Determination of Upper Limit Temperature of Strain-induced Transformation of Low Carbon Steels

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Two quenching methods were applied in the experiments carried out on Gleeble 1500 thermomechanical simulator: (1) spurting water; (2) making the samples drop into water, and the cooling rate distributed un-uniformly in the samples quenched by the former and uniformly by the latter. The critical cooling rate of preventing ferrite precipitation was determined, and the microstructure observation position in which the microstructure could be frozen to room temperature in the samples quenched by spurting water was also determined, and using this method the $A_{\gamma\rightarrow\alpha}$ was determined to be about 830°C for tested materials. Using second quenching method, single-pass deformation experiments at different temperatures were carried out, and the microstructure analysis indicates that, as the deformation temperature decreased, the ferrite morphology changed from Widmanstätten and allotriomorphs to equiaxed at about 800°C. Combining the dilation–temperature curves measured during cooling process after deformation at different temperatures, the $A_{\gamma\rightarrow\alpha}$ was determined to be also about 830°C for tested materials.

KEY WORDS: low carbon steel; ferrite; strain-induced transformation; $A_{\gamma\rightarrow\alpha}$; quenching.

1. Introduction

The ferrite grain refinement of structural steels has attracted considerable interest from metal engineering scientists due to its unique role of increasing both strength and toughness. In recent years, a lot of investigations about the grain refinement through TMCP have been carried out. For plain low carbon steel, the strain induced transformation played important role in ferrite grain refinement. Because it is very difficult to freeze the microstructures after high temperature deformation for low carbon steel, in most circumstance, it could not be determined if the ferrite in quenching microstructures was precipitated during deformation process or cooling process.

Quenching after deformation was still widely used in most investigations. In the analysis of quenching microstructures it was thought that the equiaxed ferrite couldn’t be precipitated during the quenching process, but, because the cooling rate in the sample and the critical cooling rate of preventing precipitation of ferrite were all not clear, this conclusion seems not very convincing. Recently an in situ X-ray study was carried out in Yada’s work, and the result revealed that $\gamma\rightarrow\alpha$ transformation occurred during deformation in a wide range of temperature.

In present work, two quenching methods were applied in the experiments: (1) spurting water, in which the cooling rate distributed in the samples un-uniformly; (2) making the samples drop into water, in which the cooling rate distributed pretty uniformly. Single-pass deformation at different temperatures experiments were carried out and the samples were quenched using second method, and the $A_{\gamma\rightarrow\alpha}$ was determined through the microstructure analysis and the dilation–temperature curves during cooling process after deformation. In addition, the cooling rates at different positions of the sample quenched by spurting water were measured and the critical cooling rate of preventing ferrite precipitation was determined with the aid of microstructure characterization, moreover, the microstructure observation position in the samples was also determined, and using this method the $A_{\gamma\rightarrow\alpha}$ was also determined.

2. Experimental Procedures

Two low carbon steels with the composition given in Table 1 were selected for this investigation. The austenite grain size of the steels after reheating at 1 000°C for 3 min was about 50 $\mu$m. Hot deformation and quenching experiments were carried out on Gleeble 1500 thermomechanical simulator. Microstructures were characterized by using an optical microscope. Dilation–temperature curves post-deformation were measured on Gleeble 2000 thermomechanical simulator.

(1) The samples of steel A were heated to and held at 1 000°C, and then cooled to different temperatures and deformed at 0.7 and strain rate of 1 s$^{-1}$, after

| Table 1. Composition of studied steels, (mass%) |
|-------|-------|-------|-------|-------|
|       | C     | Si    | Mn    | P     | S     |
| A     | 0.17  | 0.09  | 0.36  | 0.013 | 0.013 |
| B     | 0.057 | 0.35  | 0.91  | 0.01  | 0.006 |
deformation the samples were quenched immediately by making the samples drop into the water. The deformation temperatures were chosen to be 900°C, 850°C, 820°C and 790°C.

(2) The samples of steel A were heated to and held at 1000°C, and then cooled to different temperatures and deformed with strain value of 0.45 and strain rate of 1 s^{-1}, after deformation the samples were cooled at the cooling rate of 3°C/s and the dilation–temperature curves were measured simultaneously. The deformation temperatures were: 1000°C, 960°C, 920°C, 900°C, 870°C, 830°C.

(3) Using samples of steel B, the cooling rate distribution in sample quenched by spurting water was determined, and the critical cooling rate preventing the precipitation of ferrite was also determined combing the microstructure analysis. The single pass deformation experiments in which the samples were deformed at different temperatures and quenched by spurting water were also carried out, the deformation temperatures were 830°C, 820°C and 800°C, the strain and strain rate were 0.7 and 1 s^{-1} respectively.

3. Results

3.1. Single-pass Deformation Experiments Using Second Quenching Method

Figure 1 was the microstructures in the center of the samples deformed at different temperatures and quenched by making them drop into water. From this figure, it was shown that the ferrite amount increased as deformation temperature decreases, and the ferrite morphology was Widmanstätten and allotriomorphs when the deformation temperature was above 820°C and equiaxed when the deformation temperature was 790°C. This indicates that the ferrite precipitation mechanism of the samples deformed above 820°C and below 800°C may be different.

3.2. Dilation–Temperature Curves During Cooling Process after Deformation

It was reported that the deformation in austenite field influenced the $A_{s3}$ of the steels.\(^{13,16,17}\) The work of Ouchi et al.\(^{17}\) indicated that for Si–Mn steels the $A_{s3}$ changed slightly as deformation temperature changed when the deformation temperature was above 880°C. In present work, the relationship of $A_{s3}$ and deformation temperature in temperature range of 830–1000°C was investigated.

The dilation–temperature curves during cooling process after deformation of the samples deformed at different temperatures were shown in Fig. 2. It was shown that the $A_{s3}$,
the points at which the curves deviated from straight part, were about 790°C when the deformation temperatures were above 900°C, and when the deformation temperatures were below 900°C the \( A_{3} \) increased as deformation temperature decreased and it coincided with the deformation temperature when the deformation temperature was 830°C. The relationship of \( A_{3} \) and deformation temperature was shown in Fig. 3. This result indicates that when the samples were deformed below 830°C the ferrite would be precipitated during deformation, and this transformation is so called “Strain Induced Transformation (SIT)”\(^{6,7}\) or “Dynamic Transformation (DT)”.\(^{3,11}\) There are also other expressions for ferrite transformation accelerated by deformation, such as “Strain Assisted Low Temperature Transformation (SALT)”\(^{8}\), “Strain Induced Dynamic Transformation (SIDT)”\(^{9}\) and “Deformation Induced Ferrite Transformation (DIFT)”.\(^{10}\) According to this result, when the sample was deformed at 820°C, a little ferrite should precipitated during deformation, and in the quenching microstructure most of ferrite was precipitated during cooling process. When the sample was deformed at 790°C, the amount of strain induced ferrite increased due to the pretty lower deformation temperature, and the carbon content of austenite untransformed during deformation was increased due to the strain induced transformation so that this part of austenite could transformed into martensite and bainite and avoid ferrite precipitation during quenching. So the reason that the ferrite morphology changed at about 800°C of deformation temperature was that the ferrite of the samples deformed at 790°C was precipitated during deformation and the ferrite of the samples deformed above 820°C were mainly precipitated during quenching process.

3.3. Determination of Critical Cooling Rate of Preventing Ferrite Precipitation and Determination of \( A_{3} \) Using the Quenching Method of Spurting Water

The locations of thermocouple for determining the cooling rate distribution in the sample quenched by spurting water and the temperature–time curves at those positions were shown in Figs. 4 and 5 respectively. It was shown that the cooling rate at different position differed very largely each other. The cooling rates drop dramatically from upper surface to center in the sample. The microstructures near the upper surface and in the part of 1/4 diameter, at which the cooling rates were about 800°C/s and 200°C/s respectively, were shown in Fig. 6. Combining the microstructure observation from the upper surface to the lower surface, the critical cooling rate preventing the precipitation of ferrite after deformation was determined to be about 400°C/s, and it was only 80°C/s when the sample was not deformed. According to this result, the deformation microstructure in the part near the upper surface, about 2–3 mm, of the sample can be frozen to room temperature without the precipitation of ferrite.

Figure 7 was the result of single-pass deformation at different temperatures for steel B, it was shown that very little amount of ferrite existed in the upper surface microstructure of the sample deformed at 830°C, and the ferrite amount of that deformed at 800°C was pretty larger. This indicates that the upper limit temperature of strain-induced transformation for steel B was also about 830°C.

4. Discussion
4.1. The Morphology of Strain-induced Ferrite

The ferrite morphology was mainly affected by the undercooling. It would be inclined to be equiaxed if it were
precipitated at small undercooling, or Widmanstätten and plates if precipitated at large undercooling. \cite{15} From Fig. 1, the ferrite in samples deformed above 820°C was mainly Widmanstätten and allotriomorphs, and this indicates that the ferrite was precipitated at lower temperature due to the fast cooling. In the sample deformed at 790°C, see Fig. 1(d), the ferrite was equiaxed and precipitated along the austenite grain boundaries. According to the dilation–temperature curves after deformation, the ferrite would be precipitated from austenite during the deformation at 790°C, and this ferrite could be said to be strain-induced ferrite. So the reason of the change in ferrite morphology could be explained as follows: when the ferrite was precipitated during deformation, it would exist generally in equiaxed morphology; when the ferrite was precipitated during cooling or holding process, its morphology was mainly affected by the undercoolings, i.e., the transformation temperatures. As to the morphology of strain-induced ferrite, two factors determined its equiaxed morphology. One is the precipitation temperature. Because the deformation that could cause strain-induced transformation was generally carried out in temperature range of \(A_{g5}-A_{s3}\), the diffusion of C and Fe atoms could occur easily and the ferrite grains prefer presenting an equiaxed appearance. The other one is the deformation. As well known, at small undercoolings both the smoothly curved and faceted ferrite/austenite interfaces can exist in the transformation, \cite{15} but the strain-induced ferrite has only smoothly curved interfaces, as shown in Fig. 1(d). It was proposed that deformation could lead to the formation of austenite grain boundary ledges, serration or bulges, \cite{18-20} and the relatively movement of austenite grains could take place during deformation. According to this, if the strain-induced ferrite were precipitated during deformation, it was difficult to form semicoherent \(\gamma/\alpha\) interfaces due to the deformation effects. So the ferrite grains presented smoothly curved interfaces and had the similar migration rates at different directions. Although deformation can increase the driving force of transformation, the increased driving force can be minimized by the rapid transformation instead of creation of semicoherent interfaces and orientation relationships. But about the influence of deformation on \(\gamma/\alpha\) interfaces further investigations are still needed.

4.2. Factors of Influencing the \(A_{s3}\)

Chemical composition and deformation parameters such as strain value and strain rate influence the \(A_{s3}\). Decreasing
the C and Mn content can make the $A_{d3}$ increase and the $A_{e3}$ should be increased. But in present work, the $A_{e3}$ is 864°C for steel A and 888°C for steel B, and both these two steels have the same $A_{d3}$. The $A_{d3}$ of steel B is lower than that of steel A relative to $A_{e3}$. As shown in Table 1, except for C, the Si and Mn content of steel B is more than steel A, maybe it was this difference of Si and Mn content, which were all substitute atoms, that resulted in the difference of $A_{d3}$ relative to $A_{e3}$. But the mechanism of the effect of composition difference on $A_{e3}$ is not clear.

Strain-induced transformation is a diffusional transformation. Increasing the strain rate has a negative effect on strain-induced ferrite transformation. Decreasing the strain rate will increase the deformation time, and there would be enough time for the transformation to take place during deformation. So there would be an upper limit strain rate for the strain-induced transformation at a fixed temperature, and there would be an optimum strain rate range for the application of strain-induced transformation in the ferrite grain refinement because, if the strain rate were pretty low the strain-induced ferrite precipitates will have enough time to grow, and if the strain rate were pretty high the strain-induced ferrite amount would be small.

Increasing the strain rate has a positive effect on strain-induced ferrite transformation. This is because of not only the increase of deformation energy but also the increase of transformation time resulted from increase of strain rate. In addition, there is a critical strain value for strain-induced ferrite transformation, and this value is generally 0.3 as reported.

### 4.3. Driving Force of Strain-induced Ferrite Transformation

The strain induced transformation occurrence needs a temperature condition. According to the results of present work, the strain-induced transformation could occur in the temperature range of $A_{d3}$–$A_{e3}$. The deformation parameters such as strain value and strain rate have influence on the $A_{d3}$, but the $A_{e3}$ cannot be above $A_{e3}$. In the temperature above $A_{e3}$, although deformation can increase the free energy of austenite, the deformed austenite would change to the un-deformed state through the dynamic recovery and dynamic recrystallization process because that the free energy level of un-deformed austenite is lower than that of ferrite.

The driving force of austenite to ferrite transformation and the effect of deformation on it were shown in Fig. 8 schematically. When the temperature was below $A_{e3}$, as shown in Fig. 8(a), it was thermodynamically possible that the ferrite precipitated from austenite. At this time, the driving force of the austenite with composition of $X_g$ decomposed into the ferrite with composition of $X^{g\gamma}_{\omega}$ and austenite with composition of $X^{g\gamma}_{\omega}$ was the line AB. When the austenite was deformed, the free energy line of austenite would rise and the reaction of ferrite precipitation from deformed austenite should be $\gamma \rightarrow \alpha(X^{g\gamma}_{\omega}) + \gamma(X^{g\gamma}_{\omega})$, and the driving force of this reaction was the line CD. Obviously deformation increased the driving force of transformation.

When the temperature was above $A_{e3}$, as shown in Fig. 8(b), it was thermodynamically impossible that ferrite precipitated from austenite because the driving force (line BC) was positive. When the austenite was deformed, the free energy line would rise to the position of $G'$, and the driving force of ferrite precipitation form deformed austenite was negative (line AC), so the precipitation of ferrite from deformed austenite was thermodynamically possible. But when the free energy of austenite increased due to deformation, the dynamic recovery and dynamic recrystallization would also take place, and obviously the driving force of this process (line EF) was larger than that of ferrite precipitation from austenite. So the process that firstly took place should be dynamic recovery and dynamic recrystallization other than the ferrite precipitation from austenite during the deformation above $A_{e3}$.

### 5. Conclusion

The upper limit temperature of strain-induced transformation $A_{d3}$ could be determined through the deformation and quenching experiments, in which the samples were quenched by making them drop into water, and the experiment of measuring the dilation–temperature curves after deformation at different temperatures. The critical cooling rate preventing ferrite precipitation was about 400°C/s, so only the microstructures in the part near the upper surface of the samples quenched by spurted water could be frozen to room temperature. The strain-induced transformation could occur in the temperature range of $A_{d3}$–$A_{d3}$, and the $A_{d3}$ cannot be above $A_{e3}$.
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REFERENCES

1) M. I. Vega, S. F. Medina and M. Chapa: ISIJ Int., 39 (1999), 1304.
2) P. D. Hodgson, M. R. Hickson and R. K. Gibbs: Scr. Mater., 40 (1999), 1179.
3) H. Yada, Y. Matsumura and T. Senuma: Proc. of Int. Conf. on Physical Metallurgy of Thermomechanical Processing of Steels and Other Metals (THERMEC-88), ed. by I. Tamura, ISIJ, Tokyo, (1988), 200.
4) P. Yang, Y. Fu, F. Cui and Z. Sun: Acta Metall. Sin. (China), 37 (2001), 592.
5) P. Yang, Y. Fu, F. Cui and Z. Sun: Acta Metall. Sin. (China), 37 (2001), 601.
6) R. Priestner: Thermomechanical Processing of Microalloyed Austenite, ed. by A. J. De Ardo et al., TMS-AIME, Warrendale, PA, (1982), 78.
7) P. D. Hodgson, M. R. Hickson and R. K. Gibbs: Mater. Sci. Forum, 284–286 (1998), 63.
8) Y. Adachi, T. Tomida and S. Hinotani: Tetsu-to-Hagané, 85 (1999), 620.
9) W. Y. Choo, K. K. Um, J. S. Lee, D. H. Seo and J. K. Choi: Proc. of ISUG2001, ed. by S. Takaki and T. Maki, ISIJ, Tokyo, (2001), 2.
10) H. Dong: Proc. of ISUG2001, ed. by S. Takaki and T. Maki, ISIJ, Tokyo, (2001), 18.
11) H. Yada and C.-M. Li: ISIJ Int., 40 (2000), 200.
12) P. D. Hodgson and M. R. Hickson: Scr. Mater., 40 (1999), 1179.
13) V. M. Khlestov, E. V. Konopleva and H. J. Mcqueen. Can. Metall. Q., 37 (1998), 75.
14) R. Prienster and P. D. Hodgson: Mater. Sci. Technol., 8 (1992), 849.
15) D. A. Porter and K. E. Easterling: Phase Transformations in Metals and Alloys, ed. by Van Nostrand Reinhold Company, New York, (1981), 318.
16) B. Sun: Proc. of Int. Conf. on Physical Metallurgy of Thermomechanical Processing of Steels and Other Metals (THERMEC-88), ed. by I. Tamura, ISIJ, Tokyo, (1988), 412.
17) C. Ouchi, T. Sampei and I. Kozasu: Tetsu-to-Hagané, 67 (1981), 143.
18) I. Tamura, H. Sekine, T. Tanaka and C. Ouchi: Thermomechanical Processing of High Strength Low Carbon Steel, ed. by Butterworths, London, (1996), 42.
19) A. Sandberg and W. Roberts: Thermomechanical Processing of Microalloyed Austenite, TMS-AIME, Warrendale, PA, (1981), 405.
20) R. K. Amin and F. B. Pickering: Thermomechanical Processing of Microalloyed Austenite, TMS-AIME, Warrendale, PA, (1981), 377.