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Strengthening effect of nanoscale precipitation and transformation induced plasticity in a hot rolled copper-containing ferrite-based lightweight steel

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ABSTRACT

Effects of copper addition on microstructure and tensile property of a hot rolled ferrite-based lightweight steel are investigated. Copper addition delays bainitic transformation, resulting in higher volume fraction of retained austenite with lower stability in the steel. Addition of 1.0 wt% copper increases the yield strength and tensile strength of the steel by 17% and 19%, respectively, and maintains high ductility. The yield strength enhancement is predominantly attributed to nanoscale copper precipitation. The higher work hardening rate due to precipitation hardening and the transformation-induced-plasticity mechanism leads to higher tensile strength (805 MPa) and remarkable elongation-to-failure (~30%) for the copper-containing steel.

Aluminum-enriched transformation-induced-plasticity (TRIP) steels, also called δ-TRIP steels or ferrite-based lightweight steels, could possess lower densities, excellent combinations of strength and ductility, and appreciable producibility [1–10], and therefore have great potential for automotive applications. As a characteristic constituent phase, δ-ferrite forms during solidification and persists during subsequent thermo-mechanical processing and heat treatment due to the addition of a large amount of aluminum [2,4,6]. Moreover, it is hardly refined by the austenite-to-ferrite transformation and recrystallization. Correspondingly, the δ-ferrite is oftentimes pretty coarse, and softer than α-ferrite and fine austenite [11]. Strengths of ferrite-based lightweight steels are therefore relatively low owing to the presence of a large fraction of the δ-ferrite phase in the microstructure.

It is well known that precipitation strengthening is a useful approach to improving the strengths of steels. Microalloying with strong carbide-forming elements such as Ti, Nb, Mo and V could remarkably increase yield strengths and/or tensile strengths of ferritic steels [12–15] and TRIP-assisted multiphase steels [16,17] through precipitation strengthening as well as other microstructural modifications (such as grain refinement, bainitic transformation, etc.), however the ductility of the steels may be deteriorated. It has been another important strategy to strengthen the above types of steels by utilizing copper (Cu)-rich precipitates [18–21]. Because of limited solubility in Fe, Cu solutes could undergo continuous nanoscale precipitation during a proper isothermal hold or aging process, with the crystal structure of precipitates transitioned from coherent body-centered cubic (BCC) through twinned 9R to face-centered cubic (FCC) [22]. The effects of Cu addition on microstructure as well as strength enhancement without noticeable ductility deterioration in TRIP-assisted steels had been studied [23], however the microstructural features including the Cu precipitation, bainitic transformation, and retained austenite (RA) stability need to be further elaborately clarified to interpret the correlation between microstructure and mechanical behavior. This is one of the objectives of the present investigation which is to be performed on hot rolled ferrite-based lightweight steels containing δ-ferrite. The other objective is to confirm the feasibility of utilizing Cu precipitates to strengthen ferrite-based lightweight steels and to attain an excellent combination of strength and ductility in the meantime.

Two ferrite-based lightweight steels were fabricated for the investigation. The chemical compositions of the steels are Fe-0.35C-1.1Mn-4.1Al-0.4Si-(0\1.0)Cu (in wt%) with traces of nitrogen, sulfur and phosphor. The two steels were referred to as "Cu-free" and "1.0Cu", respectively hereafter for brevity. Steel ingots were homogenized at 1200 °C for 2.5 h, and rough rolled from 180 mm to 40 mm, followed by air cooling. The slabs were then reheated to 1200 °C for 60 min and hot-rolled into plates of 3.6 mm with a finishing temperature of approximately 880 °C. The rolled plates were immediately water quenched to 450 °C, and then held at this temperature in a furnace for 60 min, and finally furnace cooled to room temperature in order to simulate a coiling procedure. After cooling, the hot-rolled plates were pickled in a dilute

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HCl solution to remove the surface oxide film for microstructural and mechanical analyses.

Tensile test samples were cut along the rolling direction (RD) with a gauge length and width of 50 mm and 12.5 mm, respectively (according to the ASTM E8-04 standard). Tests were performed on a computer-controlled 8801 Instron machine with a constant crosshead speed of 3.0 mm/min (corresponding to an initial strain rate of 8.8×10^{-4}/s) at room temperature. At least three samples were tested to obtain average tensile property for each steel.

Microstructures on the plane of the plate RD and normal direction (ND) were observed using optical microscope (OM) and field-emission scanning electron microscope (SEM; FEI Versa 3D). Metallographic specimens were mechanically polished and etched in a 4% nital solution. Electron back-scattered diffraction (EBSD) analysis was carried out on the RD-ND plane of the hot-rolled steel plates as well, using Hitachi Su-70 Schottky field emission SEM at an acceleration voltage of 20 kV. EBSD data were post-processed by the HKL CHANNEL 5 flamenco software. Fine microstructures were further characterized by transmission electron microscope (TEM; JEOL ARM-200F) equipped with a Cs corrector, operating at an acceleration voltage of 200 kV. TEM foils were prepared by twin-jet electropolishing punched disk specimens (3 mm in diameter and 80 μm in thickness) in a 4 vol% perchloric acid ethanol solution at −30 °C and with a voltage of about 50 V. Phase identification was made by an X-ray diffractometer (XRD, Bruker-AXS D8 Discover) with Cu Kα radiation operated at 40 kV and 40 mA at room temperature. Volume fractions of retained austenite (RA) were measured from the Rietveld refinement of XRD profiles to take account of the texture effect. The carbon concentration of the RA was determined using the following equation [24], assuming that the partitioning effect of substitutional elements can be neglected,

$$a_\gamma = 3.578 + 0.033X_C + 0.00095X_{Mn} + 0.0056X_{Al} + 0.00157X_{Si} + 0.0015X_{Cu} \quad (1)$$

where a_γ is the austenite lattice parameter in Å, and X_C, X_{Mn}, X_{Al}, X_{Si} and X_{Cu} are the concentrations of C, Mn, Al, Si and Cu in steels, respectively, in wt%. The austenite lattice parameter (a_γ) was determined from the d-spacing of $(220)_\gamma$ positions. To evaluate the effect of tensile strain on the variation of the volume fraction of RA, XRD measurements were made in the uniformly-deformed regions of fractured specimens and the necked regions which are 3 mm away from the fracture surfaces. Tensile true strains, ε, were calculated as follows:

$$\varepsilon = \ln\left(\frac{w_0 t_0}{wt}\right) \quad (2)$$

where w_0 and t_0 are the width and thickness of tensile specimens before tensile test, respectively; w and t are the width and thickness of tensile specimens in the uniformly-deformed regions or necked regions after tensile test, respectively. It should be noted that the necked region is in the three-dimensional stress state (rather than the uniaxial tensile stress state as in the uniformly-deformed region), and the strain value calculated by Eq. (2) roughly represents tensile limit strain.

Fig. 1 shows typical OM and SEM micrographs, and EBSD phase maps of the Cu-free and 1.0Cu steels, respectively. The hot-rolled microstructures are basically banded structures in which lighter bands are aggregates of recovered/recrystallized δ-ferrite grains, and darker bands are transformation products of prior austenite during the isothermal hold and furnace cooling [25]. Darker bands show different variants and morphologies, depending on whether Cu is added. In the Cu-free steel, the darker bands mainly include carbide-free bainite which is composed of lath-shaped bainitic ferrite and austenite, in addition to polygonal α-ferrite grains and some blocky austenite particles (Fig. 1(a, b)). Martensite is hardly observed in the microstructure, as indicated by the EBSD phase map (Fig. 1(c)). It is thus implied that the RA after bainitic transformation reaction is quite thermally stable. In the 1.0Cu steel, the darker bands are mainly composed of blocky austenite/martensite packets (Fig. 1(d, e)). Some of prior austenite transform into feathery bainitic structure during the simulated coiling process, which is illustrated in black regions in Fig. 1(d) and by the inset magnified view in Fig. 1(e), and will be further confirmed by TEM analysis. The formation of martensite (M/A islands) as verified by the EBSD phase map (Fig. 1(f)) is ascribed to the transformation of large and relatively thermally-unstable austenite when the temperature drops below *Ms.* (martensite start temperature) during furnace cooling. According to EBSD statistical measurements from at least 100 grains for each phase, the grain sizes for δ-ferrite, α-ferrite and RA are

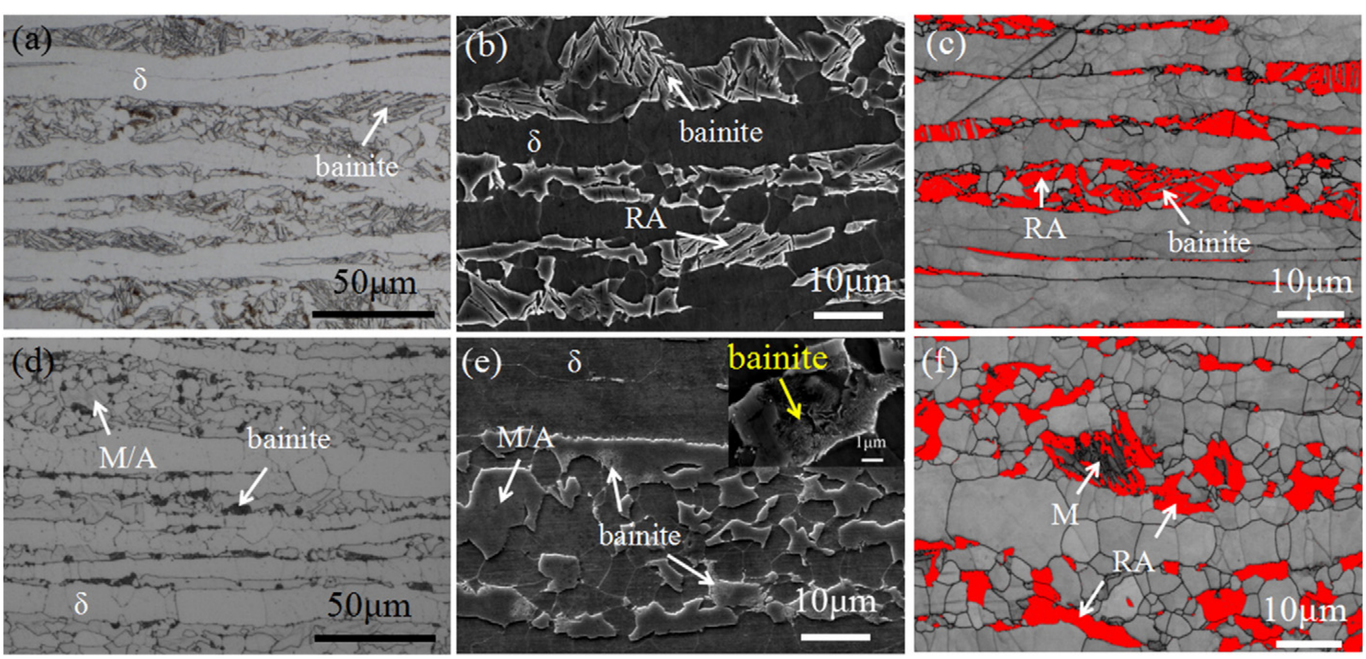

Fig. 1. OM and SEM micrographs and EBSD phase maps of the hot rolled lightweight steels: (a–c) Cu-free steel; (d–e) 1.0Cu steel. SEM micrographs and EBSD maps show transformation products of prior austenite in darker bands in more detail. The step size for EBSD mapping is 70 nm. The area in red color depicts RA. (For interpretation of the references to color in this figure legend, the reader is referred to the web version of this article.)