Original Article

Serrated flow and failure behaviors of a Hadfield steel at various strain rates under extensometer-measured strain control tensile load

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A B S T R A C T
A Hadfield steel has been investigated to clarify the serrated flow and to explore the failure behavior at room temperature. The tensile experiments were performed under extensometer-measured strain control, rather than under conventional cross-head displacement control, at strain rates ranging from \(6 \times 10^{-3}\) s\(^{-1}\) to \(6 \times 10^{-6}\) s\(^{-1}\). Three types of serrations, including type A, B and C ones, are observed. The occurrence of different types of serrations depends on both strain rate and strain level. The type C serration is identified in Hadfield steels at room temperature for the first time. At high strain rates, there is substantially higher serration density and reduction in stress compared with that observed at low strain rates. Furthermore, two different initiation modes of deformation bands are revealed. The fracture crack nucleates at a position with dense twins, and propagates primarily in the direction perpendicular to the tensile axis and deflects frequently due to the interaction with the boundary of grains and twins.

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1. Introduction

Hadfield steel, which exhibits a unique combination of properties such as high strength, high toughness, and high resistance to wear and heavy impact loads, has been widely used to manufacture railway crossings, grinding mill liners, crusher jaws and cones, impact hammers and bullet-proof helmets [1,2]. The focus of previous researches on Hadfield steel is placed mostly on their deformation mechanisms. The microstructures and mechanical properties are determined by the deformation mode [3]. Compared with other iron steels, Hadfield steel exhibits excellent work-hardening ability. The essential reasons for its remarkable work-hardening ability have been investigated by many excellent researches [4–6]. It is generally accepted that the high work-hardening ability is attributed to the transformation of \(\gamma\) to \(\alpha\) or \(\epsilon\) martensitic structures and deformation twins [5–9]. Additionally, dynamic strain aging (DSA), which results from repeated dynamic interactions of mobile dislocations with diffused solute atoms,
is also considered to increase the work-hardening rate of Hadfield steel, resulting in high uniform elongation and ultimate tensile strength finally [10–12]. The contribution of the DSA to the stress response of Hadfield steel is clearly visible, which is manifested in the form of serrated flow on the stress-strain curves. However, the exact contribution to the resulting work-hardening rate remains controversial, because various deformation mechanisms occur during straining [11,12].

For Hadfield steel with 1014Mn and 1.014 C, microstructures and interactions such as twins, stacking fault (SF) [5,13], dislocation walls [14,15], DSA of Mn-C couples [6,7], are contribute to the work-hardening behavior. The strain rate sensitivity (SRS) is strongly affected by the evolved microstructures, which results in a positive or a negative strain rate dependency. According to recent investigations [3,16], the deformation mechanisms significantly affect the microstructures and the corresponding mechanical properties of Hadfield steel. The deformation twins are formed uniformly along the initial grain boundaries (GBs) when subjected to explosion load, while they primarily appear in the form of twin bundles for samples in cold rolled state. These differences finally result in higher yield strength but lower tensile strength and elongation for samples manufactured by explosion. Furthermore, a physical model is established to illustrate the work-hardening and surface nanocrystallization behaviors [16].

Large numbers of investigations on the DSA behavior are conducted with Al-Mg alloys [17,18] and high manganese austenitic twinning-induced plasticity (TWIP) steels [19–22]. According to these studies, three types of serrations including type A, B and C, which depend on the applied strain rate and temperature, are detected. Moreover, the serrations gradually transform from type A to type B or C with the reduction of strain rate or increase of temperature. However, there are quite scarce studies on the DSA behavior of Hadfield steels. Dastur et al. [10] and Canadinc et al. [11] reported that type A and B serrations occur at relatively lower temperature in Hadfield steel, while type C serration appears only at high temperature. With increasing temperature, the serrated flows increase in magnitude and frequency, and then vanish at temperatures above 190° [11]. Therefore, it can be easily concluded that only one type of serrations is the common feature of Hadfield steels under room temperature tensile load, which is inconsistent with that of Al-Mg alloys.

For innovation of processing technology and the following application of Hadfield steel, it is extraordinarily necessary to warrant a good understanding of the deformation mechanism and the serrated flow behavior. The aim of current study is to investigate the serrated flow and failure behavior based on uniaxial tensile test in a wide range of strain rates, and special attention is paid to clarify the serration type with change of strain rates. Moreover, it should be specially emphasized that all the tensile tests were conducted under extensometer-measured strain control, while previous tensile tests were almost all carried out under cross-head displacement control. Surprisingly, entirely different types of serrations are detected during straining, as opposite to previous observations under displacement control [10,11].

**Fig. 1 – Initial OM of the studied Hadfield steel.**

<table>
<thead>
<tr>
<th>Alloy</th>
<th>C</th>
<th>Mn</th>
<th>Si</th>
<th>P</th>
<th>S</th>
<th>Fe</th>
</tr>
</thead>
<tbody>
<tr>
<td>Hadfield</td>
<td>1.23</td>
<td>12.31</td>
<td>0.12</td>
<td>0.026</td>
<td>0.025</td>
<td>Bal.</td>
</tr>
</tbody>
</table>

**Table 1 – Chemical compositions (wt %) of the investigated Hadfield steel.**

2. **Experimental procedure**

The Hadfield steel used in this investigation was melted by vacuum induction furnace. After homogenization at 1473 K for 2 h, the steel ingot was hot forged to 45 × 15 mm² blocks with finishing forging temperature 1273 K. The hot-forged plates were solution treated at 1373 K for 0.5 h followed by water quenching, and then a fully austenitic microstructure with an average grain size 76 μm was obtained (Fig. 1). The nominal chemical compositions in percentage weigh are listed in Table 1. Dog-bone specimens, with a nominal gauge section of 14 × 10 × 3 mm³, were used for uniaxial tensile tests. Tensile tests were performed on a MTS test system at room temperature under extensometer-measured strain control with applied strain rates ranging from 6 × 10⁻³ s⁻¹ to 6 × 10⁻⁶ s⁻¹. An extensometer with 10 mm gauge length was used to measure the deformation strain. It is highly possible that instability may occur during straining under extensometer-measured strain control when the testing specimens break down. Therefore, according to the elongation of Hadfield steel, the uniaxial tensile tests were lasted only to strain of 0.25. Beyond the strain for each test, the strain control was automatically transformed to crosshead displacement control, hence stress-strain behavior beyond the strain of 0.25 was not provided.

Samples for microstructural analysis were cut in a direction parallel to the tensile axis using an electro-discharging machine. For optical microscopy (OM) and electron backscatter diffraction (EBSD) observations, samples were ground and polished, then etched with solutions of 4 % Nitral and acetic-picral (5 mL acetic acid (CH₃COOH), 6 g picric acid 100 ((NO₂)₃C₆H₂OH), 10 mL distilled water and 100 mL ethanol (95 %)), respectively. The mechanical twins were identified by a Zeiss Supra 55 FEG scanning electron microscope equipped with a HKL-EBSD system. For transmission elec-
Electron microscopy (TEM) observation, samples obtained from thin slice with 0.4 mm thickness were mechanically thinned to 30 µm, and then thinned to perforation by ion polishing using a Gatan precision ion polishing system. Submicrostructures of the deformed samples were then observed by TEM (Hitachi H-800) operating at 200 kV.

3. Results

3.1. Flow stress behavior

Several typical stress-strain curves of the investigated Hadfield steel under strain control at room temperature are shown in Fig. 2. The flow stress including yield strength increases gradually with the decrease of applied strain rate, showing a negative SRS. In addition, obvious serrated flows are detected at all selected strain rates, and the type of which exhibits a visible strain rate dependency. For clarity, each stress-strain curve is shifted up one to another by 50 MPa (Fig. 3), and the marked section in Fig. 3a is magnified as shown in Fig. 3b. Clearly, the amplitude and the frequency of the serrated flow decrease significantly as the applied strain rate increase, showing positive SRS.

At strain rates ≤ 3 × 10⁻⁵ s⁻¹, type C serrations, characterized by abrupt stress drops to below the general level of the flow stress curve followed by an immediate re-rise of the stress, appear as a certain critical strain is reached. The critical strains at the strain rates of 3 × 10⁻⁵ s⁻¹ and 6 × 10⁻⁶ s⁻¹ are 0.0228 and 0.037, respectively, demonstrating a positive SRS. It should be emphasized that this type C serrations are never reported in Hadfield steels even at high temperature. Serrations appearing at the high strain rates of 6 × 10⁻⁴ and 6 × 10⁻³ s⁻¹ are consistent with the results of previous study on similar Hadfield steel [11], in which similar large and dense serrated flows appear with strain rates ranging from 1 × 10⁻⁴ to 1 × 10⁻³ s⁻¹. Detailed analysis reveals that the large and dense serrations occurring at relatively higher strain rates are composed of type A and B as shown in Fig. 3b. The type B serrations, characterized by quick stress oscillations around the envelope of stress-strain curves, are detected promptly after yielding, while the type A ones occur at increased strain levels. The type A serrations in present study are featured by stress peaks appearing between two adjoining type B serrations, differing from the conventional type A ones typified by sharp stress peaks [16]. In addition, the type B serrations are also different from the conventional type B ones defined by small magnitude oscillations [21].

Three typical sections of stress- and displacement-time curves under strain control tensile are shown in Figs. 4a, 4b and 4c for the strain rates of 6 × 10⁻⁴ s⁻¹, 3 × 10⁻⁵ s⁻¹ and 6 × 10⁻⁶ s⁻¹, respectively. The displacement-time curves at the selected strain rates are step-like, while the stress-time curves show significant differences in shape of serrations. At the high strain rate of 6 × 10⁻⁴ s⁻¹, the serrations of type A and B appear simultaneously, exhibiting arc-shaped rises of stress concomitant with high frequency and high magnitude oscillations as illustrated in Fig. 4a. Note that the arc-shaped rises are extremely fuzzy as compared with those in high manganese austenitic TWIP steel [21]. At the strain rate of 3 × 10⁻⁵ s⁻¹, type C serrations, besides type A and B ones, appear as a certain critical strain is reached, and the serration type is gradually prevailed by type C, as shown in Fig. 4b. As the applied strain rate decreases to 6 × 10⁻⁶ s⁻¹, the serrations of type A and B have disappeared, and the flow stress of the studied Hadfield steel only exhibits type C serration (Fig. 4c). In addition, completely different serration characteristics with the change of the applied strain rate are detected, which are also quite different from previous studies [3,11]. Furthermore, the stress oscillations of type A and B are scarcely consistent with the variation of crosshead displacement, in other words, they coincide with the abrupt increments of crosshead displacement occasionally, as shown in Fig. 4a. However, for type C ones, the burst-ups of stress are basically consistent with the abrupt increases of crosshead displacement, as illustrated in Fig. 4b and 4c.
Fig. 4 – Stress- and displacement–time curves of the studied Hadfield steel at the strain rates of (a) $2 \times 10^{-3}$ s$^{-1}$, (b) $3 \times 10^{-5}$ s$^{-1}$ and (c) $6 \times 10^{-6}$ s$^{-1}$.

Fig. 5 – Schematics of the deformation bands: (a) initiating outside and propagating inside the gauge length and (b) initiating within and propagating outside the gauge length.

According to previous researches [21,22], type A serrations are stem from locking of dislocations, the arc-shaped rises of stress should correspond to the initiation of type A serrations. In present work, all the tensile tests were conducted under constant strain rate control; the real strain rate was determined by the strain gauge displacement speed measured by the extensometer. Therefore, each arc-shaped rise in flow stress curves represents a sudden, high localized deformation band. Based on the nucleation location, the detected serrations can be simply divided into two categories. Under constant strain rate control tensile, the arc-shaped increments of stress occasionally coincide with the large and abrupt increase of crosshead displacement (Fig. 4a), indicating that the sudden, high localized deformation bands primarily nucleate inside as well as outside the gauge length occasionally. The deformation bands then gradually propagate over the gauge length, as schematically shown in Figs. 5a and 5b. Similar to the deformation bands of type A, the type B ones also nucleate principally inside as well as occasionally outside the gauge length. As shown in Fig. 4b, most of the abrupt stress drops at the strain rate of $3 \times 10^{-5}$ s$^{-1}$ coincide with the big burst-ups of the crosshead displacement, illustrating that the deformation
bands of type C mainly nucleate outside the gauge length, as opposite to that of type A and B. When the applied strain rate decrease to $6 \times 10^{-6}$ s$^{-1}$, the stress variation and the crosshead displacement coincide with each other (Fig. 4c), meaning that all the type C deformation bands here nucleate outside the gauge length.

3.2. Microstructure evolution

To understand the accumulation process of microstructures during uniaxial tensile deformation, the microstructural evolution with typical compression strains ($\varepsilon = 0.05$, $\varepsilon = 0.15$, $\varepsilon = 0.25$ and $\varepsilon = 0.349$) are shown in Fig. 6. Clearly, the deformation bands always appear in the form of groups, in each group the deformation bands are parallel to each other. With the increase of applied strain, the deformation bands in the same grain transform from one group to two ones with different directions, and then the two groups of deformation bands gradually cross with each other until a lattice-like deformed microstructure is formed. According to previous investigations [24-25], these deformation bands can be ascribed to the generation of mechanical twins with the aid of EBSD and TEM. The area fraction of twins assumed to be equivalent to the volume fraction ($V_T$), was calculated using a point counting analysis with a rectangular net [26]. The calculated $V_T$ at the selected strains of $\varepsilon = 0.05$, 0.15, 0.25 and 0.349 are 14.8 %, 65.4 %, 74.3 % and 76.5 %, respectively. This illustrates that, with the increase of applied strain, the increase rate of $V_T$ drops apparently, and then gradually tends to be saturated.

At the strain of 0.05, part of grains remain undeformed and a few parallel deformation bands are observed in some grains, as compared with the original microstructure in Fig. 1. When the applied strain increases to 0.15, the density of the deformation bands increases significantly and the corresponding spacing reduces. In addition, two groups of deformation bands are detected in most of the grains, and some of which are intersected to each other. As the applied strain increases to 0.25, the further increase in density of deformation bands, as well as the decrease of spacing, greatly promotes the intersection of deformation bands. Some high-density lattice-like deformation bands are detected as shown in Fig. 6c. At failure state with $\varepsilon = 0.349$, the two groups of deformation bands in most grains are intersected to each other.

In order to reveal the fracture behavior of the studied Hadfield steel, typical OM of the fractured specimens at the applied strain rates of $6 \times 10^{-3}$ s$^{-1}$ ($\varepsilon = 0.384$) and $3 \times 10^{-5}$ s$^{-1}$ ($\varepsilon = 0.349$) are shown in Figs. 7 and 8, respectively. The crack nucleates and initially propagates at a position where the direction of twins is approximately perpendicular to the tensile load, as shown in Fig. 7b. The main crack propagates primarily in the direction perpendicular to the tensile load and deflects frequently during the propagation process. As shown in Figs. 7-9, the Hadfield steel principally exhibits transgranular fracture, and a small amount of intergranular fracture along GBs and twin boundaries (TBs) also occur as labeled with various color arrows. In other words, the path of crack propagation of the studied Hadfield steel is the mix of transgranular and intergranular with direction approximately perpendicular to the tensile axis. Furthermore, there are many small cracks distributing around the main crack, as marked with arrows in Figs. 7-9, indicating that the fracture of the studied Hadfield steel is carried out by the continuous expansion of main.
crack, as well as the initiation, propagation and convergence of microcracks. Therefore, a serrated crack distributing approximate along the direction perpendicular to tension load is formed.

The deformation microstructures of the studied Hadfield steel were also characterized by EBSD and TEM, as shown in Figs. 10 and 11. Black, red, blue, pink and yellow lines represent the boundaries of grain, Σ3, Σ7, Σ9 and Σ11, respectively, and the observation region of the twin diffraction pattern in Fig. 11a is labeled with red circle. Generally, the microstructure characteristics obtained by EBSD and TEM are consistent with the OM results in Fig. 6. At the strain of 0.05, a small number of primary deformation twins (DTs) which generate in most of the grains, are detected to appear from GBs as displayed in...
Fig. 10 – EBSD micrograph of the studied steel deformed at $3 \times 10^{-5}$ s$^{-1}$ strain rate: (a) $\varepsilon = 0.05$, (b) $\varepsilon = 0.25$.

Fig. 9 – Magnified section of crack tip corresponding to 3 in Fig. 7.

Fig. 11 – Bright field TEM images of the studied material at $3 \times 10^{-5}$ s$^{-1}$ strain rate: (a) $\varepsilon = 0.05$, (b) $\varepsilon = 0.25$.

is consistent with the result in Fig. 6a. In addition, the DTs are also detected by TEM and one example is shown in Fig. 11a. The twin thickness/spacing is beyond the resolution limit of the EBSD technique; as a result, the twin density is visibly lower than that in Fig. 6a. Furthermore, there are also some dislocation tangles detected between the deformation twins as shown in Fig. 11a. As the applied strain increases to 0.25, high density DTs are frequently observed as shown in Figs. 10b and 11b. The average twin thickness in Fig. 11b is 29 nm, which is visibly thinner than that at the applied strain of 0.05 (1247 nm). In addition, secondary DTs can also be found in some grain as shown in Fig. 10b.

4. Discussion

Previous tensile tests were almost all conducted under crosshead displacement control [1,7–12], and the special type C serration detected in present work has never been reported in Hadfield steels regardless of strain rate and temperature [10,8–12]. However, the tensile tests in present study were conducted under constant strain rate control, and interestingly three types of serrations including type C ones and two initiation modes of deformation bands were revealed. These different observations may be accountable considering that the actual strain rate is the average value under crosshead displacement control; however, the strain rate under extensometer-measured strain control is monitored and controlled by the extensometer. This control method is more sensitive to local strains as compared with the crosshead displacement-controlled way.

In the previous studies, the absence of type C serrations in Hadfield steels at room temperature has been accepted as an unusual behavior as compared to Al-Mg alloys and other high manganese austenitic steels. According to Cottrell model modified by McCormick [27] and van den Beukel [28], the DSA effect represents a case where collective motion of defects readily leads to macroscopic strain and stress jumps on the flow stress via the diffusion of solute atoms to the mobile
dislocations temporarily immobilized at obstacles. It is well known that the diffusivity of solutes at room temperature, including that of interstitial carbon atoms, is extremely low and the diffusion of carbons to temporarily arrest dislocations is impossible. However, it is reported that a single diffusive hop or reorientation of carbon members of C-Mn couples in the cores of dislocations or C-vacancy point defect pairs in the stress field of dislocations is energetically feasible at room temperature [20]. Therefore, the reorientation of interstitial C atoms in these points defect couples and the corresponding interactions with dislocations can temporarily immobilize the dislocations, which finally promotes the increases of stress.

When dislocation glide is hindered by obstacles, there is an elapse of time before the immobilized dislocations overcome the obstacles. According to Orowan equation, the waiting time \( t_w \) is proportional to the density of mobile dislocations \( \rho_m \), the dislocation mean free path \( L \) and the inverse of stress rate \( \dot{\tau} \). Therefore, the waiting time decreases as the strain rate increases, leaving less time for C atoms to age the interrupted dislocations. Consequently, the aging effect cannot be formed fully at high strain rate because the waiting time is not enough. As a result, the probability of immobilized dislocations to overcome the obstacles increases. This explanation assists in understanding the phenomenon of denser serrations and larger stress drops at high strain rate. Furthermore, the process of DSA in Hadfield steels with strain ranging from \( 1 \times 10^{-5} \) s\(^{-1} \) to \( 1 \times 10^{-1} \) s\(^{-1} \) is dominated by the waiting time, considering its high self-diffusion activation energy at room temperature [11]. At strain rates of \( 1 \times 10^{-3} \) s\(^{-1} \) and \( 1 \times 10^{-4} \) s\(^{-1} \), the strain distribution within the gauge length is non-uniform, and the serration frequency and the amplitude of stress drop are obviously higher as compared with those at low strain rates [11]. The more visible inhomogeneous distribution of strain at strain rate of \( 1 \times 10^{-3} \) s\(^{-1} \) result in a stress drop with high frequency and amplitude. Based on the previous research findings, it can be safely said that the comprehensive effects of waiting time and non-uniformity of strain distribution result in the denser serrations and larger stress drops at high strain rate.

As the strain increases, not only the defects such as dislocations and vacancies rapidly proliferate assisting in the diffusion of solutes, but also deformation twins sharply increase (Figs. 6 and 11) reducing the planar slip length, since the planar slip length can neither exceed the spacing of the twin lamella nor the grain size. The pinning effect generated from the reorientation and hop of carbon members of C-Mn in the cores of dislocations or C-vacancy point defect pairs in the stress field of dislocations is gradually enhanced and therefore the magnitude of DSA hump is relatively lower at high strain levels compared with that at low strains.

The process of crack initiation and propagation is roughly as follows: microcrack initiation induced by accumulation of plastic deformation \( \rightarrow \) crack coalescence within grain \( \rightarrow \) quick crack propagation beyond GBs \( \rightarrow \) fracture. Three crack initiation modes marked with various color arrows, are detected, as shown in Figs. 8b and 9. (i) Severity pile-up of dislocations resulted from the local concentration of deformation bands due to inhomogeneous deformation, induces the initiation of cracks; (ii) The relatively low interface strength between large second phases and austenite matrix is liable to cause the pile-up of dislocations within the interface, and then the crack may initiate when the released strain energy is larger than its interface energy; (iii) GBs act as strong obstacles to the dislocation motion, also can effectively induce the pile-up of dislocations and promote crack initiation finally. Accordingly, the critical shear stress required for crack initiation and propagation along twin direction is relatively lower than those in other directions [29,30], in other words, the twins distributing perpendicular to the tensile direction can effectively induce the initiation and propagation of cracks. Consequently, the crack nucleates and initially propagates at a position where the direction of twins is approximately perpendicular to the tensile load, as shown in Fig. 7b.

The crack propagation mode mainly includes intergranular propagation along the interface of second phase/matrix, transgranular propagation at the location piled up with dislocations and propagation at the re-nucleated crack tips ([31,32]). During the propagation, the tip of main crack may be blunted due to the plastic deformation localized around the tip, which hinders the crack propagation and makes the main crack deflected to a weak direction, as marked in Figs. 7b and 9. Simultaneously, multiple microcracks occurs in the front of the main crack, especially on its propagation path as shown in Figs. 7-9. The microcracks after forming propagate preferentially along TBs and GBs, and most of which are stopped and blunted as encounter with the main crack (Fig. 7b and Fig. 9). Only when the microcrack propagates long enough and is blunted wide enough, can it affect the principal crack propagation and the distribution of local stress and plastic strain. In addition, the cracks including main crack and microcracks may continually propagate until traversing the second phase particle as the tips reach the interface of second phase/matrix due to its low interface strength, as shown in Figs. 9 and 8b respectively. Furthermore, the main crack may be deflected due to the strengthening effect stem from the accumulation of dislocations, and then propagates to the dense deformation band region. Thereby, a serrated crack distributing along the direction perpendicular to tension is formed.

5. Conclusions

The serrated flow and failure behaviors of Hadfield steel was investigated at room temperature and the following conclusions are drawn from this work:

1. Three types of serrations including type A, B and C, depending on the applied strain rate, occur at room temperature with constant strain rate control. Specifically, type A serrations concomitant with type B ones are observed in high strain rate region, whereas type C ones are identified in the low strain rate region.

2. As compared to low strain rates, the high strain rate suppresses the pin effect and eventually results in the denser serrations and larger stress drops. The deformation bands of type A, B and C are initiated both outside and inside the gauge length. When the applied strain rate decreases to \( 6 \times 10^{-6} \) s\(^{-1} \), all the type C serration are initiated outside the gauge length.
(3) The main crack nucleates at the position with dense twins, propagates primarily in the direction perpendicular to the tensile axis and deflects frequently due to the interaction with GBs and TBs.

Uncited references

[23].

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