Effects of tungsten and molybdenum on high-temperature tensile properties of five heat-resistant austenitic stainless steels

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ABSTRACT

Heat-resistant austenitic stainless steels have been intensively used worldwide for turbo-chargers requiring excellent high-temperature properties to sustain their structures at very high exhaust gas temperatures. Five heat-resistant austenitic stainless steels were fabricated by varying contents of W and Mo, and their high-temperature tensile properties were investigated by analyzing phases (liquid, austenite, ferrite, and carbides) existing at high temperatures. Effects of replacement of expensive alloying element, W, by Mo or reduction in W were also examined. The increase in contents of W and Mo resulted in linearly increased volume fractions of M_7C_3 and total carbides, while the volume fraction of MC carbide hardly showed any relation, and showed a good correspondence with high-temperature yield and tensile strengths. The steel where 2 wt% of W was replaced by 2 wt% of Mo showed the best 900 °C-tensile properties, thereby confirming the successful achievement of partly replacement of W by Mo. In addition, the low-W-containing steels showed excellent 900 °C-tensile properties, which also indicated that the reduction in 1–2 wt% of W was accepted for saving costs of alloying elements.

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1. Introduction

In order to increase fuel efficiency and decrease CO_2 emission, improvement of automotive engine performance as well as reduction in vehicle's weight have been focused in research areas of automotive steels [1–8]. For achieving these purposes, automotive parts such as turbo-chargers requiring excellent high-temperature properties to sustain their structures at very high exhaust gas temperatures up to 950 °C have been actively developed [9–11]. Heat-resistant austenitic stainless steels have been intensively used worldwide for turbo-chargers because they satisfy strict requirements and demands of high economy, environmental friendliness, and high performance. They also have advantages of excellent high-temperature hardness, strength, and thermal fatigue life over conventional stainless steels [9–16].

Since they usually contain 16–25 wt% of Cr and 7–20 wt% of Ni, together with typical carbide formers such as W and Nb, they basically have coarse carbides distributed in the austenite matrix formed during the casting [17–25]. Their high-temperature properties are greatly influenced by the kind, size, volume fraction, and

distribution of very hard and thermally stable carbides. However, most of these carbides exist along solidification cell boundaries, which may cause a serious problem in deteriorating high-temperature properties and thermal fatigue life because intercellular carbides act as major fracture initiation sites [20,26,27]. Thus, understanding of cast microstructures and their effect on high-temperature properties are essentially needed for cast steel design of high-performance high-temperature turbo-chargers. In addition, studies on how the type, volume fraction, and distribution of carbides and resultant mechanical properties are affected by the addition of alloying elements should be systematically examined. In particular, W is a very expensive element used for solid-solution hardening of austenite matrix and formation of M_7C_3-type carbides, but its detailed quantitative effects in austenitic stainless steels have not sufficiently known yet [28,29].

In this study, five heat-resistant austenitic stainless steels were designed in a way how the formation of carbides was optimized by adjusting W content and by replacing a part of W by Mo in order to achieve excellent high-temperature tensile properties, and their room- and high-temperature tensile properties were evaluated. The correlation between high-temperature tensile properties and microstructural factors such as austenite, ferrite, and carbides was investigated by actually measured fractions of microstructural factors together with thermodynamically calculated fractions.

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2. Experimental

Five austenitic steel ingots were fabricated in a Y-block shape (length: 180 mm, width: 50 mm, height: 110 mm) by a conventional sand casting route at Castec Korea, Busan, Korea, and were used for the present experiments after checking the homogeneity of internal and external cast structures. Their chemical compositions are shown in Table 1. Since nitrogen was not intentionally added in austenitic steel ingots, a very small amount of nitrogen which might be present inside cast ingots was neglected. The '3W' steel has a basic composition of (0.4–0.5)C–(1.2–1.5)Si–(1.1–1.3)Mn–(9.5–10.5)Ni–(21.0–22.0)Cr–(1.4–1.6)Nb (wt%), along with about 3 wt% of W which is an effective element in solid solution hardening of austenite matrix and formation of carbides [28,29]. 1–2 wt% of W was reduced from the 3W steel to fabricate the '1W' and '2W' steels in order to investigate the W reducing effect. Also, 2 wt% of W was replaced by 1–2 wt% of Mo to fabricate '1W1Mo' and '1W2Mo' steels, as designated in Table 1. Mo is cheaper than W, and has a strengthening effect of austenite matrix [30,31]. Considering the tungsten equivalent (W_{eq}=W+2Mo), the W_{eq} was similarly maintained in the 1W1Mo steel, while it was over-flown in the 1W2Mo steel. The samples were obtained from the interior of as-cast ingots.

The cast ingot specimens were polished and etched in a 10% HCl solution (10 ml HCl+90 ml ethanol), and their micro-structures were observed by an optical microscope and a scanning electron microscope (SEM, model; S-4300E, Hitach, Tokyo, Japan). Phases were identified by X-ray diffraction (XRD, Cu Kα radiation, scan rate; 2° min^{-1}, scan step size; 0.02°), and their volume fractions were measured by an image analyzer (model; SigmaScan Pro ver. 4.0, Jandel Scientific Co., Erkrath, Germany). Electron back-scatter diffraction (EBSD) analysis (step size; 0.8 μm) was performed by an EBSD camera system (model; OIM-4000, EDAX Inc., USA) attached in a field emission scanning electron microscope (FE-SEM, model; Quanta 3D FEG, FEI Company, USA), and the data were interpreted by an orientation imaging microscopy (OIM) analysis software provided by TexSEM Laboratories, Inc.

Round tensile specimens having gage diameter of 6 mm and gage length of 25 mm were prepared, and were tested 3 times at least for each condition at room temperature and 900 °C at a cross-head speed of 1.5 mm/min by a universal testing machine of 100 kN capacity (model; 8862, Instron, Canton, MA, USA). A cylindrical-type high-temperature chamber (diameter; 340 mm, height; 190 mm) was attached to the test machine in the case of the high-temperature tensile test. The 0.2% offset stress was determined to be the yield strength. The cross-sectional areas beneath the fracture surface were observed by an SEM to observe the fracture behavior.

3. Results

3.1. Fractions of high-temperature equilibrium phases

Fractions of high-temperature equilibrium phases were investigated by thermodynamic calculations using a commercial thermodynamic calculating program, ThermoCalc [32] and an upgraded version of TCFE2000 [33,34]. Fractions of equilibrium phases such as liquid, austenite, ferrite, MC carbide, and M_7C_3 carbide in the temperature range of 400–1400 °C are shown in Fig. 1(a)–(e). In the 1W steel, the austenite and MC carbide start to form simultaneously from the liquid at 1350–1355 °C, and their fractions continuously increase as the fraction of liquid decreases with decreasing temperature to 1255 °C (Fig. 1(a)). In the as-cast microstructures, the rest of liquid is retained at inter-dendritic regions, and then M_7C_3 carbides are formed by reactions between three solid phases (austenite, ferrite, and MC carbide) and residual liquid phase. Here, the equilibrium phase diagram (Fig. 1(a)) cannot be applied to phase analyses of cast microstructures, in which the complete diffusion cannot occur in solid phases because of dendritic solidification and rapid cooling. As a result, the assumption that diffusion or mixing does not occur in solid phases during the non-equilibrium solidification is needed. For example, the solid phases existed at 1200 °C are assumed to be homogeneous in the equilibrium state, like in Fig. 1(a), but phase fractions in the solidified microstructure cannot be determined by equilibrium phase diagrams. This is attributed to the existence of liquid in which C is segregated at inter-dendritic regions by solid/liquid partitioning. Accordingly, below the liquid dissolution temperature, solid phases formed by the C-segregated liquid should be thermodynamically recalculated.

The fraction of residual liquid ($f_{L\_residue}$) at the inter-dendritic region is marked by a blue-dotted circle in Fig. 1(a), from which fractions of equilibrium phases are recalculated, as shown in Fig. 2 (a). At 1255 °C, the austenite, ferrite, and M_7C_3 carbide start to form simultaneously, and the liquid is completely consumed at 1217 °C. Thus, phases formed before the liquid dissolution (right side of the temperature of $f_{L\_residue}$ in Fig. 1(a)) exist inside solidification cells, and phases formed after the liquid dissolution (left side of the temperature of $f_{L\_residue}$ in Fig. 2(a)) exist along cell boundaries. By using this two-step calculation method, fractions of equilibrium phases of the other four austenitic cast steels are estimated, as shown in Fig. 1(b)–(e) and Fig. 2(b)–(e). The overall phase formation behavior in the other steels is almost same to that of the 1W steel. The ferrite and MC carbide start to form at 1310–1345 °C from the liquid, while the austenite forms at 1333–1355 °C, and the fraction of ferrite before the liquid dissolution ($f_{L\to\alpha}$) as well as its formation temperature increase as W and Mo contents increase (red arrow marks in Fig. 1(b)–(e)). Particularly in the 1W2Mo steel whose W_{eq}(=W+2Mo) is highest among the five steels, the fraction and formation temperature of ferrite are highest.

When assuming that the diffusion between solid phases does not occur after the formation of solid phases from the liquid in consideration of rapid cooling during the casting, fractions of phases formed from the primary liquid (fractions at black-dotted lines of Fig. 1(a)–(e)) and the residual liquid (fractions at black-dotted lines of Fig. 2(a)–(e)) are shown in Table 2. In the 1W steel, for example, fractions of austenite, ferrite, MC carbide, and M_7C_3 carbide at the black-dotted line of Fig. 1(a) are combined with fractions of

Table 1
Chemical compositions of the five austenitic stainless steels (wt%).

Steel	C	Si	Mn	P	S	Ni	Cr	Nb	W	Mo	W_{eq}[a]
1W	0.4–0.5	1.2–1.5	1.1–1.3	~0.03	~0.20	9.5–10.5	21.0–22.0	1.4–1.6	1.08	–	1.08
2W	0.4–0.5	1.2–1.5	1.1–1.3	~0.03	~0.20	9.5–10.5	21.0–22.0	1.4–1.6	2.04	–	2.04
3W	0.4–0.5	1.2–1.5	1.1–1.3	~0.03	~0.20	9.5–10.5	21.0–22.0	1.4–1.6	3.30	–	3.30
1W1Mo	0.4–0.5	1.2–1.5	1.1–1.3	~0.03	~0.20	9.5–10.5	21.0–22.0	1.4–1.6	1.08	1.18	3.44
1W2Mo	0.4–0.5	1.2–1.5	1.1–1.3	~0.03	~0.20	9.5–10.5	21.0–22.0	1.4–1.6	1.11	2.28	5.67

[a] W_{eq}=W+2Mo (wt%).