EFFECT OF STRAIN RATE ON HOT DUCTILITY OF LOW CARBON STEEL

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Abstract
Tensile strength and reduction of area for low carbon steel (LC) were measured. The experiment was performed in the high temperature region of stabile austenite at strain rates of 1 x 10^{-3} s^{-1} and 1 x10^{-2} s^{-1} usinga thermo-mechanical simulator GLEEBLE 1500. Tensile tests were carried out at test temperatures of 1000 - 1300 °C after heating to the solution temperature of 1350 °C. The change of strain rate did not influence the strength values more significantly; however, it had effect on high temperature plastic properties at temperatures under solidification and also around 1100°C.The low ductility at 1100°C for the higher strain rates was caused probably by deformation induced precipitation of Al based particles along the prior austenite grain boundaries. It led to intergranular fracture, brought into effect by cleavage mechanism. The ratio of Mn/S is an important factor in LC steel, influencing the hot ductility. At low rates of Mn/S ≤ 20, there is a high probability of embrittlement in the stable austenite region. However, the higher ratio of Mn/S and low sulphur content, as it was in the tested steel, cannot warrant sufficient hot ductility in the stabile austenite region, and this way it excludes cracking during continuous casting. The plasticity also depends on the strain rate used. The low strain rate of 1 x 10^{-3} s^{-1} can eliminate the embrittlement in the tested low carbon steel with Al and N_2 content of 0.049 % and 0.0068 %, respectively.

Keywords: low carbon steel, strain rate, hot ductility, intergranular fracture

1 Introduction
High temperature properties of steel – strength and reduction of area are important characteristics, which have decisive influence on the formation of defects during continuous casting, thermo-mechanical and heat treatment, and welding [1-4]. As reported [5-8], in the high temperature region in low carbon and microalloyed steel the existence of 3 regions of brittleness is known:

- Region I. – near to the solidification temperature
- Region II. – in stabile austenite high temperature region(900 °C – 1200 °C)
- Region III. – in phase transformation temperature region γ → α, and α → γ (III.A – the lower temperature region of the austenite, III.B the γ → α transformation region [7,8]

Hot ductility is influenced by a large number of factors: chemical composition of the steel, microstructure and macrostructure, content of harmful additives, inclusions, thermal treatment history and selected technologic factors [1, 5, 9-11]. Temperature and strain rate belong to the more important factors.

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The influence of strain rate was intensively studied just in the Region III. [2, 3, 12-16]. It was shown in [12, 13], that the ductility is increased in this region III by the increase of strain rate to the maximum around 10^{-1} to 100 s^{-1} and then it is decreasing. Strain rate increase up to values present at straightening operations (10^{-3} – 10^{-4} s^{-1}) can improve the hot ductility by the following considerations:

- suppression of less wanted preferential deformations in grain boundaries, or in their vicinity: slip in grain boundaries is suppressed and the time is too short to form the diffusion controlled growth of voids around the precipitates located in grain boundaries.
- forming of deformation induced ferrite is inhibited
- time is too short to form carbon nitride precipitates (Nb-V) and strain induced precipitation.

By the increase of strain rate the hot ductility in the transformation region $\gamma - \alpha$ is improved, also according to Ouchi [8]. It is explained by the suppression of strain concentration in ferrite during deformation, resulting in homogeneous deformation in both phases γ and α. The minimal hot ductility can increase considerably due to coarsening of AlN and/or NbC particles in the grain boundaries [3].

The beneficial influence of strain rate increase from 10^{-3} to 10^{-1} s^{-1} in the lower temperature zone (800, 900 °C) for Nb and NbV microalloyed steel was reported also in [14]. The loss of ductility of the Nb microalloyed steel at low strain rate is caused by dynamic induced precipitation of Nb(CN) particles, respectively, both within γ grains and on the γ grain boundaries. In NbV microalloyed steel the ductility deterioration is contributed also to dynamic precipitation of V(CN) in the temperature range from 800 to 1000°C. In contrast, for high strain rate > 10^{-1} s^{-1} the dynamic precipitation is less effective and static precipitation is formed mainly. From the point of plasticity deterioration, the static precipitation is less effective than the dynamic one, because the particles are coarser. Increase of strain rate from 10^{-3} to 10^{-1} s^{-1} accelerates the dynamic recrystallization, due to denser dislocation sub microstructure, but the critical deformation necessary for the dynamic recrystallization is growing too, this way there are multiple crack formation possibilities [14]. The critical strain rate for low carbon steel is according to [17] in the range from 2.5 x 10^{-3} to 1 x 10^{-2} s^{-1} and for microalloyed steel it is to 2.5 x10^{-3} s^{-1}. Authors in [14] confirmed the quite distinct dependence of hot ductility on the thermal history of the steel, particularly for low strain rate. Tensile tests performed on in situ solidified specimens showed up to 20 % worse ductility defined by reduction of area, compared to similar tensile tests on reheated specimes.

For steel C-Mn-Nb-Al the minimum hot ductility was found at 900°C, a temperature slightly higher than Ar_3 [2]. At this temperature there is a small amount of ferrite around the austenite and a large number of NbCN particles, which were precipitated in grain boundaries. With the decrease of strain rate the hot ductility decreased as well, because the mechanism of voids growth prevailed over the mechanism of void nucleation. This allows enough time for cracks to grow causing the final failure. NbCN precipitation in the grain boundaries is the source of hot ductility deterioration according to [15], too. The authors call attention to the fact that precipitation in the matrix is supposed to strengthen the deformation along the grain boundaries. Slower strain rate decreased the mobility of grain boundaries by the fine deformation induced NbCN precipitates, which are pinning the grain boundaries, lowering this way the ductility [15].

For carbon steel the dependence of hot ductility on the Al and N content [18], or Al/N ratio [19] was confirmed. The increase of Al or N concentration led to hot ductility deterioration, by

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widening and deepening of the minimum region in the phase transformation temperature region $\gamma \rightarrow \alpha$ (III) [18]. The temperature of the maximum precipitation rate for AlN is at 815°C [20] and for NbCN it is at 950 °C [21]. For ultra low carbon steel with the austenite region in annealed condition is the tip of the curve for AlN at about 1050 °C [22]. This reference supposed precipitation of AlN first of all in the grain boundaries and the precipitation on dislocations is supposed at temperatures fewer than 900 °C. For steel with 0.055%Al and 0.024%N the tip of the curve was calculated at 1150°C [23]. In works [1, 15] it has been established, that the precipitation of AlN from austenite is a very slow and difficult process.
It was shown that for carbon steel and high strain rate > 10^{-1} s^{-1} there was no dynamic precipitation before recrystallization [24]. If the deformation temperature is higher than the recrystallization temperature ($T>T_R$), then recrystallization takes place in austenite and precipitation during recrystallization is blocked, due to the decrease of strain induced driving force. If $T<T_R$, precipitation goes on before the start of recrystallization and the recrystallization is inhibited by AlN precipitation. For a rolled low carbon steel strip (0.067 % C, 0.0338 % Al, 0.0047 % N) in work [25], it was confirmed after the first rolling reduction (at about 1100 °C) for high strain rate, that there were Al_2O_3 and MnS particles only and AlN particles were not observed. It is explained by the fact that at temperatures higher than the recrystallization temperature, during rolling precipitation of AlN is not possible, due to the high strain rate and recrystallization, but precipitation takes place immediately after rolling.
In the stable austenite region (temperature region II) on the other side a number of references showed an opposite effect of strain rate on hot ductility [5, 7, 26]. The beneficial effect of the decrease of strain rate on the increase of the hot ductility was confirmed. It is due to coarsening of precipitates in the grain boundaries and this way the sensitivity to intergranular cracking is decreased [7]. However, at very low strain rates, the atoms of additives can segregate from austenite into the matrix-precipitate boundaries, into austenite grain boundaries, and into the free surfaces of micro defects, causing the decrease of cohesion strength of interface and grain boundaries [27]. In [26] Suzuki has shown, at high temperature 1000 °C in the stabile austenite region for carbon steel 0.7%C, that the hot ductility depends on strain rate in a way that the plasticity is the highest at the lowest strain rate about 10^{-3} s^{-1} and then it is decreasing with the increase of strain rate, but a small improvement in the range from 1 to 10 s^{-1}.
In studies of different authors less attention is paid to the influence of strain rate on the loss hot ductility in region II., where stabile austenite is found. The aim of our study was this region of loss ductility. We studied influences of different strain rates on hot ductility for low carbon steel.

2 Experimental material and methods

Low carbon steel with low alloying elements content was tested in the experiment. The chemical composition of the tested steel was the following (melt analysis weight %): 0.0376 % C, 0.249 % Mn, 0.011 % Si, 0.012 % P, 0.0072 % S, 0.049 % Al, 0.001% Ti, 0.002% Nb, 0.001 % V, 0.001% As, 0.004% Sn, 0.007%Sb, 0.0068 % N_2, 0.027 % Cu.
Test pieces of Ø10 mm in diameter and 110 mm long for high temperature tensile testing were machined from the rolled product 35 mm thick. For experiments the thermo-mechanical simulator GLEEBLE 1500 was used, at temperatures 1300 °C, 1250 °C, 1200 °C, 1150 °C, 1100 °C, 1000 °C and two deformation rates 1 x 10^{-3} s^{-1} (0.03 $mm.s^{-1}$) and 1 x 10^{-2} s^{-1}(0.3 $mm.s^{-1}$). Heating to the solution temperature 1350 °C / 1 min and cooling to the test temperature were made with heating rate 10°C s^{-1} and same cooling rate. Simulator records and broken test pieces

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were analyzed and the values of UTS(ultimate tensile strength) and reduction of area RA were determined. Fracture surfaces were analyzed by macro-scope Leica WILD M 32 and by scanning electron microscope (SEM) JEOL JSM 7000 with EDX analyzer INCAs- sight. Microstructure of tested material and the fracture lines were studied in light microscope OLYMPUS TH4-200.

3 Results and discussion

The tensile test results are summarized in **Fig. 1** and **Fig. 2**. In **Fig. 1** it can be seen, the strength is decreasing with the increase of test temperature near to proportionally, from the value about 55 MPa at 1000 °C, to about 25 MPa at 1300 °C, at the same time, the influence of strain rate on strength is not significant.

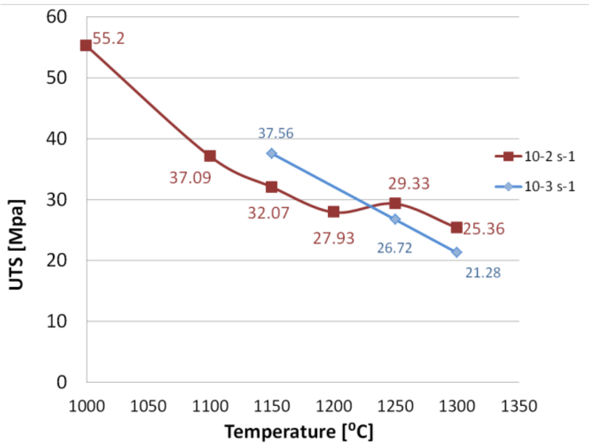

Fig. 1 Dependence of low carbon steel strength (Rm) on deformation temperature for 2 strain rates

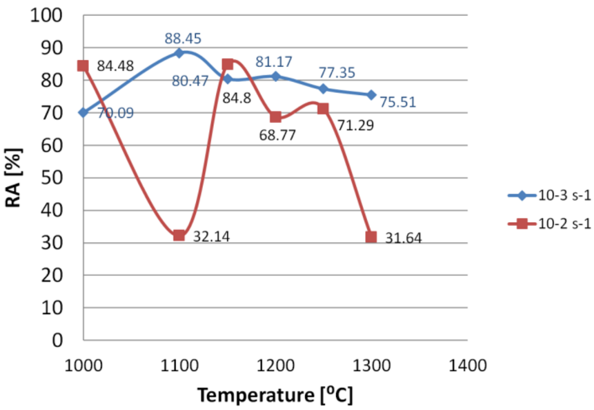

Fig. 2 Dependence of reduction of area (RA) on deformation temperature for 2 strain rates

Due to an error in recorder operation the strength values at 1000 °C and 1100 °C and deformation rateat 1 x 10^{-3} s^{-1} are missing. The influence of the deformation rate on the

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reduction of area was considerably more intense than on strength, as it can be seen significantly for temperatures 1100 °C and 1300 °C, Fig. 2. While for higher strain rate 10^{-2} s^{-1} minima of hot ductility were observed at 1100°C and 1300 °C, for the lower strain rate the ductility improved nearly in the all tested temperature range. The two decreases of reductions of area for the higher strain rate were strong. A 50 % decrease for temperature increase from 1000 °C to 1100 °C and a similar result about 40 % RA decrease for temperature increase from 1250 °C to 1300°C were recorded. The minimum at 1100°C is actually the embrittlement in II. temperature brittle region with stabile austenite.

The decrease of RA at 1300 °C can be supposed to be the transition to the region I. of hot ductility loss, which is usually at about 1350 °C. Steel with approximately similar chemical composition was studied in work [26]. The temperature of austenitizing was the same as in our experiment, the strain rate was higher – 1.2 $mm.s^{-1}$. Comparison to our results showed a good agreement in the strength values. On the other side, there was a difference in the temperature of the region I. hot ductility loss about 50 °C. In both experiments a region II. ductility minimum was discovered, in our work at 1100 °C , in work [26] at 1200 °C and in work [9] at 1200°C, too, for strain rate 0.47 x 10^{-3} s^{-1}. For theoretical and practical aspects it is important to explain the causality of the low RA at 1100 °C.

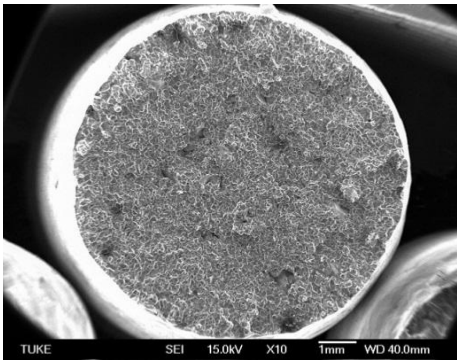

Fig. 3 Macro view of brittle fracture from ductility loss at 1 100°C and strain rate 1 x10^{-2} s^{-1}

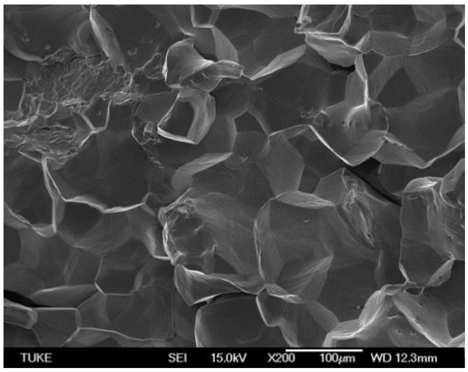

Fig. 4 Intergranularfracturefrom ductility loss at 1 100°C and strain rate 1 x 10^{-2} s^{-1}

Macro view of fracture of specimen No. 9 tested at 1100 °C is documented in **Fig. 3**. The fracture is fine grained without signs of macroscopic deformation. High magnification of the fracture surface in **Fig. 4** made it visible that the fracture is intergranular, brought into effect by cleavage mechanism. The size of intergranular facets was about 100 μm. Intergranular facets showed a secondary relief, the so called thermal faceting (**Fig.5**), generated by vacuum etching at a sufficiently high temperature and also contour line relief, exhibited typically by the stripes in the intergranular cleavage facets. **Fig. 5** and **Fig. 6** prove the next important fact: in the surfaces of intergranular facets, there were oval and stick like particles less than 50nm in size, as confirmed by EDX analysis containing Al (**Fig.7**), some of them oxygen, (**Fig.8**), too. Particles based on Al precipitated in the prime austenite grain boundaries and contributed to the development of the intergranular fracture. EDX analysis confirmed aluminum and oxygen in some particles, so it can be supposed they were Al_2O_3 particles. Al_2O_3 particles were found in work [25] at 1100 °C, too. The other particles based on Al can be AlN, though their precipitation

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in austenite is slow and difficult [1, 15]. Next analyses can prove it e.g. by transmission electron microscopy. For the deformation rate 1 x10^{-3} s^{-1} and temperature 1100 °C the fracture character was quite different, than it was for strain rate 1 x 10^{-2} s^{-1}.Fracture in **Fig. 9** showed high reduction of area and in **Fig.10** ductile character and morphology with dimples.

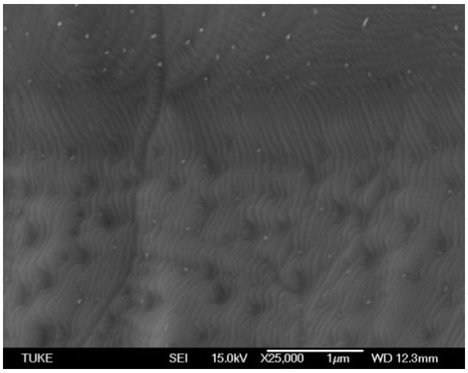

Fig. 5 Relief and thermal faceting and particles in the intergranularfacet at 1100°C and strain rate 1 x 10^{-2} s^{-1}

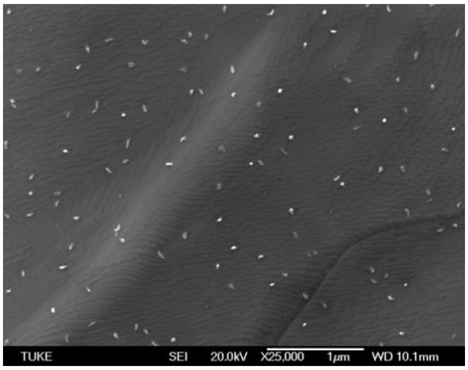

Fig. 6 Particles in the intergranularfacet surface at 1100°C and strain rate 1 x 10^{-2} s^{-1}

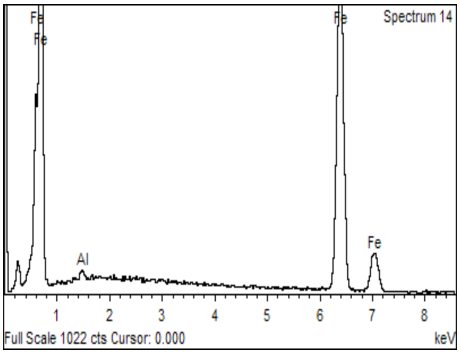

Fig. 7 EDX analysisspectrum of particle in Fig.6.

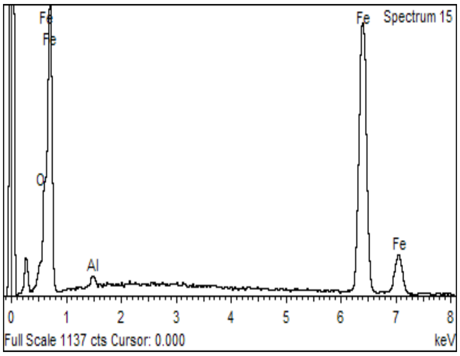

Fig. 8 EDX analysisspectrum of other particle in Fig.6.

Intergranular fracture of a tensile test piece in the thermo-mechanic simulator GLEEBLE 3500 at temperature 1000°C was obtained by the authors of work [25]. The tested LC steel contained 0.055 % C and the Mn/S ratio was Mn/S = 19. By EPMA (Electron probe micro-analyzer) analysis they showed MnS particles in the austenite grain boundaries, which initiated the intergranular fracture. They confirmed the fact, that in LC steel the ratio Mn/S has a distinctive influence on hot ductility. For low values of Mn/S < 20 the hot ductility is decreasing and the probability of cracking during continuous casting is growing. However, our experiments support the conclusion, that the high value of ratio Mn/S > 20 and a low sulphur content in LC steel cannot guarantee the high hot ductility in the stabile austenite region for the given steel composition of 0.049% Al and 0.0068% N_2. The result will depend on strain rate applied. As has been shown the strain rate 1 x 10^{-2} s^{-1} was for the tested low carbon steel with low micro-alloys

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critical from the point of embrittlement and intergranular cracking initiation in agreement with the results in references [14, 17].

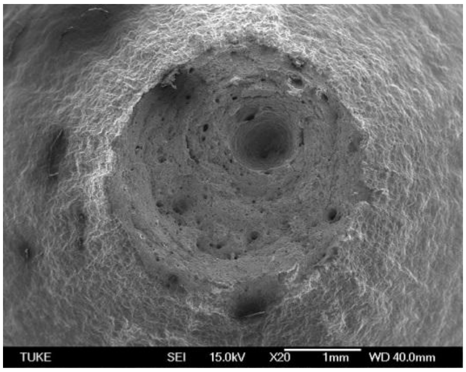

Fig. 9 Ductile fracture at 1 100°C and strain rate 1 x 10^{-3} s^{-1}

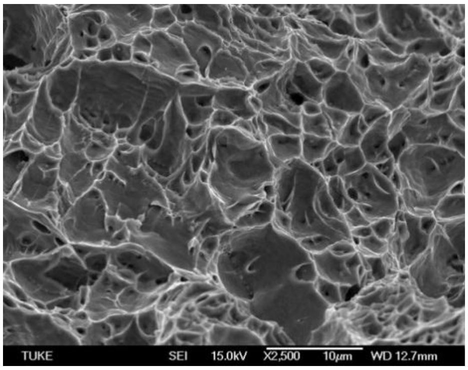

Fig. 10 Dimples in a transcrystalline ductile fracture surface at 1 100°C and lower strain rate 1 x 10^{-3} s^{-1}

4 Conclusions

The influence of strain rate on hot ductility for LC steel was studied in this work. Based on experimental results and discussion the following conclusions can be made:

1. By the increase of the deformation rate from 1 x 10^{-3} s^{-1} to 1 x 10^{-2} s^{-1} the strength in the high temperature region of stabile austenite has not changed significantly, on the other side the strain rate increase influenced the hot ductility near to the solidification temperature and in the temperature region at 1100 °C.
2. Hot ductility loss in the temperature region at 1100 °C is caused for the higher strain rate 10^{-2} s^{-1}, probably by the deformation induced precipitation of Al based particles in the prime austenite grain boundaries. This way intergranular fracture can be brought into effect by cleavage mechanism. The embrittlement can be eliminating by a lower strain rate 1 x 10^{-3} s^{-1} for the given composition of 0.049 % Al and 0.0068 % N_2 in the tested low carbon steel.

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