Fe-Mn(Al, Si) TWIP steel – strengthening characteristics and weldability

P Podany¹, M Koukolikova¹, T Kubina¹, R Prochazka¹, A Franc²
¹COMTES FHT a.s., Prumyslova 995, 334 41 Dobrany, Czech Republic
²University of West Bohemia, New Technologies - Research Centre, Univerzitni 8, 306 14 Plzen, Czech Republic
E-mail: pavel.podany@comtesfht.cz

Abstract. Twinning Induced Plasticity steel, or TWIP steel, has had increased interest in recent years from various industry sectors. This is due to it being lightweight, strong, and ductile; which are all properties that are useful in the automotive and aerospace industries. These steels potentially can offer lighter weight vehicles and parts with increased strength and other mechanical properties. This combination could offer greater fuel efficiency and performance while at the same time improving the safety features of the vehicle. This steel is characterised by being a high alloy steel, specifically having a high manganese content. It also has a fully austenitic microstructure at room temperature, which is a unique characteristic. But, for TWIP steel to be useful in various industrial sectors, it must have good weldability. This paper deals with the description of the strengthening due to the cold rolling on experimental heats of manganese steel with TRIP/TWIP effect. Impacts on microstructure, yield strength and tensile strength are described. Also, the weldability of experimental TWIP steel by studying the properties of weld joints after laser welding is described.

1 Introduction

Manganese steels with TWIP effect have attracted considerate attention recently since it has excellent tensile strength–ductility combination [1-3]. Their strengthening mechanism can be explained by the presence of alternative deformation mechanisms, such as the creation of twins (TWIP effect), phase transitions produced by strain (TRIP) and plasticity induced by shear bands [4-6]. An important parameter determining the type of deformation mechanism in these steels is their stacking fault energy (SFE).

According to Dumay et al., for SFE values greater than 18 mJ/m², the TWIP effect tends to occur, while for lower values, the TRIP effect is predominant, with an α’ martensite phase being formed for SFE smaller than 12 mJ/m². The SFE also depends on the chemical composition and test temperature of the steel. In alloys with a Mn content (wt. %) less than 15%, the TRIP effect dominates, while for a Mn content higher than 25% the TWIP effect is dominant. On the other hand, in alloys with a Mn content between 15% and 25%, the TRIP and TWIP effects can coexist [7] and combine final properties of both types of steels. It is recognised that TRIP steels show better mechanical strength whereas TWIP steels yield better elongation to fracture [8].

The role of aluminium as a solute in the manganese austenitic steels is twofold. It increases the stacking fault energy of the steel and hence reduces the probability of mechanical twinning during deformation. The higher stacking fault energy at the same time eliminates the formation of ε
martensite. Both of these effects lead to a greater resistance of the steel to hydrogen embrittlement. [9].
The role of the aluminium addition has therefore been investigated in this study for the cold rolled austenitic Fe–15.0Mn–0.1C–0.5÷1.5Al (wt. %) steel in order to establish the relation between mechanical properties, microstructure evolution, occurrences of phase transformation, strain-hardening mechanisms and deformation behaviour.

2 Experiment description
Experimental heats were manufactured in a vacuum induction type furnace and cast into a round ingot mould. The chemical composition of both heats is given in table 1. The only difference between the two heats is the amount of aluminium (1.40 % vs 0.38 %).

Table 1. Chemical composition of experimental heats

| Heat nr. | Element [wt. %] | Mn  | Si  | Al  | C   | Fe   |
|----------|-----------------|-----|-----|-----|-----|------|
| T15-81   |                 | 15.1| 1.58| 0.40| 0.12| bal. |
| T15-82   |                 | 15.0| 1.54| 1.40| 0.10| bal. |

After cooling, they were reheated in a furnace to the forging temperature of 1100 °C. In a universal hydraulic press, these ingots were then forged into slabs of 280×130 mm cross-section. The slabs were, in turn, hot-rolled to strips of a final thickness of 8 mm. The rolled strips of both heats were annealed at 850 °C for 2 hours. After annealing, the strips were cold rolled in 6 passes. The final reduction of both strips was 40 %. For the study of the microstructure behaviour and its development after annealing, the samples from cold rolled strips were then annealed at 950°C for 3 hours and at 750 °C for 1 hour.

The laser welding was done on ThruDisk 5002 with welding head BEO D70. The laser wavelength was 1030 nm (Yb: YAG) and it was connected to the fibre with 400 μm diameter. The focal length was 200 mm. Welding speed was 20 millimetres per second. Argon was used as a shielding gas at a flow rate of 24 litres per minute.

The mechanical properties were measured on standard samples (on cold rolled strips) and on mini-tensile test samples (on welded specimens). This method of mechanical properties measurement on a small amount of experimental material has proved successful in earlier research [10-16]. Standard tensile tests were carried out according to EN ISO 6892-1: Metallic materials – Tensile testing – Part 1: Test method at room temperature. The chosen strain rate was 0.001 s-1. Tests were executed using electromechanical testing system Zwick. Prior to testing, dimensions of samples were measured and recorded.

The microstructures were documented using a Zeiss Axio Observer optical microscope. EBSD analysis was performed and scanning electron micrographs taken by means of JEOL 7400F microscope and an HKL Nordlys EBSD camera from Oxford Instruments. Phase analysis by X-ray diffraction was carried out at room temperature using a Bruker D8 Discover diffractometer. The diffracted radiation was detected by means of a planar detector. A cobalt X-ray source has been used (λKα = 0.1790307 nm). The instrument was equipped with a polycapillary lens focusing the primary X-ray beam into a circular spot with a diameter of 0.5 mm.
3 Results and discussion

3.1 Microstructure evolution of sheets after cold rolling and after annealing

3.1.1 Cold rolling
The microstructure of sheets in the initial state (annealed after hot rolling) is very complex. Etching in Klemm’s reagent reveals ε martensite colourless γ (austenite) yellow to brown and α’ martensite blue to dark brown (see figure 1 a). HCP (ε) martensite forms little islands in α’ phase. The identification of all phases complies with the results EBSD analysis (see figure 1 b and caption).

![Figure 1. Microstructure of heat T15-81 in initial state a) light microscope; b) EBSD analysis - identification of phases (α’ - yellow, ε – red; γ – blue).](image)

The cold rolling affects the microstructure with deformation-induced transformation processes. Increasing strain caused by cold rolling induces twinning in austenite, twins further transforms to ε and ε transforms to α’. The microstructure of cold deformed specimens with 40% reduction is shown in figure 2. It is obvious, that the microstructure consists of the high volume fraction of martensite.

![Figure 2. Microstructures after cold rolling with 40% reduction a) heat T15-81-40%; b) heat T15-82-40%.](image)

According to the x-ray diffraction, ε martensite volume fraction in the initial state (before cold rolling) is 28 % for heat with lower aluminium content (T15-81) and 42 % for heat with higher aluminium content. BCC α’- martensite volume fraction in the initial state is similar for both heats: ca 6 – 8 %. Cold rolling leads to substantial increasing of α’ to 72 % and ε to 18 % for T15-81 and
to 62 % (α’) and to 27 % (ε) for T15-82 respectively. Austenite content drops from 64 % to 10 % (for T15-81) and from 52 % to 12 % (for T15-82).

3.1.2. Annealing
Subsequent annealing after cold rolling leads to complete recovery of microstructure and restoration of austenite fraction. Following figure 3 shows the comparison of the microstructure of both heats after annealing at 750 °C for 1 hour and after annealing at 950 °C for 3 hours.

**Figure 3.** Microstructures after annealing.
Annealing at a lower temperature for short time (1 h) leads to the microstructure with fine grains with an average size below 10 µm in comparison to rather coarse-grained microstructure after annealing at 950 °C for a longer period (3 h). The x-ray diffraction results show the decrease of austenite content due to the longer annealing for both heats. All results from x-ray diffraction are summarised in table 2. Austenite decreases mainly on the expense of increasing α’ martensite in the case of heat T15-81 and the ε martensite for heat T15-82 respectively.

**Table 2.** Volume fraction of phases according to the X-ray diffraction

|          | 750°C/1h | 950°C/3h |
|----------|----------|----------|
|          | T15-81   | T15-82   | T15-81   | T15-82   |
| α’       | 11.7     | 1.4      | 32.5     | 8.5      |
| ε        | 26.2     | 33.8     | 29.6     | 48.4     |
| γ        | 62.1     | 64.8     | 37.9     | 43.1     |
3.2. Mechanical properties
Evolution of mechanical properties during cold rolling are summarised in figure 4. Each column represents the yield and tensile strength after particular degree of reduction 0 %; 5 %; 10 %; 20 %; 30 %, 35 % and 40 %. The strengthening effect affects mainly the yield strength which increases from 258 MPa to 1472 MPa for heat T15-81 and from 304 MPa to 1278 MPa for heat T15-82.

![Figure 4. Mechanical properties of cold rolled sheets after reduction in the range 0 – 40%: a) heat T15-81; b) heat T15-82.](image)

3.3. Welding
In this experiment, laser welding was used to weld lap joints of two base metal compositions (heat T15-81 and T15-82) using laser power of 1650 Watts. In all of the welds, welding was done using a remote controlled solid state robotic laser. There were tried various metal combinations – both heats were combined, also various states – annealed and as cold rolled. The welding was done with and without a protective atmosphere. The overall look of weld line and cross section of the typical weld is shown in figure 5. The weld properties were tested by means of hardness measurement across the weld section and by means of minitensile test on samples which were taken from the top sheet, bottom sheet and the weld metal itself. The tensile strength of all samples made just from the weld metal varied from 880 to 1290 MPa with average tensile strength 1129 MPa. The average yield strength was 325 MPa. Lower values of tensile strength were achieved, when no shielding gas took place during the welding procedure.

![Figure 4. Laser welding of experimental sheets: a) configuration of weld seam, b) cross section of weld 3 with top annealed sheet + bottom cold rolled sheet and hardness values.](image)
4 Conclusion
The strengthening behaviour and microstructure evolution of two heats of steel with different aluminium levels have been studied. The following phases have been identified in both heats: γ (FCC) and two crystallographic variants of martensite, ε (HCP) + α´ (BCC). The strengthening mechanism of both steels consists in increasing martensite content which is induced by the deformation during the cold rolling. Higher yield strength and tensile strength of heat T15-81 are related to the higher content of α´ martensite in this heat (72 %) in comparison to the heat T15-82 (62 %).

The combination of alloying elements in investigated manganese steel showed the higher content of ε martensite in steel with higher aluminium content. It is on the contrary to the results of Ryu et al. [9], where the positive effect of decreasing of ε martensite content by means of Al alloying was shown. Such different behaviour can be related to the dissimilar carbon content (Ryu et al. used 0.6 % of carbon) where carbon interplay (in some way) with aluminium. This phenomenon will be further investigated in the future research.

The tensile strength of both heats exceeds 1450 MPa in as cold rolled state, the elongation A5 was above 35 % for annealed state. Both heats proved good properties after laser welding.

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