HOT DUCTILITY CHANGES OF MICROALLOYED STEEL WITH Mo ADDITION

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Abstract

The work presents results of plasticity response and strength levels of two micro-alloyed steels having different Ti, N and Mo contents. In temperature intervals of 600/650°C-750°C, 750°C-1000/1200°C and 1200°C-1350°C troughs of hot ductility expressed by reduction of areas were detected. Material with higher Mo content generally showed higher parameters of hot ductility. Reasons of the plasticity differences are expressed using chemical composition and possible bonds, using evaluation of mechanical properties such as yield strength and/or strength and by use of microstructure analysis results. In lower temperature interval, for the lower plasticity higher portion of formed allotriomorphic ferrite and bainite was typical, while the higher reduction of area was connected with higher volume fraction of acicular ferrite. The higher temperature intervals were related to higher volume fraction of precipitated particles on the austenite grain boundaries. In total plasticity response of given steel type differences in austenite and ferrite strength characteristics play an important role.

Keywords: microalloyed steel, reduction of area, allotriomorphic ferrite, acicular ferrite

1. INTRODUCTION

Strength and plastic response expressed as reduction of area (RA) play an important role at study of deformation mechanisms and defects propagation. It is known that material properties are dependent not only on temperature and realised deformation mechanisms. Transformation processes play also a role [1, 2]. During cooling process the strengthening of materials is going up, however, in most materials (especially on the base of C-Mn steels) trough formation of plastic response in the region of approx. 600-700°C and 850-1000°C can be detected [3-6]. Nucleated cracks are intercrystalline and they are propagated on the primary austenite grain boundaries. Microscopically the cracks show ductile appearance connected with micro-cavities formation, their growth and coalescence. Intergranularly nucleated inclusions, precipitates, mostly of nitride types and austenite phase transformation, especially to allotriomorphic ferrite (ATF) are the reason [2, 7]. The strain rate naturally also plays a role. During slower strain rates diffusion processes can get ahead unlike the faster strain rates supporting more favourable RA response [2, 4, 7]. At high temperatures (approx. 850-1000°C) the RA is influenced by precipitated particles on the primary austenite grain boundaries, whereas the low temperature trough (approx. 600-850°C) of the RA is connected with austenite transformation to ferrite, especially to the ATF, which forms narrow intergranular bands. The rate of yield stress between ferrite and austenite corresponds to 1:5 [4]. Afterwards, a localisation of plastic deformation in the ferrite region, cavity initiation on non-metallic inclusions and cracks formation are the results [4, 8]. One side of austenite-ATF inter-phase can be locally enriched by carbon and consequently can show lower cohesive strength and higher susceptibility to cracking [2, 6, 7]. Besides the primary grain size, chemical constitution, including residual elements, also play an important role both in narrow ferrite bands and on austenite grain boundaries at high temperatures [7, 9-15].
The aim of the work is to elucidate the differences of hot ductility in the temperature range of 600-1350°C of two low carbon-micro-alloyed and/or low alloyed steels.

### Table 1 Chemical composition of used materials (M). The * represents Cr+Cu+Ni+Sn

<table>
<thead>
<tr>
<th></th>
<th>C</th>
<th>Mn</th>
<th>Si</th>
<th>P</th>
<th>S</th>
<th>Al</th>
<th>Al</th>
<th>Ti</th>
<th>Nb</th>
<th>Mo</th>
<th>*</th>
<th>N</th>
<th>Ca</th>
</tr>
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<tbody>
<tr>
<td>A</td>
<td>0.06</td>
<td>1.61</td>
<td>0.40</td>
<td>0.013</td>
<td>0.002</td>
<td>0.028</td>
<td>0.027</td>
<td>0.021</td>
<td>0.055</td>
<td>0.160</td>
<td>0.445</td>
<td>74</td>
<td>23</td>
</tr>
<tr>
<td>B</td>
<td>0.07</td>
<td>1.61</td>
<td>0.35</td>
<td>0.010</td>
<td>0.005</td>
<td>0.030</td>
<td>0.028</td>
<td>0.053</td>
<td>0.054</td>
<td>0.024</td>
<td>0.299</td>
<td>55</td>
<td>34</td>
</tr>
</tbody>
</table>

2. EXPERIMENTAL MATERIAL AND TECHNIQUE

For experiments two manufactured continuously cast steels (marked A and B) with chemical composition presented in Table 1 were used. The calculated $\text{Ac}_1$ and $\text{Ac}_3$ temperatures corresponded to 716°C and 908°C (material A), or 715°C and 905°C (material B). From both materials tensile specimens with a section of 6 mm in diameter and gauge length of 23 mm were machined and the tensile properties were determined using the TSM 20 INOVA. For each temperature mean value was determined from two measurements. Part of the machine is consists of the loading frame FSM20 M03 and dynamo-graph module for 20 kN and 1 kN. Shift of suspension bridge and hereby the speed of loading corresponded to $1.10^{-1}$ mm.s$^{-1}$. Making-up of testing course, the control itself and recording of the test were carried out using a computer. The apparatus enables a realisation of tensile test according to the standard ČSN EN 10002-5 and the so called general test, which allows adjustment of the test conditions in six separate steps. Plasticity, respectively hot ductility is determined as RA after tensile test. The mentioned equipment includes furnace temperature regulation, measurement and temperature record controlled by computer. The used temperature interval of interest lay in range of 600–1350°C. Metallographic analysis was an integral part of solution including cleanness and type of microstructure in vicinity of the fracture after tensile test in both basic materials. Metallographic evaluation was carried out by use of the Olympus light microscope X70, as well as the electron microscope SEM JEOL JSM-6490 LV equipped with the energy dispersion analyser OXFORD INCA Energy 350.

![Fig. 1](image)

**Fig. 1.** Relation among strength and reduction of area a) material A, b) material B

3. RESULTS AND ANALYSIS

Summarised results of the obtained yield strengths and strengths on temperatures are shown in Fig. 1. In the temperature interval of 600–1000°C the material B showed 1.7-1.1 times higher strengths levels than the
material A. With higher temperature the situation changed and the differences were not higher than 1.2 times. The strengths levels of A and B could be practically taken for the same at high temperatures. Yield strengths of B were 1.4-1.3 times higher in comparison with the A. Maximal differences between strengths and yield strengths were 55 MPa (B) and 23 MPa (A). Figure 1b summarises results of the RA for both tested materials and the above mentioned temperature range. It can be seen that A generally shows the higher RA level. As the Fig. 1a, b show two principal troughs were registered in the curves shape of both materials. The first fall of the RA values is between 600/650-750°C for both curves and Figs. 2a, 2b representative micrographs show.

Fig. 2. Micrographs of material A (a) and B (b) at 700°C

Fig. 3. Micrographs of material B a) at 800°C b) at 950°C

In case of the A the second RA decrease is situated between 750-1200°C (Fig. 1a and 2a) whereas the trough of B is narrower and it is lying in the range of 750-1000°C (Figs. 1b, 2b and 3b). The 3rd RA decrease of the B is situated between 1000 and 1150°C. In the range of 1200-1350°C the A shows similar narrow and deep RA fall (Fig. 1a). Generally, the down trough levels of the A are situated at the higher RA positions and they are also wider, especially at high temperatures. In case of the A the lowest RA level reached 35% and 52% (for individual above mentioned troughs). For the B those minimal troughs levels corresponded to 15 and 3%. At 1100°C and 1300°C (A and B) deep narrow depressions were registered reaching approx. 30% of RA.
In the frame of metallographic evaluation the attention was paid to cleanliness. In both materials complex oxides and intra-, as well as intergranularly precipitated Ti(CN) particles were mostly observed. Precipitation of intergranular fine Nb(CN), Ti(CN) or TiNb(CN) cannot be also ruled out [10, 17]. However, the TEM was not carried out. In the A and B sulphides and/or oxisulphides (grade 0.5-2), sporadically aluminates with silicates (max. grade of 0.5) besides the TiN and/or Ti(CN) complexes were revealed. Microstructures of both studied materials showed different features, according to the testing temperature (Fig. 2–4). At lower temperatures noticeable ferrite discontinuous or continuous decorations of primary austenitic grains were observed, especially in the B, as it Fig. 2b shows. The A (Fig. 2a) often showed intragranular acicular ferrite (AF) morphology. Coarser, locally intergranularly formed non-metallic particles were revealed in ferrite matrix. Those were also observed on the primary austenite grains as can be seen in Figs. 2b, 3b, 4b and/or in the ends of cracks (Fig. 4). Using the EDA the particles corresponded to Ti(Nb)N and MnSiO$_3$.

Differences between the A and B are in Ti additions and N, Cu, Ni and Sn contents (Tab. 1). The B has 2.5 times higher Ti addition, while in the A the content of N is 1.35 times higher and level of (Cu + Ni + Sn) is 1.56 times higher. For the best ductility the nitrogen levels should be low in order to limit the amount of precipitation. In low nitrogen steels, the Ti/N ratio should be high 4-5:1 in order to encourage growth of the TiN particles [15]. Luo [18] reached the best ductility of steel with medium Ti content (approx. 0.06 wt. %) and Ti/N ratio corresponding to 8.9. In this case the steel contained more than 1 wt. % of Cr making the hardness of ferrite higher. Banks et al. [19] reached not so deep RA fall, even when the trough was wider, with Ti/N rate 7.5 than in case of rate 1.81. In the second case the lower RA level could be ascribe to 1.5 times higher nitrogen content. The A presented by us had the mentioned ratio 2.8 and the B 9.6. Regarding the [Nb],[N], those corresponded to 407.10$^{-6}$ (A) and to 297.10$^{-6}$ (B). The both materials also showed 0.136 wt. % Mo difference Table 1. The Mo impact on ferrite strengthening is also evident. Higher volume fraction of N leads to shift of ductility curve to lower positions and imperceptibly makes the trough brighter [15], which is obvious in Fig. 1a,b. Because of the same Nb, B level in both materials the higher nitrogen content of the A could be bound by Al, even when we were not able to detect the AlN particles on the boundary of austenite grains, where those should have precipitated preferentially and support the lower RA in the range of 600-900°C [10]. In both materials the Al contents were practically the same, too. Some aluminates (max. grade of 0.5) were detected only in the B.
Regarding the higher Ni and Cu contents in the A and higher Sn content in the B (Table 1), those also support the shift of the RA to the lower level and the ductility curves show broader trough reaching the lower temperature ranges [7, 14]. It can be also supposed that the given negative influences between the A and B were balanced. The same could be supposed in the case of C, S and P contents (Table 1). Otherwise, the RA minimum is shifted to the lower temperature with the increase of C and S content. The higher P level leads to higher susceptibility to embrittlement at the temperatures higher than 900°C [10, 11], even when in connection with Nb the P can occupy the intergranular positions and thereby limit the influence of harmful Nb(CN) precipitation, which causes the RA reduction in austenite region [9, 10, 13]. In the B the negative influence of Nb(CN) could support more intensive precipitation of Ti(CN). The elements, such as C, Mn, Al, Nb, V, B and Cr, which have stronger impact on the RA level [9, 10, 12, 20], were practically on the same level in both materials. Generally, the A showed better RA parameters than the B. The reason can be the higher Mo content. In material A the Mo level was 6.7 times higher than in the B. The presence of Mo has practically no negative influence on the RA, but the Mo decreases the Ar<sub>3</sub> temperature [10]. However, the differences in the calculated critical temperatures were minimal thanks to the influence of other elements (Table 1). The presence of Mo can secure higher hardening capacity of ferrite and it is also able to make the grain size finer as it can be seen in Fig. 2. In the A the primary austenite grain size was finer (boundaries represented greater surface) than in the B (Fig. 2). Microstructure of the B showed more noticeable ferritic decoration of primary austenite grains, consequently important differences in plasticity of intragranular and intergranular regions, because a relationship 1:5 [4] exists between the yield strength of ferrite and austenite (Rp<sub>F</sub> / Rp<sub>A</sub>). The Mo makes the ferrite regions harder and thereby decreases the above mentioned inhomogeneities, so that during deformation the plasticity is not so concentrated into the softer ferrite regions and plasticity is more balanced [4, 21].

The concept of a lower Rp<sub>F</sub> (the BCC matrix) than the Rp<sub>A</sub> level (the FCC matrix) at temperatures of 750-1050°C was advanced by Chimani and Mörwald [4]. They presented the rate Rp<sub>F</sub>(BCC) : Rp<sub>A</sub>(FCC)≈1:5 without any explanation. It could be accepted that in the BCC a dislocation bond with atoms exists (e.g. C, N) given by energy of their bond. At higher temperature, thanks to the thermally activated atom movements (facilitated), the mentioned bond is decreasing and consequently the bcc matrix becomes softer. On the contrary, in FCC matrix, with respect to the lattice properties, the bond type like in the bcc matrix including dipole induction (Fe atom-C atom-Fe atom) does not exist. Chemical bond asserts itself. In the fcc matrix the dislocations are dissociated into partial dislocations and hexagonal arrangement is formed in the stacking fault. That fault can be enriched by interstitial elements, it will be stabilised and the dislocation mobility does not decreases as much as in the BCC matrix.

4. CONCLUSIONS

Two low carbon micro-alloyed steels with different Mo (0.16-0.024 wt.%), Ti (0.021-0.053 wt%) and N (0.0055-0.0074 wt.%) contents showed various hot ductility response (RA). In temperature range of 600-1350°C, the material with higher Mo content (A) showed minimal RA of 35%, whereas the material B (lower Mo content) reached 3% RA level. The A showed the critical fall of the RA at 700°C, 1000°C and 1300°C. With exception of the lowest temperature, in the B the similar RA falls were every shifted by 200°C to the left (see Fig. 1) what can be ascribed to lower Mo content.

The 1<sup>st</sup> fall was connected with higher volume fraction of allotriomorphic ferrite (the B), while the acicular ferrite appearance in the A was reason of more favourable RA response together with the higher Mo content.
decreasing the plasticity differences between austenite and ferrite. Differences in the 2nd falls can be connected with existence of Ti(Nb)CN and MnSiO$_3$ particles on the austenite grain boundaries together with the Mo effect mentioned above. The reason of the 3rd RA falls can be elucidated by existence of MnSiO$_3$ particles first of all and sporadically appearance of Ti(Nb)N, especially at the 1100°C.

In the A the Mo content contributed to ferrite strengthening and to total more homogeneous plasticity unlike the material B, where the strengthening level reached differences between intragranular and intergranular regions. Differences of plasticity between ferrite and austenite were elucidated on the basis of dislocations interactions of the bcc and/or of the fcc structure with interstitial elements.

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