INFLUENCE OF PROCESSING PARAMETERS OF MEDIUM MANGANESE STEEL WITH 5 % MN ON DEVELOPMENT OF MECHANICAL PROPERTIES

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Abstract

Medium manganese steels belong to the group of third generation high-strength steels. These steels show an excellent combination of strength and ductility. Their manganese content ranges from 3 to 12%. After hot forming, the structure is usually martensitic. During intercritical annealing, martensite partly transforms to austenite. The choice of the correct alloying and intercritical annealing parameters, especially the heating temperature, leads to sufficiently stable retained austenite, which significantly affects the mechanical properties. The retained austenite exhibits TRIP effect during cold deformation and contributes to a significant deformation strengthening of the material.

Medium manganese steel with 0.2% C, 5% Mn and 3% Al was subjected to various intercritical annealing regimes after hot forming. To increase the stability of the retained austenite, isothermal hold at different temperatures was performed in the bainitic transformation region. To verify the effect of deformation, incremental deformation was applied during cooling. After processing, martensitic structures were obtained with varying fractions of bainite, ferrite and retained austenite. The ultimate tensile strength up to 1940 MPa was reached and the elongation was about 10%.

Keywords: Medium manganese steel, heat treatment, thermomechanical treatment, retained austenite

1. INTRODUCTION

The development of high-strength steels is currently focused on medium manganese high-strength steels, which belong to the third generation of high-strength steels. They have the advantage of chemical composition without expensive alloying elements. In addition to carbon, whose content can vary from 0.1 to 0.6%, they are alloyed with manganese in the range from 3 to 12% and also with Al, Si, Cr [1,2]. Due to their properties, such as a good combination of ultimate tensile strength (UTS) and ductility, they represent a promising material for various applications [3,4]. These steels were developed for the automotive industry, which needs materials with affordable cost and excellent mechanical properties that will lead to both weight reductions of individual parts and fuel savings, and also ensure occupant safety. Due to other properties such as excellent impact toughness especially at low temperature, they can be applied in other areas such as bridges, offshore platforms and other structural components [5].

Retained austenite plays an important role in the steel structure together with its deformation-induced martensitic transformation which occurs during subsequent cold deformation [6,7]. The most important factors affecting the fraction of austenite and its stability during subsequent deformation include the chemical composition of the austenitic phase, intercritical annealing parameters, grain size, austenite morphology, orientation, etc [8]. Compared to conventional TRIP steels, an effort is made to increase the fraction of hard phases such as martensite and to replace polygonal ferrite with acicular or bainitic ferrite, carbide-free bainite and martensite [1].

These steels are processed similarly to TRIP steels using intercritical annealing with or without isothermal hold in the bainitic transformation region [1-4]. The heat or thermomechanical processing parameters play an important role in developing a suitable microstructure that leads to the desired mechanical properties. Therefore, this paper is aimed at describing the effect of different processing parameters on the development of microstructure and mechanical properties for medium manganese steel with 5% Mn in order to obtain the best combination of UTS and ductility.

2. EXPERIMENTAL PROGRAM

For the experimental program, high-strength medium manganese steel with a carbon content of 0.2%, manganese 5% and aluminium 3 wt% was chosen (**Table 1**). Manganese as an austenite stabilizing element significantly reduces the fraction of ferrite because it shifts the transformation of austenite to ferrite in the CCT diagram to the right [1, 7, 9]. Manganese also causes together with silicon strengthening of the solid solution and retard carbide precipitation [2, 10]. Since manganese decreases the transformation temperatures of A_{c1} and A_{c3}, aluminium has been added to steel, which in turn increases these temperatures [11,12]. This makes it possible to increase the temperature of intercritical annealing and shorten holding time at this temperature. Aluminium has also been added to further increase the stability of the retained austenite [1].

Table 1 Chemical composition of the 5Mn3Al steel (wt.%) and transformation temperatures (°C)

C	Si	Mn	P	S	Cr	Ni	Al	Nb	A_{c1}	A_{c3}	M_s	M_f
0.20	0.58	5.02	0.007	0.001	0.180	0.078	2.95	0.062	1,071	682	296	175

The phase transformations were determined using JMatPro software [13]. The temperature of A_{c3} and A_{c1} were determined to be 1.071 and 682 °C respectively (**Figure 1 a, b**). It is evident from the CCT diagram that there is a large region for ferrite formation starting from high cooling rates, pearlite transformation is pushed towards lower cooling rates and without isothermal holds during cooling bainite formation will not occur (Figure 1a). Bainite formation can be promoted by isothermal holds in the region from 425 to 300 °C (**Figure 1b**).

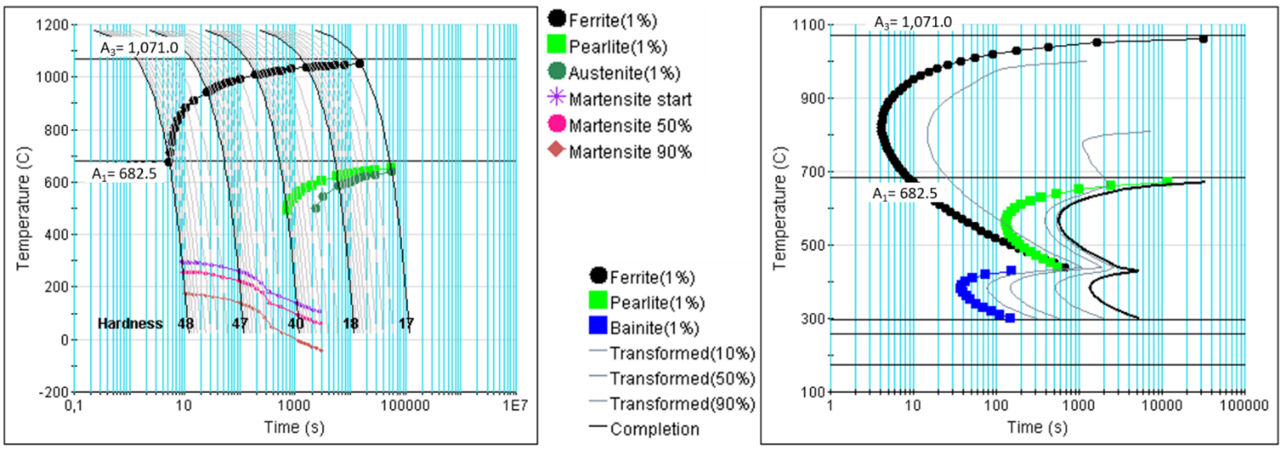

Figure 1 Austenite decomposition diagrams calculated for experimental steel in JMatPro: a) CCT, b) TTT

The experimental steel was cast into a cone-shaped ingot weighing 168 kg. After removing the head and the bottom, the ingot was split transversely in half. After surface machining, homogenization annealing was carried out at 1.200 °C for 10 hours with cooling in a closed furnace. The ingot halves were then quartered and the quarters forged into bars using a hydraulic press. The forging temperature was 1.100 °C. The forging was first carried out in straight anvils to match the triangular shape of the forging and then in shaped anvils up to a bar diameter of 20 mm. Among each forging operation, the material was reheated to forging temperature. After forging, the bars were cooled in a closed furnace. Prior to sample fabrication, the bars were annealed at 680 °C for 1 h in a furnace with a protective argon atmosphere. Cooling was carried out in a closed furnace. Threaded samples with a length of 76 mm and gauge diameter of 8 mm were made from the rods.

2.1. Thermomechanical processing

In order to accurately control the temperature profile or the embedded deformation, a thermomechanical simulator was chosen to optimize the parameters. This device features high-frequency resistive heating and enables high heating rates of up to 200 °C/s. The maximum heating temperature is 1.300 °C. Temperature measurements are made with a K-type thermocouple welded to the sample surface. Cooling can be done with water, water mist or compressed air.

In the first step, the effect of heating temperature and temperature of isothermal holds in the bainitic transformation region was investigated. The heat treatment consisted of heating to 950 or 1.050 °C with a 100 s hold. This was followed by cooling at 16 °C/s to 300 or 425 °C with a holding time of 600 s. The samples were then cooled to RT. The heating temperature was chosen to test heating to the intercritical region between the A_{c1} and A_{c3} temperatures and also to the single-phase region without the presence of ferrite.

In the second step, the effects of the embedded deformation on grain refinement, the structure evolution and mechanical properties were tested. First, the effect of the 40-fold deformation in the temperature range from 950 to 720 °C was investigated. The total magnitude of the inserted logarithmic strain corresponded to $\varphi = 5$. The strain was inserted alternately in tensile and compressive steps. This process could represent, for example, incremental rolling of tubes. In the second procedure, the compressive deformation was inserted in two steps only, one step being performed at 950 °C and the other at 720 °C. The magnitude of the logarithmic strain in one step was $\varphi = 0.095$.

Table 2 Parameters of the heat and thermomechanical treatment

Mode designation	T_A (°C)/t_A (s)	T_B (°C)/t_B (s)	Deformation	HV10 (-)
950_425	950/100	425/600	-	238
950_300	950/100	300/600	-	426
1050_425	1050/100	425/600	-	443
1050_300	1050/100	300/600	-	447
950_425_40x	950/100	425/600	40x (950 - 720 °C)	411
950_425_2x	950/100	425/600	1x (950 °C), 1x (720 °C)	434

After processing on the thermo-mechanical simulator, mini-stretch samples with a gauge length of 5 mm and a cross section of 2 mm x 1.2 mm were fabricated from the active part of the sample. The microscopic analysis was performed after standard metallographic preparation by grinding and polishing using light optical (OM) and scanning electron microscope (SEM). Nital was used for etching. Vickers hardness measurements were carried out using a load of 10 kg.

3. RESULTS AND DISCUSSION

In the first part of the experiment, only heat treatment without embedded deformation was tested (**Table** 1). In the case of the combination of a heating temperature of 950 °C with a hold at 425 °C (mode 950_425), a typical structure after intercritical annealing was obtained for TRIP steels. The structure consisted of a mixture of proeutectoid ferrite and bainite **(Figure 2)**. Bainite occurred in blocks and was formed by laths of bainitic ferrite. A number of dark particles were observed in the bainite, which in some cases looked like pearlite (**Figure 2a**). Especially those that occurred at the previous austenite grain boundaries. The scanning electron microscope also detected small areas of martensite in the structure, which were formed by transformation of insufficiently stabilized retained austenite, so-called M-A component (**Figure 2b**). Retained austenite can still be expected even between the laths of bainitic ferrite. The hardness of this structure was 238 HV10. The UTS was relatively low at only 777 MPa and the ductility was 23% (**Figure 3**). There was a significant change in the structure when the isothermal hold temperature was lowered to 300 °C. Because this temperature is already very close to the M_s temperature a large amount of martensite was detected in the structure, which could be formed both during cooling to holding temperatures and then during cooling to room temperature. Due to the isothermal hold at 300 °C, both the tempering of martensite and the formation of bainite occurred. The resulting microstructure was a mixture of tempered and fresh martensite and bainite (**Figures 2c, d**). The fraction of ferrite decreased significantly and appeared in two different morphologies. Both in the form of relatively large globular grains aligned in the matrix and in the form of thin lamellae of bainitic and acicular ferrite. Removing a large fraction of proeutectoid ferrite from the structure, the hardness value increased to 426 HV10. The martensitic-bainitic matrix also resulted in an increase in the UTS value to 1.497 MPa. The ductility decreased to 10% (**Figure 3**).

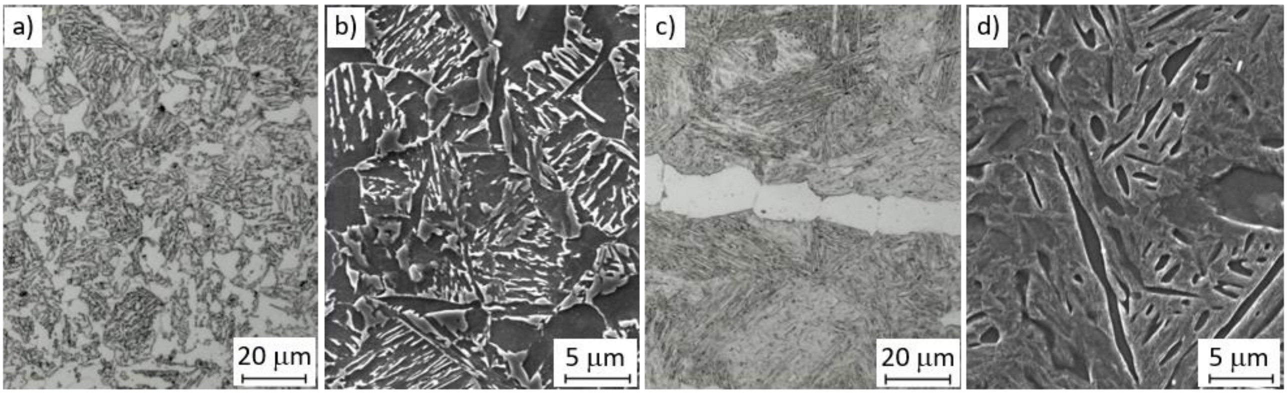

Figure 2 Microstructure after heat treatment of the steel investigated: a) 950_425 - OM micrograph, b) 950_425 - SEM micrograph, c) 950_300 - OM micrograph, d) 950_300 - SEM micrograph

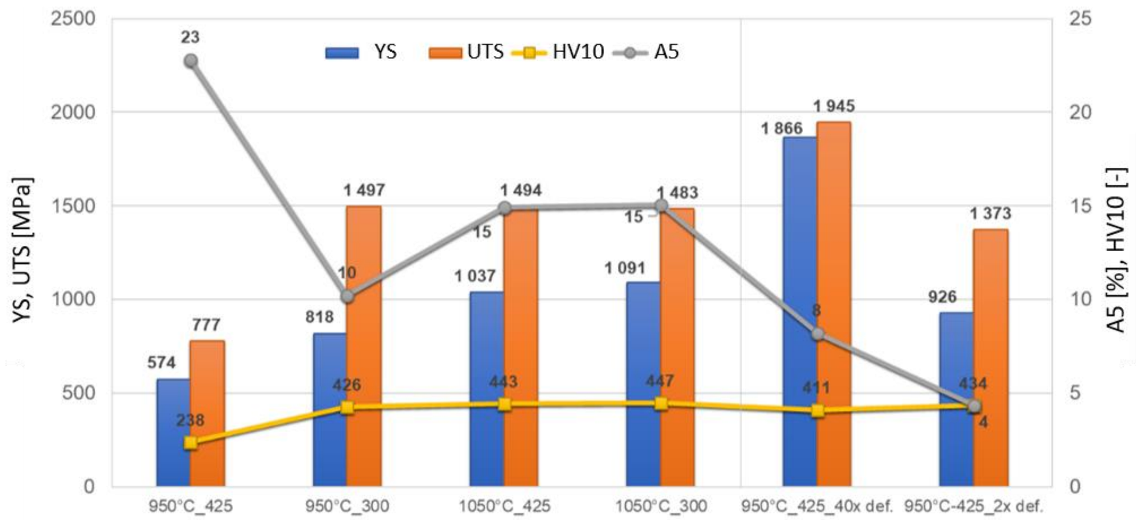

Figure 3 Mechanical properties of the steel after heat treatment and thermomechanical processing

At a higher heating temperature of 1.050 °C, a similar structure as in the previous case was obtained in both cases of isothermal hold at 425 °C and 300 °C (**Figure 4**). Proeutectoid ferrite was observed in the martensitic-bainitic matrix in the form of grains joined into elongated lined formations. Ferrite was also observed in the form of acicular or bainitic ferrite. The hardness in both modes ranged from 443 to 447 HV10. The UTS was again about 1.490 MPa and the ductility increased to 15% (**Figure 3**).

The deformation modes were tested with a heating temperature of 950 °C and temperature of isothermal hold of 425 °C. When inserting a 40-fold deformation in the temperature range from 950 to 720 °C, a very fine martensitic-bainitic structure was obtained with aligned ferrite grains and acicular ferrite distributed over the quenched matrix (**Figure 5a, b**). The hardness value was again over 400 HV10. Strain hardening occurred in the structure, as the UTS increased to 1.945 MPa with a decrease in ductility to 8% (**Figure 3**). If only two deformation steps were inserted, one at 950 °C and the other at 720 °C, a similar microstructure was obtained again (**Figure 5c, d**), but a decrease in UTS to 1.373 MPa and ductility to 4% was found.

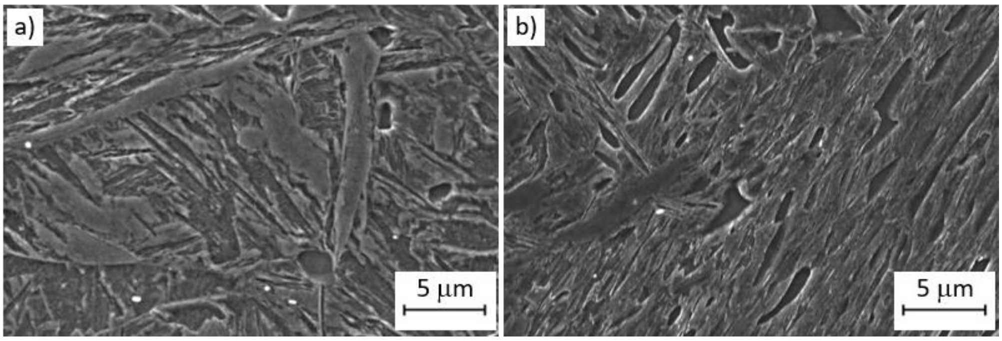

Figure 4 Microstructure after heat treatment of the steel: a) 1.050_425, b) 1.050_425 - SEM micrographs

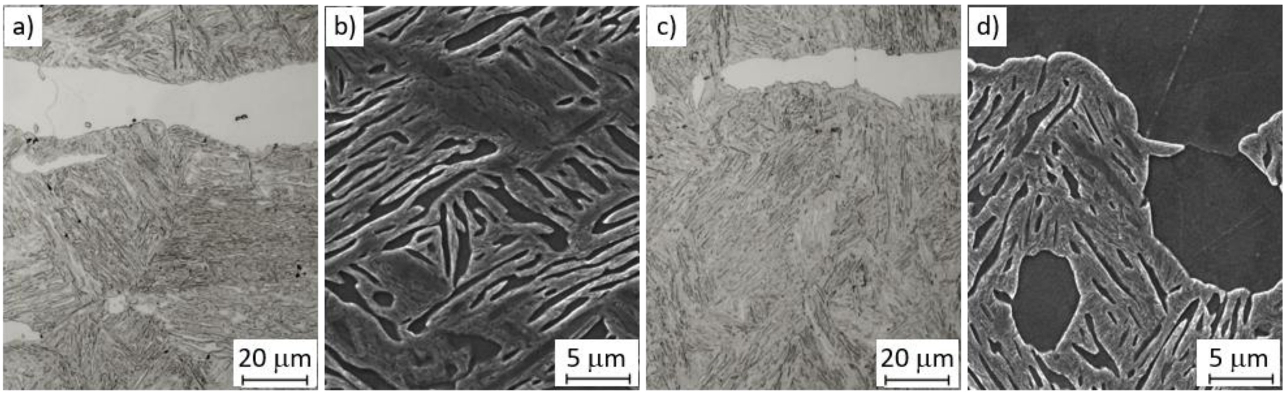

Figure 5 Microstructure of the steel after thermomechanical processing at 950 °C + 425 °C /600 s: a) 40x def - OM, b) 40x def - SEM, c) 2x def. - OM, b) 2x def. - SEM

4. CONCLUSION

For the medium manganese steels with 5% Mn and 3% Al, the processing parameters were optimized using a thermomechanical simulator. The results showed that at a lower heating temperature of 950°C, the temperature of isothermal transformation plays a significant role. If the isothermal hold temperature of 425 °C was chosen, a typical TRIP structure consisting of ferrite, bainite and a small fraction of martensite and retained austenite was obtained. Therefore, the UTS of only 777 MPa was obtained. If the isothermal hold temperature was lowered just above the MS temperature, a large amount of martensite was formed, which led to a significant increase in the yield strength up to 1.497 MPa. Increase in the heating temperature up to 1.050 °C resulted in further improvement of mechanical properties, when the UTS of about 1.490 MPa was reached, but the ductility value increased to 15%. The application of the incremental deformation in the temperature range from 950 to 720 °C again had a positive effect on UTS, which increased to 1.945 MPa, however ductility decreased to 8%.

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