Simulating Microstructure Development in High-Carbon Steel Cross-Wire Welding

A schedule for postweld heating is sought to prevent the development of brittle microstructures

BY R. PAN AND D. F. WATT

ABSTRACT. Cross-wire welding is a resistance welding process for joining steel bars into steel grids. The “wire meshes” discussed in this paper are made from high-carbon steel, which gives the product good wear resistance, but increases the difficulty of welding. Cracks sometimes develop during or after welding in high-carbon steel cross-wires unless an appropriate postweld heating schedule is applied. The purpose of our computer simulation is to predict the microstructure development, especially in the heat-affected zone, to control the properties of the final product. An empirical grain-growth equation was derived from Jominy test samples to predict the initial temperature contour in the heat-affected zone at the end of welding current input. A two-dimensional postweld heating algorithm has been developed to simulate the thermal history of the heat-affected zone during the cooling stages. A reaction kinetics algorithm based on the hardenability work of Kirkaldy and co-workers has been modified to fit the resistance-welded high-carbon steel bars. The rate and extent of the decomposition of austenite into its daughter products in the welds were calculated. Experimental observation of microstructures in welded bars with different postweld current schedules was carried out. The algorithm was found to be in general agreement with the experimental results.

The Cross-Wire Welding Process

An excellent description of the welding of crossed bar grids has been given in the AWS Welding Handbook (Ref. 1). In ordinary crossed-wire welding, electrodes apply pressure to the pieces to be welded during and after the time in which a very high electric current is passed through them. Since the largest resistance is at the interface between the two pieces, the greatest heat is produced at this interface. This causes the interface to fuse, and each of the two pieces sinks into the relatively weak austenite zones of the other. When the current is terminated, the heat-affected zone material cools rapidly because of the heat sink of the cooler base metal, and a solid weld is formed.

The welding process used in the work presented in this paper differs from that discussed in the Welding Handbook in two ways. The bars are heavier than the range given in the handbook: 15.3 mm (0.625 in.) in diameter, compared with a maximum of 12.5 mm (0.5 in.) in the handbook. More significantly, the carbon contents were much higher than those discussed in the handbook. This made it essential to apply a postweld heating cycle to avoid cracking at the weld joints during cooling. It was the determination of an appropriate postweld current input cycle designed to prevent the formation of martensite that is the topic of this paper.

Resistance Welded Cross-Wires

Cross-wires are grids made from steel bars as shown in Fig. 1. The mesh joints of the bars, which are resistance welded perpendicularly to each other, range from 5 to 25 mm (0.2–1 in.) in diameter, with mesh spacings being typically 100 to 250 mm (4–10 in.), depending on the purpose of the application. Traditionally, for high-carbon steels, these meshes have been "woven," meaning that the bars are cold formed to partially interlock at the mesh junctions to provide some stability. In the present work, the mesh joints were resistance welded in a special...
Fig. 1 -- Steel cross-wire grids where the bars have been welded with different heating schedules.

Welding Current (A)

Time (Cycles)

Fig. 2 -- Schematic welding schedule applied for the cross-wire welding.

Cra.

Fig. 3 -- Section of a resistance-welded mesh joint with a typical heat-affected zone. Note cracks often initiate at the left- and right-hand sides above the weld line.

Metallurgical Phenomena in Cross-Wire Welding

In the resistance welding of high-carbon steels, a complex metallurgical process is involved, and the quality of the final microstructures in the heat-affected zone (HAZ) is controlled mainly by the welding current and time. The machine functions are computer controlled. The novel feature of the postweld process is that the postweld current schedule is applied so that the cooling rates in the HAZ can be controlled to be sufficiently slow as to prevent the development of brittle microstructures, principally martensite. The extent to which austenite formation and fusion take place in the pieces being joined will vary with the welding cycle, as will the resultant microstructures.

To our knowledge, there is no existing literature which completely models microstructure development in the resistance welding of cross-wires, which is not surprising in view of the unusual nature of the welding machine developed for this purpose. The material used for the study reported here was a high-carbon steel rod, 15.3 mm (0.625 in.) in diameter, chosen to obtain good wear resistance. The composition of the steel is shown in Table 1.

Table 1 -- The Results of the Width of the Mixed Zone Change with Different Postweld Heating Conditions

<table>
<thead>
<tr>
<th>Sample</th>
<th>Current (cycles)</th>
<th>Time (A)</th>
<th>Mixed Zone Width (mm)</th>
</tr>
</thead>
<tbody>
<tr>
<td>No. 1</td>
<td>None</td>
<td>0.40</td>
<td></td>
</tr>
<tr>
<td>No. 2</td>
<td>5 short</td>
<td>0.58</td>
<td></td>
</tr>
<tr>
<td>No. 3</td>
<td>5 long</td>
<td>0.65</td>
<td></td>
</tr>
<tr>
<td>No. 4</td>
<td>10 long</td>
<td>0.75</td>
<td></td>
</tr>
<tr>
<td>No. 5</td>
<td>15 long</td>
<td>1.35</td>
<td></td>
</tr>
</tbody>
</table>

Modeling Microstructure Development in Welds

Although considerable effort has been put into numerical modeling of the welding process, the main focus has been on the size and shape of the weld pool, and on the thermal history of the weldment and its environs. A survey of the literature reveals that most of the thermal models that have been developed are for the heat transfer phenomena only in the resistance spot welding process. They are available to predict the temperature and voltage distributions during the nugget formation, incorporating the thermoelectric interaction at the interface in the weldment and the nugget growth. The calculations coupled to reaction kinetics.
netic theories to predict microstructural development of the HAZ are relatively limited. To date, the models developed have only dealt with arc welding processes. In the early 1980s, Ashby and Easterling (Refs. 14, 15) developed a theoretical model of fusion welding, and they were able to predict the temperature distribution and the microstructural changes in the HAZ of the fusion weld. Watt and Coon (Refs. 16, 17) have also established a model based on the hardenability algorithms developed by Kirkaldy, et al. (Refs. 18-20), to forecast the transient microstructural changes in the HAZ of arc welding. The algorithm can be used to isolate the individual effects of weld process parameters or of metallurgical kinetic factors. Later on, the model was coupled to the FEM heat transfer models established by Goldak (Ref. 21) to predict HAZ microstructure changes in arc welding for a low-alloy steel. Bhadeshia (Ref. 22) proposed another approach to model microstructures for low-alloy steels in arc welding. This model is constructed in terms of adaptations of phase transformation theory, combined with experimental information obtained from a systematic series of low-alloy steel deposits, and is capable of approximately predicting the microstructure of the fusion zone of low-alloy steel weld deposits.

In the above algorithms for modeling microstructure development, the austenite grain size is assumed a constant. In the present investigation of resistance welding, it is found that it is more realistic to assume that the austenite grain size varies with temperature and position in the HAZ. Therefore, in the present work, one of the objectives is to set up a map of austenite grain size distribution as one of the input parameters to model the microstructural changes in the HAZ in high-carbon steel cross-wire welding.

Objectives and Strategies of the Present Work

Microstructure modeling overlaps the difficult subjects of welding and phase transformation theory. The changing shape of the bars as they sink into each other during welding, combined with the complexity of the local time-temperature histories, and the general nonlinearity of the kinetics of the metallurgical reactions makes an analytical prediction of HAZ microstructures intractable. It is a difficult but interesting challenge for numerical modeling.

The main purpose of the present work is to develop a computer program using numerical methods to accurately simulate the cooling thermal history of the weld zone and its environs in the presence of a postweld heating current, and to predict the microstructure formed in the HAZ of welded crossed wires for a range of steel chemical compositions, as a function of the welding input parameters. Consequently, a desirable postweld heating schedule can be prescribed by examining the results of the simulation.

Because the main objective is to determine the form of the postweld current profile that will produce a nonmartensitic HAZ, the model time = 0 is set at the end of the welding stage. The starting point for the calculations was therefore the end of the welding stage where the geometry of the sample could be assumed to be fixed, but the starting temperatures and grain sizes at each node position within the HAZ will vary widely.

In this work, an explicit finite difference method has been adapted to develop the algorithm for two-dimensional conductive heat transfer in postweld heating calculations. Combined with an algorithm for the decomposition of austenite into its daughter phases, the algorithm can be used to simulate the temperature histories and model the development of microstructures as a function of temperature. The details of the numerical model are given elsewhere (Ref. 23), and are reviewed later in this paper. The algorithms combine heat transfer equations with metallurgical reaction kinetics to predict microstructural development. They take into account that the type and volume fraction of transformation products in a welding process depends on several factors, including austenite grain size, peak temperature, cooling rate, and
alloying elements present. Making use of established equations for each of these factors, the finite difference method calculates the local cooling rates and predicts whether austenite will transform into pearlite, bainite or martensite.

In order to map the microstructure development as a function of the temperature in the HAZ, the temperature distribution in the HAZ at the end of the weld cycle in resistance welding must be estimated. In fact, it is very difficult to obtain an exact temperature distribution by any experimental or computational method, because of the very short time period of heat input. Based on experimental results using Jominy tests, an empirical grain growth equation as a function of austenizing temperatures has been developed. The details are given elsewhere (Ref. 24) and are reviewed in the next section of this paper.

The high carbon equivalent of the base metal chosen for the cross-wire grids increases the possibility of forming martensite and the propensity to crack. The difficult part of alleviating the problem of martensite formation and cracking in resistance welding is to know how much postweld heating current is required, and for what period of time it should be applied, to coax the weld bead into forming a normalized, rather than a quenched, microstructure. To implement the program, we are attempting essentially to use the computer to overlay the cooling curves to be experienced by different areas in the weld HAZ onto the continuous cooling transformation (CCT) diagram appropriate for the steel being welded. To do this, it is necessary to develop an algorithm based on Kirkaldy's theory (Ref. 18) on austenite decomposition. This is discussed in more detail in the section after the next. To estimate the accuracy of the modeling results, the computed results of microstructure predicted in the HAZ based upon these models are compared with those obtained experimentally. Thus, the whole algorithm is composed of the modeling of the heat transfer history, isothermal transformation diagrams, and the metallurgical reactions.

**Determination of the Initial Temperature Distribution after the Welding Stage**

Estimating the temperature distribution that existed at the end of the welding cycle was the most difficult problem encountered in developing the welding algorithm.

The austenite grain size that exists when the welding current ends is largely a function of the maximum temperature that a given point in the weld structure has reached during the welding process. It also varies with time at temperature, but not nearly as sensitively as with the peak temperature reached. An additional complication is that the austenite is a high-temperature phase which decomposes during cooling so that very little or no austenite remains. The original austenite grain size becomes obscured in the low-temperature phases that form from the austenite during cooling. The rate at which these low-temperature phases form is strongly dependent on the original austenite grain size. It is the rate of these transformations that we are modeling with this algorithm. Therefore, it is important to know the original austenite grain size in the HAZ. Firstly, the grain sizes are needed as input parameters for the metallurgical equations. Secondly, if the austenite grain size can be found, then this can be used to estimate the corresponding temperature that existed at the end of the welding stage at different nodal points in the HAZ. This temperature distribution can be then used as a necessary input parameter for heat transfer calculations in the model.

What other clues are available to map the temperatures reached? The edge of the HAZ marks the region where the temperature rise was close to the ferrite to austenite transformation. It is further assumed that at the weld line, the temperature reached the melting point of the steel. For the rest of the HAZ, the prior austenite grain size distribution was revealed by etching methods reviewed by Millsop (Ref. 25). These grain sizes were measured along the vertical centerline of the weld and were used to estimate the peak temperature reached at different points along this centerline.

For the latter application, some method for transforming these grain-size measurements to austenite temperatures had to be developed. To do this, Jominy tests were carried out on the same steel as the bars that were welded. The steel bars were held at known temperatures for known lengths of time and then end quenched with water. The prior austenite grain sizes (D) were measured, then plotted as a function of temperature (T) and time (t) to produce a reference set, from which the short time welding austenite grain sizes could be related to temperature. Plotting $D^2$ vs. $t$ gave straight lines which extrapolated back to a temperature dependent $D_0$, grain size value for very short times. The empirical grain growth equation was derived based on the grain growth equation by Feltham (Ref. 26):

$$D^2 = D_0^2 + K t \exp \left[ \frac{Q}{RT} \right]$$

The final empirical equation derived for grain growth for our results becomes:
\[ D_t = D_0 + K t \exp \left( -\frac{Q_2}{RT} \right) \]
\[ K_1 \exp \left( \frac{Q_1}{RT} \right) + K_2 t \exp \left( -\frac{Q_2}{RT} \right) + 2.247 \times 10^3 \exp \left( -\frac{28966}{T} \right) + 0.427 \exp \left( -\frac{21892}{T} \right) \]  

Where \( D \) is the current grain size in \( \mu m \), \( K \) is an empirical constant, \( Q_2 \) is an empirical activation energy for the grain growth process, \( t \) is time in seconds, \( T \) is temperature in Kelvin, and \( R \) is the international gas constant. \( D_0 \) is the initial grain size in Feltham's Equation 1, but is interpreted as the short time grain size at temperature \( T \) in Equation 2. \( Q_1 \) is the empirical temperature sensitivity of the short time grain growth variation. Equation 2 is to be regarded as a simple empirical equation which gives a means of estimating the temperature from the only evidence available, the prior austenite grain size attained. The predicted temperature and austenite grain size distributions at each node in the HAZ at the end of welding current are shown in Figs. 4 and 5. The peak of the curves represents the position of the weld interface. The details of how these values are incorporated into the numerical model are given elsewhere (Refs. 16, 17).

In summary, from the Jominy experimental data, the relationship of the austenite grain growth with temperature and time was established. This has been used to derive an empirical equation of short-time austenite grain growth for the cross-wire welding case. This allowed the local peak temperature distribution in the heat-affected zone at the end of the welding cycle to be estimated using the measured austenite grain size-temperature-time relationship (Ref. 24).

**Modeling of the Decomposition of Austenite**

To simulate the decomposition of austenite in the postweld cooling stage of resistance welding, the hardenability algorithms developed by Kirkaldy and co-workers (Refs. 18-20) were used as the core for the development of the algorithm for metallurgical reaction kinetics. The algorithms can predict the rate of diffusional transformation as a function of the steel chemistry, undercooling, grain size, and the amount already transformed. The latter affects the interfacial area at which the transformation is occurring. The general equations for the diffusional decomposition of austenite have the form:

\[ \frac{dX}{dt} = B(G,T) X^m (1-X)^p \]

where \( X \) is the volume fraction of daughter product already transformed, \( G \) is the prior austenite ASTM grain size number, \( T \) is the temperature, \( m \) and \( p \) are semi-empirical coefficients set to less than one to assure convergence in a form that is derived from a point nucleation and impingement growth model (Ref. 27), which can be expressed as:

\[ m = F(1-X); p = SX \]

\[ B(G,T) = \alpha(G) D \Delta T^q \]

where \( \alpha(G) = 2^{G-1/2}, D \) is the appropriate diffusion coefficient, \( \Delta T \) is the undercooling of the austenite, and \( q \) depends on the type of diffusion appropriate for the particular reaction. The reaction rates are highly sensitive to the grain size, and

**Fig. 6** — IT diagram calculated using modified and unmodified Kirkaldy equation compared with the experimental IT diagrams for a 1060 steel (C 0.64; Mn 0.68; Si 0.22; G.S. 7.5).

**Fig. 7** — Comparison of IT Diagrams with different austenite grain-size number (C 0.63; Mn 0.83; Si 0.22).
in welds the grain sizes reach extreme values.
To use Kirkaldy's expressions in the present resistance welding problem, the amount of daughter product calculated in any time step $\Delta t$ is given as

$$\Delta X = B(G, T) X^m (1-X)^p \Delta t \quad (4)$$

To predict the microstructure development in the algorithm, the transformation points of austenite daughter products are determined as follows:

The calculation of $A_{\alpha 3}$ temperature is based on the work by Kirkaldy and Bagakis (Ref. 28). The pearlite start temperature or eutectoid is taken from Grange's early work (Ref. 29). The bainite start temperature is from Ref. 18, and finally, the martensite start temperature adapted from Steven and Haynes (Ref. 30). For the steel used in this work, the $A_{\alpha 3}$ temperature is found to be 1024 K, the pearlite start temperature is 989 K, the bainite start temperature is 847 K, and the martensite start temperature is 508 K.

Modification of Kirkaldy's Algorithm to Get a Better Empirical Fit to IT Diagram

Kirkaldy's algorithms have the advantage that they can be applied straightforwardly to predict microstructural development for a wide range of steel. Because they were developed for hardenability calculations on an extensive study of a few hundred types of plain carbon steels and low-alloy machinery steels, the algorithms are rather inaccurate at the higher carbon steel levels used in present application. It was therefore necessary to modify some aspects of the algorithms to get a better fit for the steel compositions close to those used in the present study. We concentrate only on the steels whose composition is within the following ranges: carbon, 0.35-0.95%; manganese, 0.15-1.3%; silicon, < 0.3%; Ni+Mo+Cr, < 0.2%; all others < 0.1%.

To do this, the exponents of $F$ and $S$ were modified for this narrow range of steels to achieve a better fit to the experimental diagrams from Atlas of Isothermal Transformations. For comparison, the original Kirkaldy algorithms have been run to determine what sort of systematic errors are observed. In the resistance welding case, it has been found that using Kirkaldy's equation unmodified for the higher carbon and manganese content have a tendency to delay the time of transformation start and fin-
The weld line (faying surface) be-

discussion of the same phenomenon is that if the austenite grain size in a steel is increased, a longer time is needed to complete the reaction, which shifts the C-curve to the right. This reflects the fact that if the austenite grain size in a steel is coarse, fewer nucleation sites for pearlite are available, and the diffusion-controlled transformation of the austenite is retarded. There is a shorter total interface across which austenite is transforming in the coarser material. An alternate description of the same phenomenon is that it takes longer for the advancing interface to reach the center of the austenite grains in a coarse-grained material.

Summarizing the Method of Modeling

A two-dimensional nodal network model created for cross-wire welding is composed of two bars welded perpendicularly with a HAZ along their junction. The weld line (faying surface) between the two bars is assumed to be horizontal as shown in Fig. 4A. If the bars are cut apart mathematically along this assumed horizontal weld line, then because of symmetry, no heat passes across this weld line. Therefore, only the bottom of the bar needs to be analyzed. The geometry of the bottom bar has mirror symmetry along the vertical central plane perpendicular to the axis of the bar at the weld, so only the right half of the bottom bar is needed for analysis in the algorithm. The model is then divided equally into a grid of 18 x 18 nodes, so the model extends from the centerline (weld line) to the bottom surface, and from the vertical mirror plane to a position about one bar radius beyond the right edge of the HAZ into the base metal along the length of the bar. The temperature along these latter two lines (bottom and right edge) is assumed to be fixed at room temperature.

To calculate the rate and extent of the decomposition of austenite into its daughter products in the cross-wire welding during the postweld cooling stage, an initial temperature and austenite grain size must be assigned to each node in the HAZ as discussed earlier.

In resistance welding, an electric current (I) flows creating heat (Q), which is proportional to the square of the welding current, the resistance of steel (which is temperature dependent), and time (t): $Q_{gen} = I^2R(T)t$. At time $t = 0$, we assume the fusion has already occurred at the interface of the two weld bars because of the welding current. When the postweld heating is applied, the heat generation is due to the resistance of the solid metal itself. The resistance, and therefore the heat generated, changes locally with temperature. Each element node has a
different time-temperature history so that the resistance, and therefore the heat generated, will vary with position. The amount of heat production is dependent on the amplitude and duration of the postweld heating current. For a given nodal network point at a given time step, the heat flow is analyzed using an energy balance which is simply

$$Q_{\text{stored}} = Q_{\text{in}} - Q_{\text{out}} + Q_{\text{gen}}$$  \hspace{1cm} (5)

where $Q_{\text{in}}$ and $Q_{\text{out}}$ are conductive heat flows responding to local temperature gradients, and $Q_{\text{gen}}$ is the heat generated at the node by the postweld current. The resulting temperature change in a given time step is

$$\Delta T = \rho V C_p \Delta t Q_{\text{stored}},$$

where $C_p$ is strongly temperature dependent to account for the heat release as the temperature passes through the Curie temperature (Refs. 15, 31).

As the HAZ cools, the modified Kirkaldy algorithm is used to calculate the amount (and type) of daughter product formed at each node during each time step. The rate of product formed during a given time step depends on the temperature, grain size and amount of product (length of the reaction front) already existing at the node. It is also a function of steel composition, but this is not variable in a given application. Austenite which has not transformed to ferrite, pearlite or bainite when the $M_s$ temperature is reached is assumed to transform to martensite.

**Comparison of Predicted and Observed Microstructures**

To find the best postweld heat condition, a series of different schedules was studied. The effect of postweld heating on temperature is dependent on the magnitude of the postweld heating current and on how long it is applied. The interpretation of the modeling results focuses on a few selected nodes.

As mentioned before, the purpose of supplemental postweld heat is to slow down the cooling rate in the HAZ to promote austenite decomposition into more pearlite and bainite. This means that the temperature should be kept in the temperature range between the pearlite start temperature $P_s$ ($989$ K) and martensite start temperature $M_s$ ($508$ K) long enough for all the austenite to decompose.

In choosing an optimum postweld heat schedule, it is important to remember that each node has a different thermal history, so that the influence of postweld heat on each node will be different. The temperatures at the nodes near the central part of the HAZ are higher than the others so that they experience a higher undercooling rate in the $P_s$-$M_s$ range. Also, the austenite grain size for these nodes is larger than for the others so that the reactions forming pearlite or bainite structures are much slower. If overheating occurs during the postweld treatment, the maximum temperature rise will also be located at these nodes because of the higher resistivity.

From the results of the calculation, shown in Fig. 8, the profile of the cooling curves strongly depends on the postweld
heating input schedule and peak temperature reached by an individual node in the HAZ of the weld. At a postweld heat of 15% of welding current, the temperature profiles show that the cooling curves become flatter, compared with no postweld heat, but this has a bigger effect on the nodes near the edge of the HAZ than those in the middle. Similarly, as the length of time of the postweld heat application increases, the $P_r$-$M_s$ cooling period becomes longer, with the largest change at the edge of the HAZ (Ref. 32). At a higher level of postweld heat input, 20% of the welding current, there is an early temperature increase creating a peak that is sustained as long as the current flows. The weld is overheating and will cause some of the austenite grains to coarsen. This makes it more difficult to avoid the formation of martensite.

The predictions for microstructure development by numerical modeling are shown in Fig. 9. These curves represent the volume fraction transformed to pearlite or bainite at selected nodes with different postweld heat conditions. It is noted that each node in the HAZ experiences a different temperature history. Because of different thermal histories, while some nodes are undergoing an austenite decomposition reaction, others are still in the austenite temperature range. Thus, in the algorithm, each volume fraction of transformed products is changed at each node, independently from other nodes, as the cooling time in this temperature range changes.

In the case of 15% postweld heat, as the postweld current time of application is increased, the cooling time spent in this temperature range is longer so that the amount of pearlite and bainite increases and martensite decreases at different nodes. In the case of higher postweld current inputs (> 20% of welding current), increasing the postweld heat only postpones the austenite decomposition time but the amount of pearlite or bainite transformed will not increase. If the near-weld interface nodes are still austenite at or near the end of the postweld heat cycles, then they will quench to martensite.

Figures 10 and 11 delineate the contour of modeling results of the pearlite distribution for the nodal points in the

![Fig. 12 — Microstructures along the central part of the HAZ from experimental work (no postweld heating).](image1)

![Fig. 13 — Microstructures in the HAZ at a postweld heat of welding current. A — A strip along the center part of the HAZ; B-1 — related to point A in Fig. 13A, pearlite develops at the austenite grain boundaries 200X; B-2 — same as B-1 except for magnification of 1000X; C — related to point B in Fig. 13A, feathery bainite occurs near the weld interface of the HAZ, together with martensite, 500X; D — related to point C in Fig. 13A, feathery bainite occurs near the weld line together with martensite, 500X. Light micrographs, Vilella's reagent.](image2)
HAZ at the end of welding for no postheat and 15% postheat, respectively. The pearlite is nucleated along the edge of the HAZ at the base metal (the bottom and top of these maps). When postweld heat is applied, the microstructural change happens around these nodes, and the amount of pearlite and bainite is increased around the HAZ edge nodal points. The modeling results show that the microstructure in the center part of the HAZ is still martensite. When the postweld heat applied is 15% of welding current, there is some pearlite and bainite transformed near the center part. It is known that the pearlite and bainite are very hard to nucleate and develop in the central part of the HAZ because the highest temperatures occur around the weld interface due to the high electrical resistance, and because the grain size is very large there.

It is of interest to compare the modeling results with those obtained from the observation of microstructure for actual welded samples. The model gives an adequate prediction for microstructure development. Metallographic samples for different postweld conditions were cut along the centerline of the joint, then mounted and polished. The specimens were etched by Vilella's reagent, which can bring out pearlite, bainite and martensite. The microstructures were examined using optical microscopes.

Figure 12 shows a sample where no postweld heat was applied. In this case, the microstructure is almost entirely martensite. As the postweld heating current is increased, it is apparent that the width of the mixed pearlite/bainite/martensite region is gradually increased. The measured results are shown in Table 1. The higher the postweld current (<20% of the welding current) and the longer the postweld current is applied, the wider is the mixed zone. In this terminology, the mixed zone extends to the point where anything more than 2% of the structure is non-martensitic phase.

Pearlite nodules nucleate on the grain boundaries of austenite and grow with a roughly constant radial velocity into the surrounding austenite grains. At a smaller undercooling, below gut, the nodules grow as hemispheres or spheres without interfering with each other, as shown in Fig. 13B where 15% postheating current was applied. The higher the postweld heating current (<20%), the slower the cooling, and the more perfect the pearlite that nucleates and develops.

There is another phenomenon observed in Fig. 13 along the weld interface. There is some large feathery bainite occurring around austenite grain boundaries. For a lower postheat current input, only a few small sawtooth-like bainite sheaves appear in the coarse-grained or fine-grained zone. As the postweld heat is increased, the growth of bainite speeds up so that a more completely developed sawtooth-like bainite can be observed as shown in Fig. 13C and D.

In Fig. 13A, the A, B and C positions correspond to nodes of [4,16], [3,16] and [2,16] at the cross-wire geometrical model, as shown in Fig. 4A. The computer prediction for transformed pearlite and bainite (vol-%) at nodes [4,16], [3,16] or [2,16] are [12.4%, 26.6%], [1.2%, 4.3%] or [0.1%, 0.6%], respectively. Comparing the microstructures of the HAZ at the center of the joint with the model with those observed by metallography, it is found they are in general agreement with each other.

It is clear that simply holding the current at 15% of the welding current is not the optimum schedule. Examining the cooling curves in Fig. B (and others in Ref. 32), the optimum schedule would be one which is low to begin with, allowing all nodes to cool into the range 800-600 K, at which time the current should be increased to hold the temperatures in the decomposition range until the austenite has transformed by diffusion reactions. A possible postweld schedule might be 7 s at 5%, followed by 20 s at 18% of the welding current. Unfortunately, the industrial company for which this work was being done went out of this business while the modeling work was underway, so that it has been impossible to test the next generation of postweld schedules on actual physical steel cross-wires to confirm an optimum schedule.

What has been accomplished is that an algorithm has been developed that appears to be able to simulate the decomposition of austenite in the presence of a postweld current, taking into account the wide range of temperatures and grain sizes that exist at the end of the welding cycle for these high-carbon steel bars.

Conclusions

From computer modeling results, the model created for the cross-wire welding is able to predict the microstructure development in the HAZ, making use of the finite-difference method. To do this it was necessary to estimate the temperature distribution in the heat-affected zone from the observed prior austenite grain size. It was necessary to modify Kirkaldy's algorithm to make it more suitable for modeling high-carbon steels. The austenite grain size has a profound effect on the austenite decomposition into its daughter products in the HAZ. The austenite grain size input for the calculation of reaction kinetics must change with nodal position in the HAZ. In resistance welding of these bars, the postweld heat must be carefully controlled, and this research provides one method for prescribing how this should be carried out.

The results indicate that the algorithm gives a reasonable estimate of the microstructural development which could be applied elsewhere. The accuracy is influenced by various parameters such as chemical composition, postweld heating input, and grain size. The modeling results could be used as a reference for the design of a welding procedure. An appropriate postweld heat schedule can be prescribed with the aid of an appropriate IT diagram for the steel composition, postweld computer simulation results, and an analysis of the grain-size distribution in the martensite with no postweld heat.

References

FAILURE OF WELDS AT ELEVATED TEMPERATURES

By G. R. Stevick

This WRC Bulletin presents several new insights into creep crack growth problems: 1) the potential for stress concentrations resulting from the mismatch of creep properties between weld and base metals; 2) a creep crack growth model that includes the effects of stress triaxiality; and 3) a crack initiation model based on the statistical distribution of inclusion size and spacing.

Longitudinally welded piping is used extensively in the power industry for high-temperature applications. Finite element analysis of a typical symmetric, double-V longitudinal weld showed that a material stress concentration will develop in 1 to 2 years if creep properties of the weld and base metals are different and that differences in material properties have a significant effect on the stress field after a crack has formed.

Observations based on a literature review and microscopic studies of welds in low-chromium-alloy steels indicated that crack initiation and growth models could be developed based solely on the growth and coalescence of cavities emanating from the fusion line inclusions. The models developed agree well with industry experience, accurately predicting two recent piping failures in the power industry.

This document should be viewed with the perspective of the events reported in WRC Bulletin 354, Failure Analysis of a Service-Exposed Hot Reheat Steam Line in a Utility Steam Plant and the Influence of Flux Composition of the Elevated-Temperature Properties of Cr-Mo Submerged Arc Weldments.

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