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Critical role of strain partitioning and deformation twinning on cracking phenomenon occurring during cold rolling of two duplex medium manganese steels

Binhan Sun [a,*], Fateh Fazeli [b], Colin Scott [b], Xiaojun Yan [c], Zhiwei Liu [c], Xiaoyu Qin [c], Stephen Yue [a]

[a] Department of Materials Engineering, McGill University, 3610 University Street Montreal, QC H3A 0C5, Canada
[b] CanmetMATERIALS, Natural Resources Canada, 183 Longwood Road South, Hamilton, ON L8P 0A5, Canada
[c] School of Energy and Power Engineering, Beihang University, 37 XueYuan Street, Haidian District, Beijing 100191, China

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ABSTRACT

Two ferrite-austenite duplex medium manganese steels with very similar phase fractions and grain sizes were subjected to cold rolling. The 10Mn steel presented a much better cold rollability compared with the 7Mn steel, which is believed to stem from two critical factors. Firstly, the larger strain partitioning in the 10Mn steel resulted in a less deformation accommodated by the brittle δ-ferrite, which delayed or even inhibited micro-crack formation inside the ferrite phase. Secondly, the deformation twinning activated in the austenite of the 10Mn steel during rolling effectively enhanced the ductility of the austenite-martensite (γ-α′) mixed phase.
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Medium Mn steels alloyed with 3– 10 wt.% Mn feature a large fraction of retained austenite (RA) [1,2] which can be tailored to provide various TRIP (transformation-induced plasticity) and TWIP (twinning-induced plasticity) effects [3,4]. This can result in an excellent mechanical performance, making the steel a strong candidate material for the third-generation of advanced high strength steels (AHSS). The austenite is retained by the stabilization effect of the relatively high initial Mn addition and extra Mn partitioning due to intercritical treatment [2]. However, this large amount of RA would easily transform to the hard and brittle martensite phase during deformation if its mechanical stability is not well controlled, which essentially deteriorates the ductility and the cold rollability of the steel. Currently, while great progresses have been achieved on the optimization of alloying, heat treatments and mechanical properties, the cold rolling performance of this type of steel and the corresponding microstructural evolution have rarely been reported.

A complicated deformation behavior is expected for medium Mn steels with multi-phases when they are subjected to cold rolling for two main reasons. First, the distribution of plastic strain will not be uniform due to the differences in mechanical behavior of different constituting phases. It has been reported that strain partitioning between austenite and ferrite in medium Mn steels largely influences the overall flow behavior, e.g. flow stress and strain-hardening behavior [5,6]. Second, the deformation modes activated in austenite would also affect the cold rollability. It is well established that both TRIP and TWIP effect can improve the ductility of steels, by providing large enhancements of the work-hardening rate. The objective of this study is better understanding of the microstructural response of two variants of ferrite-austenite duplex medium Mn steels during cold rolling. This allows to assess the production of medium Mn steel sheets by conventional processing lines. The dominating factors on the cold rollability was revealed by comparing the two steels produced with similar initial phase fractions and grain sizes but very different strain distribution and austenite deformation modes. The results generated in this work might also be applied to some low-density steels, which share similar microstructures [7].

The chemical compositions of the two hot rolled medium Mn steels are Fe-0.2C-7/10Mn-3Al-3Si (in wt.%); they are referred to as 7Mn and 10Mn steel. Casting and hot rolling schedules for the two steels have been presented elsewhere [2]. Small strip specimens (6 cm $*$ 1 cm) with a thickness of ~2 mm for cold rolling were machined from the hot rolled sheets. Before cold rolling, samples were intercritically annealed at 1000 °C for 10 min, followed by air cooling, in order to produce ferrite plus austenite duplex microstructures with similar phase fractions and grain sizes for the two steels. Cold rolling was conducted in a two-high laboratory rolling mill with the roll diameter of 10.2 cm and a rolling speed of around 35 rpm. The rolling direction was the same as for the hot rolling; an average of ~5% reduction in thickness was applied for each pass. After ~20% reduction, small edge cracking in the 7Mn steel occurred, and in the final stage, i.e. ~50% reduction, the edge areas were seriously cracked. However, the 10Mn steel

* Corresponding author.
E-mail address: binhan.sun@mail.mcgill.ca (B. Sun).

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showed a much better cold rollability with no edge cracks or any other defects formed up to ~50% reduction. Microstructure observations were performed with electron backscattered diffraction (EBSD), scanning electron microscope (SEM, Hitachi SU3500) and transmission electron microscope (TEM, FEI Tecnai G2 F20, operated at 200 kV). The austenite fraction of the samples was determined by X-ray diffraction (XRD) using Co K_α radiation. The hardness of each phase in the two samples was measured inside the grains using a Vickers micro-hardness tester. Compositional analysis for specific phases was performed in SEM using Energy-Dispersive X-ray Spectrometer (EDX). The strain distribution between austenite and ferrite in the two alloys during deformation was measured by an in-situ SEM tensile test. A MTI SEMtester 1000 micro load frame was used to load the samples inside the electron microscope vacuum chamber (ZEISS EV018), with a constant loading rate of 0.1 mm/min. The surface of the tensile samples (with the gage length of 12.5 mm) were polished and etched before testing. Concurrent loading and imaging of the sample for the same area were performed during each test. The microscopic strain (e) of each phase was calculated by the length change of more than five grains along the tensile direction divided by their initial length.

Fig. 1 shows the ferrite-austenite duplex microstructures of the two steels after prior annealing, which consists of mostly δ-ferrite and austenite (γ) elongated bands resulting from hot deformation in the two-phase range [10]. A very small amount of intercritically annealed ferrite (IA-α) is present inside the austenite grains, which is transformed from prior bainite and pearlite [2]. The well-known Kurdjumov-Sachs (K-S) orientation relationship (OR) exists between the IA-α and γ, with the phase boundaries highlighted as the red color in Fig. 1 (a) and (b). Detailed information of each phase in the annealed samples is listed in Table 1. The two steels exhibit very similar retained austenite fractions and grain sizes despite the different Mn additions; the fraction of δ-ferrite is also expected to be similar due to the very small amount of IA-α. This similar phase fraction is most likely due to the slow austenite formation kinetics at later stages of intercritical annealing which is controlled by the sluggish Mn diffusion [2]. Solute partitioning is observed after intercritical annealing, resulting in a higher Mn enrichment in austenite and a higher Al and Si concentration in δ-ferrite, as shown in Table 1. The austenite is softer than the δ-ferrite in both steels, although some amount of stress/strain-induced martensite might form during the indentation which essentially increases the austenite hardness value [11]. Fig. 1 (c) plots the fractions of retained austenite and strain-induced α′-martensite in the two steels cold rolled to different stages (measured by XRD); no ε-martensite was observed after deformation, either from the microstructure or the XRD patterns. The amounts of transformed α′-martensite with overall cold rolling reduction are very similar for the two steels. In the 10Mn steel, deformation twins were also formed inside austenite grains during cold rolling (Fig. 1 (d)), showing the TWIP effect; however, no such features were observed in any samples of the 7Mn steel.

Fig. 2 shows the SEM and EBSD images of the 7Mn steel samples cold rolled to ~20% reduction, which was observed as the starting stage of the edge cracking. Several micro-cracks were observed in the vicinity of the edge cracks; they are nucleated at the interface of strain-induced α′-martensite and IA-α, or inside the α′-γ mixture, as marked by an elliptic and a rectangular frames in Figs. 2 (a) and (b), respectively. The ferrite/martensite interface decohesion and martensite cracking were facilitated by the local stress concentration and low toughness of fresh martensite; they have been widely observed and well explained in dual-phase (DP) steels [12,13]. However, the cracking also initiates inside the coarse-grained δ-ferrite (Fig. 2 (c) and (d)), which is quite unexpected because the ferrite phase is normally considered as a soft/ductile phase at room temperature, and used as the soft matrix in plain carbon steels and most first generation AHSS such as DP and TRIP steels. The embrittlement of δ-ferrite could be derived from two possible reasons: firstly high enrichment of Si in ferrite; and secondly due to large grain size (110– 120 μm). It is established that Si higher than ~0.5 wt.% could worsen the ductility and fracture toughness of ferrite [14,15], due to the high degree of solid solution strengthening, although the overall tensile elongation of the steel might be enhanced depending on the microstructural change by Si addition [16]. This deteriorating effect becomes rather strong when the Si level is above certain value (3.5– 4 wt.%) [17], mainly because of the occurrence of some

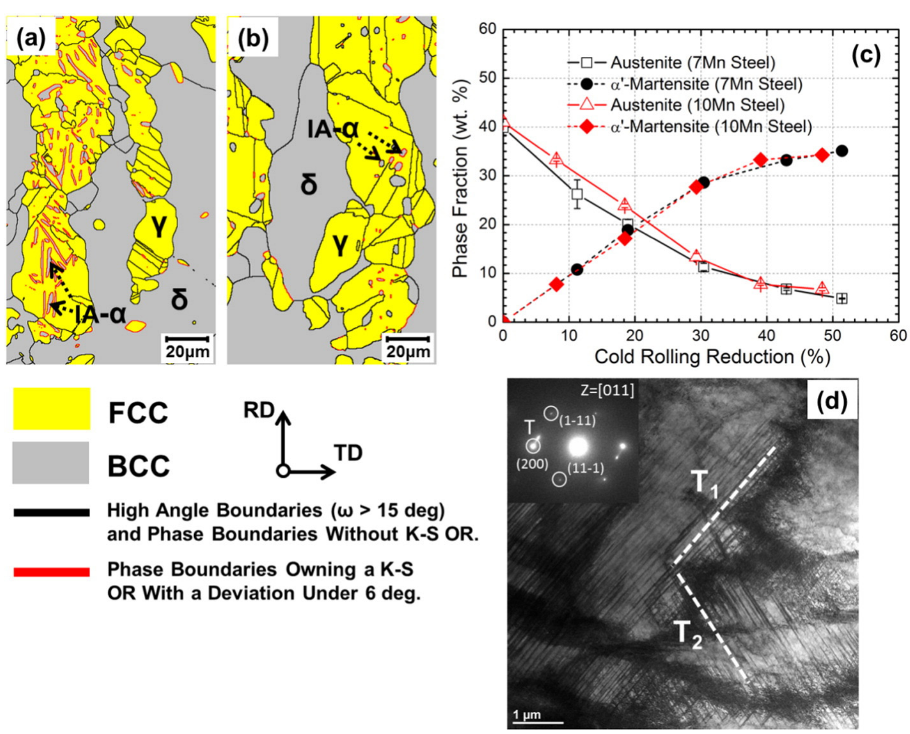

Fig. 1. EBSD phase mapping of (a) the 7Mn steel and (b) the 10Mn steel after prior heat treatment at 1000 °C for 10 min, followed by air cooling; (c) fractions of strain-induced α′-martensite and retained austenite as a function of cold rolling thickness reduction; (d) deformation twins in the 10Mn steel cold rolled to 10% reduction.

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