1.2 ADVANCED BAINITIC STEEL DEVELOPMENT FOR FUSION STRUCTURAL APPLICATIONS—Y. Yamamoto (Oak Ridge National Laboratory)

OBJECTIVE

This task aims to develop new bainitic steel, based on 3Cr-3WV(Ta) steels originally developed at ORNL, with mechanical properties of both base metal and weldments superior to those of existing commercial bainitic steels or ferritic-martensitic (F-M) steels, together with no requirement for post-weld heat treatment (PWHT). The target applications are high temperature structural components in fusion reactors such as vacuum vessel, structural ring which supports the blanket modules, and magnet shields, to be used at or above the 400-500ºC range. Improvement of long-term creep properties by introducing additional fine, stable second-phase dispersions, as well as maintaining good weldability, is targeted via optimization of alloy composition and thermo-mechanical heat treatment.

SUMMARY

A new alloy design strategy for PWHT-free bainitic steel has been proposed; reducing the as-normalized hardness without losing high hardenability, in order to reduce potential property inhomogeneity across the weldment. Four new heats of 3Cr-3WVTa base bainitic steels with low carbon combined with high Mn and Si were prepared with guidance from computational thermodynamics. The steel with the combination of 0.05C and 2 or 3Mn additions successfully showed fully bainitic microstructure together with less hardness difference between as-normalized and tempered conditions, which will reduce the potential property inhomogeneity in the weldments. High temperature tensile properties also supported potential improvement of mechanical properties at elevated temperatures. Preparation for creep-rupture testing is currently underway.

PROGRESS AND STATUS

Introduction

Development of new bainitic steels was initiated as part of the Fusion Energy Materials Program in FY2014, as a modification of the original 3Cr-3WV(Ta) steels developed at ORNL[1,2]. The alloy design strategy is to improve creep performance through minor alloying additions of Mn, Si, and N ("Gen I" steels). Computational thermodynamics tools were used to predict optimized minor alloying additions to promote the formation of stable MN (M: mainly V) in a wide temperature range below ~1100ºC and increase the stability of $M_{23}C_6$ (M: mainly Cr) relative to M_7C_3 (M: mainly Cr). Improved oxidation resistance was also expected through the Si addition [3]. The calculated continuous-cooling-transformation (CCT) diagram predicted a wider austenite stable region (retarding the transformation to ferrite or pearlite, $\mathbf{F}_s$ or $\mathbf{P}_s$) with lower bainitic transformation start temperature ($\mathbf{B}_s$) compared to the base steels, indicating that the transformation would occur in a relatively lower temperature range even with air cooling, equivalent to the lower bainite formation during isothermal annealing in the TTT diagram (see **Fig. 1**). The experimentally measured CCT diagrams of the developed steels exhibited lower $\mathbf{B}_s$ temperature, and improvement in the creep performance of the steels was observed, as expected, especially in relatively lower creep-stress testing [4]. However, it was also found that the hardness of the as-normalized material was extremely high compared to the original 3Cr-3WV(Ta) steels, whereas the hardness after tempering was smaller than the original, indicating that the variation of the properties across the welded material (in as-weld condition) could be significant compared to the original 3Cr-3WVa(Ta) steels. This suggests that the alloy design strategy of the Gen I steels failed to meet the PWHT-free requirement.

Based on the results of Gen I steels, it was suggested that reduced hardness in the normalized conditions could be a key for the PWHT-free material design strategy, since it could potentially reduce the cross-weld property inhomogeneity compared to that of the base steels. In order to decrease the hardness in the normalized condition without losing high "hardenability" to promote the carbide-free acicular bainite

ferrite formation, a new alloy strategy with lower carbon and higher Mn, together with Si for improved oxidation resistance, was proposed for "Gen II" steels. In this report, the suitability of the new alloy strategy and selection of the alloy compositions is discussed. Initial microstructure characterization results are also shown.

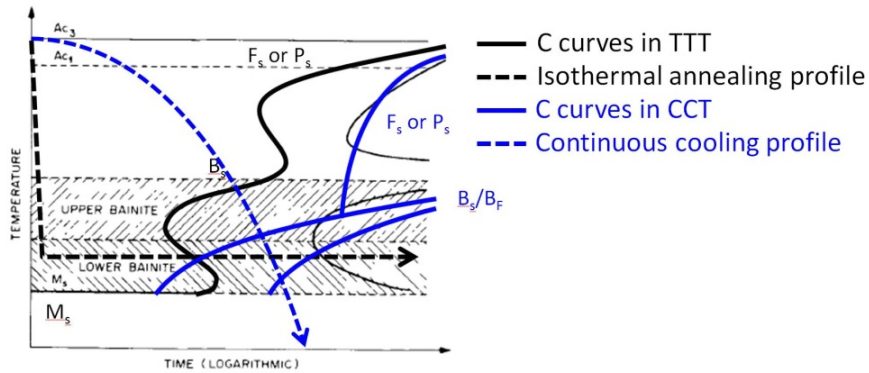

Figure 1. Schematic illustration of TTT and CCT curves for a bainitic steel

Experimental Procedure

The alloy compositions of the base and newly selected steels are listed in Table 1. The phase equilibrium and the CCT diagrams were calculated by JMatPro version 9 with Fe database. The compositions were selected based on the new alloy design strategy to produce low hardness (via 0.05C) and high hardenability (high Mn additions), as well as supplemental oxidation resistance (1Si). The ~400 g steel ingots were arc-melted and drop cast into a Cu mold with size 13 mm x 25 mm x 125 mm. The ingots were homogenized at 1200ºC for 2 h, and then water-quenched. The ingots were then hot-forged and -rolled at 1100°C to produce plate with 6 mm thickness, and then air-cooled to room temperature. Normalization at 1100°C for 30 min, followed by air-cooling, was applied to all plates. The plates were sectioned, tempering at 700°C for 1 h, followed by air cooling. Cross-sectional microstructure characterization used optical metallography, OM, and the Vickers hardness was measured with 300 g load. Tensile tests were conducted in a temperature range from room temperature to 600°C in laboratory air. Sub-sized sheet tensile specimens with gage section 0.76 mm thickness x 1.5 mm width x 7.6 mm length were used. The tests were performed with cross-head control at a constant strain rate of 10^{-3}/s.

Table 1. Nominal compositions in weight percent, balanced Fe

Heat	Composition, wt.%							Remarks
	C	Mn	Si	Cr	V	W	Ta	
Base	0.10	0.40	0.16	3.0	0.2	3	0.1	3WVTa
LC	0.05	0.40	0.16	3.0	0.2	3	0.1	Low C
MSLC1	0.05	1.00	0.5	3.0	0.2	3	0.1	Low C + 1Mn
MSLC2	0.05	2.00	0.5	3.0	0.2	3	0.1	Low C + 2Mn
MSLC3	0.05	3.00	0.5	3.0	0.2	3	0.1	Low C + 3Mn

Results

The calculated phase equilibrium of the base 3Cr-3WVTa steel and two new steels with lower C steel (LC), and Mn+Si+lower C steel (MSLC2) are shown in Figure 2. The three steels show similar matrices and second-phases, although there were two major changes produced by the alloy modifications: the A_{e1} temperature (austenite to ferrite transformation) and the amounts of the second-phases. The A_{e1}

temperature was slightly increased by lowering the carbon content, whereas the further addition of Mn (+Si) resulted in extending the austenite phase region compared to that of the base steel. This indicates that low C + high Mn would improve the phase stability of austenite relative to ferrite, so that the increased hardenability is expected to promote the formation of carbide-free acicular bainite ferrite. On the other hand, the amounts of the second phase carbides such as MC and M_6C (and M_7C_3) decreased significantly due to lower C addition, so that less hardness compared to the base steel is expected.

Figure 3 shows the calculated CCT curves of the base, LC, and MSLC2 steels. LC steel showed almost no changes in the $\mathbf{B}_s$ temperature compared to the base steel, and the ferrite formation ($\mathbf{F}_s$) was slightly promoted, consistent with what can be expected from the calculated phase equilibrium in Figure 2. On the other hand, the Mn addition in MSLC2 steel lowered the $\mathbf{B}_s$ temperature and retarded the $\mathbf{F}_s$ and $\mathbf{P}_s$ curves, indicating that improved hardenability is expected.

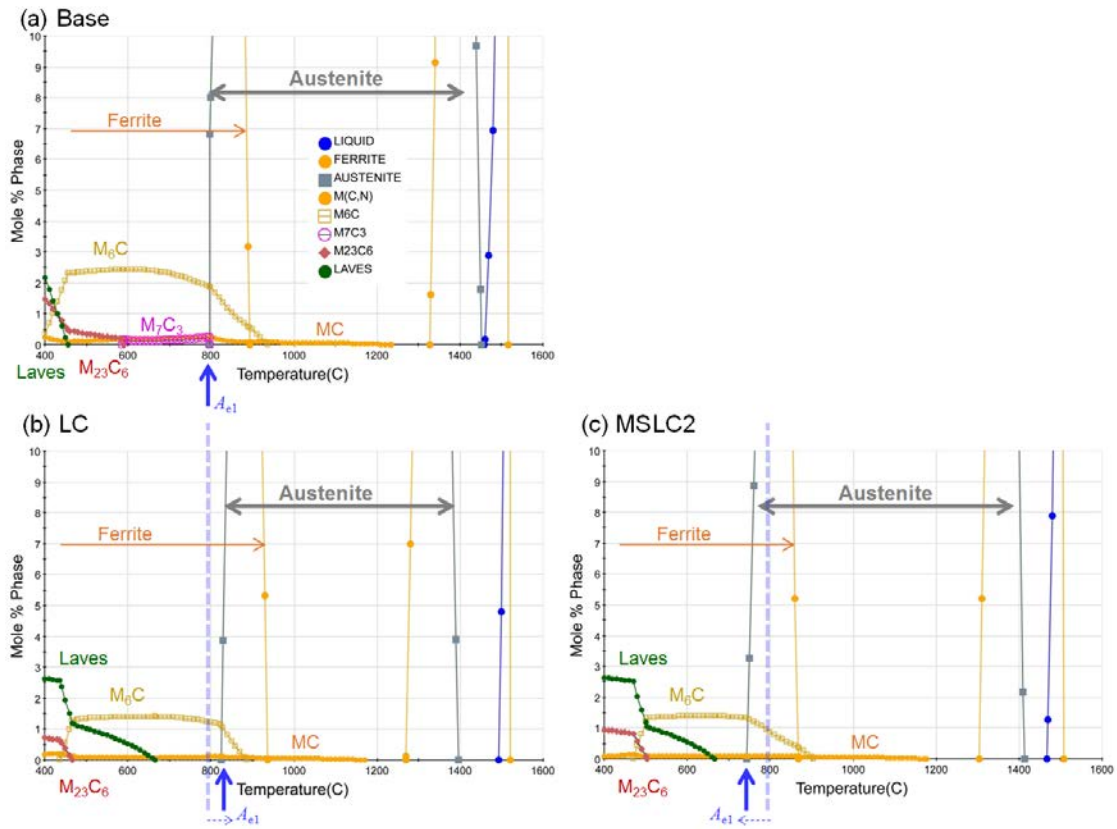

Figure 2. Calculated phase equilibrium of (a) Base, (b) LC, and (c) MSLC2 steels.

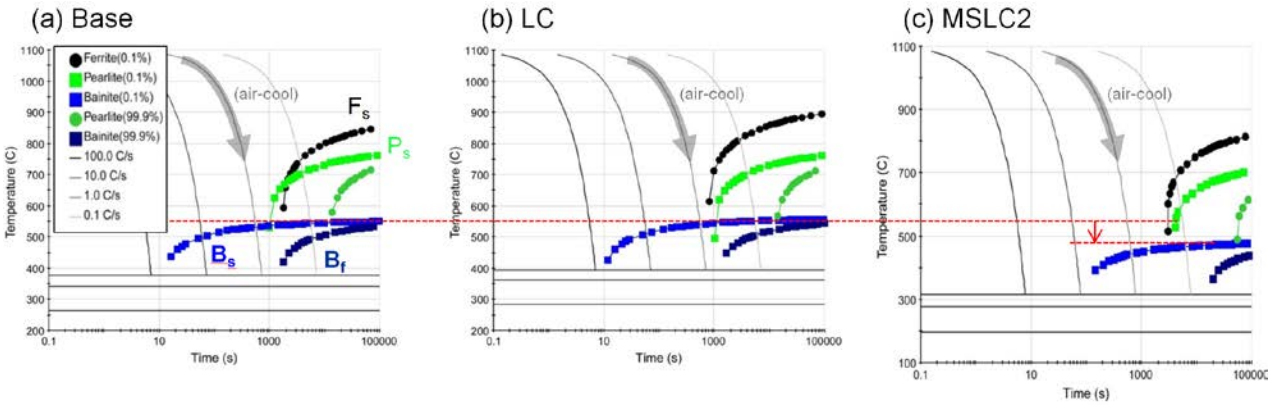

Figure 3. Calculated continuous cooling transformation diagrams of (a) Base, (b) LC, and (c) MSLC2 steels.

Optical micrographs of the as-normalized and tempered steels are shown in Figure 4. Micro-Vickers hardness of the material is also given on each micrograph. In the as-normalized condition, lowering the carbon content (LC) resulted in forming a small amount of ferrite grains and decreasing the hardness significantly. After tempering, the hardness of the base steel became ~290HV, whereas that of LC steel did not show a significant changes from as-normalized material and stayed low. On the other hand, as-normalized MSLC2 steel showed similar bainitic microstructure to the base steel, and the hardness was slightly lower than the base. The tempering resulted in similar hardness to the base steel. This result indicated that only lowering the carbon content would result in dropping the hardness at any conditions, but a careful control of the hardenability though Mn addition would allow comparable hardness of the tempered materials to that of the base steel, even with low C content.

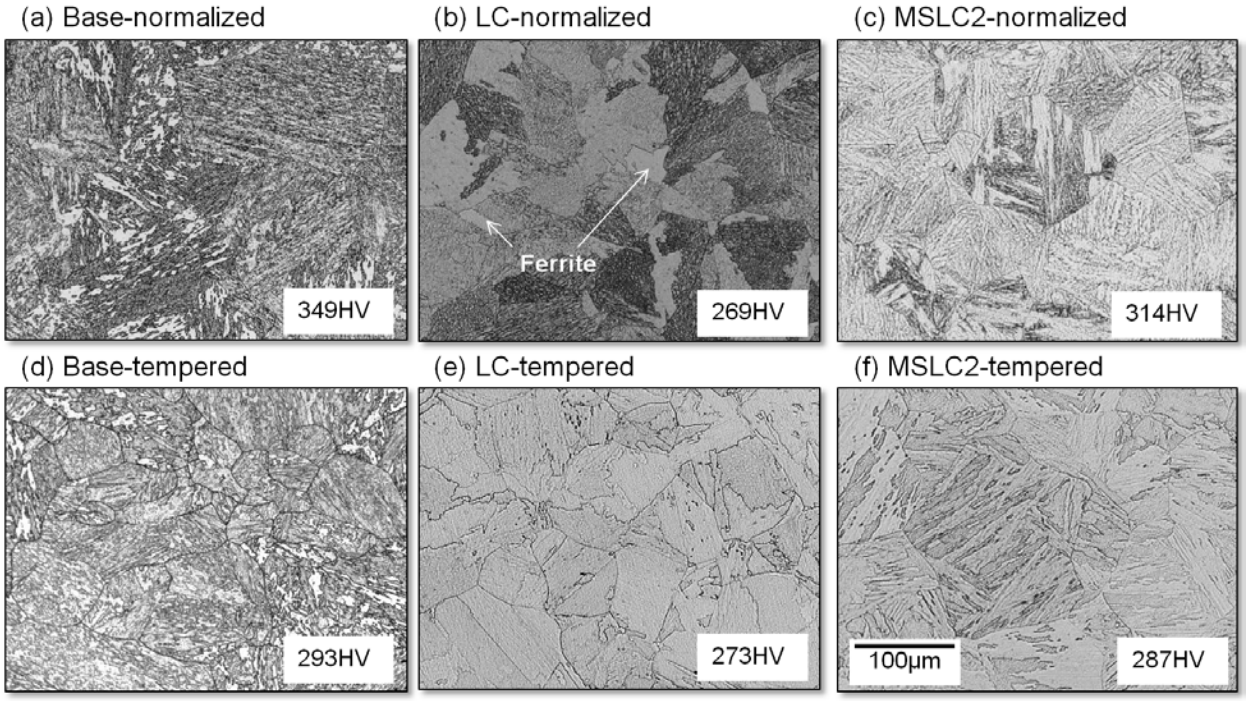

Figure 4. Optical micrographs of the 3Cr-3WVTa base steels after normalization and tempering; (a) Base-normalized, (b) LC-normalized, and (c) MSLC2-normalized, (d) Base-tempered, (e) LC-tempered, and (f) MSLC2-tempered.

Figure 5 summarizes the hardness test results of as-normalized and tempered conditions of all new steels compared with the base and Gen I steels. In order to reduce the inhomogeneous properties across the welded materials, it was considered that the smaller hardness gap between the as-normalized and tempered steels was better for PWHT-free alloy design. It should be noted that the as-normalized hardness was expected to be close to that of the as-welded fusion material since the fusion material passed through the same transformation pathway as the normalization during the weld procedure. Gen I steel showed the increased hardness in the normalized condition and the lower hardness in tempered conditions compared to those of the base steel, indicating that the hardness characteristics was opposite to the current alloy design strategy. LC and MSLC1 steels showed almost no changes in the hardness before and after tempering, which might reduce the variation of cross-weld properties. However, the lower hardness of the tempered materials compared to the base steel would result in weaker mechanical properties such as poor creep deformation resistance. High Mn additions (MSLC2 and MSLC3 steels) resulted in lowering the as-normalized hardness whereas the tempered hardness was comparable to that of the base metal.

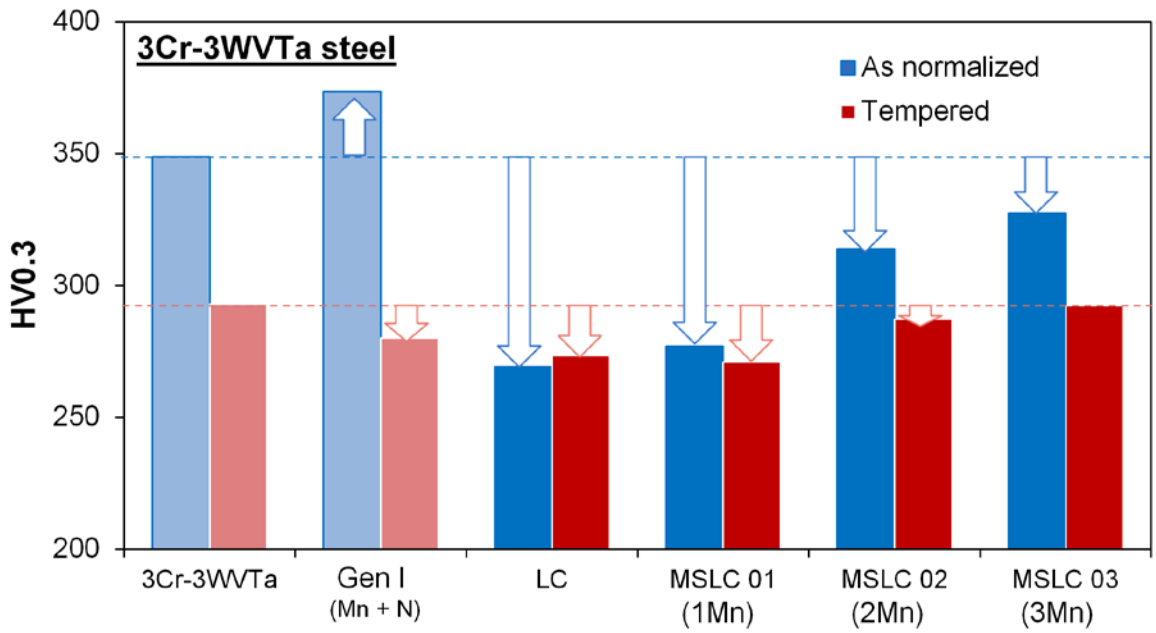

Figure 5. Micro-Vickers hardness of the bainitic steels after normalization and tempering.

Tensile properties of Gen II steels were evaluated. Figure 6 illustrates the test results of the modified steels, LC and MSLC2, as a function of test temperature. For comparison, the properties of base 3Cr-3WVTa steel reported by Jawad [5] are also shown. Although the tempering temperature of the modified steels was slightly lower than the reference material (700°C vs. 730°C) which typically made the strengths higher, the results showed obvious improvement of the yield and the ultimate tensile strengths, YS and UTS, in Mn modified alloys (MSLC2) at both RT and temperatures up to 600°C. The results indicate that the alloy design strategy for the Gen II steels (the improvement of hardenability to promote the formation of carbide-free acicular bainite ferrite) is appropriate in terms of high-temperature tensile properties. Preparation for creep-rupture testing of the materials is in progress, with testing to be initiated by the end of FY2016.

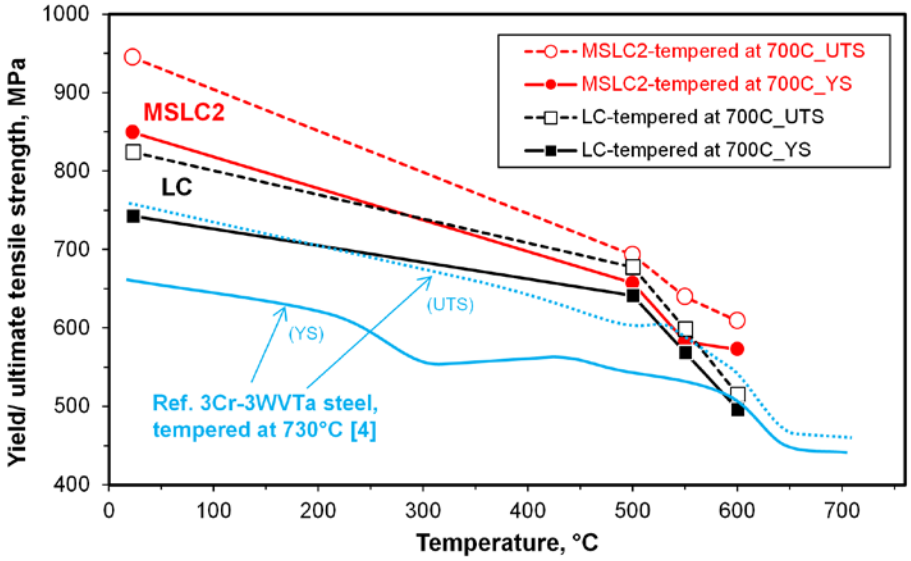

Figure 6. Tensile properties of 3Cr-3WVTa base steels (LC and MSLC2) as a function of test time, together with those of original 3Cr-3WVTa steels reported by Jawad [4].

References

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