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Effect of annealing process on microstructure and mechanical properties of high manganese austenitic TWIP steel

M. M. Karkeh Abadi[1*], A. Najafizadeh[2], A. Kermanpur[3] and Y. Mazaheri[4]
Department of Materials Engineering, Isfahan University of Technology, Isfahan 84156-83111, Iran

Abstract
In the present study, the influence of annealing temperature on mechanical properties and the microstructure of a high manganese austenitic steel (Fe-30Mn-4Al-4Si-0.5C) was investigated. X-ray diffractometry, optical and scanning electron microscopy, hardness and tensile tests were used to analyze the relationship between mechanical properties and microstructure after annealing process. The results indicated an optimum combination of strength and elongation under the condition of high cold rolling reduction with subsequent annealing in the partial recrystallization region.

Keywords: High manganese steel, Stacking fault energy, Mechanical twinning, Mechanical properties.

1. Introduction

In automotive industries, safety, weight reduction and consequently fuel efficiency are still goals and researchers are looking for newer process to provide these demands. Therefore, the development of steels for automotive applications is focused on increase in strength (to amplify body structure and to reduce car weight) combined with the preservation or improvement of ductility (for more complex car design at room temperature and crash absorbing) [1-4].
Whereas almost 70 to 80% of car parts is made of steel [2], high work hardening during plastic deformation is needed to achieve mentioned mechanical properties. This manner can be illustrated by either phase transformation (transformation induced plasticity (TRIP)) or microstructural changes (twinning induced plasticity (TWIP)) in high-Mn steels (15–30 wt% Mn) [3, 5-8].
There are many differences between TRIP and TWIP steels. As an important difference, usually the austenite phase is stable in both under cooling but not under mechanical load. It occurs phase transformation in TRIP steel under mechanical load, whereas in TWIP steel, the mechanical twinning predominates in austenite grains during cold deformation (even at high strain/strain rate levels) [9, 10]. These modes of deformation attributed to their stacking fault energy (SFE)[1]. Although these steels have low SFE, TWIP steel has higher SFE than TRIP steel. So that, $\gamma_{fcc}\rightarrow\varepsilon_{hcp}^{Ms}$ transformation occurs when γ_{SFE} values <20 mJ m^{-2}. At lower SFE, $\gamma_{fcc}\rightarrow\varepsilon_{hcp}^{Ms}\rightarrow\alpha'^{Ms}_{bcc}$ becomes predominant. For steels with higher SFE (20 mJ m^{-2}< γ_{SFE} <40 mJ m^{-2}), twinning is the main deformation mechanism [11-13]. Chemical composition, temperature and grain size are effective parameters on SFE [14, 15].
Some methods have been suggested to achieve high strength-ductility. Bouaziz et al. [16] have suggested a recovery treatment on cold rolled steel that contains high density of mechanical twins. Mechanical twins, like grain boundaries, lead to strengthening. On the other hand, ductility is improved because of decreasing produced dislocations'/ twins' density in recovery treatment. Wang et al. have suggested a distribution of bimodal grain size including a mixture of both ultrafine and fine grains. Mi et al.[17] have reported annealing treatment and subsequently large cold rolling reduction.
In the present study, an attempt has been made to achieve optimum combination of strength and elongation by large cold rolling reduction and subsequently annealing treatment in the partial recrystallization region.

2. Experimental procedure

In this research, TWIP steel has been used chemical composition of 30% Mn, 3.75% Al, 4.04% Si, 0.544% C, 0.0537% P, <0.005% S and balance Fe (in wt%). It was vacuum melted in an electromagnetic induction furnace protected by argon atmosphere, and it was cast to slabs. Slabs were homogenized at 1200°C for 120 min and were cut to small pieces for the process of 85% cold rolling at room temperature. The cold rolling was done without reversing rolling. The cold-rolled plates (with ~1mm thickness) were annealed in temperature range 500–900°C for 30 min and they were followed by argon cooling. A schematic processing diagram of steel is shown in Fig. 1. In order to microstructural observations, the specimens were mechanically polished and etched by 2% Nital. Transverse sections were examined by optical

* *Corresponding author:*
Tel:+98 (912) 3316943 , Fax:+98 (311) 3912752
E-mail: mm.karkehabadi@ma.iut.ac.ir
Address: Department of Materials Engineering, Isfahan University of Technology, Isfahan, Iran.
1. M.Sc. Student
2. Professor
3. Associate Professor
4. Ph.D. Student

microscopy (OM) and Scanning electron microscopy [1] (SEM) in a. The phase composition of samples was analyzed by X-ray diffraction [2] (XRD) using Cu Kα radiation (λ = 0.15406nm, V=40kV and I=30mA). The XRD patterns were recorded in 2θ range of 20–100° (step size of 0.05° and time per step of 1s). Hardness measurement was performed using a Vickers hardness tester under 30Kg load and dwell time of 5s. The average of 5 indentations was calculated and reported as hardness value.

Tensile specimens were prepared by wire-cut electrodischarge machining. Tensile tests were carried out in strain rate $10^{-3}s^{-1}$ with 25 mm gage length, 6 mm gage width and 1 mm gage thickness at room temperature.

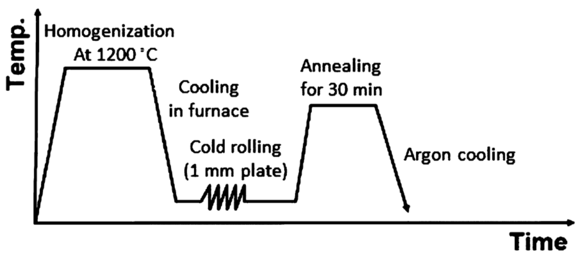

Fig. 1. The Schematic diagram of performed rolling and heat treatment processes on desired steel.

3. Results and discussion

X-ray analysis, shown in Fig. 2, indicated that the alloy has mono-phase austenite before and after cold rolling and even after tensile test. The SFE of prepared steel was calculated 67 mJ/m^2 [11, 18-20)], but it was determined about 45 mJ/m^2 [21)]. The value of 67 mJ/m^2 cannot explain the behavior of steel during cold deformation and the formation of a high volume of twins (Fig. 3a), but the model of Grässel et al.[21)] for manganese steels has good agreement with the behavior of this steel during cold rolling and microstructure observations.

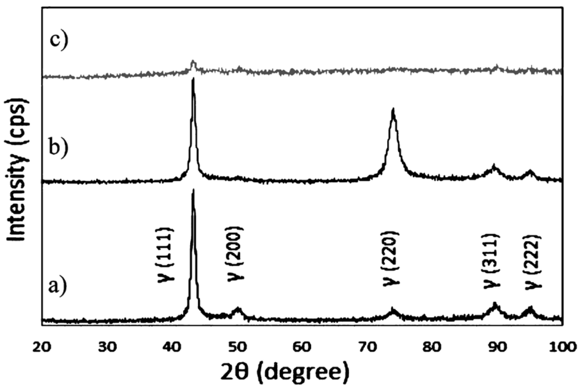

Fig. 2. X-ray diffraction results; a) for homogenized sample, b) for 85% cold rolled sample, c) for failed sample after tensile test.

[1]SEM model Philips XL30
[2]XRD model Philips X›Pert-MPD

As seen in Fig. 3, many deformation markings can be revealed after deformations. It can be seen the operation of the second twin system in 30% reduction in Fig. 3a and deformed slip bands in Fig. 3b, respectively. At large strains, it is difficult to observe the deformation twins in the grains because of the formation of slip bands. In Fig. 3c, a grid pattern has been influenced by slip bands.

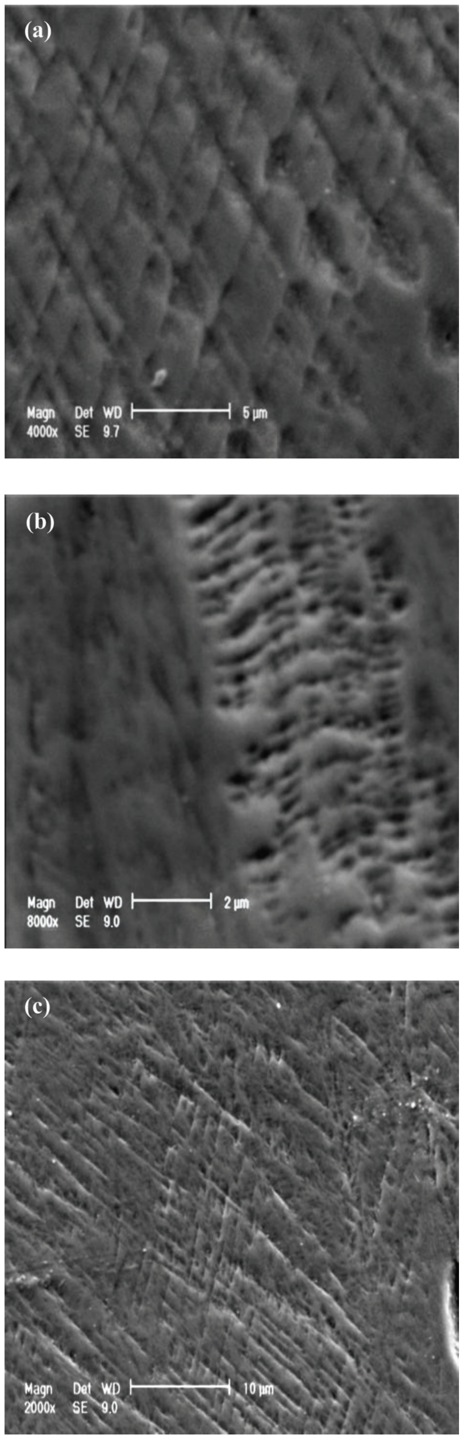

Fig. 3. Deformation microstructure, a) grid pattern caused by second twin system operation at ε=0.35, b) deformed slip bands at ε=1.89, and c) mixture of grid pattern and slip bands at ε=1.89.

Vickers hardness values for annealed steel have been presented in Fig. 4. The hardness value decreases with increase of annealing temperature, from 511 to 209Hv. The fraction of recrystallization area has been plotted vs. annealing temperature in Fig. 5 using extracted data from Fig. 4 and Eq. 1.

$$X_v = \frac{H_{500} - H_i}{H_{500} - H_{850}} \quad (1)$$

Where H_i is hardness value for each annealing temperature, H_{500} and H_{850} are hardness values of samples that have annealed at 500°C and 850°C, respectively. Assuming that recrystallization starts in some critical softening fraction value which is 0.2 [22], the recrystallization has been started at ~660°C and consequently, recrystallization temperature with 0.7 softening fraction is attributed to 750°C.

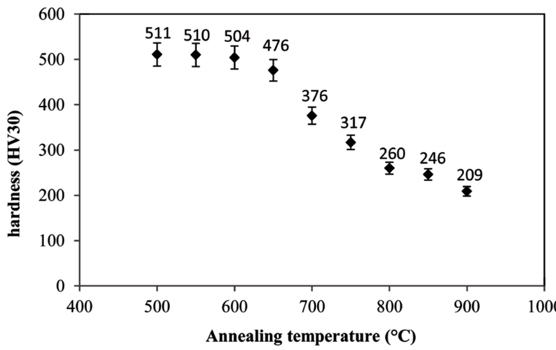

Fig. 4. Relationship between various annealing temperatures and hardness values.

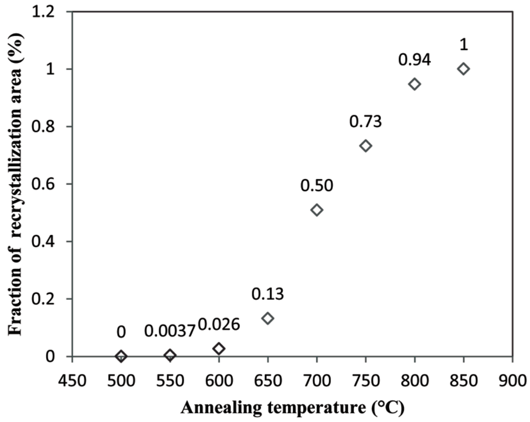

Fig. 5. The Fraction of recrystallization area vs. annealing temperature.

As shown in Fig. 6, there are many inclusions in this steel because of high amount of manganese. At 650°C, no recrystallized areas are shown and the microstructure contains dislocations and mechanical twins, similar to as-cold rolled. It means situated in recovery region. A few decreases in YS, UTS, and hardness values may be attributed to microstructural changes due to the annihilation of some dislocations and mechanical twins.

The recrystallization is started at ~660°C and 50% of it is completed at 750°C. The fully recrystallized microstructure is obtained at 850°C.As shown at this temperature, more equiaxed and fine grains contain some annealing twins. At higher temperatures, grain growth can be seen. In Fig.7 annealing twins adjacent with grain boundary were presented at 800°C.

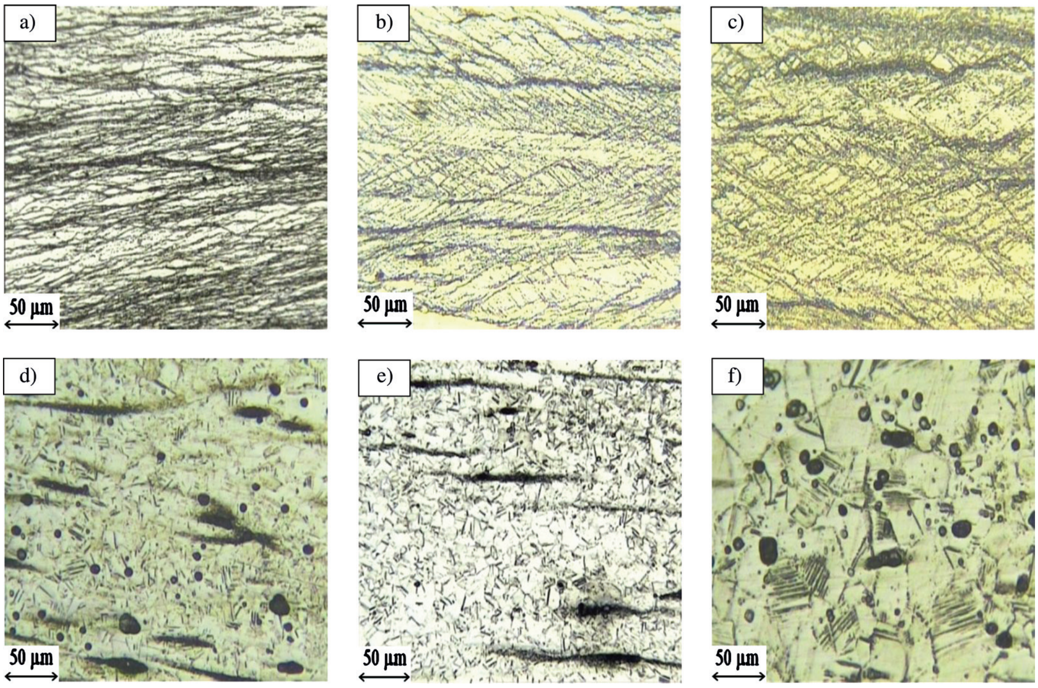

Fig. 6. Optical micrographs showing microstructures of specimens developed at various annealing temperatures, a) 650°C, b)700°C, c) 750°C, d) 800°C, e)850°C, f) 900°C.

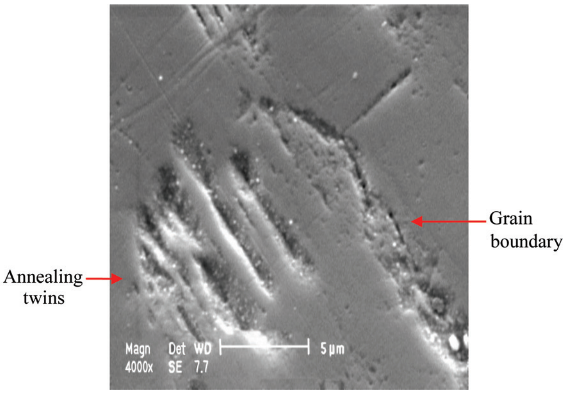

Fig. 7. Annealing twins adjacent of grain boundary.

Mechanical properties of this steel have been presented in Fig. 8. All engineering curves exhibited continuous yielding and extensive strain hardening. In as-rolled sample, there are high values for yield stress and ultimate tensile strength but it has no uniform elongation value. The value of total elongation for this sample is lower than 5%. The yield stress and ultimate tensile strength decrease with increasing the annealing temperature from 1300 to 300 MPa, and 1340 to 670 MPa respectively. At the same time, the total elongation increases from 5% to 72%.

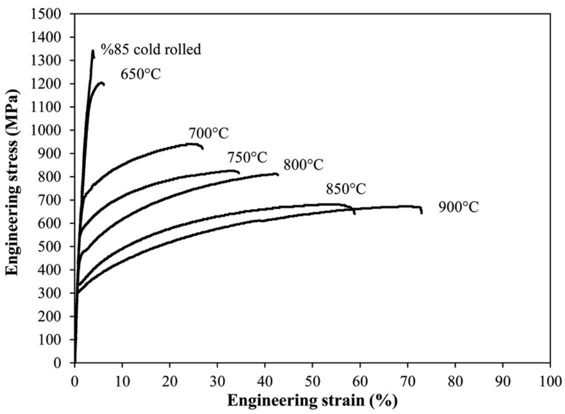

Fig. 8. Engineering stress-strain curves of 85% cold rolled specimens; and developed curves at various annealing temperatures for 30 min.

4. Conclusion

Because of low SFE and consequently limited dynamic recovery for TWIP steel, annealing process has a strong effect on mechanical properties and microstructures of this steel. The hardness, yield stress and ultimate tensile strength values decrease with increasing annealing temperature, while the elongation value increases. It was shown that the used steel had better mechanical properties when annealed at the temperature of 750 °C. By the condition, a mixture of recrystallized and unrecrystallized areas with high density of the mechanical twins (i.e., partial recrystallization region) was observed.

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