lower than 950 °C) in figure 17, the higher the strain rate is, the shorter the softening time is and the lower the degree of dynamic recrystallization is. Meanwhile, dislocations caused during high-temperature plastic deformation leads to the decrease of the reduction of area. Unusually, the reduction of area within the relative higher region II has a tendency to decrease at first within the region of 1000 °C–1150 °C and then increase within the region of 1150 °C–1200 °C. This is because, with the increase of temperature, only a small amount of dynamic recrystallization occurs at 950 °C when the strain rate is 0.005 s⁻¹.

The measured elongation δ and reduction of area ψ of 316LN steel are in table 3.
From Table 3, within the region of 1000 °C–1200 °C, the reduction of area $\psi$ and the elongation $\delta$ at the strain rate of 0.005 s$^{-1}$ is greater than the values at the strain rates of 0.05 s$^{-1}$ and 0.5 s$^{-1}$. When at the lowest strain rate of 0.005 s$^{-1}$, the low strain rate and the long pull-off time result in the initial creep behavior [30, 32, 44]. Under the temperature range of 1000 °C–1150 °C (seen in region II in figure 17), dynamic recovery, dynamic recrystallization and work hardening exist at the same time, and the work hardening is dominant, which counteracts dynamic recrystallization and dynamic recovery. When the temperature is higher within the region of 1150 °C–1200 °C, the complete dynamic recrystallization has taken place, resulting in a large number of grain boundaries, which hinder and disperse the crack propagation [43].

In all, the mechanism of crack initiation and propagation of 316LN steel at high temperature have both scale effect and interface effect. Except for the size change of crack caused by external stress, fracture properties is also related with the the crack caused by the interface effect where the dynamic recrystallization occurs during the plastic deformation [13, 25, 33, 46]. The micro-recrystallization promotes crack growth, and the small grain and sub-grain boundaries formed by large amount of dynamic recrystallization have the local strengthening effect which disperses the driving force of crack growth [46, 47].

### 5. Conclusion

The thermal tension experiment is carried out on the Gleeble-1500D during the high-temperature plastic deformation process, and the fracture behaviors by the analysis of the metallographic microscope, SEM and EDS are investigated. Results are as follows:

1. The fully recrystallization occurs within the region of 1100 °C–1200 °C and 0.05–0.5 s$^{-1}$. The higher the strain, the greater the recrystallization is and the easier it is to crack the sample.

2. The reduction of area within the region of 1000 °C–1200 °C at the strain rate of 0.005 s$^{-1}$ is greater than the values at the strain rates of 0.05 s$^{-1}$ and 0.5 s$^{-1}$, due to the initial creep behavior.

3. During the tensile process at 900 °C–1200 °C under the strain rate of 0.05–0.5 s$^{-1}$, the intergranular ductile fracture is commonly observed and the crack propagation mainly proceeds among the tricrystal boundary along the initial austenite grain boundaries and the non-metallic inclusion (Al$_2$O$_3$).

4. The initial creep behavior leads to the decrease of the area reduction within the region of 1000 °C–1200 °C at the 0.005 s$^{-1}$, which is due to the creep-fatigue interaction phenomenon that increases the stress distribution at the crack tip and strengthens the driving force of crack propagation.

5. The source of crack initiation appears and expands till to rupture, with the maximum elongation $\delta$ (9.35%) and area reduction $\psi$ (78.11%) at 1200 °C under 0.5 s$^{-1}$, where the brittle inclusions in cast ingots significantly increases the tendency of the forging cracking.

Furthermore, for a small-scale degree of dynamic recrystallization, the path for the crack propagation at the crack source is through the grain boundaries, which promotes the crack propagation. While the large-scale degree of dynamic recrystallization can not only disperse the crack growth path but also reduce the stress distribution at the crack tip, which weakens and restrains the crack growth.
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