Materials Science Forum Vols. 571-572 (2008) pp 393-398
online at http://www.scientific.net

Online available since 2008/Mar/07

In situ monitoring of weld transformations to control weld residual stresses

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Keywords: residual stress, martensitic transformations, synchrotron diffraction

Abstract. The level of residual stresses generated in fusion welds has been a major area of interest for many years. For steels, a major influence on the final state of stress is through martensitic transformation. This is because the martensitic transformation is accompanied by significant shear and volume strains. One way to mitigate the development of residual stress is by controlling the onset of the transformation such that the associated strain is able to compensate for thermal contraction all the way down to ambient temperatures. In the past it has only been possible to follow the evolution of the phase transformation during cooling of the weld metal using indirect methods such as dilatometry and differential scanning calorimetry. This paper describes the first work in which the phases present are characterized directly during the cooling of reheated weld metal at conditions typical of those encountered during welding by installing a thermomechanical simulator on a synchrotron diffraction beam line at ESRF.

Introduction

Tensile residual stresses close to the yield strength of the metal are commonly encountered in welded components. These may compromise their structural integrity through reduced fatigue life or increased susceptibility to environmentally assisted failure mechanisms. Life-limiting residual stresses can sometimes be reduced by post weld heat treatments (PWHT), but this may be impractical with large or inaccessible components, for example, in the construction of submersibles.

The interplay between the complex thermal gradients and the transient thermomechanical properties of the alloy can profoundly influence the final state of stress. Strains due to phase transformations can drastically affect the picture, giving localized stress relaxation and local material properties which are hard to predict *a priori*. Displacive transformations such as those associated with bainite or martensite formation are associated with anisotropic shape changes characterized as invariant–plane strains with large shear components in addition to dilatational strain. The shear strain manifests on a macroscopic scale when the microstructure becomes non-random, i.e., certain crystallographic variants are favoured. This opens up the potential for controlling the development of residual stresses by controlling the temperature regime over which the phase change occurs. For example, weld transformations have prevented the generation of tensile weld stresses in the fusion zone for both weld fillers in Fig. 1. Here the parent plates are identical, the only difference being the transformation temperatures for the two fillers (~421ºC for OK75.78: 200ºC for LTTE), with the former being bainitic and the latter martensitic [1].

Residual stresses have been measured in welded joints by diffraction for almost as long as diffraction techniques have been used to study strains deep within materials [2]. However, where complex material interactions occur during the welding process it can be difficult to ascertain how the measured stress profiles arose solely from an examination of the final residual stress profiles. It is therefore important to understand the evolution of these stress distributions if such transformation

related effects are to be exploited for stress management. One approach is to use synchrotron diffraction to study the stress and phase distributions around the weld pool under the transient temperature conditions experienced during welding. Elmer et al. [3-5] undertook welding trials on the synchrotron diffractometer table. Whilst these studies have successfully characterized the phase distribution, uncertainties in the temperature at the observation location due to steep transient thermal gradients make it difficult to determine unambiguously the effect of the thermomechanical conditions on the material response and the phases that form.

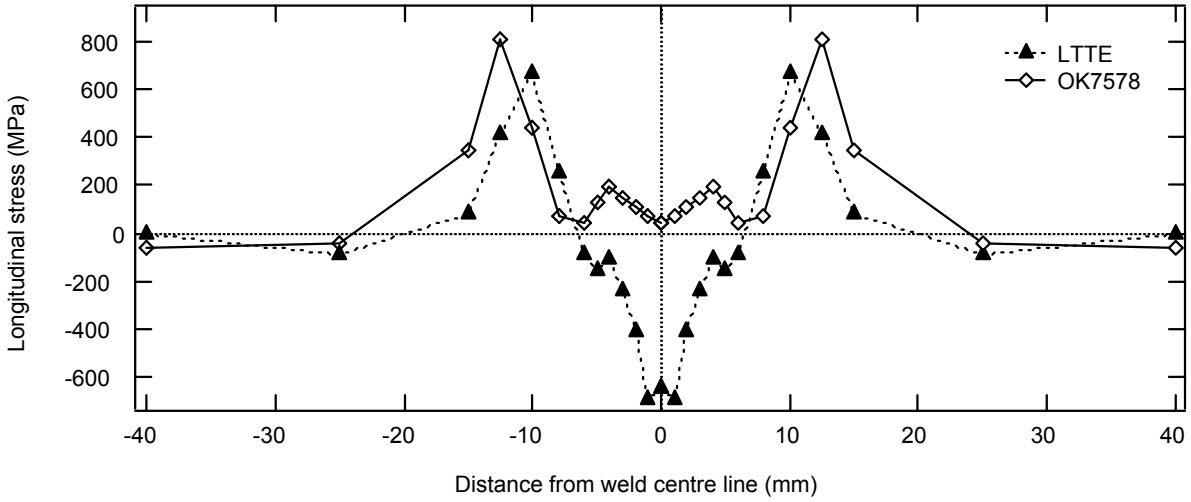

Fig. 1. Longitudinal stress as a function of distance from the weld centerline at a depth of 2.5 mm in two Weldox 960 high strength ferritic steel plates (375 x 200 x 12 mm) each containing a 5 mm deep Vee-groove into which a single weld bead of LTTE or OK75.78 was deposited using manual-metal arc welding [1]. The welding was undertaken in the down-hand position with a heat input ~1.1-1.4 kJ mm^{-1} with a preheat temperature of 125ºC. The HAZ extends laterally from between 11 & 6 mm on each side. Neutron diffraction measurements either side of the weld have been averaged.

As residual stresses form principally in the solid state on cooling from high temperatures, an alternative approach is to neglect weld solidification and focus on the material response to controlled thermo-mechanical cycles. It then becomes feasible to separate the contributions of stress, temperature and composition. By using in situ X-ray diffraction, it is possible to acquire crystallographic, textural, stress information and a wealth of other data as a function of time and temperature. This is in contrast to dilatometry and fast DSC, which provide only indirect data.

In this paper we report the first experiments using a thermo-mechanical simulator to study the evolution of phases, and potentially stresses, during cooling cycles representative of those encountered during welding.

Experimental Description

Experimental apparatus. To produce thermo-mechanical conditions typical of those experienced during welding, an Instron/NPL electro-thermal mechanical testing (ETMT) rig was used. This comprises a 4 kN screw driven mechanical testing stage with sample heating accomplished by Ohmic heating via a direct current applied across the sample under feedback control from a type-R thermocouple attached to the centre of the sample. Heating and cooling rates up to 100ºC s^{-1} may be readily achieved, with mechanical loads superimposed, if required. Samples having a 1-2 mm square or rectangular section and a length of the order of 40 mm can be accommodated. The simple geometry of these samples ensures that testing costs are kept to a minimum whilst the thickness is also ideally suited for diffraction measurements conducted in transmission with hard X-rays at synchrotron facilities. For our experiments the ETMT was mounted on the ID11 beam-line of the

European Synchrotron Radiation Facility (ESRF) in Grenoble, France (Fig. 2). $1.5 \times 1.5 \times 40$ mm samples of the bainitic steel, weld filler alloy designated OK75.78, were cut from an ISO joint prepared by multi-pass manual metal arc welding by ESAB AB, Gothenburg, Sweden. To avoid dilution of the weldment, the faces of the joint were buttered prior to welding.

To limit the effect of temperature gradients across the diffraction gauge volume from the parabolic temperature distribution, slits positioned in the incident beam path were used to define a 400 μm wide diffraction gauge volume at the centre of the sample. A variation of no more than 2ºC is anticipated over the gauge volume at a temperature of 900ºC. To ensure correspondence between the temperature measured by the thermocouple and that of the diffraction gauge volume, the test rig was translated across the beam until the centre of the thermocouple bead was determined from the observed diffraction patterns. The sample was then translated vertically such that the diffraction gauge volume was immersed in the sample directly below the centre of the thermocouple bead, at a point at which no further diffraction signal was detected from the thermocouple bead.

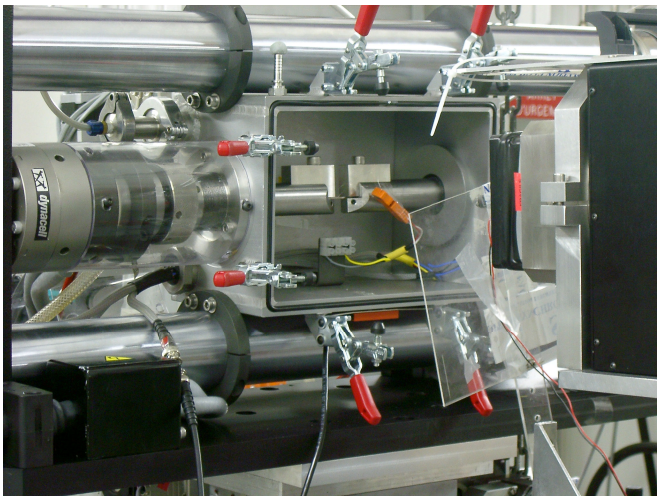

Fig. 2. Photograph of the Electro-Thermal Mechanical Test rig on the ID11 beam line at the ESRF facility in Grenoble, France. The sample environ-mental chamber is photon transparent so that X-rays can pass through the back of the chamber (from behind the rig on the left hand side), pass through the sample gauge volume and reach the detector which can be seen on the right-hand side.

A photon energy of 80 keV (0.1555 Å) was selected using a double Si {111} monochromator. Diffraction data were acquired with a FReLoN (Fast-Readout Low-Noise) camera system having 2048×2048 pixels, each corresponding to 46.8×48.0 μm mounted approximately 258 mm downstream of the sample with its centre aligned with the transmitted beam. To increase data acquisition rates and enable diffraction images to be acquired at a sufficient rate to allow the phase transformations to be adequately characterized at the highest cooling rates, only the central half of the CCD was used. In addition, the data were binned in 4 pixel blocks in the vertical direction. This enabled images of the Debye-Scherrer rings to be acquired with exposures as short as 30 ms at approximately 30 Hz. Calibration of the sample-detector distance, position of the straight-through beam and tilt of the detector were made through measurements of the diffraction pattern obtained from a LaB_6 standard sample of known lattice parameter placed in the sample position in the ETMT. Additionally, to eliminate the contribution from the background signal, exposures were acquired in the absence of the beam and subtracted from the acquired diffraction images.

To assess the effect of cooling rate on the phase transformation, the samples were subjected to thermal cycles comprising: heating to 850ºC at 10ºC s^{-1}, isothermal dwell for 60 s at 850ºC, followed by continuous cooling to room temperature at cooling rates of 100, 90, 80, 70, 60, 50, 40, 30, 20, 10, 5, 2, 1, 0.5, 0.2 and 0.1ºC s^{-1} . The mechanical stage was run under load control with a set-point of zero load. This permitted free thermal expansion and contraction of the samples throughout the thermal cycles.

Raw Data Processing. Processing of the raw diffraction images was performed using the Fit2D image processing software including corrections for the spatial distortion and detector efficiency

along with subtraction of the background signal [6]. The images were then integrated within 20º either side of the horizontal to obtain one-dimensional intensity versus 2θ data sets. A plot of intensity versus 2θ data acquired from the experiment conducted with a cooling rate of 10ºC/s is given in Fig. 3. Rietveld refinement of the one-dimensional diffraction data were performed using the General Structure Analysis System (GSAS) package [7]. The structures of ferrite (Im3m) and austenite (Fm3m) were fitted with the phase fractions and lattice parameters permitted to vary. The background was approximated using a 4-term shifted Chebyshev polynomial.

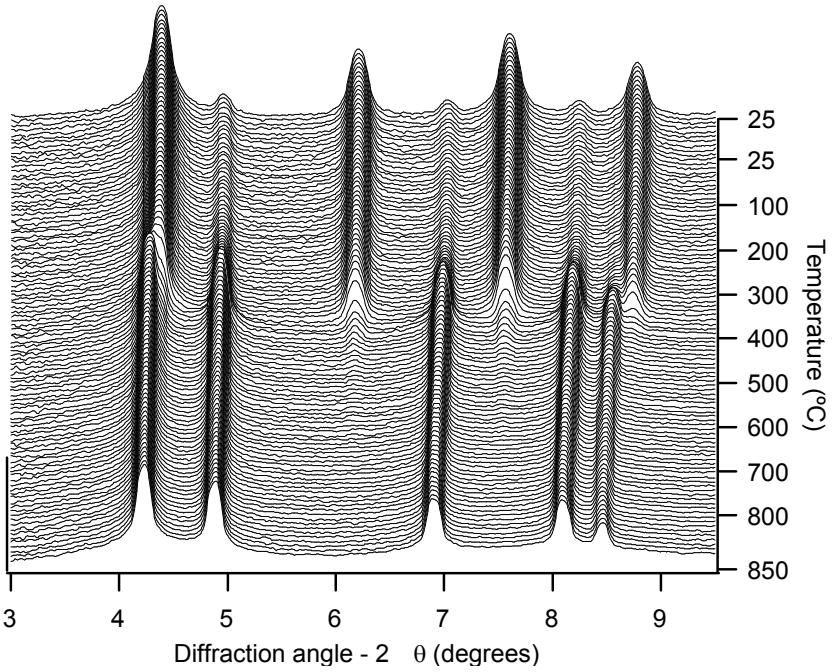

Fig. 3. The transformation of the weld filler metal OK75.78 from austenite to ferrite as a function of temperature at a cooling rate of 10 ºC/s. Note intensity is displayed on a logarithmic scale. For clarity, every 5th data set is shown.

Results

The evolution of phase fractions as a function of temperature under zero applied load at cooling rates of 100, 10, 1 & 0.1ºC s^{-1} are given in Fig. 4. The fraction of ferritic phases evolves in an approximately sigmoidal manner during cooling. The temperature at which ferrite first appears is suppressed as the cooling rate increases. The driving force available when the transformation is suppressed is greater, so the initial rate of austenite decomposition is greater at higher cooling rates. As a consequence, the majority of transformation occurs over a narrower temperature range at higher cooling rates. Notice that the method does not permit the detailed nature of the ferritic phase (allotriomorphic, Widmanstätten, bainite or martensite) to be determined. The variation in phase fractions at the different cooling rates can also be plotted as a function of temperature and time to give a continuous cooling transformation plot (CCT) (Fig. 6).

The effect of the phase transformation and the associated volume and shear strains on the development of stresses during constrained cooling are well illustrated by the Satoh test results given in Fig. 5. These show the stresses that develop as a constrained bar is cooled from its austenitic state. The form of these curves is explained elsewhere [8], but in essence, the bar shrinks as it cools, causing tension to develop, first at a rate determined by the coefficient of expansion of the austenite, then by the yield locus as a function of temperature. As the transformation takes place the transformation strains as well as any associated transformation plasticity act to reduce the tensile strains which can even reverse sign. Once transformation is complete (c.f. Fig. 4), the bar continues to shrink with the result that tensile stresses are introduced at a rate determined by the expansion coefficient of the ferrite. As is evident by comparing LTTE and OK75.78, the lower the transformation temperature, the less tensile is the final constrained weld stress.

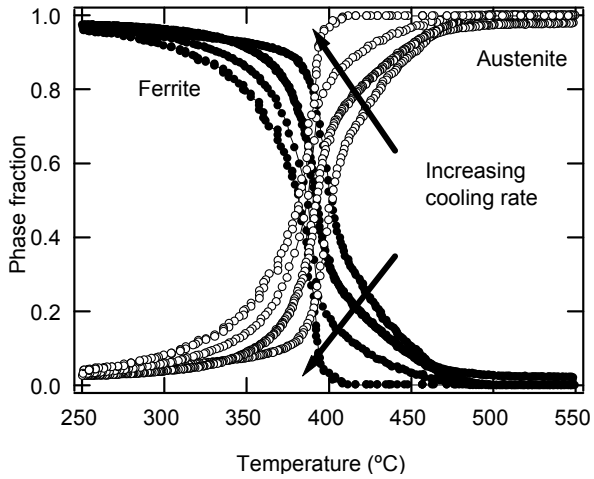

Fig. 4. Austenite and ferrite phase fractions for OK75.78 as a function of temperature and cooling rates (100, 10, 1 and 0.1 C s^{-1})

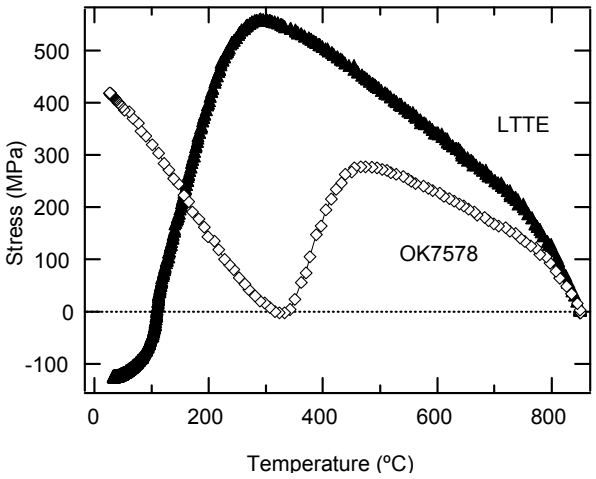

Fig. 5. Results of Satoh tests on LTTE and OK75.78 weld filler metals cooling from 900 ºC at 10 ºC s^{-1} [1].

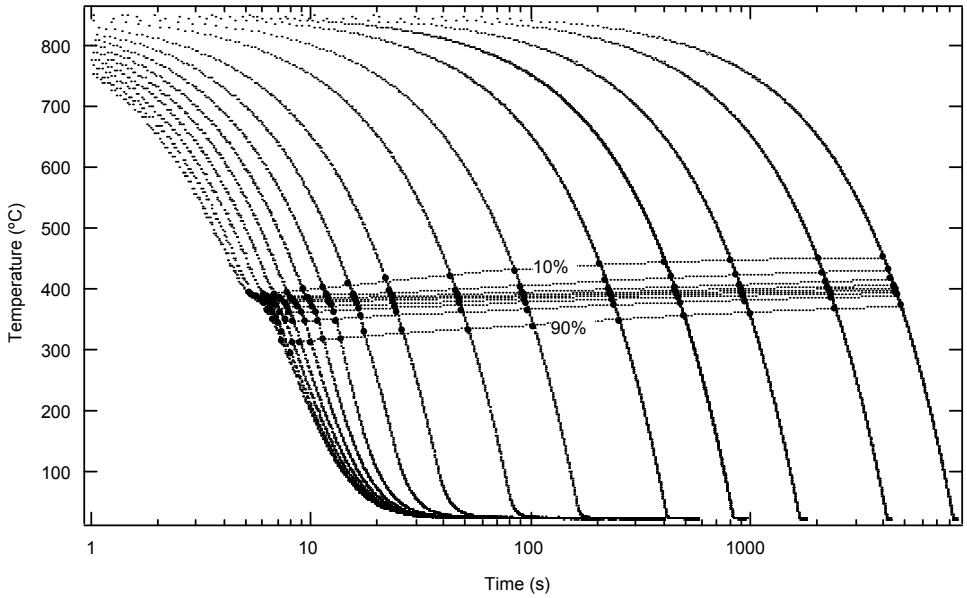

Fig. 6. Evolution of ferrite phase fraction (in tens of %) under continuous cooling for OK75.78 weld filler for cooling rates of 100, 90, 80, 70, 60, 50, 40, 30, 20, 10, 5, 2, 1, 0.5, 0.2 & 0.1C s^{-1}

Discussion

Fig. 5 shows with clarity that in the Satoh tests, the stress state at ambient temperature is more compressive when the transformation temperature is suppressed. This is well understood in that the capacity for the material to transform must not be exhausted before ambient temperature is reached. Otherwise, the thermal contraction, which continues after the parent phase has disappeared, leads once again to the accumulation of tensile stress. This effect has been well understood since the early work of Jones and Alberry [9, 10]. The advantage, however, in conducting the *in situ* experiments using the X-ray source is of course that the diffraction data carry a wealth of additional information which are too extensive to be discussed in this brief paper.

Turning now to actual welds in Fig. 1, in the absence of any phase change, a broad tensile peak (± 15 mm) would be expected in the weld and heat affected zone due to the constrained cooling of the weld from an essentially stress-free condition at high temperature where plastic strain occurs to relieve the stress down to the yield criterion. The phase transformation of the plate and weld filler

within 10 mm of the weld centre line for OK75.78 has lowered the longitudinal weld stress to around zero. For the LTTE weld filler on the other hand the lower transformation temperature has introduced compressive longitudinal stresses within 6 mm or so from the weld line.

Conclusions

1. We have demonstrated that thermo-mechanical simulation combined with fast synchrotron diffraction allows the characterization of the phases under conditions representative of those encountered in welding. The advantage of using a thermomechanical simulator is that the conditions can be very carefully controlled and the thermal history at each location is well known. The technique was used successfully to determine the cooling rate dependence of the phase transformation in the weld filler alloy OK75.78.
2. The data that can be acquired with this technique include the temporal and thermal evolution of phase fractions, preferred orientation, lattice parameters and, potentially, residual stress as a function of both cooling rate and applied mechanical constraint.
3. The data may be used to both rationalize the development of residual stresses during welding and provide data for the numerical modelling of welding processes.
4. For multipass welds the situation will be somewhat more complicated than discussed here, partly because the response on heating is not the same as for cooling. As a result some material will transform only during one pass while other regions closer to the weld deposit may transform more than once. Further work is needed to understand these effects.

Acknowledgements

We acknowledge Dr. L. Karlsson of ESAB AB for the provision of weld filler materials. This work was undertaken when HJS was a research fellow at the University of Manchester. The provision of synchrotron beam time by the ESRF and the help of Drs Matthew Peel, Gavin Vaughan, Jonathan Wright, Magnus Bostrom and Andrew Gotz in undertaking the experiments is gratefully acknowledged; the ETMT was made available through Fame38.

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