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MEASUREMENT OF RESIDUAL STRESSES BY X-RAY DIFFRACTION NEAR SIMULATED HEAT AFFECTED ZONES IN AUSTENITIC STAINLESS STEELS*

V.S. Iyer[†], R.W. Hendricks[†] and S.A. David[‡]

[†]Materials Engineering Department
Virginia Polytechnic Institute and State University
Blacksburg, Virginia 24061

[‡]Metals and Ceramics Division
Oak Ridge National Laboratory**
Oak Ridge, Tennesse 37830

ABSTRACT

Simulated heat-affected zones (HAZs) were made in austenitic stainless steel specimens using a Gleeble. The samples were heated to temperatures as high as 1100°C by computer controlled resistance heating. By controlling the heating rate, maximum temperature, and cooling rate, a wide range of residual stresses were introduced in the specimens. Stress measurements were made using X-ray diffraction. It was found that significant stress gradients were produced in the simulated HAZs, and that all stresses were compressive in nature, both in the longitudinal and transverse directions. These results are not representative of the residual stresses determined in the HAZs of real welds, thus calling into question some aspects of the role of the Gleeble in such simulations.

INTRODUCTION

Residual stresses are developed in the heat affected zone (HAZ) during welding. These stresses may be both tensile and compressive in nature and may have detrimental effects on the mechanical and environmental properties of the welded material. There has been a great deal of work done to understand the relationship between welding parameters and environmental response of the weld, but very little has been done to study the effect of the state of residual stresses developed during welding on the mechanical/environmental behavior of materials. Thus there is a need to accurately measure these stresses. Among the various non-destructive techniques X-ray diffraction

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has been widely used. Preferred orientation and grain size problems that frequently occur in real welds test the validity of the X-ray data. To counter this problem, we have created simulated HAZs in a series of austenitic stainless steels using a Gleeble. Through such treatments, it is possible to eliminate the problems associated with directional solidification, grain size coarsening, and alloy compositional effects which are common in and around real welds.

The Gleeble is a product of the new and exciting field of simulation. It is essentially a large, computer-controlled power source for either resistance or induction heating of a sample, and plays an important part in recent advances in the science of welding. The beauty of the Gleeble lies in its flexibility to accommodate a wide range of mechanical and thermal variables occurring at different rates [1]. It is unique in the fact that it may be used simultaneously for mechanical and thermal testing. The extent to which the simulation of the properties of the HAZ of a real weld can be achieved in a specimen depends on the type of simulation used (i.e., resistance or induction heated samples). In the work reported here, the specimens were heated using resistance heating. In resistance heating method, uniform current is passed through the specimen. Resistance to the flow of the current generates heat. The ends of the sample are tightly gripped by two water cooled copper jaws. Thus a steep thermal gradient exists along the length of the sample with the center being the hottest.

Residual stresses were measured using the X-ray diffraction technique. By measuring the lattice strain, the stress can be determined if the X-ray elastic constants for the alloy are known. The crystal lattice spacings were determined from Bragg's law [2]

$$n\lambda = 2d\sin\theta \tag{1}$$

where n is a positive integer, λ is the wavelength of the incident radiation, d is the lattice spacing and θ is half the diffraction angle. Tilting the X-ray source and detector through an angle ψ with respect to the surface of the specimen results in sampling various grains at different orientations near the surface. The changes in the d-spacings were examined by measuring shifts in the diffracted peak positions. In the case of biaxial stresses, the d-spacing of the lattice planes is given by [3]

$$\frac{d_{\phi\psi} - d_0}{d_0} = \frac{1+\nu}{E}\sigma_\phi \sin^2\psi - \frac{\nu}{E}\{\sigma_{11} + \sigma_{22}\} \tag{2}$$

where $d_{0,0}$ is the d-spacing at zero degrees tilt angle (ψ) and zero degrees specimen rotation (ϕ), $d_{\phi\psi}$ is the d-spacing at sample orientation (ψ, ϕ), ν is Poisson's ratio, E is Young's modulus, $\sigma_{11}$and σ_{22} are the normal stresses in the x- and y-directions, respectively, ψ is the diffractometer tilt angle, and d_0 is the lattice parameter of the stress-free material and is approximated by $d_{0,0}$.

MATERIALS AND EXPERIMENTAL WORK.

The materials used in the experiments were austenitic stainless steel of types 310, 316 and 20Cb-3 supplied by Carpenter Technology [4]. After machining, the final specimen dimensions were $1/2 \times 1/2 \times 3$ inches. All of the specimens were solution annealed for 20 minutes at 1050°C in an argon atmosphere furnace and then furnace cooled. Prior to heating in the Gleeble, a thermocouple was percussion welded to the center of the specimen. Water-cooled copper jaws gripped the specimen while it was

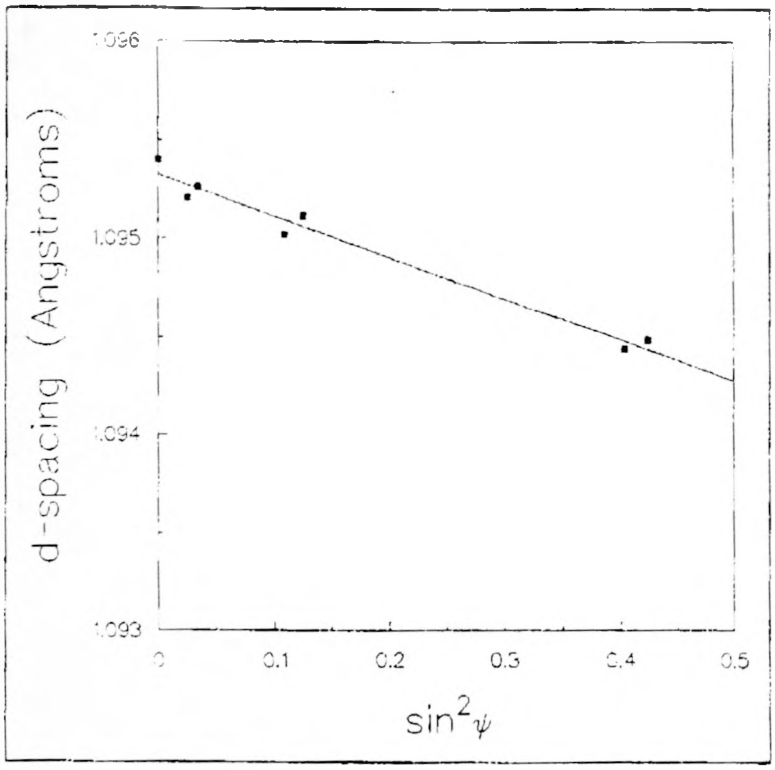

Figure 1: Typical d versus $\sin^2\psi$ plot.

heated and cooled. The jaws were placed in such a way that they created a free span of about 1 inch in length in the center of the specimen. As will be seen later, this was the region in which the maximum residual stresses were created. The Gleeble parameters were adjusted to raise the temperature of the specimen to 1100°C in as little as 13 seconds. Different peak temperatures and heating rates were employed to study their effect on the resulting state of stress [5].

The residual stress measurements were made using a Technology for Energy Corporation (TEC) Model 1610 mobile X-ray stress analysis system. The {311} planes of austenite were measured with CrK_β radiation at approximately 147° 2θ over a ψ range from -40 to $+40$ degrees. The diffraction peaks were corrected for the usual effects such as background, sample absorption, and Lorentz polarization. A parabola was fitted to the top 20% of the peak to determine the peak position, and the 2θ position was calculated using calibration data obtained for the position-sensitive detector. The d-spacings were computed from Bragg's law, and the stress was then calculated from a linear regression of the d versus $\sin^2\psi$ plot. A typical plot is shown in Figure 1. Fits of the data to a more general equation which accounts for shear stresses were also made, but no statistically significant shear stresses were discerned in any of our samples. Stresses were measured in both the transverse and longitudinal directions on all four sides of each sample.

RESULTS AND DISCUSSION.

The objective of our experimental work was to investigate the correlation between the state of residual stress present in the simulated HAZs in the specimens and their corresponding susceptibility to stress corrosion cracking. The environmental implications of the stress distributions reported here are presented elsewhere [6]. Alloys

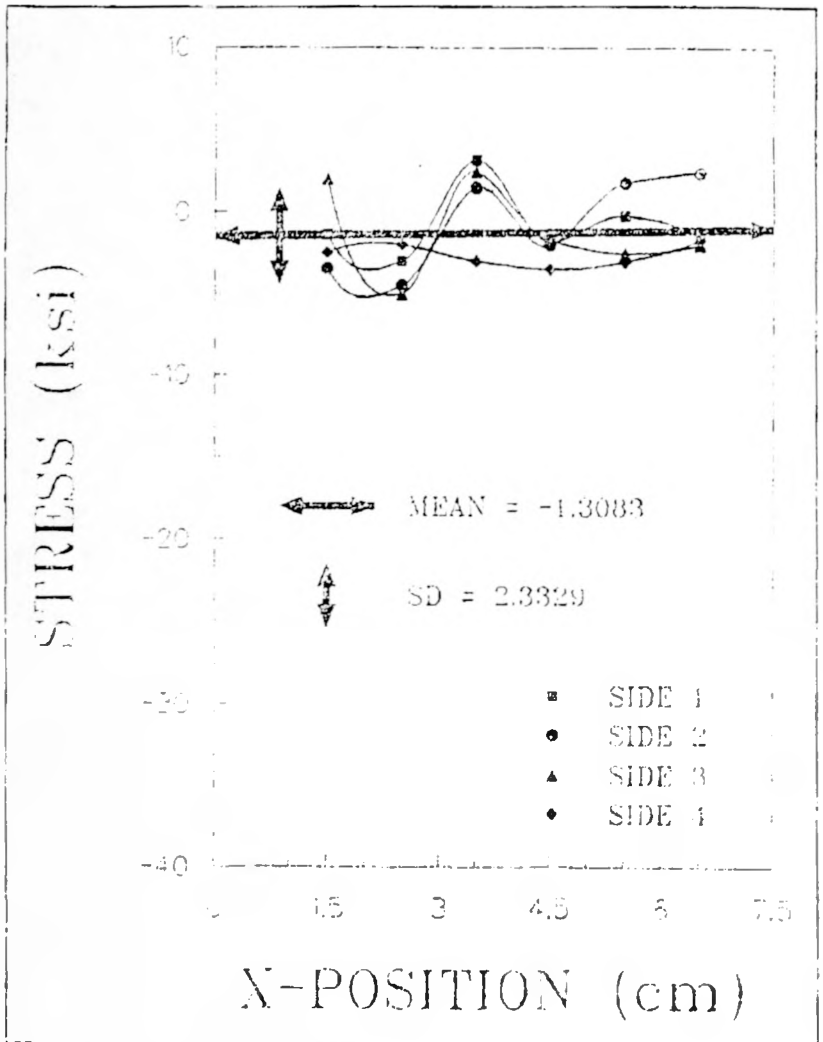

Figure 2: Stress distribution for a stress-free sample of 316 austenitic stainless steel.

of varying nickel content were chosen and different thermal cycles were employed to control the state of residual stress developed in the specimen.

A plot of the stress distribution versus location on the specimen for a stress-free sample is shown in Figure 2. These data are representative of the results from every one of the thirty samples prepared for this work. The mean of the data for the sample is within the counting statistics error of the stress analysis measurements. Thus, there is no statistical significance to any differences along the length or from face to face of samples. The samples are, within experimental accuracy, isotropically stress free. During the Gleeble operations different heating rates and peak temperatures were employed to change the state of stress developed in the specimen. The stresses developed were plotted against positions measured on the samples. A typical sequence for samples of 316 stainless steel heated to various temperatures in 26 seconds is shown in Figures 3 through 6. These stresses are typical of those generated at different heating rates and in different alloy compositions. In fact, there was no statistically significant difference in the states of stress generated in any alloy. This is most likely because the thermal conductivities, electrical resistivities, and emissivities of the various alloys are

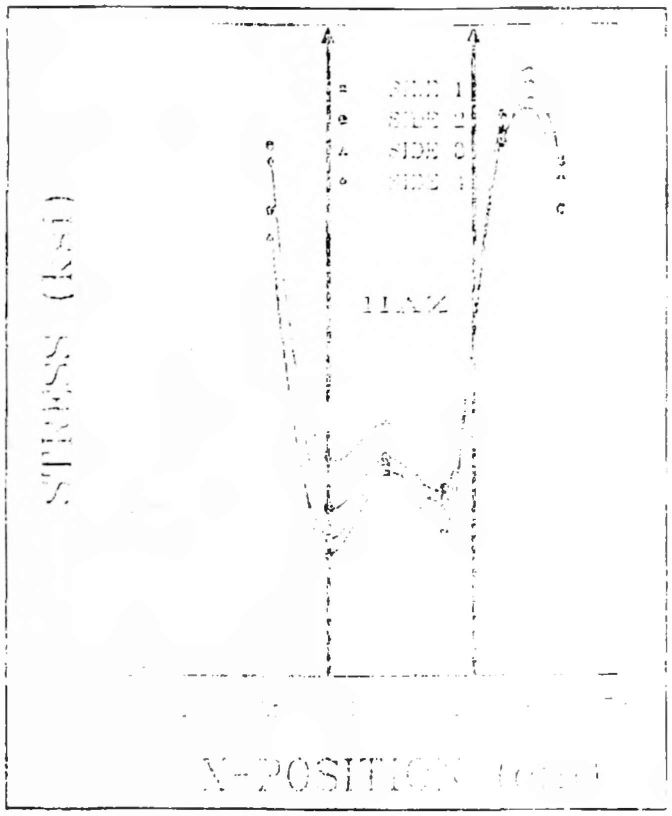

Figure 3: Stress versus position for 316 SS specimens heated to 800°C in 26 seconds.

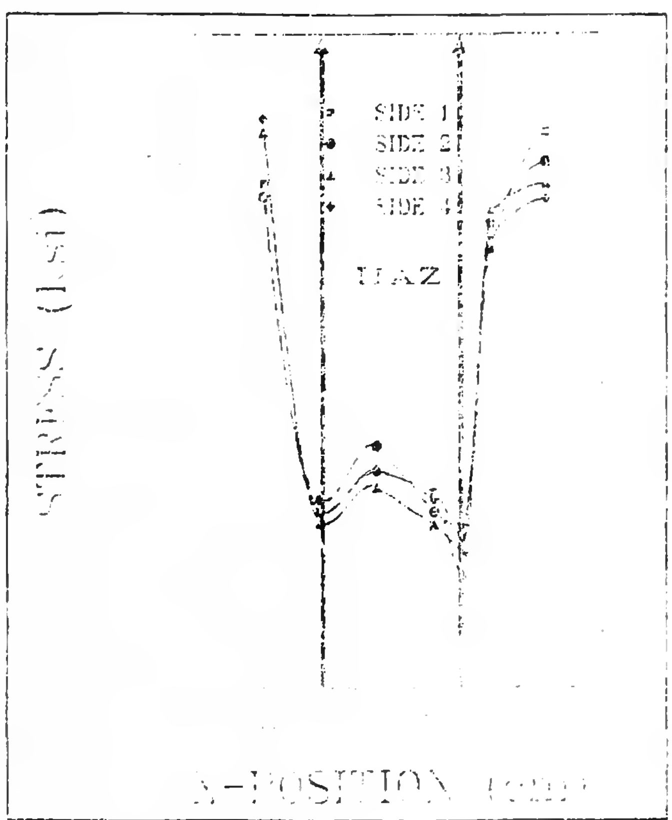

Figure 4: Stress versus position for 316 SS specimens heated to 900°C in 26 seconds.

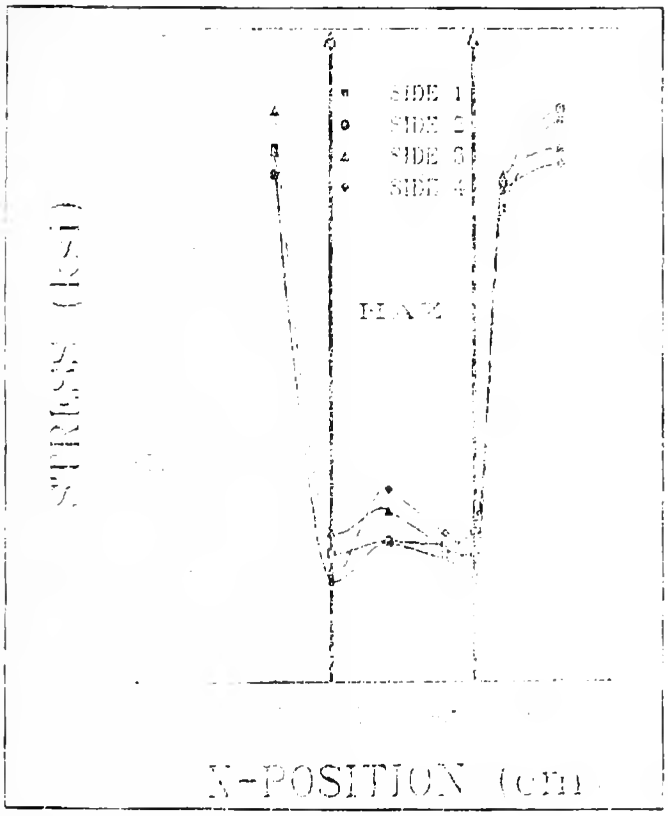

Figure 5: Stress versus position for 316SS specimens heated to 1000°C in 26 seconds.

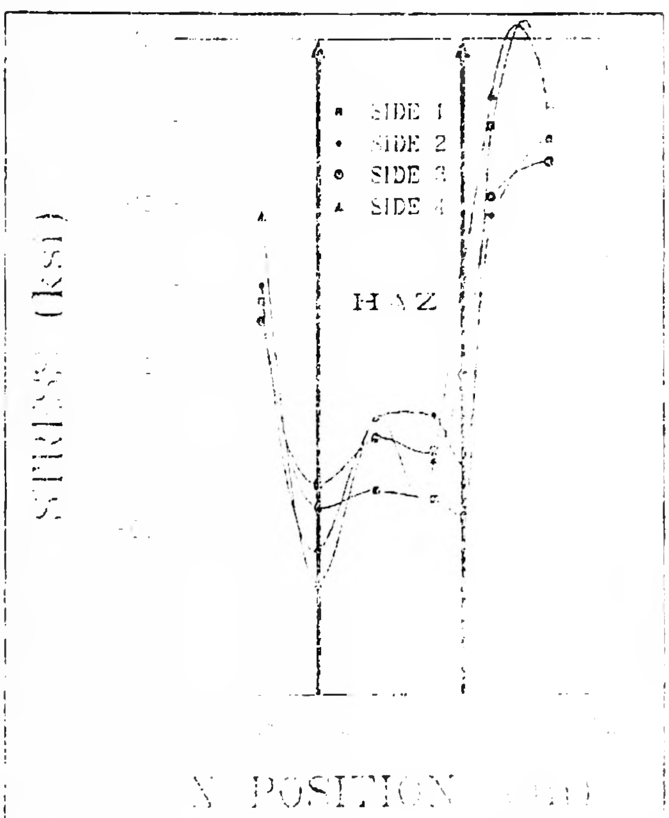

Figure 6: Stress versus position for 316SS specimens heated to 1100°C in 26 seconds.

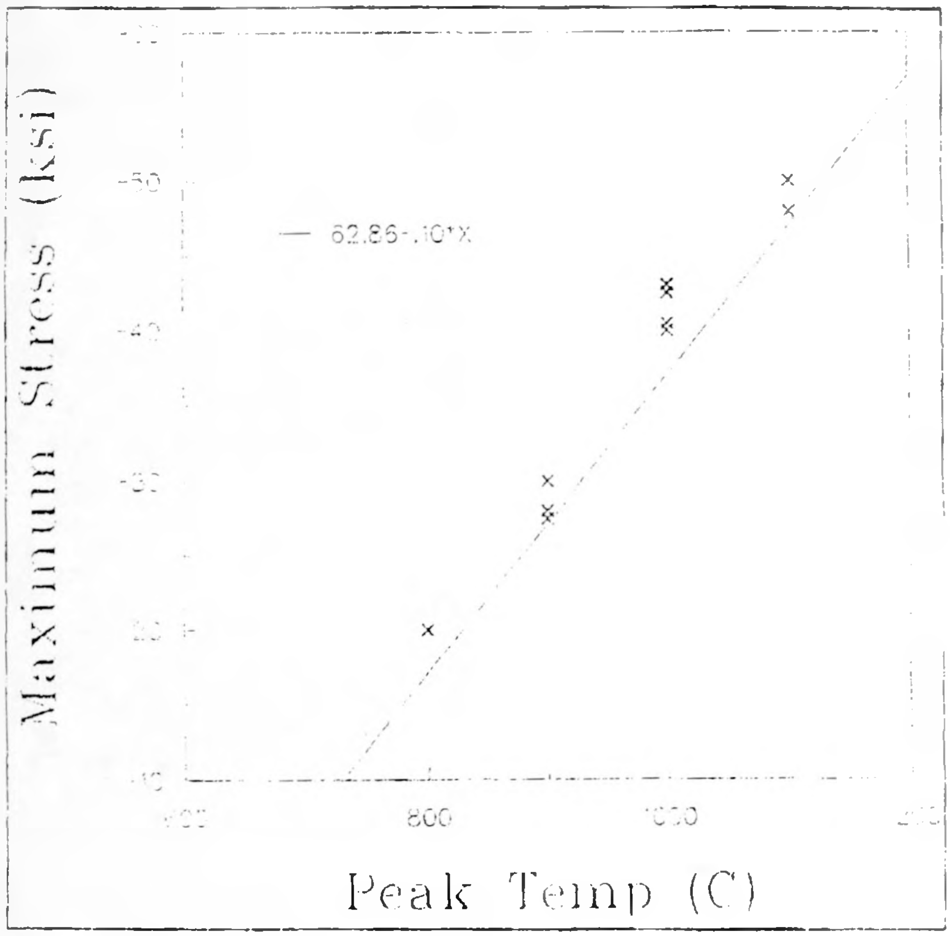

Figure 7: Maximum stress versus of peak sample temperature.

not very sensitive to composition. Thus, it is expected that the thermal gradients, and hence the residual stress, should be determined for the most part by the Gleeble power parameters rather than by alloy compositional variables. That the peak stresses developed were not a function of the sample heating rates is consistent with our argument, to be presented later, that the residual stresses result from the thermal gradients at the sample surface which result from radiative losses.

As is seen in these data, the stresses in our simulated HAZs are compressive in nature. Compressive stresses of magnitude as high as 55 ksi were generated in the specimens heated to the highest temperatures. Correspondingly, as the peak temperature decreased the magnitude of the stresses also decreased as shown in Figure 7. Also, a steep gradient in the stress distribution is observed with the regions in the HAZ being the most compressive and regions under the jaws being close to stress free. As the peak temperature decreased the gradient became more shallow. Again, as with the stress distributions, the peak stress developed in the sample is independent of the sample alloy composition.

It is important to note that these compressive stresses were observed in both the transverse and the longitudinal directions. Nowhere, within the spatial resolution of our X-ray beam (about 1×5 mm), were any tensile stresses found. However, it is clear that tensile stresses must exist below the surface of the samples. In an attempt to find such stresses, $25 \mu m$ thick layers of material were removed by electropolishing and the residual stresses remeasured. These data were corrected for the effects of material

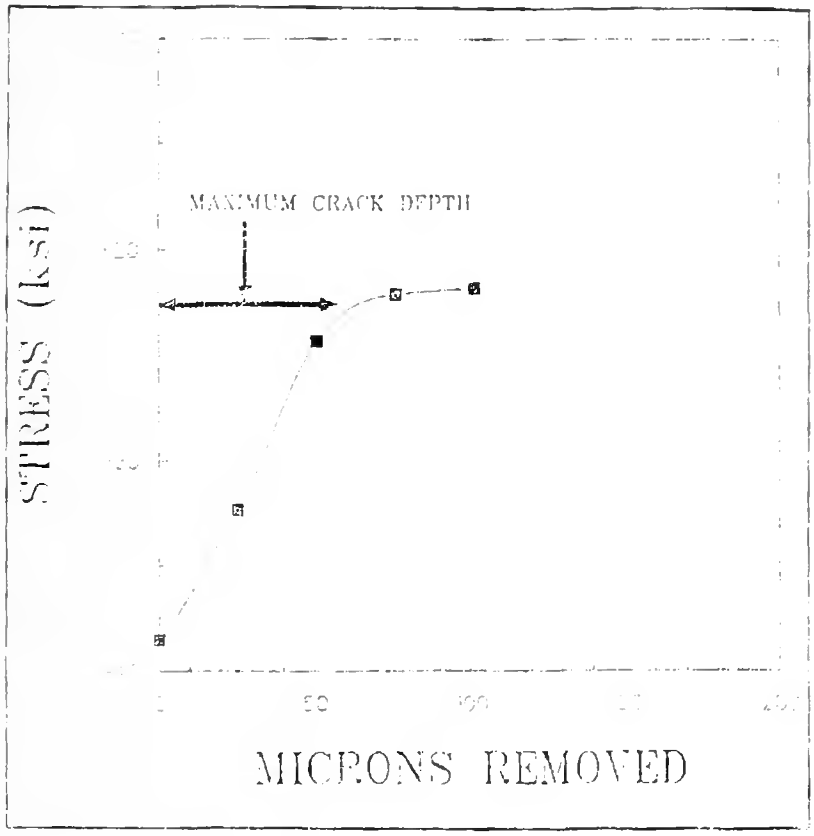

Figure 8: Residual stress profile as a function of depth below the surface as determined after material removal by electropolishing.

removal and the results are shown in Figure 8. It is important to note that, although the cracks which developed in these alloys when exposed to boiling $MgCl_2$ were typically about $60\mu m$ deep (and none were greater than $100\mu m$ deep), the compressive nature of the residual stresses in these alloys persisted to depths of more than $125\mu m$ [6].

In a real weld the residual stresses in regions in and close to the HAZ are always tensile in nature and as the distance from the weld centerline increases it becomes more compressive [7]. This situation is significantly different from the results presented here. Simulated HAZs have not previously been carefully mapped for residual stresses. It is clear that the simulated HAZ created under the conditions used here does not represent the HAZ of an actual weld.

The results of this investigation have significant implications since literally thousands of Charpy impact samples with simulated HAZs have been prepared in Gleebles in an attempt to characterize the mechanical properties of welded samples under controlled conditions. The present work brings the correlation of these results with the properties of "real" HAZs into serious question. It is clear that the distribution of stresses in the simulated HAZs, especially in the region of the stress gradients, is very complex.

A finite element model is being developed to try to understand the distribution of the temperature in the depth and along the length of the specimen given the heat input and peak temperature and the specimen geometry [8]. From the temperature distribution in the specimen the strain values can be calculated and correspondingly the stress distribution throughout the specimen developed. From our work to date, we have determined that the most likely explanation of the thermal gradients which result in the residual stress distributions reported here lies in the large radiative losses from the surfaces at the hottest part of the sample.

The samples with the stress distributions reported here have been subjected to accelerated stress corrosion cracking in boiling $MgCl_2$ following the ASTM G36-87 standard [9] with surprising results. It has been found that the low nickel-content alloys (Ni $\leq 20\%$) all crack even though there is, as yet, no evidence for tensile residual stresses. These results are reported elsewhere [6] and remain the subject of intensive further investigation.

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[4] We are indebted to Mr. N.B. Schmidt of Carpenter Technology Corp. for providing the materials used in this study.

[5] VSI and RWH would like to express their gratitude to the staff of the Welding Group of the Metals and Ceramics Division of ORNL for their generous support and especially for access to the Gleeble 1500 which was so crucial to the success of this project.

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