Materials Science and Engineering A 527 (2010) 2725–2732

Contents lists available at ScienceDirect

Materials Science and Engineering A

journal homepage: www.elsevier.com/locate/msea

Effect of microstructural features on the hot ductility of 2.25Cr–1Mo steel

X.-M. Chen [a], S.-H. Song [a,*], Z.-C. Sun [a], S.-J. Liu [a], L.-Q. Weng [a], Z.-X. Yuan [b]

[a] Shenzhen Key Laboratory of Advanced Materials, Department of Materials Science and Engineering, Shenzhen Graduate School, Harbin Institute of Technology, Xili, Shenzhen 518055, China
[b] School of Materials and Metallurgy, Wuhan University of Science and Technology, Wuhan 430081, China

ARTICLE INFO

Article history:
Received 21 October 2009
Received in revised form
20 December 2009
Accepted 14 January 2010

Keywords:
Hot ductility
Steel
Dynamic recrystallization
Grain boundary segregation

ABSTRACT

Some factors contributing to the hot ductility losses of a 2.25Cr–1Mo steel were identified over the temperature range 750–950 °C, after the specimens were austenitized at 1000 °C, furnace cooled to different temperatures, and held there for sufficient periods of time, followed by tensile testing. There were two types of ferrite present in the microstructure, namely, pro-eutectoid ferrite and deformation-induced ferrite. The pro-eutectoid ferrite was only formed below Ar_3 (~825 °C), which was nucleated on the inclusions and distributed uniformly. Nevertheless, the deformation-induced ferrite was formed in a much wider temperature range. It was distributed mainly along austenite grain boundaries above Ar_3, and around the pro-eutectoid ferrite below Ar_3. The deformation-induced ferrite had a primary effect on the hot ductility, which was mainly responsible for a hot ductility trough. There was a peak in the quantity of deformation-induced ferrite between 800 and 900 °C, which was just corresponding to the hot ductility trough. The morphology of ferrite was also essential. The net-like structure of ferrite formed along austenite grain boundaries was the most deleterious to the hot ductility.

1. Introduction

Transverse cracking on the surface of continuously cast products is still a serious problem which causes ductility losses in the austenite–ferrite region for low alloy low carbon steels. Cracks can nucleate and propagate during the straightening operation which is carried out normally in the range 700–950 °C. Previous studies [1–6] have indicated that hot ductility deterioration in the austenite–ferrite region is caused by the following factors: (1) formation of a thin pro-eutectoid ferrite layer along austenite grain boundaries; (2) segregation of undesirable elements, such as S, Sb, Sn, As, P, Cu and others, at ferrite/austenite interfaces or at austenite grain boundaries; and (3) intergranular precipitation of carbides, carbonitrides, or nitrides.

Mintz et al. [1] suggested that the hot ductility trough is controlled by the austenite–ferrite transformation. It was confirmed [2] that the hot ductility trough in austenite–ferrite region depends primarily on the thickness of pro-eutectoid ferrite formed along austenite grain boundaries. Nachtrab and Chou [3] put forward that grain boundary segregation of Sn, Cu and Sb can seriously reduce the hot ductility of C–Mn steel and deformation at high temperatures can enhance the grain boundary segregation. Harada et al. [4] pointed out that the origin of surface transverse cracks is the micro-segregation, particularly of phosphorus. A study by Guo et al. [5] demonstrated that a 2.25Cr–1Mo steel containing 0.06 wt.%P has a lower hot ductility than the steel containing 0.01 wt.%P at temperatures between 750 and 1000 °C. However, as reviewed by Mintz [6], phosphorus is likely to improve the hot ductility and reduce transverse cracking during straightening if the segregation to grain boundary is controlled.

Until the present time, there have been few studies that have distinguished the effects of deformation-induced ferrite and pro-eutectoid ferrite on the hot ductility. It was the aim of the present work to investigate the effects of the morphology and magnitude of these two different ferrites on the hot ductility of a 2.25Cr–1Mo steel. Dynamic recrystallization and non-equilibrium segregation of phosphorus during high temperature deformation were also addressed.

2. Experimental procedures

The experimental steel was prepared by vacuum induction melting with an ingot of 50 kg. The chemical composition of the steel (wt.%) was 0.13C, 0.40Si, 0.53Mn, 0.028P, 0.0097S, 0.023Cu, 2.45Cr and 0.97Mo. Obviously, the steel was doped with P. The resulting ingot was hot rolled in the range 900–1000 °C into a plate 20 mm in thickness. The tensile specimens were machined from the plate with their axes perpendicular to the rolling direction. The specimens had a gauge length of 50 mm with a gauge diameter of 10 mm.

* Corresponding author. Tel.: +86 755 26033465; fax: +86 755 26033504.
E-mail address: shsonguk@yahoo.cn (S.-H. Song).

0921-5093/$ – see front matter © 2010 Elsevier B.V. All rights reserved.
doi:10.1016/j.msea.2010.01.047

In this work, two groups of specimens were prepared. One group was used for hot tensile tests, and the other was undeformed and used merely for metallographic observation after undergoing the same thermal history as the former for comparison. Hot tensile tests were performed using a high temperature tensile testing machine. Under an argon atmosphere, the specimens were austenitized at 1000 °C for 15 min, then furnace cooled to different test temperatures between 750 and 950 °C where the specimens were held for adequate time (750 °C for 60 min, 800 °C for 30 min, 812 °C for 30 min, 818 °C for 30 min, 825 °C for 30 min, 832 °C for 30 min, 840 °C for 30 min, 850 °C for 30 min, 900 °C for 15 min, and 950 °C for 15 min) to enable phosphorus grain boundary segregation to reach equilibrium, followed by tensile testing. According to McLean [7], Song et al. [8] and Erhart et al. [9], the holding times at different temperatures were long enough to acquire equilibrium grain boundary segregation. The tensile tests were carried out with a constant strain rate of 10^{-3} s^{-1} which is close to that used during straightening operation [6], and the reduction in area (RA) was employed to evaluate the hot ductility. After the specimens were fractured, they were immediately quenched by a high speed argon flow with a cooling rate higher than 30 K/s [10] to prevent further austenite–ferrite transformation during cooling. The undeformed specimens were maintained for the same time as the deformed ones at corresponding temperatures, and then quenched.

Microstructures near the fracture surfaces for the deformed and undeformed specimens were metallographically examined using optical microscopy. In addition, the fraction of ferrite in the microstructure was measured. In the measurement, 15 fields with an area of 2.7×10^5 μm^2 each were randomly selected for each condition and the mean value of data points obtained was taken as the measured result. Fractographic observations were performed using a Hitachi S-4700 field emission gun scanning electron microscope (FEG-SEM).

The austenization temperature used in the present work was 1000 °C, which is lower than the temperature usually used for the research of hot ductility. This is because the highest operation temperature of the hot tensile testing machine used is just above 1000 °C. The austenization temperature is usually between 1320 and 1350 °C for the investigation into hot ductility in order to explore transverse cracking present in continuous casting. Nevertheless, a study by Yasumoto et al. [11] suggests that on raising the solution treatment temperature, the width and depth of the ductility trough are greatly increased for low carbon steels, but the temperature corresponding to the minimum ductility does not change apparently. Transverse cracking on the surface of slab is most likely produced around the minimum ductility temperature. Accordingly, the present work, which focuses on the effect of microstructural features on the hot ductility, is still of implication in exploring the susceptibility to transverse cracking in continuous casting although a lower austenization temperature was employed.

3. Results and discussion

The hot ductility values at different temperatures between 750 and 950 °C are presented in Fig. 1. Since there were no data points between 750 and 800 °C and between 850 and 900 °C, a dashed line was used to link the data points from 750 to 800 °C and from 850 to 900 °C. As seen, there is an evident ductility trough in the range 750–900 °C and the minimum ductility with an RA value of 63.3% occurs at 825 °C. Fig. 2 shows typical microstructures after hot tensile testing at different temperatures. Obviously, there is no ferrite formed at 900 °C while there is some at 850 °C (see Fig. 2c and d), indicating that there is a temperature between 850 and 900 °C above which no austenite–ferrite transformation has taken place both during isothermal holding prior to tensile testing and in the process of the testing. From that temperature, the magnitude of ferrite in the microstructure increases gradually with decreasing test temperature and the corresponding hot ductility drops. At 825 °C, almost perfect ferrite networks along austenite grain boundaries are formed by ferrite films, which corresponds to the minimum RA value (see Fig. 1). With further decreasing test temperature, the magnitude of ferrite increases continuously with the network morphology of ferrite dying away gradually and the corresponding hot ductility starts to recover. For example, at 818 °C there are few ferrite precipitates with a film-like morphology, and most of them are isolatedly distributed (see Fig. 2a), which would be beneficial to the ductility. Based on the above, it is claimed that the net-like structure of ferrite along austenite grain boundaries plays an important role in the reduction of ductility. Fig. 3 represents the microstructures before deformation at 825 and 818 °C, respectively. Clearly, there is almost no ferrite formed at 825 °C while there is some at 818 °C. Hence, without deformation the highest temperature at which the ferrite can be formed is around 825 °C. In addition, the morphology of ferrite is much different from that after deformation. There are few ferrite precipitates formed preferentially along grain boundaries.

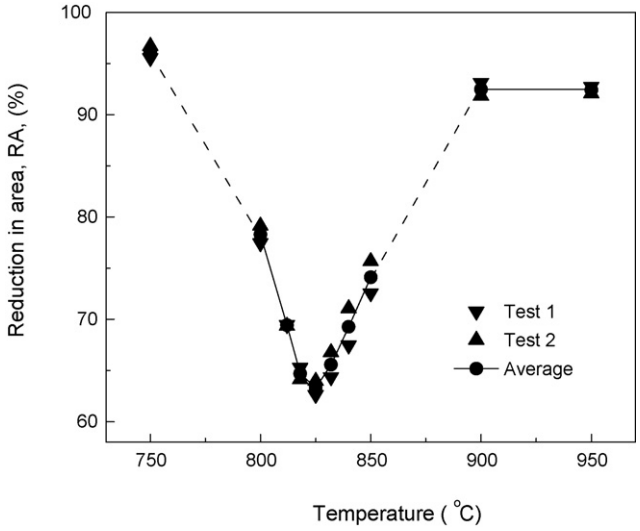

Fig. 1. Hot ductility as a function of temperature for the experimental steel.

Three types of ferrite, namely, pro-eutectoid ferrite, deformation-induced ferrite and Widmanstätten ferrite, may precipitate during the experiment. According to Du et al. [12], the morphology of Widmanstätten ferrite is mainly Widmanstätten and allotriomorphs, which is not present in Figs. 2 and 3. Thus the ferrite existing in the present case can be divided into pro-eutectoid ferrite produced during isothermal holding before tensile testing and deformation-induced ferrite produced during tensile testing. Previous work [12] demonstrated that not any kind of ferrite can be formed in the microstructure if deformation occurs at temperatures higher than Ae_3, and also at temperatures higher than Ar_3 there can be no pro-eutectoid ferrite formed. Accordingly, it is seen from Figs. 2 and 3 that Ae_3 is between 850 and 900 °C and Ar_3 is around 825 °C. Using the Andrews's empirical formula [13], we estimate Ae_3 for the experimental 2.25Cr–1Mo steel as approximately 885 °C, which is in good agreement with the present observation. Therefore, the ductility shelf on the right side of the trough, as shown in Fig. 1, could start from this temperature. As described by Mintz [6], the hot ductility trough may cover the temperature range of about 30 °C below Ar_3 to Ae_3, which well supports the present results (see Fig. 1).

The magnitude of ferrite in the microstructure is shown in Fig. 4 as a function of temperature before and after deformation.