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Study on the ferrite grain refinement during intercritical deformation of a microalloyed steel

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ABSTRACT

In the present research, the ferrite grain refinement during intercritical deformation of a low carbon microalloyed steel within the two phase ($\alpha+\gamma$) region was investigated using hot torsion testing. The physical processes that occurred during intercritical deformation were discussed by observing the optical microstructure and analyzing the flow curve responses. The shape of the flow curve suggests that the certain dynamic softening mechanisms take place during deformation. Dynamic softening mechanisms compensate for the hardening effect of deformation and gradually keep balance with it. This flow softening is the result of deformation-induced ferrite transformation and continuous dynamic recrystallization of ferrite. Strain increasing promotes both of the softening mechanisms. Consequently, ultrafine ferrite grains continuously nucleate not only at grain boundaries but also inside austenite and pre-eutectoid ferrite. As a result, ultrafine ferrite with average grain size of ~1.8 μm achieved. It is concluded that with strain increasing, in addition to deformation-induced ferrite transformation, continuous dynamic recrystallization of ferrite contributes to the further ferrite grain refinement.

1. Introduction

In recent years, several thermomechanical processing methods in which ultrafine ferrite (UFF) grains has been produced in microalloyed and plain carbon steels investigated [1–6]. One of the methods with most promising for producing UFF steels is intercritical rolling within the two phase ($\alpha+\gamma$) region. This method is one of the hot rolling strategies for making hot rolled steel strip. Moreover, it has received much attention due to its potential to broaden the product range and reduce the production costs. However, the control of the sheet thickness and the final sheet properties are more complex than in the standard austenitic rolling strategies. The origin of the increased complexity is in the concurrency of several possible microstructural events. A better knowledge and control of the microstructural events could contribute to improve the rolling schedule of such materials. Deformation-induced ferrite transformation (DIFT) is one of the microstructural events that can take place during intercritical deformation of steels and leads to the extensively microstructural refinement [7–9]. Although the processes of restoration and transformation from the single phase deformed austenite have been relatively well understood, however, deformation of austenite/ferrite mixtures and the subsequent transformation of the remaining austenite and pre-eutectoid ferrite have received very little attention. In particular, the dominant mechanisms controlling the formation of ultrafine ferrite grains through intercritical deformation of austenite have yet to be clearly defined. Accordingly, the aim of the present research is to study the mechanisms of ferrite grain refinement during intercritical deformation of a microalloyed steel using torsion testing, with particular attention to understand how to attain a homogeneous, equiaxed, and ultrafine ferrite microstructure.

2. Experimental procedures

A low carbon Nb–Ti microalloyed steel was used in the present research. The chemical composition (wt%) was 0.09C, 0.4Si, 1.5Mn, 0.014P, 0.008S, 0.031Nb, 0.013Ti, 0.028Al, and balance Fe. Torsion samples with a gauge length of 20 mm and diameter of 6.7 mm were machined from the as-received plates with the longitudinal axis parallel to the rolling direction. The deformation tests were performed on a torsion equipment described elsewhere [10]. Samples were enclosed in a quartz tube with a positive pressure of Argon gas atmosphere to prevent decarburization during induction heating. The temperature was monitored by two thermocouples inserted in the drilled ends of the samples. The temperature accuracy was within ±5 °C. Initially a continuous cooling torsion test was conducted to measure the critical transformation temperatures. The A_{r1} and A_{r3} were found to be 742 and 780 °C, respectively.

The deformation schedule for the study of the microstructure evolution involved austenitization at 1200 °C for 2 min, followed

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by cooling at a rate of 1 °C/s to a deformation temperature of 750 °C (within the two phase ferrite plus austenite region) and held for 120 s. Then, the deformation conducted at 750 °C up to different amount of strain with a constant strain rate of 0.1 s^{-1}. To follow the evolution of the ferrite microstructure as a function of strain, the tests were interrupted at the selected strains and the microstructure was retained by quenching the samples with water jet sprays immediately, 0.2 s, after deformation. Then, deformed samples with different strain were mounted so that the microstructure of the plane containing the shear direction and the shear plane normal could be examined. Specimens were prepared for optical metallography in the standard manner.

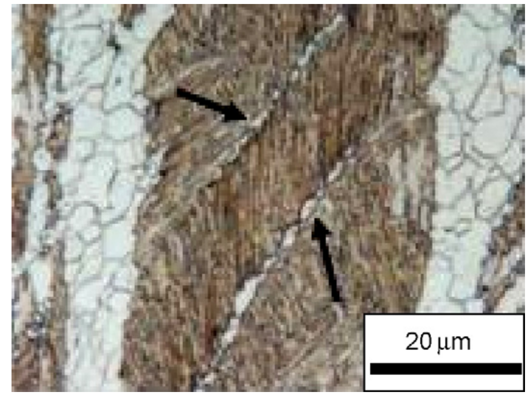

Fig. 1. Intragranular ferrite nucleation sites in deformed austenite (i.e., deformation bands shown by arrows).

3. Results

3.1. Microstructural development during intercritical deformation

The microstructure of as-received material was consisted of near-equiaxed primary ferrite grains having approximately 95% volume fraction with average grain sizes of 13 µm and the rest was pearlite and carbides. Plastic deformation can, in principle, influence the kinetics of ferrite formation via an increase in nucleation site density, change in grain shape and an increase in the stored energy within the grains. As a typical example, Fig. 1 indicates that for deformed austenite more ferrite nucleation sites are available. As can be seen, the austenite grain boundaries are completely decorated by ferrite followed by intragranular ferrite nucleation on deformation bands (shown by arrows). It is seen that in deformed austenite there are additional nucleation sites inside the austenite grains, such as deformation bands, twin bands and also dislocation arrays, which increase the effective grain boundary area.

Fig. 2(a) presents typical example of the true stress–true strain flow curve obtained from the torsion testing of investigated steel within the two phase ($\alpha + \gamma$) region. As can be seen, the flow curve increases and reaches to a maximum value at the early stage of deformation, and the slope of the flow curve is decreased gradually. The shape of the flow curve suggests that the certain dynamic softening mechanisms compensate for the hardening effect of deformation and gradually keep balance with it. This softening behavior is the result of both DIFT occurrence and continuous dynamic recrystallization of ferrite, which will be discussed in more detail in the following section. In addition, it should be mentioned that the dynamic softening occurred could be partly due to dynamic recovery of austenite as well.

Fig. 2(b) shows the microstructure of a sample which after reheating at 1200 °C and cooling down at the rate of 1 °C/s was isothermally held at 750 °C for 120 s, without deformation, and then water quenched immediately. As is seen, the microstructure is consisted of polygonal pre-eutectoid ferrite grains and the rest is austenite which has been transformed to martensite during quenching.

Fig. 2(c) through (f) shows the development of the ferrite grain refinement with increasing of strain during intercritical deformation. The microstructures consisted of martensite (gray areas) and ferrite (white areas). It is worth remarking that the morphology of ferrite grains formed during deformation is characteristically different from that of ferrite grains formed in undeformed sample (Fig. 2(b)). With increasing strain, the volume fraction of elongated ferrite grains is reduced, while on the contrary the volume fraction of equiaxed fine ferrite grains is noticeably increases.

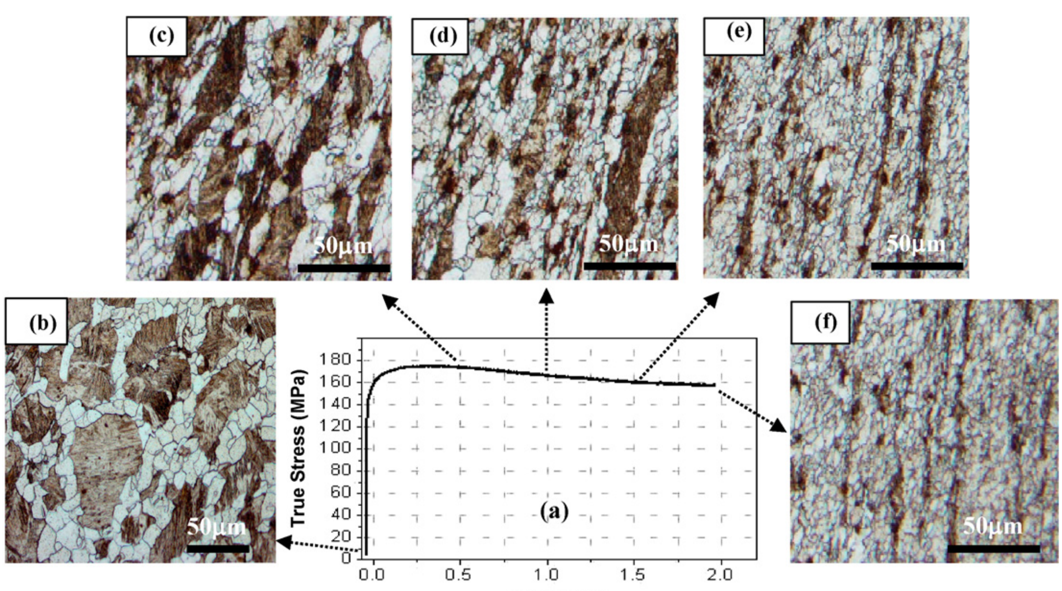

Fig. 2. Flow behavior and microstructural development of Nb–Ti steel during intercritical deformation with different strain at 750 °C and with a strain rate of 0.1 s^{-1}. (a) True stress–true strain flow curve, (b) $\varepsilon = 0.0$, (c) $\varepsilon = 0.5$, (d) $\varepsilon = 1.0$, (e) $\varepsilon = 1.5$ and (f) $\varepsilon = 2.0$.