HOT DUCTILITY BEHAVIOR OF A CONTINUOUSLY CAST Ti-Nb MICROALLOYED STEEL

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Abstract

The hot ductility behaviour of a Ti-Nb micro alloyed steel was investigated to evaluate the probability of surface crack formation during the continuous casting process, by performing hot tensile tests. The testing temperatures are ranging from 1100 °C to 700 °C. The effects of testing temperature range and deformation rate on hot ductility were investigated. The results show that this steel exhibits poor ductility over almost the whole testing temperature range. The ductility starts to decrease at 1000 °C in the single phase γ-region, characterized by grain boundary sliding and surface cracks, reaches a minimum in the two-phase α-γ-region at 750 °C and slightly increases with decreasing testing temperature. Furthermore, low deformation rates severely decrease the ductility behaviour. Microstructural examinations and supplementary thermo-kinetic computer simulations revealed distinct Ti-Nb precipitation throughout the microstructure being responsible for the deteriorated materials hot ductility.

Keywords: Metallurgy, steel, continuous casting, hot ductility, low alloyed, testing methods

1. INTRODUCTION

In the straightening operation of the continuous casting process, high mechanical and thermal stresses are affecting the steel slab. Theses stresses may cause internal tears and surface cracks that lead to a loss of productivity and quality in the steelmaking process [1-3]. The sensitivity to transverse cracking of continuously cast steels can be primarily adressed to the poor ductility behaviour at a temperature range of 700 °C - 1000 °C. The straitening operation takes place at these temperatures [4]. Poor ductility behaviour is caused by the strain concentraitions at the ferrite films along the austenite grain boundaries and precipitations. Ferrite, which is the lower strengh phase, prefers to form at the austenite grain boundaries and absorbs the majority of deformation. If the potential of ferrite to accommodate the strain is exeeded, poor ductulity values occur [5]. In the austenitic high tempartuer region, a high amount of fine precipitates are the main damaging mechanism. These precipitates can lead to local hardening and severe stress concentrations, which can trigger the nucleation of wege-type cavities that lead to interconnections of cavities surounding the precipitates [6].

The hot ductility can be evaluated by measuring the percentage reduction of area (RA) after fracture in a hot tensile test of samples, by using similar thermomechanical conditions [7,8]. To avoid transverse cracks, the RA value should exceed the limit of 40 % [9]. To apply the thermomechanical conditions of continuous casting, in-situ melted and solidified specimens are used in the hot tensile test. This approach allows the dissolution and nucleation of precipitation, such as (Ti,Nb)(C,N) and MnS, which weaken the grain boundaries and considers the segregation of alloying elements [10,11].

In this work, the hot ductility of a micro alloyed Ti-Nb steel was investigated in a temperature range of 600 °C to 1200 °C to predict the temperatures at which the hot ductility behaviour deteriorates.

2. EXPERIMENTAL

2.1. Chemical composition

The chemical composition of the investigated steel is given in **Table 1**. To obtain phase transformation temperatures, thermodynamic equilibrium calculations have been performed using the thermokinetic software

MatCalc (version 6.02) [12]. The ferrite start temperature (A_{e3}) was found to be 837 °C and the ferrite finish temperature (A_{e1}) 674 °C.

Table 1 Chemical composition of the investigated steel in wt%

C	Cr	Mn	Nb	N	Ni	P	S	Ti	Al	Fe
0.081	0.037	1.43	0.046	43 ppm	0.02	0.0096	0.0033	0.066	0.041	bal.

2.2. Hot tensile tests

An in-house developed vertical thermomechanical simulator BETA 250-5 (**Figure 1a**) was used to conduct hot tensile tests. The experiments were performed in a vacuum atmosphere with a pressure of 0.2 mbar. Cylindrical tensile specimen (**Figure 1b**) were machined from the continuously cast and hot rolled billed with their axis parallel to the rolling direction. The upper end of the specimen was clamped to the tension arm which is responsible for the displacement in upwards direction. The lower end was clamped to a special extractor, which has a fixed position and is equipped with a spring and three gripper arms, which have three functions: they hold the sample when pulling it upwards, they compensate for the thermal expansion of the specimen and support it during the melting cycle of the temperature curve, so it doesn´t break in this unstable experimental phase. A Pt/Pt-Rh thermocouple was spot-welded to the body center of the specimen to measure the surface temperature. An inductive heating system was used to perform the temperature cycle. The induction coil is physically linked with the tension arm and moves with half of the speed of the upwards displacement to ensure, that with ongoing deformation, the induction coil system stays positioned at the center of the specimen.

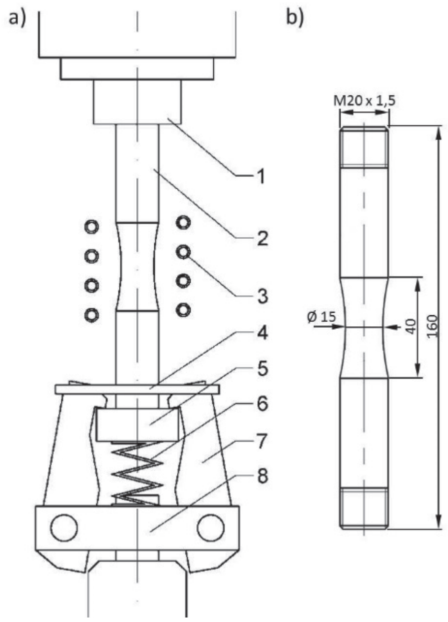

Figure 1 Tensile testing equipment “BETA 250-5” (1. tension arm, 2. tensile specimen, 3. induction coil, 4. securing ring, 5. holding ring, 6. steel spring, 7. gripper arm, 8. extractor unit); b) geometry and dimension of the tensile specimen in mm [13]

The temperature cycle for the experiment is shown in **Figure 2**. The measured surface temperature during the melting phase is 1445 °C, which was held for 60 seconds. The melting phase is followed by a first cooling phase with 5 °C/s till 1250 °C and a second cooling phase with 1 °C/s till the testing temperatures, which range from 1100 °C to 600 °C. As a standard the tensile tests were performed at a strain rate of 1 x 10^{-3}/s. For each testing temperature, a total number of 3 samples were tested. To evaluate the hot ductility, the reduction of area was measured graphically by a stereo microscope. Metallographic analyses were carried out on longitudinal cross sections of ruptured specimen near the fracture surface using light optical microscopy (LOM) and scanning electron microscopy (SEM).

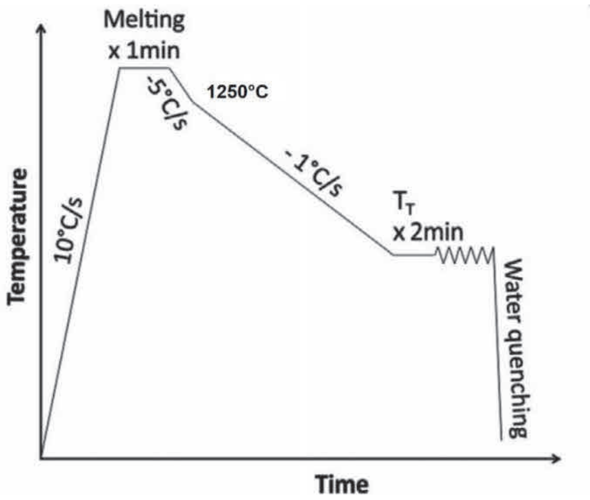

Figure 2 Temperature cycle for the experiment

3. RESULTS

3.1. Hot ductility behaviour

The hot ductility curve represented by the reduction of area as a function of the deformation temperature for the in-situ melted steel is shown in **Figure 3**. The ductility minimum is at 750 °C but the critical proposed RA-value of 40 % by Mintz is exceeded at a temperature range from 700 °C to 900 °C. This critical value is displayed as a horizontal dashed line in **Figure 3**. The ductility starts decreasing in the pure austenitic region with decreasing temperature, however below 700 °C the ductility recovers.

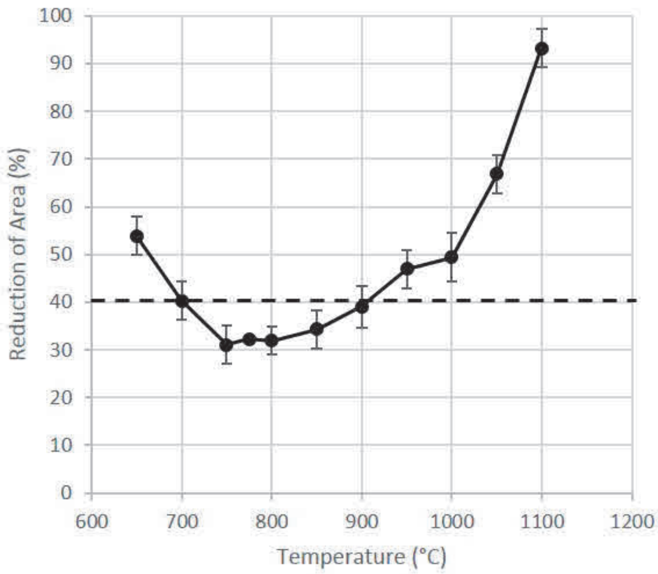

Figure 3 Hot ductility curve

3.2. Metallography

LOM images of water quenched specimens after pulling to fracture show a large number of cracks and pores, especially at temperatures between 750 °C to 1000 °C [15]. In specimens pulled at 900 °C and 800 °C, the pores, which form preferably at the grain boundaries at the former austenite grains, are already interconnected and form cracks along the grain boundaries. In the specimen pulled at 775 °C these cracks are elongated over multiple grain boundaries and a small amount of ferrite appears at the former austenite grain boundaries. At a deformation temperature of 700 °C however, we see the formation of ferrite and a smaller number of cracks and pores. Furthermore, primary precipitates are visible in all specimens. These precipitates are in the size of 1-10 µm and have been identified as primary (Ti,Nb)(C,N) precipitates and MnS precipitates [15].

3.3. Simulation of secondary precipitation

Figure 4 shows the results of the precipitation kinetics simulation, which include the mean particle radii, the phase fraction and the number densities of (Ti,Nb)(C,N) and MnS precipitates. Additionally, AlN is included in the simulation, but none is formed according to it. The initial temperature is 1471 °C to simulate the continuous casting process. The subsequent heat treatment is the same as shown in **Figure 2**. The deformation temperature of the shown simulation is 800 °C, where there is a very poor ductility behaviour. The simulated strain rate is 1 x 10^{-3}/s, in accordance with the experiment. The initial grain size of the austenitic grain is set to 500 µm and the nucleation sites are set to grain boundaries, dislocations and additionally (Ti,Nb)(C,N) can nucleate on MnS precipitates. (Ti,Nb)(C,N) starts nucleating at 1360 °C mainly at dislocations and MnS at 1105 °C at the grain boundaries. At a running time of 505 s, the deformation starts severely increasing the number densities and phase fractions of the precipitations. Shortly after the deformation starts, a massive drop in the mean radius can be observed. This is due to the severe increase of new nucleated precipitates which can be seen in the increase in number density. After a few seconds, number density and mean radius are staying constant and so is the phase fraction.

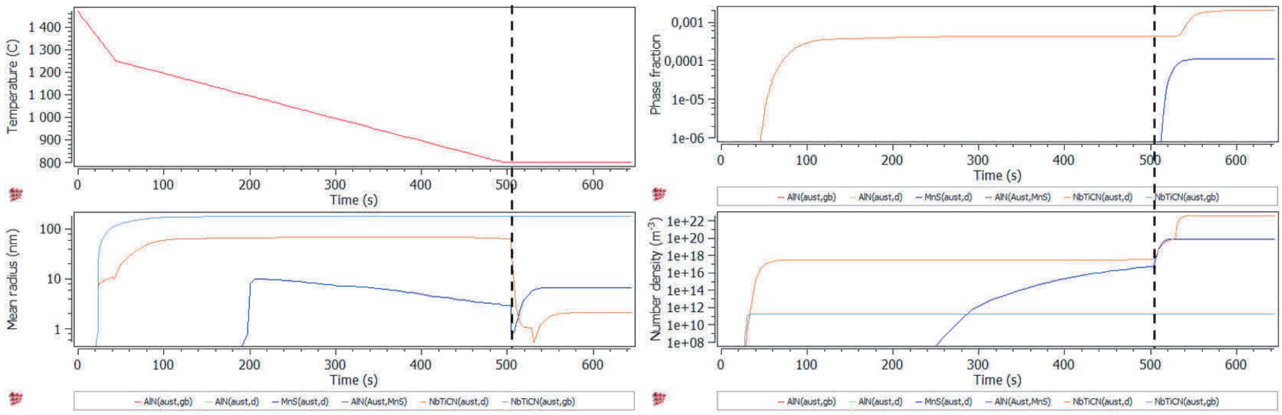

Figure 4 Calculated evolution of mean particle radius, phase fraction and number density of (Ti,Nb)(C,N) and MnS precipitation during the cooling phase and the deformation temperature of 800 °C

4. DISCUSSION

In the literature, two major mechanisms are described where the precipitates reduce the hot ductility in the austenitic region. Precipitates on grain boundaries favour the nucleation and interconnection of cavities around them. At low strain rates, the nucleation, growth and coalescence of voids around second phase precipitates lead to a poor ductility behaviour [16,17].

The second mechanism describes the presence of fine precipitates at dislocations, which increases the strength of the grains due to dislocation pinning. Hence grain boundary sliding is hindered. Cavities form by the migration of vacant lattice sites into the triple point region of grain boundaries, where high stress concentrations occur. Wedge cracks are the result of accelerated linking of growing cavities leading to a poor ductility behaviour [18,19].

Both effects are causing the ductility drop in the austenitic temperature region. Lückl et al. [20] reported a similar behaviour in their experimental studies with steel grads with lower Ti-content compared to our investigated chemical composition.

The increase in ductility at lower temperatures can be attributed to the formation of sufficient ferrite and the suppression of the strain localization effect [9].

5. CONCLUSION

The hot ductility of an in-situ melted and solidified micro alloyed Ti-Nb steel was investigated. The region of poor ductility was found to be between 700 °C and 900 °C, with a minimum at 750 °C. This behaviour is the result of a high density of fine secondary precipitates (MnS) at the grain boundaries and in the bulk at dislocations ((Ti,Nb)(C,N)) of the material. The recovery of the ductility at testing temperatures of 700 °C and below occurs due to sufficiently formed ferrite preventing intergranular stress concentrations.

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