Effect of Hot-rolling Parameters on the Microstructure and Property of Fe–Mn–Al–C Micro–laminated Dual Phase Steels

Mingda ZHANG,1,2)# Haifeng XU,1) Wenquan CAO1) and Han DONG1)

1) Special Steel Institute, Central Iron and Steel Research Institute, Haidian District, Beijing, 100081 China.
2) School of Materials Science and Engineering, Tsinghua University, Haidian District, Beijing, 100084 China.

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In this study, two kinds of micro-laminated dual phase steels with different strength were produced by simple hot rolling and air cooling processes. High carbon and aluminum contents contribute to the high strength in this kind of steel, which has close relationship with the phase ratio and micro-hardness of the ferrite and martensite phases. They were hot rolled at two temperatures (950°C and 1 100°C) with different reductions (30% and 70%). The mechanical property results reveal that the effects of hot-rolling temperature and reduction on tensile properties are small. The fractography of the samples was examined to explore the relationship between the micro-laminated microstructure and the mechanical properties. It is worth noting that the impact toughness of the rolling plane is enhanced obviously at low hot-rolling temperature with large reduction. The increase of phase interface density along the thickness direction contributes to the ultrahigh impact toughness of rolling plane.

KEY WORDS: dual phase microstructure; laminated material; mechanical property; toughness.

1. Introduction

Traditional dual phase (DP) steel is a kind of typical multiphase steel, which is usually composed of ferrite and martensite/austenite (MA) phases.1) Hard granular MA islands and soft ferrite matrix are the characteristic microstructure of the traditional DP steel. The composite microstructure contributes to the good comprehensive mechanical properties, such as low yield strength, continuous yielding, good plasticity and good formability.2–4) By changing the shape and the ratio of ferrite and martensite/austenite phases in DP steel, both high strength and ultrahigh toughness can be obtained in the new kind of micro-laminated DP steel.5) The micro-laminated dual phase microstructure has close relationship with the formation of δ-ferrite and the extension of δ+γ dual phase region at high temperature due to the addition of aluminum.6,7) The microstructure and mechanical properties of this kind of steel were discussed in our previous work.2) An excellent combination of high strength and ultrahigh impact toughness was obtained in this kind of steel. However, the effects of hot rolling process on the microstructure and mechanical properties are still not researched systematically.

In this study, the microstructure and the mechanical properties of the micro-laminated dual phase steels were examined to understand the influence of hot-rolling parameters. Both the hot-rolling temperature and reduction have great effect on the ultrahigh impact toughness of the rolling plane, which is related to the special micro-laminated dual phase microstructure.

2. Methodology and Experiment

Two steels were designed in this study, the chemical compositions of steel A and steel B are shown in Table 1. These steels were first smelted in a 50 kg vacuum induction Table 1. Chemical composition of the investigated alloys (mass%).

| steel | C  | Mn | Al | Fe  |
|------|----|----|----|-----|
| A    | 0.05 | 5  | 3  | bal. |
| B    | 0.20 | 5  | 4  | bal. |

Fig. 1. The schematic illustration of the hot rolling process.
furnace and cast into ingots. Then the ingots were hot forged into square slabs with the thickness of 40 mm (A/B-1 and A/B-2) and 20 mm (A/B-3 and A/B-4) at 1 200°C. These slabs were homogenized at 1 100°C (A/B-1 and A/B-3) and 950°C (A/B-2 and A/B-4) for 2 hours (h), and subsequently hot rolled to 12 mm, followed by air cooling to room temperature. The hot-rolling start temperatures are close to the homogenization temperatures. The temperature drop during the hot rolling process is less than about 100°C. Figure 1 shows the schematic illustration of the hot rolling process.

The samples for microstructure analysis were polished and etched in 4% nital. The microstructure of the samples was observed by optical microscopy (OM) and scanning electron microscopy (SEM). The micro-hardness of different phases was measured in the transverse plane under a load of 50 g with a loading time of 10 s. Two groups (L–S and L–T) of standard samples for Charpy V-notch impact tests were cut with different sampling directions to examine the anisotropy of the impact toughness. L–T (resp. L–S) configuration was investigated in which L corresponds to the axial direction and T (resp. S) to the notch direction. The impact tests were conducted at both −40°C and 25°C. The dog-bone shaped tensile specimens were cut with the axial orientation parallel to the rolling direction (L direction). The tensile tests were conducted in a universal testing machine at room temperature with a strain rate of 10⁻³ s⁻¹. Four impact samples (two for L–T and two for L–S) and two tensile samples were prepared for each hot rolled state. More details of the sampling information are shown in Fig. 2. The fracture surface of Charpy V-notch impact samples and tensile samples were examined. The phase interface numbers of unit length along the thickness direction were also counted according to the metallographic photograph.

3. Results

3.1. Phase Diagram

(0–1)C5Mn3Al phase diagram and (0–1)C5Mn4Al phase diagram were calculated by using the Thermo-Calc software with TCFE 7 database as shown in Figs. 3(a) and 3(b). Aluminum has a great effect on promoting the formation of ferrite phase. It not only increases the stabilization of α-ferrite but also expands the δ-ferrite phase region. The α+γ and δ+γ dual phase regions are expanded and connected by adding aluminum in these steels. All the hot working processes in this study, including hot forging and hot rolling, were carried out in the dual phase region. No other phase formed at these hot working temperatures. While the ratio of δ phase and γ phase varies with different temperatures.

3.2. Microstructure

Figure 4 shows the optical microstructure of these hot rolled steels. The bright and dark regions correspond to ferrite phase and martensite phase, respectively. Different from the inhomogeneous two phase microstructure before heating and rolling, the ferrite and martensite phases in the hot rolled steels arranged alternatively along the short transverse direction. The volume fraction of ferrite lamellae in A steel is between 45–50%, while 30–35% in B steel, which are much smaller than the traditional DP steel.2–4,8,9) The phase interface between ferrite and martensite phases are flat and smooth. The thickness of ferrite and martensite lamellae decreases with the hot-rolling reduction. A large number of small white regions are found in the original martensite lamellae in the steels hot rolled at 950°C as shown in Figs. 4(b), 4(d), 4(f) and 4(h). Plastic deformation occurred more severely in the matrix along the longitudinal direction than along the transverse direction through the comparison of the images in the LxS plane and the TxS plane. The ratio of length to width of the different phases in the TxS plane is less than that in the LxS plane as shown in Figs. 4(a), 4(e), 4(i) and 4(j).

The SEM micrographs show more details of the martensite lamellae in these steels as shown in Fig. 5. The morphology of martensite phase in B steel is different from that in A steel. Higher carbon and aluminum contents contribute to the fine microstructure of martensite phase in B steel. New ferrite phase is found in the original martensite phase in the steels hot rolled at 950°C. The new ferrite phase exhibits the similar deformation characteristic with the ferrite and
martensite lamellae. The new ferrite phase volume fractions are about 10% in A steel and 20% in B steel. The distribution of the new ferrite phase in B steel is more dispersed and uniform than that in A steel.

3.3. Mechanical Properties

Figure 6 shows the micro-hardness of the ferrite and martensite phases in these steels. With different hot-rolling parameters, the micro-hardness of the ferrite and martensite phases in A steel keeps at around 185 HV and 425 HV, respectively. The micro-hardness of the ferrite phase in B steel is about 205 HV, which is a little higher than that in A steel. The micro-hardness of the martensite phase in B steel is extremely high. It changes heavily at different hot-rolling temperatures with different reductions. High carbon content induced the high sensitivity of martensite proper-
ties in B steel. The micro-hardness of the martensite phase decreases with hot-rolling temperature mainly due to the newly formed ferrite. The small reduction of hot rolling can also decreases the micro-hardness of the martensite phase.

The high micro-hardness of both ferrite and martensite phases in B steel indicates its high strength. The mechanical properties of these steels change little with different hot rolling processes.

The impact test results are given in Table 3. The impact toughness in L–S direction is much higher than that in L–T direction. The impact toughness in the L–S direction increases with the hot-rolling reduction and decreases with the hot-rolling temperature in these steels, while the fluctua-

| Table 2. Mechanical properties of these steels @25°C. |
|-----------------------------------------------|
| steel | Yield strength (MPa) | Ultimate tensile strength (MPa) | Uniform elongation (%) | Total elongation (%) | Reduction of area (%) | Yield ratio |
|-------|---------------------|---------------------------------|-----------------------|---------------------|-----------------------|-------------|
| A-1   | 519                 | 853                             | 9.8                   | 22.3                | 65.0                  | 0.608       |
| A-2   | 509                 | 899                             | 10.3                  | 20.5                | 55.5                  | 0.566       |
| A-3   | 488                 | 805                             | 10.8                  | 22.8                | 67.0                  | 0.606       |
| A-4   | 517                 | 872                             | 10.3                  | 20.3                | 59.5                  | 0.594       |
| B-1   | 700                 | 1,468                           | 8.8                   | 12.0                | 23.0                  | 0.477       |
| B-2   | 574                 | 1,374                           | 10.3                  | 12.0                | 19.5                  | 0.418       |
| B-3   | 690                 | 1,355                           | 9.5                   | 11.0                | 19.0                  | 0.509       |
| B-4   | 586                 | 1,373                           | 10.8                  | 13.8                | 20.0                  | 0.427       |

| Table 3. Impact toughness @−40°C and 25°C. |
|-----------------------------------------------|
| steel | Charpy V-notch impact absorbing energy (KV2) |
|       | @−40°C (J) | @25°C (J) |
|-------|-------------|-----------|
| A-1   | 302         | 60        | 320       | 101        |
| A-2   | 379         | 49        | 385       | 63         |
| A-3   | 19          | 16        | 111       | 81         |
| A-4   | 285         | 53        | 305       | 71         |
| B-1   | 65          | 8         | 174       | 23         |
| B-2   | 123         | 12        | 201       | 27         |
| B-3   | 34          | 6         | 53        | 11         |
| B-4   | 51          | 9         | 106       | 17         |
tion of the impact toughness in the L–T direction is small. Hot rolling at 950°C with 70% reduction contributes to the highest impact toughness in the L–S direction, nearly 400 J for A steel and 200 J for B steel at room temperature.

4. Discussion

The microstructure of these steels is composed of ferrite and martensite lamellae. During the cooling process after hot forging, the austenite phase transformed into martensite phase due to the low stability of austenite in these steels. During the heating process before hot rolling, austenite nucleated and grew in original martensite phase. The ratio of δ phase and γ phase varies with different temperatures in these steels. According to the phase diagrams as shown in Fig. 3 and the lever law, the ratio of δ phase and γ phase at 1 100°C (hot rolling) is close to the ratio at 1 200°C (hot forging). Almost all the original martensite phase transformed into austenite during the heating process. While the ratio of δ phase and γ phase at 950°C is higher than that at 1 200°C. Austenite also nucleated and grew in original martensite phase at 950°C, at the same time, partial original martensite phase transformed into new ferrite phase. The distribution of new ferrite phase has close relationship with the morphology of the original martensite phase. The dispersed and uniform distribution of the new ferrite in B steel is mainly due to the fine microstructure of martensite phase. During the hot rolling process, the new ferrite was compressed and extended together with other phases. As the martensite transformation occurred again during the air cooling process, both the simple ferrite and martensite micro-laminated microstructure (hot rolling at 1 100°C) and the ferrite and martensite micro-laminated microstructure with new ferrite (hot rolling at 950°C) were obtained finally. The new ferrite increased the number of phase interface between ferrite and martensite along the thickness direction. The property of the nearby martensite phase was also affected by the formation of new ferrite, and then changed the micro-hardness of the original martensite phase. The competition relationship between the nearby martensite phase and the new ferrite caused the change of the original martensite phase micro-hardness or not as shown in Fig. 6.

Figure 7 shows the engineering stress-strain curves of these steels. As can be seen, the alloy composition is the most important factor of the mechanical properties in these steels. Hot-rolling temperature and reduction have very small effect on the tensile properties. In these micro-laminated dual phase steels, the strength mainly depends on the phase ratio and the phase property of different phases, which are sensitive to the composition. Carbon and manganese increase the martensite phase volume fraction. Aluminum promotes the formation of ferrite phase significantly. Two groups of elements control the phase ratio of these steels. The ferrite volume fraction of A steel is higher than B steel. In addition, carbon usually increases the hardness of ferrite and martensite phases by solid solution strengthening and precipitation strengthening. Aluminum also increases the hardness of martensite phase by solid solution strengthening. High carbon and aluminum contents contribute to the high strength of B steel.

The plasticity of DP steel has close relationship with the plasticity of each phase and the deformation coordination between these phases. Different from the traditional DP steel, these steels in this study with high martensite volume fraction still exhibit good plasticity. The micro-laminated microstructure limits the formation of micro voids and propagation of cracks along the cross-section. However, cracking along the lamellar interface may occur during the tensile deformation. The fractography of tensile specimens proves this conjecture as shown in Fig. 8. Lamellar fracture is the fracture feature of these micro-laminated dual phase steels, which is similar to the laminated materials. The weak phase interface between ferrite and martensite lamellae is the main reason of the lamellar fracture. The stress-strain relationship is in complex stress condition after the necking starts during the tensile process. The stress perpendicular to the axial direction first destroyed the weaker phase interface between ferrite and martensite phases. Then the separate lamellar broke respectively. Tearing morphology was found in fractography of tensile specimens. The feature of brittle fracture was found in the tensile samples of B steel as shown in Figs. 8(c) and 8(d). Lower plasticity of a certain phase and large performance difference between different phases induced the relatively low plasticity of B steel.

Lamellar fracture feature is also found in the macroscopic fractography of the impact specimens as shown in Fig. 9. Part of the L–S direction sample remains connected after impact at −40°C and 25°C. Apparent splits can be observed in the fracture surface of the impact samples which were impacted in the L–T direction. The macroscopic fractography of A steel shows more plastic deformation characteristic compared with B steel, which is similar to the tensile test. More white areas appeared in the samples impact at −40°C. The loss of plasticity caused by low tem-

![Fig. 7. Engineering stress - engineering strain curves of (a) A steel and (b) B steel.](image-url)
Figure 10 shows the comparison of the impact toughness between these steels in this study (L–S sample) and other steels as a function of the ultimate tensile strength at room temperature. For most steels, the impact toughness decreases with the ultimate strength. The micro-laminated steel in this study breaks through the limit of this rule. The impact toughness of these steels is 2–3 times as high as that of other steels at the same strength grade. The impact toughness of the samples with large reduction (A/B-1 and A/B-2) is much higher than those with small reduction (A/B-3 and A/B-4). It increases further with the relatively low hot-rolling temperature at 950°C (as shown by the arrow). The product of the ultimate strength and impact toughness could reach 346 GPa·J and 276 GPa·J for A steel and B steel, respectively.

Kimura et al. investigated the similar laminated microstructure and the delamination behavior of the tempformed steel. A yield strength of 1364 MPa and a V-notch Charpy absorbed energy of 125 J were obtained at room temperature in the sample that was tempformed at 500°C. They believed that the ultra refinement of the transverse grain structure was the key to enhancing both the yield strength and the toughness of the tempformed samples. Different from fine transverse grain structure in deformed martensite structure, the soft ferrite and the hard martensite lamellar microstructure in this study also exhibited delamination toughening behavior.

Distribution density of phase interface along the thickness direction is considered as the key factor on the ultrahigh
impact toughness. The phase interface number of unit length is measured and counted to serve as an assessment factor. Both the large hot-rolling reduction and the low hot-rolling temperature contribute to the increase of the phase interface distribution density: the former increases the phase interface number of unit length along the thickness direction increases the impact toughness of the rolling plane can be further enhanced by increasing the phase interface distribution density along the thickness direction.

5. Conclusion

In the present study, Fe–5Mn–3Al–0.05C and Fe–5Mn–4Al–0.20C micro–laminated dual phase steels were hot rolled at various temperatures (950°C and 1 100°C) with different reductions (30% and 70%). The microstructure changes induced by hot rolled parameters and their effects on mechanical properties were investigated. The property changes induced by hot-rolled parameters and their effects related to phase ratio and phase property. High carbon and aluminum contents contribute to the high strength of ferrite and martensite phases, which result in the high ultimate strength and low plasticity in B steel. The effect of hot-rolling parameters on the tensile properties is small.

(2) High distribution density of phase interface along the thickness direction increases the impact toughness of phases and martensite phases, which result in the high ultimate strength. At the same strength level, the impact toughness of the rolling plane can be further enhanced by increasing the phase interface distribution density along the thickness direction.

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