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Effects of interphase TiC precipitates on tensile properties and dislocation structures in a dual phase steel

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ABSTRACT

In this work, titanium-free and titanium carbide-strengthened dual-phase steels were studied. While both had the same base composition of Fe–0.1C–1.5Mn–0.5Si–0.6Cr (wt%), the latter included a 0.1 wt% Ti addition for interphase precipitation. Utilizing the Gleeble thermal simulations, both two dual-phase steels were produced with the same ferrite grain size and the same ferrite/martensite volume fractions of 0.7 and 0.3, respectively. Their corresponding microstructures (including deformed structures) and mechanical properties were examined. The results indicate that the titanium carbide-strengthened dual-phase steel possesses an excellent combination of the strengthened ferrite and the weakened martensite, which brings about increased yield and tensile strengths without sacrificing total elongation.

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1. Introduction

Dual-phase steels (hard martensite embedded in soft ferrite matrix) have been widely applied in the automobile industry since 1970's due to their suitable mechanical properties and relatively easy fabrications [1–4]. The general mechanical properties of dual-phase steels are mainly associated with the parameters such as ferrite grain size [5–7], ferrite/martensite volume fractions [8,9], and phase distribution (including martensite morphology) [5,8–10]. Yield strength, tensile strength, and elongation are mainly dependent on ferrite/martensite volume fractions and ferrite grain sizes [5–9]; elongation is also influenced significantly by martensite distribution [5,9,10]. Processing to produce the dual-phase steels is usually done by intercritical annealing from room temperature to two-phase ($\alpha + \gamma$) region [1,2,4,8]. Dual-phase steels have also been produced from the as-hot-rolled condition [3]. The initial microstructure prior to intercritical annealing may differ depending on whether the steel has been normalized (ferrite + pearlite) [1,2,4] or quenched (martensite) from austenite [5,8,10]. Martensite as the initial structure can provide abundant heterogeneous nucleation sites including both martensite lath boundaries and prior austenite grain boundaries, which cause a strong effect of grain refinement to provide a good combination of strength and elongation [5,10]. However, the quenching of steel plates from austenitization temperatures usually causes distortion, cracking and residual stress, which could give rise to further problems in subsequent processes. On the other hand, several approaches [5,6,9,11–13] have been attempted for producing fine-grained and ultrafine grained dual-phase steels by intercritical annealing following cold/warm deformation of ferrite-pearlite [6,9,11–13] or cold deformation of martensite [5]. The fine or ultrafine grained dual-phase steels are attractive because of their ultrahigh strength [6,11]. However, the steels possess a relatively low uniform elongation. It is well known that fine-grained metals undergo a distinct reduction in ductility as the grain is reduced in the nanometer range. On the other hand, heavy cold-rolling unavoidably increases the risk of cracking. For further strengthening, microalloying additions of carbide-forming elements of Ti, V and Nb [11–13] have been utilized with the process of cold-rolling (using the initial structure of ferrite-pearlite) followed by intercritical annealing. The related studies reported that ultrafine-grained ferrite (with a size about 1–2 μm) with nanometer-sized carbides [12,13] in dual-phase steels can be obtained. However, the studies did not reveal the formation mechanism of these tiny carbides in ferrite nor the detailed information for the density of carbides, which could be used to evaluate the corresponding contribution to the strength. It is clear that at the intercritical annealing temperature, new ferrite recrystallized from the deformed ferrite matrix, which was lean of carbon, so it was not expected to precipitate copious carbide particles in the newly formed ferrite grains.

Although the “as-hot-rolled” process has difficulty producing the fine ferrite grains, it possesses the advantage of reduced energy

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consumption and increased process efficiency [14]. Recently, some research studies have dealt with the influence of V, Nb and Ti precipitates on mechanical properties of the "as-hot-rolled" dual-phase steels [15–17]. It has been confirmed that the strong carbide-forming elements played a crucial role in strengthening ferrite in dual-phase steels. One study [17] investigated the effects of cooling schedules for the same hot-deformation austenite on the evolution of the transformation products: ferrite, precipitated carbides, and second phases (which involved with martensite and bainite), and revealed that yield ratio (YS/UTS) can be improved through interphase precipitation of carbides in the "as-hot-rolled" dual-phase steels, as compared to that of the traditional dual-phase steel.

After the successful development of 780 MPa-grade Nanohiten steel in JFE Steel [18], several research works dealing with interphase precipitation of (Ti, Mo)C have been conducted in ferritic steels [19–22]. In a recent study [23] through a simple heat treatment, the concept to strengthen ferrite via interphase-precipitated nanometer-sized VC in dual-phase steels was performed. However, in that study it did not control the ferrite grain size nor the volume fraction of the ferrite/martensite. As a result, the effect of interphase-precipitated carbides on mechanical properties in dual-phase steels has not yet been clearly clarified. In the previous study by Freeman and Honeycombe [24], it has been clearly indicated that the addition of Ti can also make a significant strengthening contribution to ferritic steels by means of interphase-precipitated nanometer-sized carbide particles under the lower intercritical temperatures. From the economic point of view, the development of nano-precipitated TiC-strengthened dual-phase steels appears to be attractive. In the present study, two dual-phase steels with and without interphase-precipitated TiC carbides were produced to have the same ferrite grain size and the same volume fraction of ferrite/martensite. Thereby, the mechanical properties of the steels were compared to identify the role of interphase-precipitated TiC carbides. Furthermore, scanning transmission electron microscopy (STEM) was employed to elucidate the development of the deformed structures in these steels during tensile test.

2. Experimental Procedure

In this study, two experimental steels had the same base composition of Fe–0.1C–1.5Mn–0.5Si–0.6Cr (wt%); one had no addition of Ti (labelled Ti-free steel) and the other one contained 0.1 wt% Ti (labelled Ti-bearing steel). These two alloy steels were casted and hot-rolled to a thickness of 4.5 mm. The strips were cut into rectangular-shaped specimens of 10 cm in length and 2 cm in width and then capsuled in evacuated quartz followed by homogenization at 1200 °C for 3 days to eliminate segregation. After removing decarburization layer, the specimens were machined to 2 mm thick. A series of trial heat treatments were conducted on the Gleeble 1500 equipment in order to control dual-phase morphology. Finally, two heat treatments were selected: for Ti-free steel, specimens were austenitized at 1000 °C for 30 s and cooled at a rate of 20 °C/s to the holding temperature of 650 °C for 2 min; for Ti-bearing steel, specimens were austenitized at 1050 °C for 30 s and also cooled at a rate of 20 °C/s to the holding temperature of 650 °C for 2 min. Both specimens were then quenched to room temperature for transforming the retained austenite to martensite. In this study, the difference in austenitization conditions between Ti-free and Ti-bearing steels brought about the same grain size of ferrite and volume fraction of ferrite/martensite in dual-phase structures (as shown later). Although austenitization at 1050 °C employed for Ti-bearing steel is not high enough to dissolve all TiC carbides, the interphase precipitation of TiC carbides in Ti-bearing steel can still be achieved (the related results will be shown later). After heat treatments, tensile tests were conducted according to ASTM E8 standard [25]. A strain rate of $1.0 \times 10^{-3}\ s^{-1}$ was employed on an MTS 810 testing machine equipped with an extensometer at room temperature. Optical metallography (OM) samples were prepared from Gleeble specimens, and ferrite grain size in the corresponding samples was measured by the intercept method. Volume percent of ferrite was determined by point counting method via the Photoshop software [26]. For each sample, the Vickers hardness measurement of ferrite and martensite was performed on 50 individual ferrite grains and 50 martensite islands respectively, with a 10 g load. Transmission electron microscopy (TEM) samples were produced by cutting slices from the Gleeble specimens and thinned mechanically to 60 μm on SiC papers. Afterwards, the specimens were twin-jet electropolished using a solution of 5% perchloric acid, 15% glycerol and 80% ethanol, and the potential used was 37 V. For field-emission-gun transmission electron microscopy (FEG TEM), Tecnai F30 with high-angle angular dark-field (HAADF) detector was used at 300 kV.

3. Results and Discussion

Nearly the same prior austenite grain sizes (about 16 μm as shown in Fig. 1a and b) for Ti-free and Ti-bearing steels generated the corresponding dual-phase structures (as shown in Fig. 1c and d): where white-etched phase was ferrite and dark-etched phase was martensite. Both types of dual-phase steels had approximately the same grain size of ferrite, which was measured to be 14 and 16 μm in Ti-free and Ti-bearing steels, respectively. Furthermore, holding at 650 °C for the same period of time yielded nearly the same volume percent of martensite, which was measured to be 30% and 29%, respectively. Through carefully controlling grain sizes and volume percentages, the difference of mechanical properties between the two steels can be ascribed to the addition of Ti as it is the only variable.

The formation of interphase precipitation in ferrite grains of dual-phase Ti-bearing steel was revealed by TEM as shown in Fig. 2. Sheet spacing, inter-carbide spacing, and carbide size of interphase precipitation were measured to be 18.2 ± 0.5 nm, 42.6 ± 8 nm and 4.7 ± 0.3 nm, respectively, by the method proposed in [19]. After interphase precipitation was introduced into ferrite phase, the hardness of ferrite was enormously strengthened from 198 ± 7 Hv (Ti-free steel) to 257 ± 8 Hv (Ti-bearing steel), while at the same time the hardness of martensite decreased dramatically from 497 ± 16 Hv (Ti-free steel) to 364 ± 20 Hv (Ti-bearing steel). The increase in ferrite hardness is expected due to interphase precipitation hardening; however, the significant decrease in martensite hardness is beyond expectation. Since martensite hardness depends primarily on carbon content and substructure [27], it is appropriate to conclude that the decrease in martensite hardness resulted from the carbon consumption of interphase precipitation during $\gamma \rightarrow \alpha$ transformation, causing less carbon content in martensite in Ti-bearing steel, as compared to that in Ti-free steel. However, since precipitation behavior varies with the progression of $\gamma \rightarrow \alpha$ transformation, carbon distribution in austenite would be inhomogeneous. Therefore, morphology of martensite in these two different dual-phase steels should be carefully observed in TEM. Results from TEM observation are shown in Fig. 3, and they reveal that in the area adjacent to the interface of ferrite/martensite, twinned martensite is usually observed in Ti-free steel, while lath martensite dominates in Ti-bearing steel. The morphology change corresponds well with hardness variation.

The stress–strain curves for both steels are presented in Fig. 4a. The related yield strength (YS), ultimate tensile strength (UTS), and elongation, along with microstructural information are listed in Table 1. The benefit of interphase precipitation increases YS from 377 MPa (Ti-free steel) to 539 MPa (Ti-bearing steel), by 162 MPa as shown in the stress–strain curve (Fig. 4a), where the corresponding YS is determined by 0.2% strain offset; it also promotes UTS from 622 MPa (Ti-free steel) to 786 MPa (Ti-bearing steel), by 164 MPa. The yield ratio of Ti-bearing steel is 0.69, higher than the 0.61 of Ti-free steel, owing to the increase in the ferrite strength and simultaneously the decrease in the difference between the martensite strength and the ferrite strength. The most intriguing part is regarding the elongation, where Ti-bearing steel shows almost the same uniform and total elongation as that of Ti-free steel.

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