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PREDICTION OF DEFORMATIONS
DURING GAS-TUNGSTEN-ARC STATIONARY WELDS

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ABSTRACT

Local temperature measurements on the heated and unheated surfaces, and strain measurements on the unheated surfaces of unrestrained circular weld specimens of annealed and cold-rolled Nitronic 40 stainless steel during stationary welding, are compared with values predicted from finite-element programs for temperature and strain variations. Experimental and predicted temperature histories agree within 10%. Predicted and measured hoop strain profiles (using a moiré fringe technique), for the unheated surface are compared, showing significant deviations near the central region. Transient deflection measurements of the unheated specimen surfaces show good agreement with theory during the period the arc is operating. Close agreement in deflection behavior was observed during the cooling portion of the weld cycle for the annealed specimen, whereas substantial deviations occurred for the cold-rolled specimens.

KEY WORDS: Gas-tungsten-arc welding, moiré effects, numerical analysis, stainless steels, thermoplasticity, finite element analysis.

INTRODUCTION

The familiar process of arc welding is a combination of rather complex and rapidly interacting phenomena. A concentrated source of energy (the high temperature plasma of the arc), supplies heat to melt a limited amount of material on either side of a joint. Joining is achieved as this material fuses together and solidifies. Although this process is simple in concept, the heating must be very intense in order to limit the amount of material melted, the thermal damage to surrounding material, and the energy required. It is apparent that very high spatial and temporal gradients will occur during both melting and solidification.

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The complex thermal and metallurgical phenomena in a weld zone are accompanied by equally complex stress and strain variations. After initial straining due to thermal expansion, the melted region goes through a stress-free state followed by contraction during solidification and cooling of the fusion zone. Temperatures in the region surrounding the molten zone the heat-affected zone (HAZ), are sufficient to cause some recrystallization in cold-rolled materials. This means that properties in this region will change significantly. Only in the region surrounding the HAZ can the commonly used physical properties be expected to apply during the entire process.

The interactions during melting and solidification cause complex stress and strain fields, which are very important characteristics of a welded joint. Hence, considerable effort has been devoted to understanding the basic process with the objective of being able to predict temperature variations and stress and distortions for specified welding conditions.

The application of heat-source solutions to calculate weld-zone temperatures[1] provided very helpful qualitative information. Using these results to calculate stresses and strains was very limited because of the complexity of the problem and the question of accuracy for the temperatures in the fused and heat-affected zone. The recent availability of large computers has made possible a more precise and integrated study of the problem. Both finite-element and finite-difference techniques are being used to analyze the thermal-mechanical processes occurring during welding. Recent attention has been devoted to the finite-element (FE) method. This approach appears to be more adaptable to systems in which there is concurrent interest in both thermal and mechanical behavior.

A recent review [2] of reported analyses of gas-tungsten-arc (GTA) welding showed that the applicability of current theoretical models has not been established. This is in part due to the lack of definitive experimental measurements and required thermal and mechanical property values. The present investigation was therefore undertaken to study the behavior of a simple system so that a minimum number of factors would be involved and measurements could be clearly interpreted and compared with predictions. Experiments consisted of stationary GTA welds on circular unrestrained plates. Temperature and deformation measurements were made in a controlled-atmosphere chamber, and are compared with predictions from thermal [3] and mechanical [4] finite-element computer codes.

PROCEDURE

The general arrangement of the test apparatus is shown in Fig. 1. The welding electrode and round plate specimens were mounted in a cylindrical chamber. The chamber was evacuated to a pressure of 300 μm , back-filled with argon, reevacuated and then again filled with argon to atmospheric pressure. A screen-covered, toroidal orifice surrounded the electrode to direct a flow of argon gas along the axis of the electrode, causing the necessary purging of metal vapor from the arc plasma region. It was possible to form and operate an arc with the conventional shroud around the lower end of the electrode. Thus, there was an unobstructed view of the arc region. A precision XYZ mechanism was used to position the electrode tip. Power was supplied by a programmable power supply.

Fig. 1. Schematic of experimental system.

During solidification, color motion pictures were taken with a 16-mm motion picture camera at 64 frames per second, clearly revealing the location and movement of the liquid-solid interface. Temperatures on the upper and lower surfaces were measured with Chromel-Alumel thermocouples 0.125-mm in diameter spot welded to the specimens. The fusion zone size and shape were determined from postweld sectioning of specimens.

All specimens were made from Nitronic 40 stainless steel and were from the same heat to assure nearly identical chemical composition, which was approximately 21% chromium, 6% nickel, and 9% manganese. The nickel content is large enough to maintain a stable austenitic structure throughout the weld temperature cycle. Cold-rolled specimens were cross rolled to minimize in-plane anisotropic effects. The specimens were all 76.2-mm in diameter, and 4.75-mm thick.

The strain field on the unheated sides of the specimens were measured using a moiré fringe technique. A rectangular grid was made on a lapped and polished surface of the specimen. The grid pattern was contact-printed on a specimen that was coated with photosynthetic etch resist, from a glass slide that had a rectangular grid image. The photosynthesized areas were then etched off leaving a grid pattern such as shown in Fig. 2. Each dot was approximately 0.025-mm on a side; the distance between centers was approximately 0.050-mm.

Fig. 2. Moiré grid etched on Nitronic 40 specimen.

The image of the specimen grid was projected through a reference grid onto a ground glass as shown in Fig. 1. Focal distances were selected such that the specimen and reference grid frequencies were slightly mismatched, thus creating an interference pattern. Since there was no angular mismatch between the reference and specimen grids, the initial image is a magnified version of the specimen grid pattern. Initial fringe spacing was approximately 25 times the 0.050-mm dot pattern spacing.

Strains were determined by observing changes in the moiré fringe spacing using the relationship:

$$\epsilon = \frac{P}{F} - \epsilon_0 \quad (1)$$

where P is the grid pattern spacing, F the moiré fringe spacing, and $\epsilon_0 = P/F_0$. Thus, the initial apparent strain created by mismatching the specimen and fringe spacing is subtracted from the measured strain to obtain the actual strain. The changes in the moiré fringe pattern was recorded every 0.4-s using a time-controlled 35-mm camera.

The dynamic out-of-plane displacement behavior was measured using inductance-type displacement gauges. Spherical contacts were used on the displacement gages to minimize thermal losses into the transducers. Residual contours were measured using a profilometer.

Since it was impossible to make all the measurements desired during a single weld, data were obtained during several tests. Every effort was made to provide identical energy inputs and flux distribution into each of the specimens by using identical weld parameters. The current was 160 A for a 3-s arc duration. The same electrode material and electrode shape was maintained, namely a 3.2-mm-diameter thoriated tungsten electrode with a tip machined to a slender, sharp point. The electrode gap was carefully set to be the same from one experiment to the next. The argon shielding gas flow rate was 990 litres/hr, with a 10-s pre- and postpurge, and gas compositions were unchanged throughout the experiments.

Temperature data were recorded on an oscillograph; displacement measurements were made with a strip-chart recorder. A recording of the time of arc initiation was produced by a signal from the welding power supply.

THERMAL AND DEFORMATION MODELING

Although thermal and mechanical response are coupled during welding, simultaneous calculation of the temperature and stress variation is a formidable task. Fortunately, coupling effects are not controlling, and it is possible to treat the thermal and mechanical problems separately. A general-purpose FE heat transfer program (TACO), recently developed by Mason [3] was used to calculate temperature variation. This is an implicit program which solves the nonlinear heat conduction with arbitrary heat input and boundary conditions. Since Lagrangian coordinates are used, only relatively small movement can be tolerated. This requires the weld pool to be modeled as stationary. Pool convection is accounted for by using a high value of the thermal conductivity. The heat of fusion is simulated with a step increase in specific heat capacity. Thermal property variations that were used are presented in Fig. 3, and are based on handbook and manufacturer's data. The radial variation in heat input was specified by (from Ref. 5)

$$q(r) = \frac{3Q}{\bar{r}^2} \exp \left[-3 \left(\frac{r}{\bar{r}} \right)^2 \right] \quad (2)$$

Here, $\bar{r}$ equals the radius within which 95% of the energy is transferred. This was determined as the radius of the discolored area on the heated surface outside the fusion region. The total heat transferred, Q , was determined by requiring the predicted depth of penetration to approximately match the value determined from a cross-sectioned specimen. The reasoning involved was to match approximately the volume of metal melted by the arc with that predicted from the assumed heat flux distribution.

The deformation behavior was calculated using an implicit FE program (NIKE2D), developed by Hallquist. [4] The temperature variation calculated by TACO has been incorporated as the input to NIKE2D, which accounts for the thermal strain behavior and temperature-dependent material properties by means of a thermoplastic constitutive model. The code compares current stress levels against the following yield criterion:

$$0 \leq \frac{1}{2} \left| \sigma_{ij} - \frac{\sigma_{kk}}{3} \delta_{ij} \right| \left| \sigma_{ij} - \frac{\sigma_{kk}}{3} \delta_{ij} \right| - \frac{\bar{\sigma}^2}{3} \quad (3)$$

where σ is the effective yield stress, including work-hardening effects. When plastic behavior is considered, the following differential expression describes the thermo-mechanical behavior:

$$d\sigma_{ij} = C_{ij,kl} \left(d\epsilon_{kl} - d\epsilon_{kl}^P - d\epsilon_{kl}^T \right) + \frac{dC_{ij,kl}}{dT} \left(\epsilon_{kl} - \epsilon_{kl}^P - \epsilon_{kl}^T \right) dT \quad ,$$

Fig. 3. Variation of thermal properties with temperatures (based on Refs. 6 and 7 data).

where $C_{ij,k\lambda}$ is the temperature elastic constitutive matrix, and $d\epsilon_{k\lambda}$, $d\epsilon_{k\lambda}$, and $d\epsilon_{k\lambda}$ are the incremental total, plastic, and thermal strains, respectively. NIKE2D calculates incremental stresses and strains so the constitutive implemented in the code is formulated accordingly.

Since the mechanical properties are a function only of temperature (as illustrated in Fig. 4 for Nitronic 40), the behavior is modeled as being a function only of the current temperature state without accounting for the effect of the temperature history. Thus, annealing, high temperature creep, recrystallization, phase change behavior, and solidification substructure effects are not accounted for. Also, the secant modulus is used for the plastic region, based on ultimate strength, yield strength, and elongation to failure. It is recognized that the tangent modulus would be more accurate, but necessary data were not available.

Fig. 4. Variation of mechanical properties with temperature (based on Ref. 7 data).

Illustrative calculated results for the residual effective stress and strain are present in Fig. 5.

A = 0.014 (mm/mm)

C = 0.042

E = 0.071

G = 0.099

I = 0.127

a) Effective strain for annealed specimen.

A = 0.016 (mm/mm)

C = 0.049

E = 0.082

G = 0.11

I = 0.15

b) Effective strain for cold-rolled specimen.

A = 50.0 (MPa)

C = 147.0

E = 245.0

G = 342.0

I = 439.0

c) Effective stress for annealed specimen

A = 120.0 (MPa)

C = 357.0

E = 594.0

G = 827.0

I = 1069.0

d) Effective stress for cold-rolled specimen

Fig 5. Contour plots of effective stress and strain for annealed and cold-rolled specimens. The plots show a portion of a cross section, the axis of the circular plate is the left edge of the plots.

DISCUSSIONS OF RESULTS

Measured and predicted temperature variations at three locations are shown in Fig. 6 for an annealed Nitronic 40 specimen. The weld current of 160 A and time of 3-s were selected to produce a partial penetration weld and cause as little damage as possible to the grid pattern on the unheated surface. The general trends of the measured and predicted curves are in good agreement. Maximum deviations on the order of 50°C are thought to be primarily due to difference between actual thermal properties and those shown in Fig. 3, and the approximate modeling of the convection in the molten fusion zone.

Fig. 6. Temperature variation in Nitronic 40 stainless steel.

Lorentz-force-induced convective cells in the molten pool and the stagnation point flow of the arc plasma gas on the pool surface are primary contributors to the fusion zone shape. Fortunately, the length-squared nature of heat conduction causes the errors between predicted and observed isotherms to dampen out quickly in the HAZ. This can be seen by comparing the shape of the recrystallized region of the cold-worked specimen in Fig. 7 with the overlayed isotherm. Local errors in the temperature gradients in and near the fusion zone cause relatively small errors in the deformation predictions, since materials at near-melt temperatures have relatively little mechanical strength.

Fig. 7. Micrograph of weld on cold-rolled specimen with overlaid isotherms representing the, (a) predicted fusion zone, and (b) approximate recrystallization region.

Illustrative frames of the moiré fringe pattern photographs are shown in Fig. 8. Hoop strain distributions determined from the photographic sequences for annealed and cold-rolled specimens are compared with the predicted variations in Figs. 9 and 10 at 2.5 and 3.6-s after arc initiation. A similar comparison of residual hoop strain variation is presented in Fig. 11.

Fig. 8. Photographs of moiré fringe patterns taken at 2.45, 3.6-s, and 2-min. The center of specimen is indicated on the cross on the photographs).

(a)

(b)

Fig. 9. Hoop strain profiles for (a) the unheated surface of annealed and (b) cold-rolled specimens at 2.4-s.

(a)

(b)

Fig. 10. Hoop strain profiles for (a) the unheated surface of annealed and (b) cold-rolled specimens at 3.6-s.

(a)

(b)

Fig. 11. Residual hoop strain profiles for (a) the unheated surface of annealed and (b) cold-rolled specimens.

Referring to Figs. 9, 10, and 11, it can be seen that the moiré fringe measurements show the unheated surface to experience tensile strain initially, but this is gradually changed to a compressive strain. In contrast, the calculated distributions for the annealed specimens remain tensile in the central region. The possibility of this being due to inappropriate mechanical properties investigated by acceptable variations of the coefficient of expansion, the yield strength, and the plastic modulus in the computer calculations. Results still indicated a tensile strain in the central 2.50-mm region. It is therefore, hypothesized that the reason for this is due to substantial changes of the mechanical properties of the fusion and heat-affected zones. It is also possible that a different constitutive relation or relations are needed to describe the deformation of this material when it is at temperatures near melting. The fact that the predictions of the residual hoop strain for the cold-rolled specimens becomes compressive in the central region is attributed to the much higher strength of this material and to the increased ease of creep in the high temperature regions of the weld.

As a first step to improving the modeling of material behavior, the inclusion of a maximum-temperature-controlled constitutive model should be considered. After a region exceeds a temperature where a structural property change occurs (i.e., recrystallization, melting, etc.), the code would switch to the material behavior description for that region to a new set of property versus temperature curves. Thus, during the heating cycle one set of property curves is used; during cooling, different material property curves are followed. The rationale for this type of model is based upon the concept that many material structure changes occur at high temperatures where the mechanical strength is low. Thus, the region is nearly stress free. This type of model would not be valid for low temperature structure changes, the near melt temperature strain behavior, nor would it better model creep behavior. In the following section, note how annealing and recrystallization had virtually no effect on the deformation during the heating portion of the weld thermal cycle.

Measured and predicted vertical displacement variations with time at the specimen centerline, and 18.0-mm from the center are plotted in Figs. 12 and 13. Comparisons of the final vertical distortions with radial distance are shown in Fig. 14.

(a)

(b)

Fig. 12. Measured and predicted vertical distortion for (a) annealed and (b) cold-rolled specimens at the centerline.

(a)

(b)

Fig. 13. Measured and predicted vertical distorting for (a) annealed and (b) cold-rolled specimens at 18.0-mm from the centerline.

(a)

(b)

Fig. 14. Comparison of final measured and predicted distortion profile for (a) annealed and (b) cold-rolled specimens.

The time variation of the predicted and observed vertical deformations at the center and 18.0-mm from the center can be seen in Figs. 12 and 13 to be in acceptable agreement during the time the arc is operating. However, for the cold-rolled specimens, shortly after the arc is extinguished and cooling begins, the predicted and measured curves start to deviate. This trend continues through the cooling period, leading to a dramatic difference between predicted and measured residual deformations shown in Fig. 14. This behavior, attributed to creep in the cold-rolled specimens, can occur due to the high density of dislocations present. It is interesting to note that this effect continues long after the temperature gradients in the plate thickness direction have become negligible.

In contrast, the predicted and measured vertical deformation variations of the annealed specimens are in relatively good agreement during both the welding and cooling periods. The residual deformations shown in Fig. 14, illustrate that little creep behavior occurred throughout the cooling period, as might be anticipated for an annealed microstructure.

CONCLUSIONS

Temperature variations as measured with thermocouples adjacent to the fusion zone were satisfactorily predicted by the FE thermal code used. This can be attributed to the program calculating the general shape of the fusion zone. However, local differences between the shape of the actual and predicted molten regions occurred, and are due to incorrectly accounting for molten zone convection and inaccurate high temperature thermal property values. Errors in temperature gradients in this region contributed to observed differences between predicted and measured hoop strains at the center of the unheated surface of the specimens.

However, this had relatively little effect on the accuracy of deformation predictions as was evidenced by the experiments on annealed specimens. Failure to account for time-dependent behavior of the cold-rolled material introduced significant errors in the deformation predictions. The use of a thermoplastic constitutive model was satisfactory for the prediction of deformation behavior in welds on an annealed specimen.

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