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Factors influencing the tensile behavior of a Fe–28Mn–9Al–0.8C steel

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ABSTRACT

High Mn–Al–C steels were recently reappraised as the promising material for advanced auto body system by their excellent combination of strength and ductility over 60,000 MPa% associated with microband-induced plasticity. This study was intended to examine the factors affecting the tensile properties of high Mn–Al–C steels. For this purpose, tensile tests and microstructural analyses were performed on a Fe–28Mn–9Al–0.8C steel at various conditions. As usual, strength increased and elongation decreased with decreasing the grain size. The strain hardening rate of the fine-grained steel remained unchanged to the medium strain level but that of the coarse-grained steel continuously increased to the high strain level, resulting in exceptional ductility. Fully austenitic steel obtained by high temperature solution treatment exhibited the continuous increase of strain hardening rate while that of the steel solution-treated at low temperature decreased with increasing strain due to the bimodal grain size distribution and the presence of submicron ferrite. In the strain rates of $2 \times 10^{-4}\,s^{-1}$ to $10^{-1}\,s^{-1}$, strength was relatively insensitive to the strain rate but elongation decreased with increasing the strain rate, indicating that plastic deformation was mainly achieved by thermal activation in this strain rate range. In the temperature range of 25–450 °C, both strength and elongation decreased with increasing temperature except at 300 °C where serrated flow occurred with extended ductility. The flow characteristics at 300 °C were rationalized in terms of dynamic strain aging associated with carbon exhaustion.

1. Introduction

Steels utilizing the austenitic phase have received great attention as the auto body material. The transformation-induced plasticity (TRIP) aided steel and twin-induced plasticity (TWIP) aided steel are the two representatives [1]. In the former, a small amount of retained austenite in ferrite and/or bainite matrix transforms into martensite during plastic deformation, resulting in enhancement of strength and ductility. In the latter, mechanical twins formed in austenite matrix with the high Mn content during deformation act as the effective obstacles for dislocation movement leading to high strain hardenability. The occurrence of TRIP and TWIP primarily depends on the stacking fault energy of austenite directly tied to its stability. It is known that TRIP and TWIP prefer to take place when the stacking fault energy of austenite is <20 mJ/m^2 and 20–40 mJ/m^2, respectively [2–4].

In addition to TRIP and TWIP steels, a class of age-hardenable austenite-based steels containing the high content of Mn, Al and C was recently reappraised as the promising auto body material [5]. This steel class was first developed in late 1950s for the use of cryogenic or highly corrosive environments by replacing the high Ni–Cr steel [6]. The mechanical properties of high Mn–Al–C steels in both solid solution and age hardened states are reported to be comparable to those of TWIP steels [5]. Besides, the high Al content makes this steel class remarkably advantageous to weight saving. The addition of Al into high Mn austenitic steels considerably increases the austenite stability and in turn stacking fault energy, leading to suppression of either TRIP or TWIP [7–9]. Instead, plastic deformation of high Mn–Al–C steels is to be achieved by some forms of dislocation gliding [4].

In the past decade, TRIP and TWIP steels have been comprehensively studied for their commercial application. As a result, some TRIP steels have already been commercialized [10] and hot coils of a TWIP steel was successfully fabricated in the commercial production line of POSCO Kwangyang Mill very recently [11]. By contrast, in spite of several previous investigations [12–15], the research regarding commercialization of high Mn–Al–C steels is still far behind compared to TRIP and TWIP steels. Considering its potential as the advanced auto body material, it is worth examining the factors influencing the mechanical properties of the high Mn–Al–C steel class. The present work was undertaken to investigate the effect of the grain size, constituent phase, strain rate and testing temperature on the tensile properties of the high Mn–Al–C steel class.

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2. Experimental procedure

25 kg ingots of Fe–28Mn–9Al–0.8C (wt.%) were prepared by induction melting in an argon atmosphere. Ingots were homogenized at 1200 °C for 2 h and then hot-rolled. After annealing at 1000 °C for 1 h, the hot-rolled plates were cold-rolled with the thickness reduction of 50–65%. The cold-rolled plate was solution treated and water-quenched. Solution treatment was performed at 700–1000 °C up to 1 h. The specific weight of the present steel was measured as 6.87 g/cm^3, much lighter than ordinary commercial steels.

Tensile samples were machined from the water-quenched plates. Room temperature tensile tests were performed on the samples with the gage length of 25.4 mm up to failure using an INSTRON machine (model 4484) with the initial strain rates of 2×10^{-4} to 10^{-1} s^{-1}. High temperature tensile tests at 100–450 °C were carried out on the steel solution-treated at 1000 °C for 1 h using a MTS machine (model 810) with the initial strain rate of 10^{-3} s^{-1}. For high temperature tests, the samples were held at the testing temperature for 10 min before straining for equilibrating temperature. At least, three tests were done at each condition and the average were cited here.

Microstructures were optically examined by etching the mechanically polished samples with 2% Nital. The grain size measurement with/without considering annealing twin boundaries was conducted by electron back scattered diffraction (EBSD) (Oxford INCA system) equipped in a field emission gun scanning electron microscope (FEG-SEM, JSM 6500F). X-ray diffraction (XRD, Rigaku, D/MAX 2500H) as well as phase mapping by EBSD were employed on the samples before and after tensile tests for phase identification. Submicrostructures were observed by transmission electron microscopy (TEM, JEOL 1010) operating at 200 kV. Thin foils for TEM observation were prepared by a twin-jet polishing technique using a mixture of 10% perchloric acid and 90% methanol at −35 °C with an applied potential of 20 V.

Table 1
Room temperature tensile characteristics of Fe–28Mn–9Al–0.8C with three different grain sizes.

D_A (D_{AT}), μm	YS, MPa	UTS, MPa	e_u, %	e_f, %	UTS × e_f, MPa%
5 (3.5)	633	955	59.8	70.9	67,710
8 (6.7)	539	903	69.8	82.3	74,317
38 (14)	440	843	89.3	100.3	84,553

e_u, uniform elongation; e_f, elongation to failure.

3. Results and discussion

3.1. Grain size

Three different grain sizes were obtained by changing the solution treatment time at a fixed temperature of 1000 °C. Their SEM band contrast images are shown in Fig. 1a–c. Regardless of the grain size, most grains contain annealing twins. Without considering annealing twin boundaries, the grain size was 5 (±2.1) μm, 8 (±2.5) μm and 38 (±9.2) μm for 1 min, 10 min and 1 h solution treatment, respectively. When annealing twin boundaries were considered in the grain size measurement, the grain size was reduced to 3.5 (±12.1) μm, 6.7 (±1.5) μm and 14 (±4.4) μm, respectively. The XRD analysis (Fig. 1d) and phase mapping by EBSD (not shown here) revealed that solution treatment at 1000 °C made the steel fully austenitic.

The representative engineering stress–strain curves of the steel with three different grain sizes are plotted in Fig. 2a and the tensile properties inferred from Fig. 2a are listed in Table 1. All engineering curves exhibited continuous yielding and extensive strain hardening from the onset of plastic deformation.

As listed in Table 1, the steel showed exceptionally high elongation with reasonably high strength. As usual, yield strength (YS) and ultimate tensile strength (UTS) increased with decreasing the grain size and the trend was opposite for elongation. The combination of strength and elongation, which is often expressed in terms of the

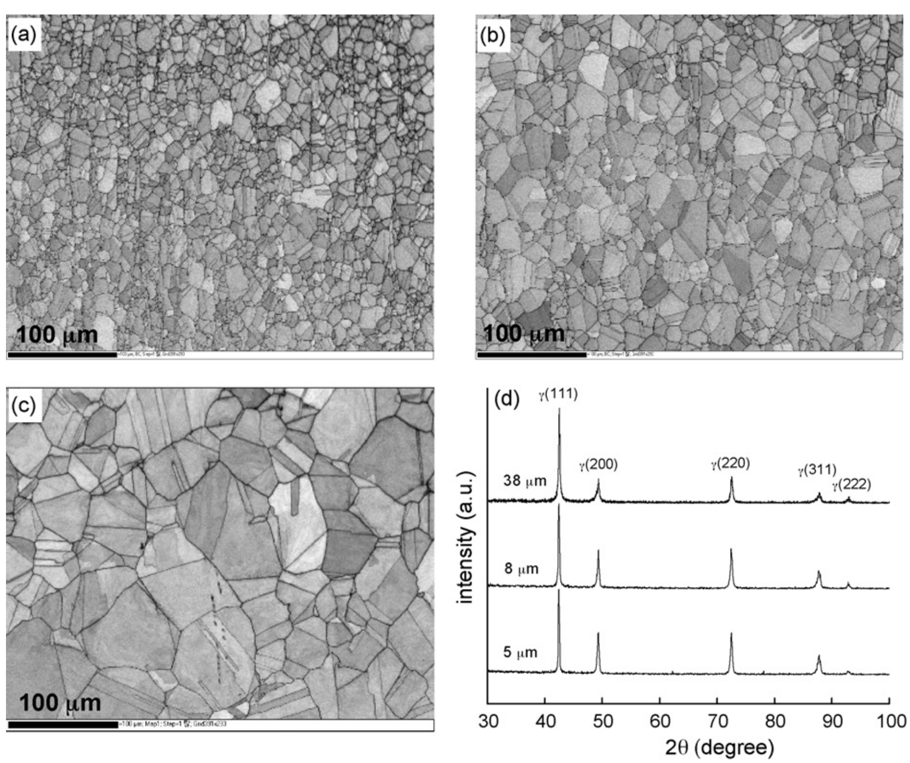

Fig. 1. SEM band contrast images of Fe–28Mn–9Al–0.8C solution-treated at 1000 °C. (a) D_A = 5 μm and D_{AT} = 3.5 μm for 1 min, (b) D_A = 8 μm and D_{AT} = 6.7 μm for 10 min, (c) D_A = 38 μm and D_{AT} = 14 μm for 1 h, (d) XRD profiles of (a), (b) and (c).