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# Hot ductility of medium carbon steel with vanadium

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ABSTRACT

Hot ductility of medium carbon steel containing 0.52 wt% of carbon and 0.11 wt% of vanadium was investigated using a hot tensile test performed up to fracture. The hot ductility was evaluated by measuring the reduction of area of the fractured specimens, which were strained at a variety of test temperatures in a range of 600–1100 °C at a strain rate of $2\times10^{-3}$/s. The hot ductility was excellent in a temperature range of 950–1100 °C, followed by a decrease of the hot ductility below 950 °C. The hot ductility continued to drop as the temperature was lowered to 600 °C. The loss of hot ductility in a temperature range of 800–950 °C, which is above the $Ae_3$ temperature, was due to V(C,N) precipitation at austenite grain boundaries. The further decline of hot ductility between 700 °C and 750 °C resulted from the transformation of ferrite films decorating austenite grain boundaries. The hot ductility continued to decrease at 650 °C or less, owing to ferrite films and the pearlite matrix, which is harder than ferrite. The pearlite was transformed from austenite due to relatively high carbon content.



## 1. Introduction

The continuous casting of steels has been widely studied and significant efforts have been made to solve industrial problems related to continuous casting since it was commercially applied in the 1960's on the basis of the high mass productivity and quality of steels [1,2]. Nevertheless, some industrial issues remain, including transverse cracking of the slab edge that can occur during the straightening operation in continuous casting at a temperature region associated with poor hot ductility. Such transverse cracks can be a more serious problem in steel with higher carbon content and/or microalloying elements such as Ti, Nb, V, B, etc. [1–16].

It has been generally reported that the hot ductility curves of low carbon steels yielded by hot tensile testing, as shown in Fig. 1, show three distinct regions: (i) a high ductility high temperature region, (ii) a deep ductility trough – embrittlement region, and (iii) a high ductility low temperature region. The high ductility high temperature region has excellent hot ductility due to low concentrations of flow stress and strain resulting from dynamic recrystallization during the tensile test. Meanwhile, a deep ductility trough, that is, embrittlement region, appears around the $Ae_3$ temperature where austenite transformed to ferrite. In this region, the formation and coalescence of voids occurs preferentially in ferrite films during the tensile test, leading to intergranular facture. This is possibly attributable to localized deformation of ferrite films, which are relatively soft compared to the austenite matrix. In the high ductility low temperature region where the fraction of ferrite that is transformed from austenites is higher, stress/strain concentration in ferrite regions is no longer effective and hence hot ductility can be recovered in a lower temperature range. Many researchers have studied hot ductility of microalloyed steels containing Ti, Nb, and V [1–10]. In the case of steels containing Ti, Nb, and V, the ductility trough may deepen and broaden toward higher temperatures above the $Ae_3$ temperature due to MX precipitates such as Ti(C,N), Nb(C,N), and V(C,N) at austenite grain boundaries; these precipitates are frequently accompanied by the formation of weak precipitate free zones adjacent to austenite grain boundaries or by the prevention of dynamic recrystallization, resulting in intergranular failure. In particular, Cho et al. [10–13] paid much attention to boron containing steels and have shown that precipitation of BN at austenite grain boundaries deteriorate the hot ductility, based on the experimental results on the effect of various metallurgical factors such as cooling rate, thermal cycle, alloying elements including N, Nb, and Ti on the hot ductility of boron-bearing steels.

Research on hot ductility has been widely performed in relation to carbon steels and/or microalloyed steels [1–19]. However, most studies have been limited to low carbon steels below 0.35 wt% of carbon and there have been few studies on medium or high

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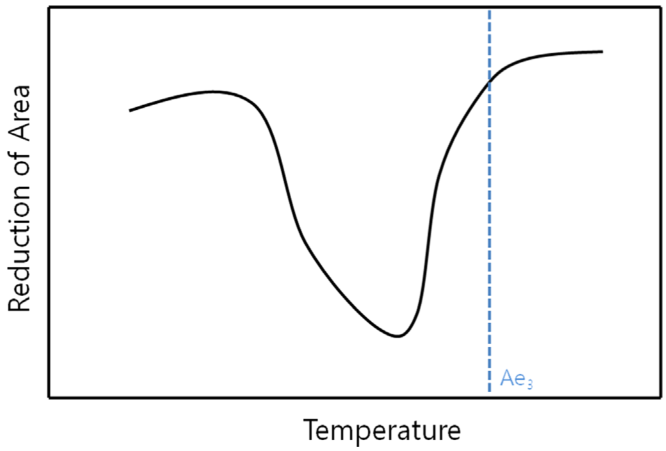


**Fig. 1.** General hot ductility behavior of low carbon steels.

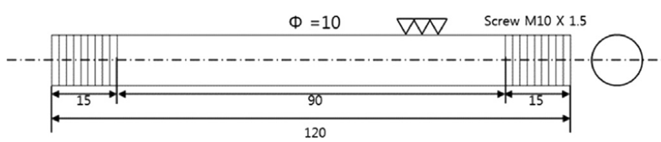


**Fig. 2.** Dimensions of tensile specimens.

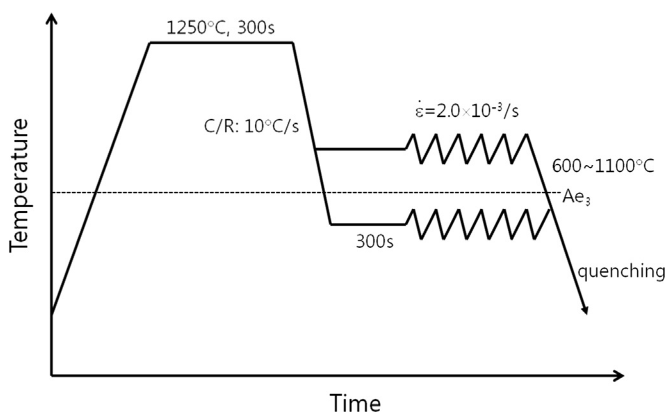


**Fig. 3.** Schematic diagram of thermo-mechanical conditions for hot ductility test.

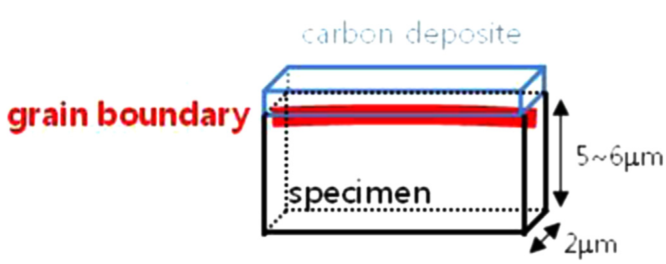


**Fig. 4.** Schematic diagram of sample for TEM extracted by FIB.

carbon steels. The purpose of the present work is to understand the hot ductility of medium carbon steels with 0.5 wt% of carbon and 0.1 wt% of vanadium, as well as to optimize continuous casting conditions for these steels based on the understanding of the hot ductility.

## 2. Experimental

The steel composition used in this experiments is Fe–0.52C–0.26Si–1.0Mn–1.1Cr–0.11V–0.019Al (wt%) which is similar to the specifications of DN-51CrV4 steel. The phosphorus and sulfur

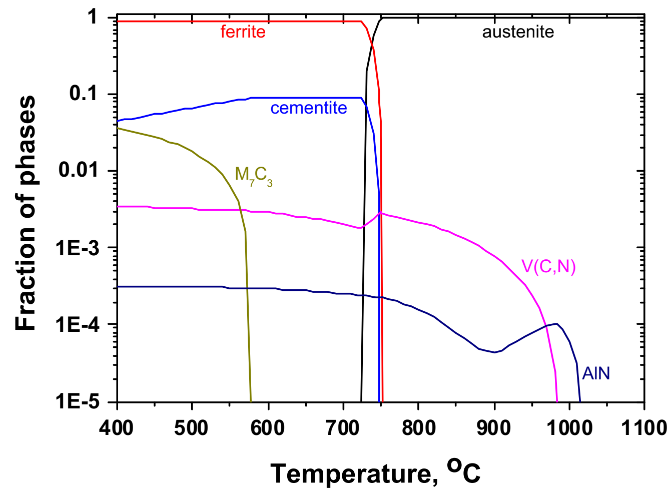


**Fig. 5.** Thermodynamics calculation of the steel used in the present study using Thermo-Calc 3.0 with TCFE7 database.

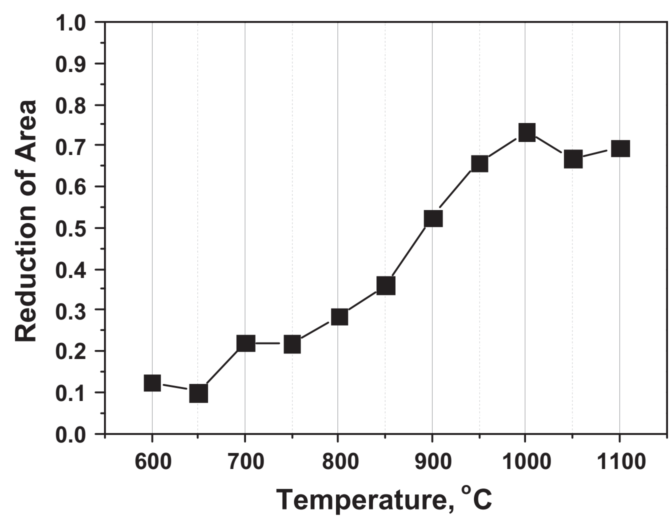


**Fig. 6.** Hot ductility behavior of the steel used in the present study.

content are 0.00141 and 0.0017 wt%, respectively, and the oxygen and nitrogen concentrations are negligible below 50 ppm. The sample was cut from a slab produced in a continuous casting mill. Cylindrical tensile specimens for hot ductility were machined from the sample in Fig. 2. Hot tensile tests were carried out using a Gleeble 3800 machine (thermo-mechanical simulator) which can control the temperature of the specimens and the uniaxial load and strain. Hot ductility was evaluated by measuring the reduction of area of the tensile specimens strained to failure at various temperatures. This method is widely applied to assess slab surface cracking during continuous casting [3–14]. Fig. 3 illustrates the thermo-mechanical cycles used in the experiments. Specimens were heated to 1250 °C and held for 300 seconds, and then cooled to various test temperatures in a range of 600–1100 °C at a cooling rate of 10 °C/s. It is expected that total vanadium could be soluble during the heat treatment at 1250 °C for 300 seconds [20,21]. The specimens were held for 300 seconds and strained to failure at a strain rate of $2.0 \times 10^{-3}$/s, and then quenched to room temperature. These thermo-mechanical conditions are known to represent the straightening operation of continuous casting process [1–8].

The specimens were characterized using optical microscopy (OM), scanning electron microscopy (SEM, JSM-7001F, JEOL), and transmission electron microscopy (TEM, JEM-2100F, JEOL) as well as the electron backscatter diffraction (EBSD) technique.