Plastic Deformation and Damage Behaviors of Fe-18Cr-18Mn-0.63N High-Nitrogen Austenitic Stainless Steel under Uniaxial Tension and Compression

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Uniaxial tensile and compressive deformation behaviors of Fe-18Cr-18Mn-0.63N high-nitrogen austenitic stainless steel were investigated at different strain rates. It is found that, with increasing strain rate, the yield strength of the steel increases either under tension or compression, while the ultimate tensile strength and the total elongation decrease under tension. The plastic deformation behavior prior to necking under uniaxial tension at different strain rates can be well described by the modified Ludwik relation. Under tensile deformation at the low strain rate of 10⁻¹ s⁻¹, microcracks prefer to initiate around the Al₂O₃ particles in the steel, whereas cracks nucleate at grain boundaries or along slip bands at the high strain rate of 10⁻² s⁻¹. Compressive deformation behavior of the steel is not so sensitive to the strain rate. The surface fluctuation is more serious under compressive deformation rather than tensile deformation. The tensile plastic deformation is mainly governed by the formation of planar slip bands and deformation twins, and deformation by twinning becomes more prominent with increasing strain rate, while dislocation slip preferred by planar slip bands, dislocation bands and twin-like bands basically controls the compressive plastic deformation, and deformation twins are rarely formed. [doi:10.2320/matertrans.M2014262]

(Received July 14, 2014; Accepted October 10, 2014; Published November 29, 2014)

Keywords: high nitrogen austenitic stainless steel, plastic deformation, tension, compression, strain rate, microstructure

1. Introduction

High nitrogen austenitic stainless steels (HNASSs) exhibit excellent mechanical properties, corrosion and oxidation resistances, wearability, etc, so great efforts have been made to develop such steels over the past decades.¹⁻³ Moreover, it is known that traditional austenitic stainless steels normally contain Ni element, which would unfortunately cause allergic reactions in the human body during implantation; therefore, Ni-free HNASSs would also have the great potential in practical medical applications.⁴,⁵

Apparently, a clear understanding of the mechanical deformation behavior of such HNASSs is of particular importance for their engineering and medical applications. It has been recognized that, as a comparatively low stacking fault energy (SFE) alloy, HNASS exhibited typical deformation features, i.e., planar glide at low strain, and deformation twinning at intermediate or high strain.⁵,⁷ Byun⁸ summarized the deformation microstructures of austenitic stainless steels as the following categories in terms of the equivalent stress range: (i) dislocation tangles at the equivalent stress of less than 400 MPa; (ii) small and isolated stacking faults in the range from 400 to 600 MPa; and (iii) large stacking faults/twin bands at more than 600 MPa. Several investigators studied the ductile-to-brittle transition phenomenon in such kind of steels with a face centered cubic (F.c.c.) lattice, and obtained some unusual results.⁷,⁹ Müllner et al.⁷ proposed a model of crack nucleation and crack propagation to account for the brittle fracture behavior in the steel. However, to the best of our knowledge, the relationship between plastic deformation features and strain rates is still seldom elaborated. In the present work, the deformation and damage behavior of a Ni-free HNASS of Fe-18Cr-18Mn-0.63N under uniaxial tension or compression will be experimentally investigated focusing on the strain rate effect.

2. Experimental Procedures

The experimental steel used in present study was melted by induction furnace and electroslag remelting furnace both filled with N₂. The cast ingot was forged into a plate with 30 mm thickness at 1200°C. The chemical compositions of the experimental steel are shown in Table 1 (the steel is termed Fe-18Cr-18Mn-0.63N steel).

The forged plate was machined into the tensile specimens with a gauge section of 5 mm × 3 mm × 25 mm and compression specimens with a size of 4 mm × 4 mm × 6 mm. Before mechanical tests, all specimens were electropolished in a solution of perchloric acid and glacial acetic acid with a volume ratio of 1 : 9 under 18–20 V for 150–180 s, in order to obtain a strain-free and smooth surface for microscopic observations. Tensile and compressive tests were all carried out at room temperature (RT) with the initial strain rates of 10⁻⁴ s⁻¹ and 10⁻² s⁻¹. Compression tests, for which the strain was just measured by cross-head displacement, were performed up to a deformation strain of 45%. The initial microstructure of the raw materials was observed by optical microscope (OM), and surface deformation features and fracture surfaces of the deformed specimens were observed by scanning electron microscopy (SEM). Thin foils used for

Table 1 Chemical compositions of the Fe-18Cr-18Mn-0.63N steel (mass%).

<table>
<thead>
<tr>
<th>elem</th>
<th>%</th>
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<tbody>
<tr>
<td>Cr</td>
<td>17.97</td>
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<tr>
<td>Mn</td>
<td>18.0</td>
</tr>
<tr>
<td>N</td>
<td>0.63</td>
</tr>
<tr>
<td>C</td>
<td>0.056</td>
</tr>
<tr>
<td>Al</td>
<td>0.02</td>
</tr>
<tr>
<td>Fe</td>
<td>Balanced</td>
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transmission electron microscopy (TEM) observations are sliced from the gauge part of deformed specimens by spark-cutting parallel to the loading direction, then mechanically thinned down to dozens of micron thick and finally polished using a twinjet electrolytic polisher operated at 20 V in an electrolyte of perchloric acid and alcohol at −25°C. TEM observations were carried out using a TECNAI G2 20 electron microscope operated at 200 kV.

3. Results and Discussion

Figure 1 shows the three-dimensional microstructures of the raw Fe-18Cr-18Mn-0.63N steel material. Apparently, the initial microstructure consists mainly of f.c.c. austenitic grains with an average grain size of about 30 µm, and the presence of a number of annealing twins and a few precipitates due to pre-thermal-mechanical treatment were also detected in the raw material.

3.1 Stress-strain behavior

The influence of the strain rate on the tensile engineering stress-strain curves of the Fe-18Cr-18Mn-0.63N steel is shown in Fig. 2(a), and the data for tensile properties are listed in Table 2. It is apparent that the yield strength ($R_{p0.2}$, 0.2% proof stress) of the specimen deformed at the strain rate of $10^{-4}$ s$^{-1}$ is lower than that at $10^{-2}$ s$^{-1}$, while the ultimate tensile strength ($R_m$) and total elongation ($\delta$) decline with increasing strain rate due to an obvious decrease in the work hardening capability. Similar phenomenon was also found in other HNASSs. As seen in Fig. 2(b), yield strengths measured at these two strain rates from compressive tests exhibit the same tendency as the case from tensile tests.

Actually, the total elongation can be divided into two parts, i.e., elongation of uniform deformation ($\delta_u$) and elongation of inhomogeneous (or localized) deformation ($\delta_i$). $\delta_u$ represents the total plastic strain before reaching the max-engineering stress, and reflects the strain hardening exponent of the material, which indicates the resistance of material to plastic deformation, while $\delta_i$ and the reduction in area ($\psi$) are usually associated with the final fracture morphology.

Table 2 gives the corresponding measurements of $\delta_u$, $\delta_i$ and $\psi$ at these two strain rates. It is found that $\delta_u$ generally decreases with increasing strain rate, indicating that work-hardening is also influenced by the strain rate. In contrast, the values of $\delta_i$ and $\psi$ are almost independent of the strain rate.

As is well known, for many other metals and alloys, the flow curve in the region of uniform plastic deformation can be expressed by the Ludwik relation:

$$\sigma = K_1\varepsilon^n$$

(1)

where $\sigma$ and $\varepsilon$ are the true stress and the true plastic strain, respectively, $K_1$ is a proportionality constant termed the strength coefficient, and $n_1$ is an exponent termed the strain hardening exponent. Although this equation adequately describes the strain hardening behavior of many engineering materials, it is not applied to stable austenitic stainless steels and other f.c.c. metals with low stacking fault energies. In

![OM image of microstructures of the Fe-18Cr-18Mn-0.63N steel.](Fig. 1)

![Tensile (a) and compressive (b) engineering stress-strain curves of the Fe-18Cr-18Mn-0.63N steel at two strain rates of $10^{-4}$ s$^{-1}$ and $10^{-2}$ s$^{-1}$.](Fig. 2)
view of this, a second term, which accounted for deviations at low strains, had been added to the Ludwik model by Ludwigson\textsuperscript{12}) to construct the modified Ludwik relation:

\[ \sigma = K_1e^{n_1} + \Delta, \text{ where } \Delta = e^{(K_2+n_2)e} \]  \hspace{1cm} (2)

Here, \( \exp(K_2) \) defined as the true yield stress is the value of true stress extrapolated to a true strain of zero, and it corresponds to the short-range stress inducing the onset of plastic flow, i.e., the movement of the first mobile dislocations.\textsuperscript{13}) The parameters \( K_2 \) and \( n_2 \) in eq. (2) can be determined from a plot of \( \ln(\Delta) \) versus \( \varepsilon \).

Soussan \textit{et al.}\textsuperscript{13}) and Simmons\textsuperscript{14}) have ever used the modified Ludwik relation to simulate the work hardening behavior of 316LN alloys containing nitrogen up to 0.25 wt% and Fe-17Cr-(8-10)Mn-Ni alloys with nitrogen contents varying from less than 0.2 to almost 1 wt%, respectively. It was reported that the work hardening behavior of all these materials could be simulated using the modified relation, and the influence of nitrogen content on the modeling parameters (i.e. \( K_1, n_1, K_2, n_2 \)) has been discussed in detail. Also, Soussan \textit{et al.}\textsuperscript{13}) defined an additional parameter, the transition strain (\( \varepsilon_{t_1} \)), as the strain, at which the departure parameter (\( \Delta \)) is very small in comparison with the Ludwik term. They obtained \( \varepsilon_{t_1} \) by using the following equation:

\[ r = \frac{\Delta}{K_1\varepsilon_{t_1}^{n_1}} = \frac{\exp(K_2+n_2e)}{K_1\varepsilon_{t_1}^{n_1}} \]  \hspace{1cm} (3)

The transient stress (\( \sigma_{t_1} \)) corresponding to \( \varepsilon_{t_1} \) is calculated using the following equation:

\[ \sigma_{t_1} = K_1\varepsilon_{t_1}^{n_1} \]  \hspace{1cm} (4)

To quantify the uniform plastic deformation stage of the current Fe-18Cr-18Mn-0.63N Ni-free HNASS, the modified Ludwik relation was also adopted to simulate its flow curves at different strain rates. Here, an \( r \) value of 0.02 was selected to evaluate \( \varepsilon_{t_1} \), that is, \( \varepsilon_{t_1} \) refers to the true plastic strain accommodated by the steel, as \( \Delta \) was decreased down to only 2% of true stress.

As seen in Fig. 3, the uniform plastic deformation behavior under uniaxial tension at two strain rates of \( 10^{-4} \) s\(^{-1} \) and \( 10^{-2} \) s\(^{-1} \) can be well described by the modified Ludwik relation, and the flow curve parameters are shown in Table 3. The \( n_1 \) value decreases with increasing strain rate, implying that the strain hardening of the steel under tension is more pronounced at low strain rate. The true yield stress, \( \exp(K_2) \), decreases significantly with increasing strain rate, in contrast to the engineering yield stresses (\( R_{0.2} \) in Table 2), and the values of the engineering yield stress are much higher than the true yield stress values (\( \exp(K_2) \) in Table 3). Compared to Fe-17Cr-(8-10)Mn-Ni HNASS with a similar nitrogen content (0.52 wt%) investigated by Simmons,\textsuperscript{14}) the current Fe-18Cr-18Mn-0.63N Ni-free HNASS shows a lower \( n_1 \) value (a lower work hardening ability) and a decrease in the transition strain (\( \varepsilon_{t_1} \)), but a comparable tensile strength. Such a difference in mechanical properties results mainly from the existence or non-existence of Ni element in these two steels.

### 3.2 Surface deformation and damage characteristics

Figure 4 shows the surface deformation and damage features near the fracture of Fe-18Cr-18Mn-0.63N steel specimens uniaxially tensioned at different strain rates, where the loading direction is vertical. It is obvious that dislocation slip is the major tensile deformation characteristic. A transgranular microcrack of 50 \( \mu \)m in length forms along slip bands (SBs) at the low strain rate of \( 10^{-4} \) s\(^{-1} \) (Fig. 4(a)). In a wider field of view (Fig. 4(b)), one can see that these microcracks tend to connect with each other, accompanied by the formation of some shear bands. On the contrary, almost no obvious microcracks can be found on the damaged surface at the high strain rate of \( 10^{-2} \) s\(^{-1} \), and only a few non-propagation microvoids appear at grain boundaries (GBs) or slip bands (Fig. 4(c)). Under a low magnification (Fig. 4(d)), only a few shear bands can be seen on the surface. A possible explanation is that there is no sufficient time for these voids to extend along SBs at such a high strain rate, even though the flow stress reaches a certain level.

Figure 5 shows the typical surface deformation and damage features under compressive tests, where the loading direction is vertical. It is clear that more serious surface deformation takes place and some inclusions are pushed out of the matrix during compression at the high strain rate of \( 10^{-2} \) s\(^{-1} \) (Fig. 5(c),(d)) if compared with the case at the low strain rate of \( 10^{-4} \) s\(^{-1} \) (Fig. 5(a),(b)). Actually, the effect of the strain rate on the number of pushed-out particles and the plastic flow behavior (e.g., yield strength, work-hardening, plastic instability, etc.) can both be explained by occurrence of grain rotation during deformation. The deformation-induced local stress concentration can be more readily relaxed by sufficient grain rotation during compression at a low strain rate.

Two kinds of particles were found on the surface of compression specimens (Fig. 5(d)). Further inspections of
these particles by energy spectrum analysis (EDS) are shown in Fig. 6. One kind of these particles is Al₂O₃ particle with an irregular geometric morphology, while the other one is AlN precipitate with a hexagonal structure, which was also seen in other nitrogen-containing austenitic stainless steels.¹⁵,¹⁶

Actually, when melting the raw material, slag containing Al₂O₃ was added during electroslag remelting to the steel. Considering that Al₂O₃ will hardly precipitate in HNASSs at normal procedure, it can be sure that Al₂O₃ particles observed in specimens are not precipitates but inclusions introduced during melting, and they can appear both in grains and at GBs. Clearly, Al₂O₃ particles can hinder dislocation slipping to enhance the strength of the steel, but stress concentrate may also easily occur around Al₂O₃ particles, leading to the formation of cracks (Fig. 5(d)).

The AlN phases are usually found in grains, and they may move under the action of dislocation slips. Precipitates with geometrically favorable orientations will be pushed out of matrix during compression without any obvious cracking around them, due to their interactions with SBs, as seen in Fig. 5(d).

### 3.3 Fracture surface features

Figure 7 shows fracture features of tensile specimens of the Fe-18Cr-18Mn-0.63N steel at two strain rates. A slight necking phenomenon is observed similarly at the both strain rates (Fig. 7(a),(d)), but the fracture surface at 10⁻² s⁻¹ seems to be a little flatter than that at 10⁻⁴ s⁻¹. On the whole, the fracture surface of the experimental steel consists of three areas, i.e., fibrous zone, radiation area, and shear lip. Secondary cracks are generally found in the fibrous zone, and the radiation area occupies a larger proportion of the fracture surface (Fig. 7(b),(e)), indicating a comparatively limited ductility of the steel.
Further insight into the strain rate effect on deformation and fracture characteristics is obtained from magnified observations of the fibrous zone (Fig. 7(c),(f)). It is obvious that the morphological features in fibrous zone are quite different at these two strain rates. The fibrous zone consists of dimples with a bimodal size at the low strain rate of $10^{-4}$ s$^{-1}$, and some particles are usually found in large dimples, as indicated by arrows in Fig. 7(c). The bimodal dimples were often observed in the fracture of Al alloys containing precipitates, and it has been generally believed that the formation of the bimodal dimples is closely related to the second-phase particles in the material, i.e., debonding and fracturing of large particles induce the formation of large dimples, and fracturing of the matrix cause the presence of small dimples.\textsuperscript{17,18} Undoubtedly, the formation of bimodal dimples in the current steel at the low strain rate should be
also related to debonding and fracturing of Al2O3 (maybe also AlN) particles. Moreover, the ‘snake slips’ and ‘ripples’ can be observed in the inner walls of the larger and deep dimples (Fig. 7(c)). For the case at the high strain rate of $10^{-2}$ s$^{-1}$, the characteristic of bimodal dimples is not obvious, and nearly no particles and ‘snake slips’ can be detected in dimples, which are relatively shallow (Fig. 7(f)). The difference in dimple features is well consistent with different tensile elongations achieved at these two strain rates (Fig. 2(a) and Table 2), and also indicates that the deformation and damage mechanisms should be different, depending upon the strain rate applied.

On stretching at the low strain rate of $10^{-4}$ s$^{-1}$, microcracks usually initiate around the Al2O3 particles, and propagate along GBs and SBs (Fig. 4(a)). As cracks connect with each other, final fracture occurs. Since microcracks may grow up and propagate along SBs in any preferentially oriented grains in necking area, the final fracture morphology is a little rugged (Fig. 7(a)). In contrast, at the high strain rate of $10^{-2}$ s$^{-1}$, almost no obvious surface cracks are found near the fracture (Fig. 4(c),(d)), and also particles are seldom found in the large dimples on the fracture surface (Fig. 7(f)). Actually, at such a high strain rate, slips become harder to activate sufficiently. Once cracks nucleate at GBs or along SBs, they would propagate along the direction macroscopically perpendicular to the tensile axis, forming a radiation area with more brittle morphology (Fig. 7(e)), so that the final fracture surface is relatively flat (Fig. 7(d)).

### 3.4 Deformation microstructures

Figure 8 shows the typical tensile deformation microstructures of Fe-18Cr-18Mn-0.63N steel specimens at two strain rates. Planar slip bands, which dominate slip deformation characteristics in low stacking fault energy metals, were widely found in the deformed specimens, as seen in Fig. 8(a) and (d). At the low strain rate of $10^{-4}$ s$^{-1}$, vein-like structure, which is a typical kind of wavy-slip type dislocation structure, tends to form (Fig. 8(b)), in addition to a few deformation twins (Fig. 8(b),(c)). At the high strain rate of $10^{-2}$ s$^{-1}$, much more deformation twins were found to form in grains. As seen in Fig. 8(f), a large number of deformation twins pass through a low-angle GB. Furthermore, due to the inadequate deformation compensation provided by planar dislocation slipping and twinning, second-order twins (SOT) were also found to form in the host deformation twin (Fig. 8(e)). The formation of these SOT decreases the average twinning strain and increases the work hardening rate. Such structural twins have also been found in other deformed metals such as Cu-Ti alloys, ß-phase titanium alloys, Co-Fe, and TiAl alloys. The above TEM observations strongly demonstrate that twinning deformation become more prominent with increasing strain rate.

Byun summarized typical features of the deformation structure of 316 and 316LN series alloys and put forward that the critical stress for twinning in polycrystalline in terms of the equivalent or uniaxial stress can be given by

$$\sigma = 6.14 \cdot \frac{\gamma_{SF}}{b_p}$$

where $\gamma_{SF}$ is stacking fault energy (SFE), and $b_p$ represents the Burgers vector. For the experimental steel, $\gamma_{SF}$ is as low as $\sim10 \, \text{mJ/m}^2$ considering the effect of nitrogen on $\gamma_{SF}$. For a low SFE stainless steel, if we take $b_p = 0.145 \, \text{nm}$, the critical uniaxial (or equivalent) tensile true stress is estimated to be 423 MPa, which is a little higher than the true yield stress (exp($K_2$) in Table 3), indicating that deformation twins should form in the early stage of plastic flow deformation after yielding.

Figure 9 shows the typical dislocation patterns observed during compression deformation at two strain rates. Unlike the situation under tension, planar slip bands along three slip systems were generally observed under compression (Fig. 9(a),(d)), especially at the low strain rate of $10^{-4}$ s$^{-1}$, some coarse dislocation slip bands can be clearly observed (Fig. 9(b)). That means dislocation slip deformation is greatly enhanced under compression rather than under tension, primarily owing to the existence of a more favorable stress state to shear plastic deformation under uniaxial compression. In this case, the plastic deformation can be
almost accommodated by dislocation slipping, so that the proportion of deformation twins decreases notably, and only few twins were found at the high strain rate of $10^{-2}\text{s}^{-1}$ (Fig. 9(f)). It is interesting to note that many deformation bands were detected to form in both specimens (Fig. 9(c),(e)). The relevant diffraction patterns (Fig. 9(c),(e)) have indeed demonstrated that these bands are not deformation twins (DTs) but deformation bands, which exhibit similar configurations to DTs. The size (or thickness) of these twin-like bands changes from dozens of nanometers at $10^{-4}\text{s}^{-1}$ (Fig. 9(c)) to micrometer scale at $10^{-2}\text{s}^{-1}$ (Fig. 9(e)). It is inferred that the formation of these twin-like bands should be related to the planar dislocation glide process. For the current high-nitrogen steel with a low SFE, plastic slip deformation is accommodated mainly by glide of extended dislocations comprising pairs of partial dislocations bounded by a stacking fault ribbon. During plastic flow process, the sustained glide activity of extended dislocations on parallel sets of planes (see Fig. 9(a)) would yield overlapping stacking faults, and ultimately lead to the formation of bands with configurations akin to deformation twins. In fact, DTs can be treated as a result of partial dislocation glide on every \{111\} layer, but this kind of dislocation glide is well-regulated and the process of this regular dislocations glide is still unclear. The formation mechanism of DTs remains somewhat controversial (i.e., the pile-up/glide mechanisms,31-34 the pole mechanisms,35 the homogeneous nucleation mechanisms,36 etc.37). It is believed that these twin-like bands containing partially regular dislocations glide might be the precursors of deformation twins. Jin et al.31,32 and Singh et al.33,34 have brought forward the dislocation pile-ups/glide mechanisms for the twin formation, i.e., the generation and glide of twin dislocations can be treated in a similar manner to those of ordinary dislocations. Byun30 also illustrated several aspects based on the successive glide of partial dislocation to support the above twin formation mechanisms. In view of above analyses, the twin formation process in Fe-18Cr-18Mn-0.63N steel can be inferred to occur according to the following sequence: dislocation pile-ups/glide $\rightarrow$ dislocation bands $\rightarrow$ twin-like bands $\rightarrow$ DTs. A further evidence to support the above twin formation mechanisms is given in Fig. 9(c), from which one can see that some large twin-like bands (marked by arrows) tend to be separated into small ones. In fact, Xiao et al.39 also reported quite similar results that with increasing strain, the average twin layer thickness decreased slightly, while the average twin boundary spacing decreased significantly in a Cu-Zn alloy.

4. Conclusions

Plastic deformation and damage behavior of Fe-18Cr-18Mn-0.63N high nitrogen austenitic stainless steel was investigated under uniaxial tension and compression at different strain rates. The obtained results are mainly summarized as follows.

(1) With increasing strain rate, the yield strength of Fe-18Cr-18Mn-0.63N steels increases either under tension or compression, while the ultimate tensile strength and the total elongation decrease under tension. This is closely related to the strain rate-dependent deformation and damage micromechanisms. The uniform plastic deformation behavior under uniaxial tension at different strain rates can be well characterized by the modified Ludwik relation.

(2) At the low strain rate of $10^{-4}\text{s}^{-1}$, microcracks usually initiate around Al$_2$O$_3$ particles in the Fe-18Cr-18Mn-0.63N steel, and propagate along GBs and SBs, while cracks nucleate at GBs or along SBs, and extend along the direction macroscopically perpendicular to the tensile axis at the high strain rate of $10^{-2}\text{s}^{-1}$. The formation of planar slip bands and deformation twins is
the major microscopic deformation features; however, deformation by twinning becomes more remarkable with increasing strain rate.

(3) Compressive deformation behavior of the Fe-18Cr-18Mn-0.63N steel is not so sensitive to the strain rate. Compared with tensile deformation, the surface fluctuation is more serious as the specimen is subjected to compressive deformation. Dislocation slip featured by planar slip bands, dislocation bands and twin-like bands dominates the plastic deformation under compression, and deformation twins are rarely found.

Acknowledgements

This work was financially supported by the National Natural Science Foundation of China (NSFC) under Grant nos. 51231002, 51271054 and 51201027, and also by the Fundamental Research Funds for the Central Universities of China under Grant nos. N110105001, N120405001 and N120505001.

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