High ductility of bainite-based microstructure of middle carbon steel 42SiMn

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Abstract. Heat and thermo-mechanical treatments with various processing parameters were applied to middle carbon low alloyed 42SiMn steel. The aim of the treatment was to obtain multiphase microstructure typical for TRIP (Transformation induced plasticity) steel and to achieve the best combination of ultimate tensile strength and ductility. TRIP steels typically possess about 5-15% of metastable retained austenite, which can transform to martensite during plastic deformation. The gradual phase transformation during loading postpones the onset of necking, thus increasing ultimate tensile strength and ductility at the same time. Manganese and silicon, used as the main alloying elements of the experimental steel, are employed to increase austenite stability and to hinder cementite precipitation during the treatment. All proposed methods of heat and thermo-mechanical treatment contain bainitic hold at 400 °C or 425 °C. The final microstructures were very complex, consisting of bainite, ferrite, very small areas of extremely fine perlite lamellas, about 10% of retained austenite and M-A constituent (austenitic islands partially transformed to martensite). Even though pearlite and martensite are undesirable microstructure in TRIP steel, the tensile strength ranged from 850 to 1065 MPa and ductility A5mm from 26 to 47 %.

1 Introduction
Various grades of advanced high strength steels have been proposed in the last two decades [1]. Many of them require particular chemical compositions, based either on heavy alloying or on special micro alloying. Each group of AHSS has slightly different combination of mechanical properties, starting from relatively low strength and high ductility of dual phase steels [2] to very high strengths with lower ductility of martensitic steels after Q-P (quenching and partitioning) processing [3-7]. Somewhere in the middle are TRIP (transformation induced plasticity) steels which can possess interesting combinations of relatively high strengths and very good ductility [8, 9]. Another advantage of TRIP steels is that they can reach these properties even without special alloying. The most typical chemical composition of TRIP steels consist of 0.2-0.4% C, 1-2%Mn and 1-2%Si [10, 11] and the first conventional TRIP steels kept their carbon contents at the lower boundary of the interval, around 0.2%. Many other alloying concepts have been also investigated to improve mechanical properties or workability of the steel.

The main principle which is responsible for good strength to ductility balance is mechanically induced transformation of retained austenite to martensite, which can occur during cold plastic deformation. The gradual transformation postpones the necking of the steel, which increases homogeneous plastic deformation. It has been shown that about 10% of properly stabilized retained austenite is ideal to ensure the utilization of TRIP effect. To retain sufficient amount of austenite to room temperature in low alloyed steel, special heat or thermo-mechanical treatment has to be applied. The aim of the treatment is not only to produce desired semi-product but mainly to produce suitable
multiphase microstructure and also to increase carbon content in remaining austenite during the cooling and thus decrease its martensite start temperature below room temperature. Higher carbon content in the steel would naturally help to achieve this goal and it would also increase the hardness and strength of the steel. Conventional steel with middle carbon contents usually pay for increase in strength by deteriorating ductility. This does not have to be the case for TRIP steels, as this article is going to suggest.

2 Experimental program

2.1 Characterization of used steel
Middle carbon, low alloyed steel with very simple alloying concept was used in this work. Besides 0.4% C, it was further alloyed only by 0.6% Mn and 2% Si. Manganese was added to support austenite stabilization, as an austenite-forming element. Silicon can hinder cementite precipitation during the cooling, which in the end effect also promotes austenite stabilization, as more carbon is than available to diffuse into remaining austenite. TTT (time temperature transformation) diagram of the steel was calculated in JMatPro software (Figure 1).

![Figure 1 TTT diagram of 0.4C-0.6Mn-2Si steel.](image)

Mechanical properties after the processing were obtained by tensile test on small flat samples with the length of active part being 5 mm and the cross section 2 x 1.2 mm. Volume fraction of retained austenite was determined by X-ray diffraction phase analysis using automatic powder diffractometer AXS Bruker D8 Discover with HI-STAR detector and Co lamp ($\lambda_{K\alpha} = 0.1790307$ nm). Focusing polycapillary lens was used to achieve X-ray spot with 0.5 mm diameter. Resulting microstructures were analyzed by scanning electron microscope Zeiss EVO.

2.2 Heat and thermo-mechanical treatment
Three methods of thermo-mechanical treatment and six methods of heat treatment were tested on experimental steel (Table 1). Two soaking temperatures were chosen, the higher temperature of 1000 °C ensures full austenisation of the steel, while the lower temperature of 850 °C lies in the intercritical austenite - ferrite region. Soaking hold of 100 s was performed for all processing methods used in this work. Two compressive deformations were carried out during the cooling of the samples from soaking temperature. The size of each deformation corresponded to 10% of actual size of the sample (i.e. the second deformation was slightly smaller). Deformation temperatures of 900 °C and 800 °C or 720 °C were chosen on the base of previous experiments [5, 6]. Two cooling rates were tested, 15 °C/s and 30 °C/s, to cool the samples down to bainitic hold. Bainitic hold lasted 600 s for all of the methods and was applied either at 425 °C or 400 °C.
Processing of all samples was carried out at thermo-mechanical simulator, which allows precise control of thermal and deformation parameters and good repeatability of processing parameters. The active part of processed samples is cylindrical with 8 mm diameter and the length of 16 mm. The geometry of the sample was optimised to achieve homogeneous thermal field distribution in active part of the specimen.

Table 1. Processing parameters (Ts – soaking temperature, Tdeformation – temperature of compressive deformations, TB – bainitic hold temperature), mechanical properties and retained austenite (RA) volume fraction.

| Ts [°C] | Tdeformation [°C] | Cooling rate [°C/s] | TB [°C] | Rm [MPa] | Asmm [%] | RA [%] |
|-------|------------------|-------------------|-------|--------|--------|-------|
| 1000  | 900, 800         | 30                | 425   | 992    | 37     | 13    |
| 1000  | 900, 720         | 30                | 425   | 925    | 42     | 12    |
| 1000  | 900, 720         | 30                | 400   | 963    | 34     | 14    |
| 1000  | -                | 30                | 425   | 1013   | 29     | 14    |
| 1000  | -                | 30                | 400   | 1068   | 26     | 13    |
| 850   | -                | 15                | 425   | 891    | 37     | 11    |
| 850   | -                | 15                | 400   | 869    | 34     | 5     |
| 850   | -                | 30                | 425   | 932    | 45     | 9     |
| 850   | -                | 30                | 400   | 934    | 47     | 11    |

3 Results
Thermo-mechanical treatment with higher temperature of the second deformation 800 °C resulted in very coarse bainitic microstructure with small amount of fine grains of proeutectoid ferrite and small pearlitic areas (Figure 2). The microstructure had high strength of 992 MPa with very good ductility Asmm of 37%. There was 13% of retained austenite, which formed the laths of upper bainite.

Lowering of the temperature of the second deformation to 720 °C slightly decreased tensile strength to 925 MPa and improved ductility Asmm to 42%. There was higher amount of free ferrite in the microstructure, bainitic blocks were finer even though individual bainitic laths were generally thicker than in the previous case (Figure 3). Very fine pearlitic areas were also observed in the microstructure.

Lower bainitic hold temperature of 400 °C further prolonged bainitic laths of M-A constituent and increased the amount of bulk M-A constituent islands (Figure 4). Despite of 14% of retained austenite in the microstructure, the ductility Asmm of this sample reached only 34%.

Heat treatment was first repeated with the same thermal schedule, which was used in thermo-mechanical processing. Rather coarse microstructures made of mainly upper bainite were found in the samples soaked at 1000 °C and cooled by 30 °C/s to either 420 °C or 400 °C (Figure 5). In both cases, distinctively lamellar bainitic microstructures possessed high strengths above 1000 MPa. Lower temperature of bainitic hold increased the strength to 1068 MPa at the cost of relatively low ductility Asmm 26%. No traced of pearlite were observed in either microstructure suggesting deformations applied during the cooling accelerated pearlite transformation in all the samples during thermo-mechanical treatment.

To increase ferrite fraction in the final microstructure without risking pearlite formation at lower cooling rates, soaking temperature of the next four treatments was kept at 850 °C. The combinations of two cooling rates, 15 °C/s and 30 °C/s, and two bainitic holds, 425 °C and 400 °C, were tested.
Lower soaking temperature resulted in all the cases in visible refinement of the final microstructure (Figure 6, Figure 7). Slower cooling rate 15 °C/s caused pearlite formation in the sample held at 425 °C. Small pearlitic colonies with typical size below two micrometers were quite frequent in the microstructure (Figure 6). It should be however noted that inter-lamellar distance was very small. Tensile strength was just below 900 MPa with ductility 37%. Mechanical properties deteriorate further to 869 MPa tensile strength and 34% ductility $A_{5mm}$ with the drop of bainitic hold to 400 °C. These relatively small changes in mechanical properties are accompanied by more significant changes in the final microstructure, which is now predominantly ferritic-pearlitic with a few bulky islands of M-A constituent (Figure 7). The pearlite blocks are coarser than in the previous case and the laths in them are sometimes thicker. There is only 5% of retained austenite in the microstructure, mostly of the bulky type.

On the other hand, lower soaking temperature combined with higher cooling rate of 30 °C/s resulted into the best combinations of mechanical properties and austenite volume fractions around 10%. The microstructure obtained after bainitic hold at 425 °C had very homogeneous distribution of fine equiaxed grains of free ferrite and bainitic laths with occasional occurrence of M-A constituent islands (Figure 8). Few laths of very fine pearlite were also sporadically found in the microstructure. The combination of high ultimate tensile strength of 932 MPa with ductility $A_{5mm}$ of 45% prove that the presence of pearlite in this form and amount does not necessarily have a negative effect on mechanical properties.

Decrease of bainitic hold temperature to 400 °C further increased the ductility $A_{5mm}$ to 47% without the change in tensile strength. The microstructure consisted of higher amount of free ferrite, most of the bainitic blocks are smaller than in previous case, however there are also heterogeneously distributed blocks with the length of bainitic laths around 10 micrometers (Figure 9).
Figure 4. Thermo-mechanical treatment: 1000°C/100s, deformations at 900°C and 720°C, 30°C/s, hold at 400°C/600s.

Figure 5. Heat treatment: 1000°C/100s, 30°C/s, hold at 400°C/600s.

Figure 6. Heat treatment: 850°C/100s, 15°C/s, hold at 425°C/600s.

Figure 7. Heat treatment: 850°C/100s, 15°C/s, hold at 400°C/600s.

Figure 8. Heat treatment: 850°C/100s, 30°C/s, hold at 425°C/600s.

Figure 9. Heat treatment: 850°C/100s, 30°C/s, hold at 400°C/600s.

4 Conclusions
Nine methods of heat and thermo-mechanical treatments were applied to low alloyed middle carbon steel 0.4C-0.6Mn-2Si. In most of the cases, the resulting complex multiphase microstructures possessed interesting combination of relatively high ultimate tensile strength above 900 MPa with ductility $A_{5mm}$ above 35%.
The very best ductility $A_{5\text{mm}}$ of 47% combined with very good strength of 934 MPa were achieved after simple heat treatment consisting of heating to 850 °C with 100 s hold at the temperature and subsequent cooling by 30 °C/s to 400 °C. High ductility $A_{5\text{mm}}$ of 45% obtained after similar heat treatment with the hold at 425 °C suggest that the first result is no exception, but rather a rule for this kind of the treatments and resulting microstructures. The final microstructures with high ductility consisted of bainite-ferrite mixture, with about 10% of retained austenite and small amount of extremely fine pearlitic areas.

Acknowledgement
The present contribution has been prepared within project LO1502 ‘Development of the Regional Technological Institute’ under the auspices of the National Sustainability Programme I of the Ministry of Education, Youth and Sport of the Czech Republic aimed to support research, experimental development and innovation.

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