IMPROVING THE PERFORMANCE CREEP STRENGTH-ENHANCED FERRITIC STEELS

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ABSTRACT

This project aims to better understand how chemical composition and microstructure of CSEF steels influence mechanical properties. The approach relies on a combination of fundamental and applied studies of the effects of heat-treatment, welding, and process control on microstructural evolution and material properties. Recent testing at 650°C/70 MPa confirmed that reducing the tempering temperature from 760°C to 650°C for 9 Cr steel plate increased rupture life of cross-weld specimens from 519 h to 2,414 h. Also, specimens used for synchrotron diffraction experiments were made into test specimens. Tensile testing at 650°C showed that a Type-IV-resistant experimental 9 Cr steel had consistently higher strength than commercial Grade 92 steel. Stress rupture testing of similar specimens is underway at 650°C/70 MPa.

BACKGROUND

Creep strength-enhanced ferritic (CSEF) steels such as the 9 Cr steel, ASTM A387 Grade 91, have become the key to the realization of increasingly efficient coal-fired power plants. The ability of these high-strength materials to withstand increased temperatures and pressures and/or allow the use of decreased tube wall thicknesses, at a cost that is significantly lower than that of austenitic steels of equivalent strength, is a primary reason for recent increases in the efficiency of fossil power plants worldwide. Currently, CSEF steels are used up to approximately 600°C, and are increasingly being specified and used for superheater tubing and main steam piping in coal-fired steam boilers, as well as in heat-recovery steam generators used in combined cycle gas turbine units. Until recently, the widely-held expectation was that the capabilities of CSEF steels could be pushed to 650°C (or higher), which would allow the economic use of Ni-based alloys for the higher-temperature components in planned advanced steam cycles. If this objective were realized it could eliminate the need for austenitic steels along with the problems associated with the performance of austenitic steel-to-ferritic steel weld joints.

The performance of CSEF steels, however, does not always meet expectations, and there have been reports of numerous failures of CSEF steels after only a few years in service. This practical experience apparently results from two main causes: (1) long-term properties that are not in accord with the projections made from the measurements used to qualify the alloys; and (2) an inability to attain the alloy microstructures required to achieve the desired properties in structures that have experienced certain fabrication or repair procedures. The optimum properties in CSEF steels are achieved by producing a

fully-tempered martensitic microstructure through a specific normalization and tempering sequence. However, traditional welding and subsequent post-weld heat treatment (PWHT) sufficiently alters this microstructure and results in welded joints with a variety of microstructures that includes weld metal with large compositional gradients, as well as both coarse-grained (CG) and fine-grained (FG) heat-affected zones (HAZ). The FG-HAZ regions are sources of a number of failures because the microstructures that develop there can significantly reduce creep strength. Also, the size and orientation of the FG-HAZ lead to high triaxial stress states which also accelerate creep damage. This combination results in so-called 'Type-IV' failures in CSEF steels, and studies show that the strength reduction of such joints can be on the order of 50%.

The implications of such failures include disruptions of electrical supply, increased cost of electricity, and the potential for catastrophic failure endangering the safety of power plant personnel. There is an urgent need to understand the mechanisms of failure of components made from these materials, and to devise materials solutions to improve component performance.

OBJECTIVES

The goal of this program is to improve the performance of CSEF steels, with the specific goals of providing guidelines for the maximum safe use temperatures of this class of alloys and, from understanding the causes of the current temperature limitations, developing approaches for increasing the current practical use temperature limits.

Two broad technical issues are being addressed in this effort. One relates to control of mechanical properties and microstructure, and how they depend on chemical compositions and heat treatments. The second emphasizes understanding the causes of Type IV failure and aims to develop strategies to minimize or eliminate it. The overall approach will rely on a combination of fundamental and applied studies of the effects of heat-treatment, welding, and process control on microstructural evolution and material properties.

CONTROL OF PROPERTIES AND MICROSTRUCTURE

Current activities related to microstructure control in CSEF steels are reexamining previously published work[*] indicating that modification of plate tempering and PWHT procedures had the potential of minimizing behaviors contributing to Type IV failures. The observations were based on testing a gas-tungsten arc weld made in Grade 91 plate using Grade 9 filler metal. The plates used for the welding were normalized as usual but tempered at 621°C (1150°F) rather than above the ASME Code specified minimum of 732°C (1350°F). (For reference, contemporary practice for Grade 91 tends to tempering temperatures of 760°C and above.) Pieces of the weld made with these plates were then given PWHT at temperatures of either 732°C or 760°C. Specimens of both PWHT

[*] Brinkman, C. R., Sikka, V. K., Horak, J. A., and Santella, M. L., "Long-Term Creep-Rupture Behavior of Modified 9Cr-1Mo Steel Base and Weldment Behavior," ORNL/TM-10504, 1987

pieces were then prepared for metallographic examination, hardness testing, and creep-rupture testing.

The results of microhardness testing done on a cross-section of the weldment are shown as Figure 1. The hardness indentations extend from base metal, across the HAZ, and into the weld deposit. The hardness profile for the as-welded condition (without PWHT) is shown for reference, and it indicates that regions near both the base metal/HAZ boundary and the weld metal/HAZ boundary are characterized by relatively low hardness values. These regions are often referred to as either soft or overtempered zones and they are accepted as an indication of weakened HAZ regions that are the source of Type IV failures. The 732°F PWHT reduced the hardness values particularly in the regions where they were highest, but a soft zone persisted in the base metal/HAZ vicinity. As expected, the 760°C PWHT reduced hardness values further, but, surprisingly, it also virtually eliminated the soft zone.

In creep-rupture testing at 593°C, cross-weld specimens from the 621°C temper/760°C PWHT coupon failed in base metal with cup-cone characteristics of relatively high ductility. In contrast, cross weld specimens from 760°C temper/732°C PWHT coupon failed in HAZ regions under that same temperature and stress conditions. These promising results were never thoroughly investigated due to closure of the Breeder Reactor and associated programs.

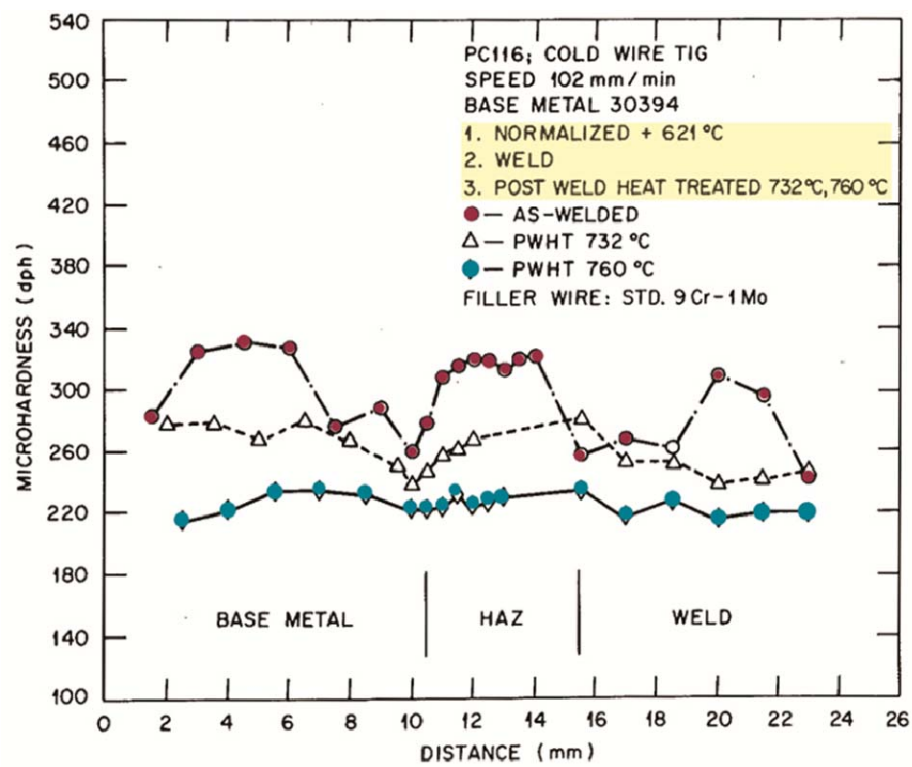

Figure 1: Plot showing hardness indentation traces across a weldment made Grade 91 steel for conditions of: (1) as welded, (2) PWHT at 732°C, and (3) PWHT at 760°C. The Grade 91 plate was tempered at 621°C.

RESULTS

The results discussed above clearly suggest that modified procedures for the combination of plate tempering and weld PWHT can influence the intensity of HAZ soft/overtempered zones and potentially benefit resistance to Type IV failure. Current activities aim to comprehensively examine these possibilities. As an initial screening, three Grade 91 plates with dimensions of 300 mm x 150 mm x 25 mm were normalized for 2 h at 1080°C. The normalized plates were then cut in half, and individual pieces were tempered for 1.5 h at 600, 650, 700, 760, or 800°C. On each 150-mm-x-150-mm face a 150-mm-long x 6-mm-deep groove was machined and welded with Grade 91 filler metal by manual gas-tungsten-arc welding. After welding, the plates were sectioned into 5 1-inch segments or weld length to use for subsequent PWHT. A segment of each plate was to be PWHT for 4 h at the same temperatures used for the original tempering of the plate. A photograph of one of the sectioned plates is shown as Figure 2.

Figure 2: Photograph showing welded Grade 91 plate that was tempered at 760°C before welding.

Each of the finally-heat-treated segments was large enough to yield 4 tensile-creep test specimens with gage dimensions of 50.8-mm length x 3.2-mm diameter.

Tensile and stress-rupture testing was begun for the 5 segments tempered at the 5 different temperatures, welded, and then PWHT at 760°C. The results from stress-rupture testing at 650°C/70 MPa are shown in Figure 3. A rupture life of 519 h was found for the specimen taken from the weld where both the plate tempering and PWHT was done at 760°C to reflect commonly used commercial practices. In contrast, a rupture

life of 2,414 h was found for the specimen where the plate was tempered at 650°C and PWHT was 760°C. At the time of reporting, a specimen was running past 1,046 h from plate tempered at 600°C followed by PWHT at 760°C. Overall, these results show that cross-weld rupture life can be improved by reducing plate tempering treatments to the range of 600-650°C. The results also support the earlier findings shown in Figure 1 of improved creep performance for reduced plate tempering treatments.

The screening tests will continue and be complimented by microstructure analysis furnished through subcontract for a graduate student at Ohio State University. Based on the encouraging initial results more systematic creep testing and analysis will proceed.

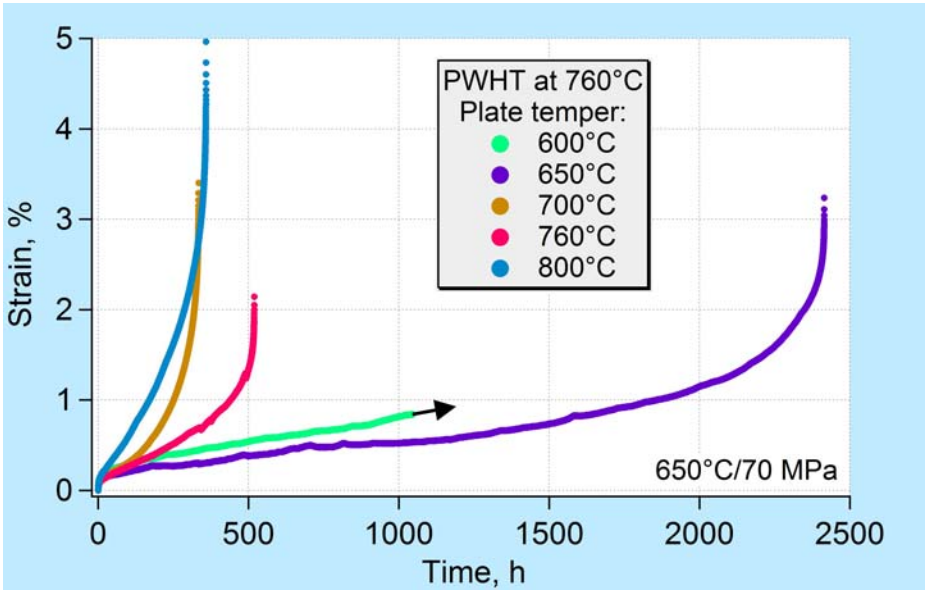

Figure 3: Plot of total strain with testing time for cross-weld specimens from Grade 91 plates tempered at 600, 650, 700, 760, or 800°C and then PWHT at 760°C

UNDERSTANDING THE CAUSES OF TYPE IV FAILURE

The second major activity on this task emphasizes determining and understanding the effects of chemical composition and microstructure on Type IV failure in 9 Cr steels. One way this is being addressed is through basic studies to examine the underlying mechanisms that lead to this behavior. Since Type IV failure is associated with weld heat-affected zones the phase transformation behavior of 9 Cr steels under the dynamic conditions associated with welding and heat treating is being characterized using advanced techniques and analyzed using thermodynamics and kinetics modeling.

Previous work established that phase transformation could be monitored in real-time using high-speed, high-energy x-ray diffraction at the Advanced Photon Source synchrotron. The results of initial work confirmed that the HAZ transformation behavior of an experimental 9 Cr steel known to be resistant to Type IV failure was significantly different than that of a commercial heat of Grade 92 steel. For instance, both steels were heated to a peak temperature near 900°C which would produce microstructures

characteristic of FG-HAZs. Interpretation of those diffraction data indicated that the Grade 92 transformed to about 85% austenite and 15% of the original tempered martensite. During cooling the austenite transformed to about 60% untempered martensite, 22% new ferrite, and about 2% of retained austenite. For the experimental steel, 9Cr-B, heating to the 900°C peak temperature formed only about 44% austenite with the balance being 56% of tempered martensite. On cooling, this austenite transformed to a mixture of 21% new ferrite + 23% untempered martensite. Similar experiments were also done for heating to temperatures near 1100°C characteristic of CG-HAZ regions of weldment microstructures.

RESULTS

To determine how these different temperature conditions might influence local properties in weld HAZs the diffraction specimens were fabricated into specimens for tensile and creep testing, Figure 4. This involved machining reduced sections into the specimens where the centers of the gage lengths corresponded to the locations of the thermocouples used to monitor specimen temperatures during diffraction. Additionally, the specimen ends were reinforced to prevent failure outside of the gage lengths during testing.

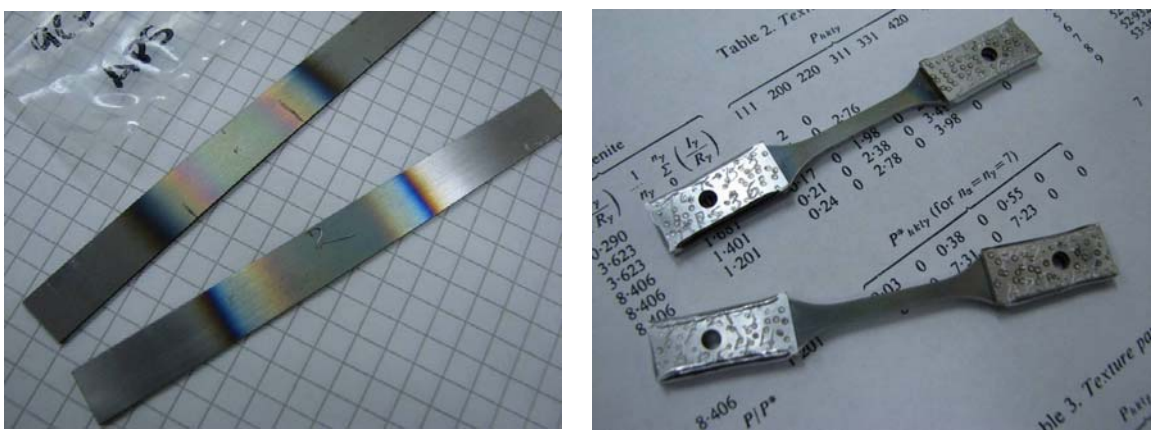

Figure 4: Photographs showing specimens used for synchrotron diffraction experiments (left), and after their preparation for tensile and creep testing (right)

The results from tensile testing the specimens at 650°C are shown as Figure 5. For both HAZ peak temperatures of 900°C and 1100°C the experimental steel, 9Cr-B, was stronger than the Grade 92 (P92). Similar specimens are currently being creep tested at conditions of 650°C and 70 MPa. The specimens will subsequently be analyzed to determine the possible sources of the property differences such as phase transformation behavior, solid solution strengthening, and carbide particle stability.

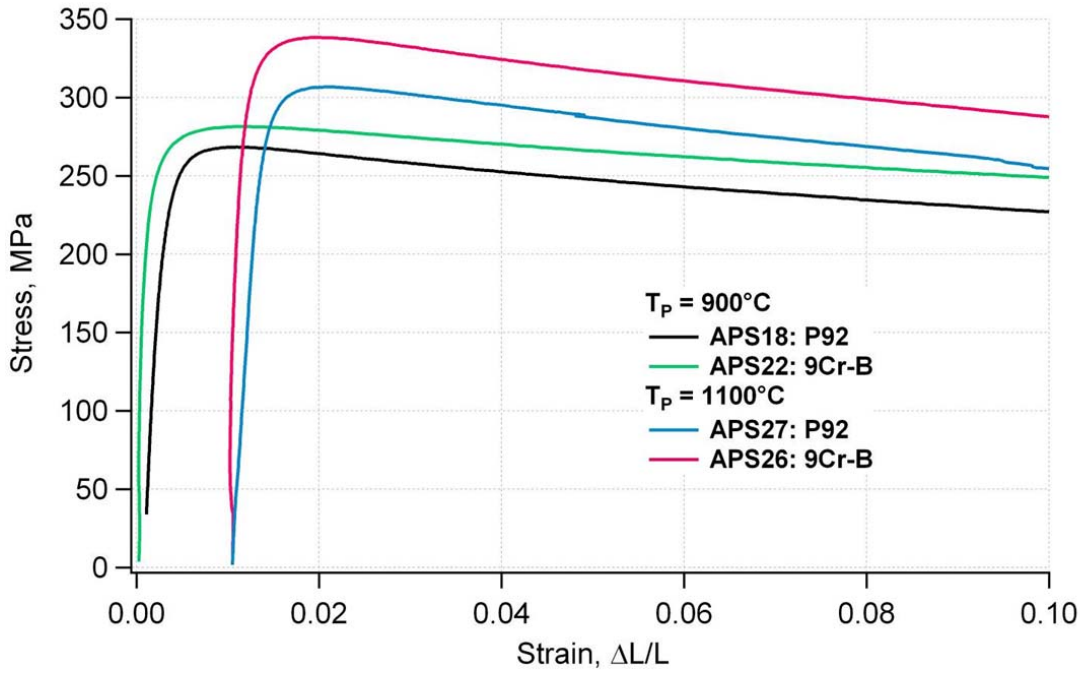

Figure 5: Tensile test results from 650°C for specimens used for synchrotron diffraction