Residual Stresses and Plastic Deformation in GTA-Welded Steel

Postweld inspection of the plastic deformation that occurs during welding leads to a better understanding of residual stresses near welds

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ABSTRACT. Residual stresses and plastic deformation in single pass GTA welded low-carbon steel were studied by means of x-ray diffraction in combination with optical microscopy and hardness measurements. The residual stresses and the amount of plastic deformation (microstrain) were obtained from x-ray diffraction line positions and line broadening. Since the plates were polished before welding, it was possible to observe in the optical microscope two types of Lüders bands. During heating curved Lüders bands and during cooling straight Lüders bands perpendicular to the weld are formed. The curved Lüders bands extend over a larger distance from the weld than the straight Lüders bands. The amount of plastic deformation as obtained from the x-ray diffraction analysis is in agreement with these observations. An explanation is offered for the plastic deformations observed. It is concluded that in the present experiments plastic deformation is the main cause of the residual stresses.

Introduction

It is well known that during arc welding residual stresses are produced in the vicinity of a welded joint (Refs. 1-3). These stresses can give rise to distortion and under certain circumstances even to premature failure. Thus, residual stresses play an important role as far as the quality and reliability of a welded construction are concerned.

The thermal cycle accompanying the welding process results in phase transformations, melting, solidification and nonhomogeneous cooling of the welded material. These phenomena usually have mutual interaction. Therefore, residual stresses near welds are in most cases highly complex in nature.

Residual stresses near welds only arise when plastic strains, i.e., irreversible changes in shape or volume, occur during the thermal cycle. Normally only stresses and not plastic strain near welds are monitored.

By measuring the residual stresses and the amount of plastic deformation simultaneously, a better understanding can be obtained about the origin of the stresses and also about the behavior of the welded material as a construction element.

X-ray diffraction provides a valuable tool to determine nondestructively the residual stresses (line shift) as well as the degree of plastic deformation (line broadening) as a function of position in a welded specimen. No other method is capable of providing both types of information simultaneously.

The x-ray diffraction method is highly accurate, but is limited by the fact that only information is obtained about a relatively thin (~5 μm) surface layer.

In the present study, the x-ray diffraction technique is applied to the specific case of single pass GTA welds in steel plate. The aim of the study was to show that the x-ray diffraction results are helpful in obtaining a better insight into the origin and nature of residual stresses and plastic deformation in welded material, and therefore, may be of value in advancing the control of weld residual stress and distortion.

Experimental

The specimens to be welded were cut from hot-rolled Type Fe 510 steel plate (160 X 120 X 4 mm; 6.3 X 4.7 X 0.16 in.). This type of steel contains about 0.21 wt-% C, 0.5 wt-% Si, 1.5 wt-% Mn, 0.045 wt-% P, 0.045 wt-% S, and is practically single phase. The plates were ground on both sides to a final thickness of 3.70 ±0.02 mm, using a superfinishing grinding disk (up to mesh 240). After grinding, the zones on which stress measurements were to be carried out were polished in a mixture of 3 parts H₃PO₄ (85 wt-%) and 5 parts H₂O₂ (30 wt-%). To prevent excessive heating during polishing, the nonrelevant surface area was covered with an acid-proof lacquer. In order to remove stresses produced during fabrication and subsequent handling, the plates were stress relieved for 6 h at 650°C (1202°F) in a helium atmosphere, followed by slow furnace cooling (after 65 h the temperature was 45°C; 113°F).

After this annealing treatment, the grain size was about 8 μm, whereas the residual stress level was about 10 MPa (1.5 ksi).

Welding

Autogenous GTA welding of the samples was carried out in an airtight box containing pure argon. The welding torch was mounted on a transporting device and could be moved mechanically.
at a constant speed with respect to the sample to be welded. Single pass bead-on-plate welds were made over about one-third of the sample length — Fig. 1.

Apart from the welding current (heat input) all welding parameters were kept constant throughout the experiment. The welding parameters used are listed in Table 1.

X-Ray Diffraction Measurement of Stresses

The x-ray diffraction measurements of stresses near welds are based on the following equation (Ref. 4):

$$\frac{d_{t, y}}{d_0} = S_{\sigma_L + \sigma_T} + \frac{1}{2} S_2 \sigma_T \sin^2 \psi$$  (1)

where $d_{t, y}$ is the spacing of the (hkl) lattice planes parallel to the welding direction at an angle $\psi$ with the specimen surface, $d_0$ the stress-free spacing of the (hkl) lattice planes, $S_1$ and $S_2$ are the so-called x-ray elastic constants for the constant throughout the experiment, and $\sigma_L$ and $\sigma_T$ are the stresses parallel and perpendicular to the welding direction, respectively — Fig. 1. The lattice spacings are calculated according to Bragg's law from the observed positions of diffraction peaks. Lattice spacings measured as a function of the tilt angle $\psi$ (around an axis parallel to the welding direction) are plotted vs. $\sin^2 \psi$ (a so-called $\sin^2 \psi$ plot). $\sigma_T$ is obtained from the slope of a straight line through the data points in the $\sin^2 \psi$ plot and $\sigma_L + \sigma_T$ is calculated from the intercept. As is readily seen from Equation 1, the accuracy with which $\sigma_L$ can be obtained is not critically dependent on the accuracy of $d_0$, whereas this is not the case for $\sigma_T$. In the present investigation, however, the stress-free (hkl) lattice spacing of the as-annealed specimens (see above) can be measured with great precision. Thus, reliable values for both $\sigma_L$ and $\sigma_T$ are obtained from a single $\sin^2 \psi$ plot.

The x-ray diffraction measurements were performed on a computer-operated Siemens F-40 diffractometer with a diffracted-beam curved-graphite monochromator. The tilt of the specimen over the angle $\psi$ was performed about the $\theta$ axis. The standard specimen holder was replaced by a one-axis specimen shift device. After each stress measurement, on command of the measurement computer, the specimen was transported in such a way that at constant $\psi$ the $x$ coordinate of the irradiated area of the specimen was changed — Fig. 1. The specimen shift device was equipped with adjustment screws, which made it possible to align the specimen in the diffractometer, i.e., to set the specimen at $\psi = 0$ in the symmetrical Bragg-Brentano focusing geometry, and to correct for a displacement of the "average" specimen surface with respect to the $0/2\theta$ axis of the diffractometer (due to the stresses present, the specimen surfaces were not completely flat).

For each specimen position where stresses had to be measured, the local remaining specimen displacement $s$ was obtained from the measured peak position $2\theta_m$ of the (220) reflection of a Si powder (NBS SRM 640a), which was attached to the specimen surface by allowing to dry a suspension of Si powder in alcohol. The reference for the (220) Si peak position $2\theta_0$ was obtained from a measurement with the normal specimen holder in the well-aligned diffractometer. The specimen displacement $s$ then followed from $2\theta_m - 2\theta_0 = 180 \times \sin \theta / \sin 2\theta R \sin (\theta + \psi)$ where $R$ is the radius of the diffractometer circle. This formula also served to correct for the effect of the so found $s$ on the measured positions used in the stress determinations.

The conditions and settings for the diffraction measurements and the con-
The transverse stress $\sigma_T$ as a function of the distance $x$ to the weld centerline, at $y = 0$, for a welding current of 150 A. The shaded area represents half of the weld.

The longitudinal stress $\sigma_L$ as a function of the distance $x$ to the weld centerline, at $y = 0$, for a welding current of 150 A. The shaded area represents half of the weld.

The measured profiles were corrected for background, for angular dependence of the Lorentz factor, the polarization factor and the absorption factor (Ref. 5) and for the presence of Kα2 radiation (Ref. 6). Subsequently, the peak position was determined by matching a parabola to the top part of the profile. Finally, the peak positions were corrected for the effect of local, remaining specimen displacement (see above).

By plastic deformation, crystal imperfections, especially dislocations, are introduced. The nonuniform strain fields around dislocations contribute to x-ray diffraction line broadening. From an analysis of the line broadening, information can be obtained about the volume average of the strain variations. This is expressed in a microstrain parameter $\varepsilon$, which in turn is indicative for the amount of plastic deformation a specimen has endured. In fact, the energy stored in the material due to plastic deformation is closely proportional to $\varepsilon^2$.

To obtain the microstrain as a function of $x$ in the present specimens (Fig. 1), the broadening of the line profiles obtained at $y = 0$ during the stress measurements was analyzed. The measured profiles were corrected for background, for angular dependence of the Lorentz factor and the polarization factor, and for the Kα2 presence in the same way as with the peak position determinations. Details of the applied single-line Voigt analysis, which is performed on the basis of line width measurements, are pub-

Fig. 2 — The transverse stress $\sigma_T$ as a function of the distance $x$ to the weld centerline, at $y = 0$, for a welding current of 150 A. The shaded area represents half of the weld.

Fig. 3 — The longitudinal stress $\sigma_L$ as a function of the distance $x$ to the weld centerline, at $y = 0$, for a welding current of 150 A. The shaded area represents half of the weld.

Fig. 4 — A — The maximal values of the stress in the heat-affected zone, $\sigma_{L,\text{max}}$ and $\sigma_{T,\text{max}}$; B — the positions $x$ at which they are observed vs. the welding current $I$ ($y = 0$). In Fig. 4B the positions of $\sigma_L = 0$ and of the fusion boundary (FB) are also indicated.
lished elsewhere (Refs. 7, 8).

Apart from the microstrain, this analysis also yields an effective crystallite size. However, for the present specimens, the crystallite size contribution to the line broadening proved to be negligibly small, and therefore the full line width was attributed to microstrain. By analyzing the line broadening with reference to the line broadening of profiles measured far away from the weld, it was possible to obtain a microstrain parameter $e$ which gives solely information about the plastic deformation induced by welding. Crystal imperfections, left after the annealing treatment of the specimens, do not contribute to the $e$ thus obtained (Ref. 9).

### Optical Microscopy and Hardness Measurements

To obtain information about the microstructure of the welded samples, transverse cross-sections were made of all welds. After grinding, polishing and etching in 2% nital, the cross-sections were studied by means of a Leitz Neophot type microscope. The hardness was measured by a Leitz Durimet Vickers hardness tester employing a load of 98 N.

### Results

Using the x-ray diffraction technique described in the previous section, stress measurements were carried out on a stress-relieved, unwelded sample and on a series of samples welded with different welding currents — Table 1.

The stress measurements carried out on the unwelded stress-relieved sample reveal a stress level of about 10 MPa, a value which is of the same order of magnitude as the uncertainty in the measurements. Typical examples of the results obtained on the welded samples are given in Figs. 2 and 3 (welding current 150 A).

Figure 2 shows the transverse stress $\sigma_T$ as a function of the distance $x$ to the weld centerline. It appears that $\sigma_T$ reaches a minimum value just outside the weld and a maximum value ($\sigma_{T,\text{max}}$) at a somewhat larger distance from the weld. With further increasing distance, $\sigma_T$ levels off to zero value. In Fig. 3, the longitudinal stress $\sigma_L$ is plotted as a function of the distance $x$ to the weld centerline. The figure shows that after an initial increase $\sigma_L$ reaches a maximum ($\sigma_{L,\text{max}}$), followed by a decrease to negative values, the rate of decrease changing significantly at $\sigma_L = 0$.

Similar results as those presented in Figs. 2 and 3 (welding current 150 A) were obtained for the other specimens (welding current 50, 75, 125 and 175 A). The influence of the welding current on the magnitude and distribution of the stress is summarized in Fig. 4.

In order to determine the degree of plastic deformation due to welding, line broadening was measured for all samples. From the measured line broadening, the microstrain $e$ was calculated using the method described in the previous section. In Fig. 5 the obtained value of $e$ is given as a function of distance to the weld centerline for the sample welded with 150 A. It was found that for all samples the microstrain decreases...
with increasing distance to the weld. Furthermore, it appears that Leds, being a measure of the total amount of plastic deformation, rises with increasing welding current.

In Fig. 6 the macrostructure of one of the welded samples (upper surface) is given. The presence of Lüders bands next to the weld is a rough indication where plastic deformation has taken place. The hardness as a function of distance to the weld centerline for the 150 A sample is given in Fig. 7.

Discussion

Considering the results presented in the previous section, it appears that during arc welding, residual stresses of relatively large magnitude are produced in the vicinity of the weld. As is shown in Figs. 2 and 3, the longitudinal stress \( \sigma_L \) is tensile at a small distance from the center of the weld becoming compressive at larger distances, whereas the transverse stress \( \sigma_T \) remains tensile, its value decreasing with increasing distance from the center of the weld. Both the longitudinal and the transverse stress patterns are characterized by the occurrence of a maximum value at some distance from the weld centerline. The exact locations of these maxima depend on the welding current — Fig. 4.

As will be shown below, the obtained stress distribution can be explained qualitatively in terms of plastic deformation occurring during heating (primary plastic deformation) and subsequent cooling (secondary plastic deformation).

Assume that during welding the plate to be welded is free to move in its plane only. To start with, suppose that the welding is carried out in such a way that the highest temperature reached at the weld centerline is just below the melting point of the material and that only elastic strains occur and no phase transformations whatsoever take place.

During heating, the heated part of the plate will try to expand, which is resisted by the relatively cold surroundings. As a result, compressive stresses in and near the weld will develop and these are balanced by tensile stresses in the outer parts of the plate. The stresses will be biaxial and the highest compressive stress at each position \( x \) will be roughly proportional to the highest temperature \( T(x) \) reached at \( x \). In Fig. 8 a schematic plot of \( \sigma_L \) at \( T(x) \) is given as a function of distance \( x \) to the weld centerline. (The behavior of \( \sigma_T \) will be discussed later.)

After complete cooling, the plate returns to its original state of stress, which means that nothing is changed by welding — Fig. 8.

With the steel used in the present investigation, this will never happen. From the coefficient of thermal expansion and the temperatures occurring during welding, it is easily derived that at the end of the heating period compressive stresses occur that exceed the yield stress. Therefore, the plate material in the vicinity of the weld will deform plastically (primary plastic deformation). Since the yield stress decreases drastically with increasing temperature, the stresses in and very near the weld will hence be practically zero. Melting and solidification of material in the weld does not change this situation. Directly after solidification, i.e., at the start of the cooling period, a stress distribution will exist as depicted in Fig. 9A, the highest compressive stress occurring outside the weld.

During cooling, again stresses will build up which now are tensile and add to the already existing stresses (a rough indication of the stress introduced by cooling is obtained from Fig. 8). In and near the weld, the temperature is still high, i.e., the yield stress is still low, and again plastic deformation occurs (secondary plastic deformation). For each position \( x \), this goes on until a temperature \( T^*(x) \) is reached, below which the actual stress can no longer exceed the yield stress (which increased with decreasing temperature). The situation resulting after cooling from \( T(x) \) to \( T^*(x) \) without and with plastic deformation is shown in Fig. 9B. Stresses developing during further cooling are no longer relaxed by plastic deformation. The final result is shown in Fig. 9C, which compares well with the experimental results — Fig. 3.

The above discussion concerns the behavior of the longitudinal stress \( \sigma_L \). The development of the transverse stress \( \sigma_T \) can be explained in a similar way. The behavior of \( \sigma_T \) as a function of \( x \) must be such that \( \sigma_T \) vanishes at both edges of the plate. Furthermore, the magnitude of \( \sigma_T \) is influenced by the length of the weld with respect to the length of the plate and by the travel speed. In the limiting case of a full length weld, made at infinitely high travel speed, no transverse stresses will develop because of the absence of plastic deformation in the transverse direction. The situation in the present experiment is half way between that of a spot weld and that of a weld of full plate length, and it must be expected therefore that some transverse stresses will develop.

The form of the Lüders bands associated with the primary plastic deformation can be understood as follows: During heating, the only stress present in the base metal at the edge of the weld pool is circumferential. The plastic deformation should lead to relaxation of this stress. The most effective way to relax the circumferential stress is to form a Lüders band in a direction between the longitudinal and transverse directions. This direction will also be influenced by the local yield strength of the material. The Lüders band makes an angle of 45 deg with the surface of the plate. Far away from the weld, where \( \sigma_L \) becomes the most important compressive stress, the bands will bend toward a direction perpendicular to the weld, giving maximum relaxation in the longitudinal di-
Fig. 9 — A — The longitudinal stress \( \sigma_L \) as a function of distance \( x \) to the weld centerline at the highest temperature \( T_h \) reached locally, without and with plastic deformation. The local yield stress \( \sigma_y \) is indicated; B — The longitudinal stress \( \sigma_L \) as a function of distance \( x \) to the weld centerline at an intermediate temperature \( T^*(x) \) (see text) without and with plastic deformation. The local effective yield stress (see text) is indicated; C — The longitudinal stress \( \sigma_L \) as a function of distance \( x \) to the weld centerline after complete cooling to room temperature.

..image:: Fig9a.png
:alt: Longitudinal stress as a function of distance to the weld centerline at the highest temperature.

..image:: Fig9b.png
:alt: Longitudinal stress as a function of distance to the weld centerline at an intermediate temperature.

..image:: Fig9c.png
:alt: Longitudinal stress as a function of distance to the weld centerline after complete cooling.

The longitudinal stress, thus relaxing the longitudinal stress.

Unfortunately, it is difficult to apply the Von Mises flow criterion, which states that the plastic deformation vector is always perpendicular to the local flow surface. As the principal axes of the stress tensor in this stage of the process can hardly be predicted, an appropriate Von Mises plot cannot be constructed.

For the secondary plastic deformation, the Von Mises construction of Fig. 10 can be used. In the case of \( \sigma_L > \sigma_T \) and \( \sigma_y \) being both positive, the flow bands tend to be perpendicular to the weld as the deformation vector is parallel to the weld.

From Fig. 9A and B, it can be concluded that the primary plastic deformation zone (curved bands) should be larger than the secondary plastic deformation zone (perpendicular bands), which is in agreement with the experiment — Fig. 6.

In addition to plastic deformation (primary and secondary), quenching of the plate surface in the vicinity of the weld can lead to the production of residual stresses. As proposed by Wohlfahrt (Ref. 3), the plate surface is so hot that it loses a considerable amount of heat by radiation. If the amount of radiated heat is larger than the amount transported by heat conduction from the interior of the plate to the surface, the surface will cool faster than the interior. This will give rise to residual stresses of the interior, which upon cooling will lead to a compressive stress in the surface layer. As far as the present experimental situation is concerned, it must be realized that the stress measurements were carried out under unstrained conditions (i.e., during stress measurements no external forces were acting on the specimens). Under these conditions \( \int \sigma_L \, dx \, dz \) should be zero for all values of \( y \). If \( \sigma_L \) is independent of \( z \) it follows that \( \int \sigma_L \, dx \, dz = \sigma_L \, dx = 0 \) for all \( y \). In Fig. 11 the value of \( \sigma_L \, dx \) at \( y = 0 \) is plotted as a function of welding current. The figure shows that the integral does not deviate significantly from zero over the range of welding currents used. Apparently, the quench effect does not lead to noticeable residual stresses.

It has also been pointed out by Wohlfahrt (Ref. 3) that in the case of welds in steel, the \( \alpha \)-\( \gamma \) and \( \gamma \)-\( \alpha \) phase transformations can lead to the production of residual stresses.

During heating, ferrite transforms to austenite at \( T_o-\gamma \). Since austenite has a smaller specific volume than ferrite, a tensile stress will be produced in the austenite, which will be partially relaxed.
by means of plastic deformation. When cooling down, the austenite will retrans­form to ferrite at $T_{\gamma-a}$. Due to the corres­ponding volume change, a compres­sive stress will be produced in the fer­rite, which again will be relaxed by plastic deformation. Since in general $T_{\gamma-a} > T_{\gamma-a}$ and the yield strength decreases with increasing temperature, the plastic deformation taking place at $T_{\gamma-a}$ will not fully compensate the plastic deformation taking place at $T_{\gamma-a}$. The net result of transformation and retransformation there­fore is a compressive stress.

Stresses due to transformation and retransformation are found to play a role in the case of alloyed steels (Ref. 3). They are believed to be of minor importance, however, in the case of unalloyed and low-alloy steel, since the transforma­tions in these steels take place at such a high temperature that all developed stresses immediately relax.

An attempt to detect the plastically deformed region by means of increased hardness was not successful. The intru­sions of a Vickers diamond in the plas­tically deformed region showed cracks perpendicular to the longitudinal direc­tion, which apparently are caused by the high residual stress level (tension) in this region. Because of this, the average di­agonal length of a Vickers intrusion be­comes much larger, resulting in deviat­ing (lower) hardness values — Fig. 7.

The line profile analysis, however, indicates the presence of plastic de­formation (up to the x-coordinate where $\sigma_L$ changes sign; compare Figs. 3 and 5), but it does not reflect directly the pri­mary and secondary plastic deforma­tion. This could be due to the recovery of the deformed metal during the thermal cycle.

The measured microstrain curve of Fig. 5 can well be explained with the help of Fig. 12. In Fig. 12A the microstrain due to primary plastic deformation, including the effect of recovery, is shown, whereas in Fig. 12B the effect of secondary plastic deformation is shown. Both contributions can be added to form the pattern of Fig. 12C, which is in qual­itative agreement with Fig. 5.

Conclusions

1) During welding, Lüders bands of different geometry are formed in the vicinity of the weld. An analysis indi-

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**Fig. 10** — Intersection of the Von Mises flow cylinder and the $\sigma_L$, $\sigma_T$ plane.

**Fig. 11** — $\sigma_L$ as a function of welding current $I$.

**Fig. 12** — Effect of plastic deformation and subsequent recovery on the microstrain $e$ as a function of the distance to the weld centerline. $A = e_1$ due to primary plastic deformation; $B = e_2$ due to secondary plastic deformation; $C = e_t = e_1 + e_2$ with recovery after complete cooling of the welded plate.
cates that the plastic deformation associated with these bands is the main cause of the resulting residual stress.

2) The Lüders bands can be easily observed by visual inspection of plates polished before welding, and the amount of plastic deformation can be determined quantitatively from the broadening of x-ray diffraction line profiles.

3) Hardness measurements do not give reliable information about the amount of plastic deformation in the vicinity of welds.

4) The conventional measurement of stresses by x-ray diffraction can be more fully exploited by also dealing with the broadening of the diffraction lines measured. In this way, additional information is obtained about the degree of plastic deformation.

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References