Development of New Ferritic/Martensitic Steels for Fusion Applications

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Abstract—Martensitic steels are considered for structural applications for fusion power plants, but they are limited by strength to operating temperatures below 550–600°C. For increased plant efficiency, steels for >650°C service are sought. Based on the science of precipitate strengthening, a thermo-mechanical treatment (TMT) was developed that respectively increased strength at 700°C of commercial nitrogen-containing steels and new steels designed for the TMT to 80 and 200% greater than for commercial steels with a conventional heat treatment. Precipitates in the steels after the TMT were up to eight-times smaller at a number density four orders of magnitude greater.

Keywords—structural steels; martensitic steels; precipitate strengthening

I. INTRODUCTION

Ferritic/martensitic steels are at present the only viable structural material for fusion power plants [1]. Their major advantage is good thermal properties relative to other elevated-temperature alloys. A major shortcoming is high-temperature strength, which places a limit on the maximum service temperature capability of 550–600°C. This has led to work to develop oxide dispersion-strengthened (ODS) steels [1–3]. These steels, strengthened by small oxide particles, are produced by complicated and expensive mechanical-alloying, powder-metallurgy techniques, as opposed to conventional processing. Despite being around for almost 40 years, ODS steels are still in the development stage [2,3]. This paper presents the science and technology of the development of new and different dispersion-strengthened steels for applications at 650°C and higher using conventional processing techniques.

II. EXPERIMENTAL PROCEDURE

A thermo-mechanical treatment (TMT) was developed to produce fine dispersions of nano-sized particles in commercial Cr-MoVNbN-type high-temperature steels and in steel compositions developed especially for the TMT. The TMT involves heating the steel to convert ferrite to austenite and dissolve existing precipitates, after which the steel is cooled to a hot working temperature of 700–1000°C. Hot working introduces a high density of dislocations into the matrix. Dislocations act as nucleation sites for a fine distribution of MX (V- or Nb-rich nitride and/or carbonitride) precipitates. The hot-working procedure used involved hot rolling steel plates, after which the steel was annealed to grow precipitates to optimum size for hardening, followed by air cooling to convert the austenite matrix to martensite.

III. RESULTS AND DISCUSSION

A normalized-and-tempered (N&T) steel such as commercial modified 9Cr-1Mo (nominal composition Fe-9.0Cr-1.0Mo-0.20V-0.08Nb-0.05N-0.10C, composition in wt %) consists of martensite laths (elongated subgrains with average width ≈ 0.25–0.5 µm) with a high dislocation density \(10^{13}-10^{15} \text{ m}^{-2}\) and precipitate particles on prior-austenite grain boundaries, on lath boundaries, and in the matrix. Dominant precipitates are “large” (60–200 nm) \(M_23C_6\) particles located mainly on lath boundaries and prior-austenite grain boundaries. If vanadium and/or niobium are in the composition, then smaller (20–80 nm) MX precipitates at a low number density also form. Fig. 1(a) is a transmission electron microscope (TEM) photomicrograph of N&T modified 9Cr-1Mo steel with average particle size and number density of \(M_23C_6\) precipitates estimated at 130–150 nm and \(7-8 \times 10^{18} \text{ m}^{-3}\), respectively, and the MX is estimated at 30 nm and 7-8 \(10^{18} \text{ m}^{-3}\), respectively.

After a 25.4-mm-thick plate of modified 9Cr-1Mo steel was processed by the new TMT, the microstructure was dramatically different from N&T steel [Fig 1(b)]. The MX particle size was four-times smaller (7–8 nm), and the number density was three orders of magnitude greater (2-9 \(10^{18} \text{ m}^{-3}\)). The effect of the TMT can be controlled by changing (1) austenitization temperature and time, (2) hot-rolling temperature, (3) amount of reduction by hot-rolling, and (4) annealing temperature and time.

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A commercial 12% Cr steel HCM12A (nominal composition Fe-12.0Cr-2.0W-0.40Mo-1.0Cu-0.30Ni-0.25V-0.05Nb-0.06N-0.10C) was given two different versions of the new TMT, one rolled to 750°C and the other at 800°C. In contrast to modified 9Cr-1Mo steel, no fine precipitates were visible by TEM. However, after tempering 1 h at 750°C, fine precipitates (4.2 nm, 2.4 x 10^{21} m^{-3}) were detected by dark-field TEM in the one rolled at 800°C (Fig. 2). Evidently, the higher vanadium and nitrogen in HCM12A promoted precipitates during the TMT that were too fine to observe by TEM. During tempering, these precipitates coarsened in the steel rolled at 800°C to the point they could be observed by TEM.

Based on thermodynamic calculations, small (400-g) heats of new compositions were produced to maximize the effect of the TMT. To demonstrate the effect of nitrogen, two new steels with nominal composition Fe-9.0Cr-1.0Mo-1.0Ni-0.30V-0.07Nb-0.05C were compared, one containing 0.035% N (9Cr-MoNiVnBnN4) and the other 0.065% N (9Cr-MoNiVnBnN5). For both, average MX precipitate size was smaller and number density greater than for the commercial steels: about 4.0 nm, 1.0 x 10^{22} m^{-3} and 3.3 nm, 7.2 x 10^{22} m^{-3} for 9Cr-MoNiVnBnN4 and 9Cr-MoNiVnBnN5 (Fig. 3), respectively.

The most important properties for the new steels are elevated-temperature tensile and creep strengths. Tensile tests were conducted on modified 9Cr-1Mo steel after a TMT (hot-rolled 20% at 750°C) and after the TMT plus a 1 h temper at 750°C and compared to average values for the steel in the N&T condition. After the TMT (no temper), the 0.2% yield stress and ultimate tensile strength from room temperature to 800°C were considerably greater than those for the N&T modified 9Cr-1Mo steel, with the relative difference increasing with increasing test temperature. For tests at 600 and 700°C, the yield stress of the steel with the TMT was 61 and 88% greater than for the N&T steel. After tempering the steel with the TMT, the yield stress was about 7 and 69% greater at 600 and 700°C, respectively. Total elongation after the TMT was less than in the N&T condition, but given the normal trade off of strength and ductility expected, ductility was excellent with total elongation of 16 and 22% at 600 and 700°C, respectively, and increasing further after tempering at 750°C to 24 and 30%.

The 12% Cr steel HCM12A after the two TMTs (one hot-rolled 50% at 750°C, the other hot-rolled 50°C at 800°C) and a temper showed large increases in yield stress relative to the N&T HCMA12A: 47 and 64% increases occurred at 600 and 700°C, respectively, with little difference between the two TMTs. An increase of 22% was observed for tests at 800°C. The strength of the HCM12A with the TMT plus temper was greater than that of modified 9Cr-1Mo steel with just a TMT, a reflection of the smaller precipitates at a higher number density formed in HCM12A. The strength up to 700°C of the HCM12A was greater than PM 2000, which is the best commercial ODS steel available (Fig. 4).

To demonstrate the excellent strength properties of the new steel compositions developed to take advantage of the TMTs, the yield stress of the new 9Cr-MoNiVnBnN (0.042% N) composition was compared to an experimental ODS steel labeled 12YWT (Fe-12.0Cr-2.5W-0.4Ti-0.25Y_{2}O_{3}). The strength of this ODS steel had superior strength compared to
any available commercial ODS steel [4]. The TMT produced yield stress values comparable to those of the 12YWT up to 700°C—the highest temperature that 9Cr-MoNiVNbN was tested [Fig. 5(a)]. Total elongations for 9Cr-1MoNiVNNbN in the TMT condition and the TMT-and-tempered condition were also comparable to those of the 12YWT ODS steel [Fig. 5(b)].

The most important mechanical property for elevated-temperature steels is creep resistance. Limited creep tests have been conducted. In Fig. 6, creep curves to rupture are shown for modified 9Cr-1Mo steel after a N&T and a TMT. Rupture life for the steel with TMT was ≈80 times greater than for the N&T steel. Even with this large difference in strength, fracture ductility was excellent; total elongation was 21%. Note that modified 9Cr-1Mo steel had a lower yield stress after TMT than HCM12A and the new 9Cr-MoNiVNbN steels, and therefore these latter steels are expected to have significantly higher creep strengths.

These results indicate clearly that thermo-mechanical treatments can be devised to produce a dense dispersion of nano-scale MX precipitates in vanadium- and nitrogen-containing steels. The new TMT involves three distinct steps, each of which involves a range of experimental conditions that need to be optimized to produce the most favorable precipitate microstructure for increased elevated-temperature strength. Along with optimization of the new TMT process, steel compositions that fully exploit the TMT need to be developed further. The initial efforts on the small heats of steel developed for the new processing procedure indicated that it should be possible to develop compositions that will be significant improvements over the commercial N&T steels or these steels with the new TMT.

The high number density of precipitates produced by the new TMT using conventional processing methods is similar to the particle number densities in the best ODS steels produced by the much more complicated and expensive powder-metallurgy and mechanical-alloying procedures. Dispersion strengthening by these nitrogen-rich MX precipitates should allow such steels to be used at 650–700°C, which is a significant improvement over upper-use temperatures of 550–600°C for the elevated-temperature ferritic and martensitic steels now available.

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Figure 5. (a) Yield stress and (b) total elongation of new 9CrMoVNb steel after TMT and after TMT and temper compared to modified 9Cr-1Mo steel after a conventional normalize and temper (N&T) and a high-strength experimental ODS steel 12YWT.

Figure 6. Creep curves of modified 9Cr-1Mo steel after the conventional N&T heat treatment and the TMT tested at 138 MPa at 650°C.

REFERENCES