THE UNIFORMITY OF SHOT PEENING INDUCED RESIDUAL STRESS

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ABSTRACT

Two rectangular samples of ASTM SA 508 Class 2 steel, stress relieved and shot peened to 14-16A intensity, were examined in detail to determine the principal macroscopic residual stress distribution. The uniformity of the shot peening induced macroscopic residual stresses with orientation in the plane of the surface and as a function of depth were examined and compared. The microscopic residual stress (plastic deformation) distribution was determined as a function of depth, and compared for the two samples.

The calibration technique to determine the single crystal elastic constants in the (211) direction and verification of the values obtained by comparison with mechanically measured applied stress are discussed.

The results indicate variation in the magnitude of the subsurface compressive macroscopic residual stress with direction in the plane of measurement for either sample of less than 12 ksi. The mean value of the macroscopic stress distributions for the two samples examined differed by less than the same amount at any depth examined. The microstress distribution was found to vary essentially linearly as a function of depth, reaching a negligible amount immediately beneath the microscopically compressive surface layer. The microstress distributions in the two samples examined were identical within the limits of experimental error.

INTRODUCTION

The data presented here were obtained as part of a joint study conducted with the Babcock & Wilcox Corp., under EPRI funding, to determine the most reliable method of residual stress measurement to be applied to large steel pressure vessels. The ultimate goal of the study is to provide a definitive comparison between the residual stress distribution determined in the near surface layers of pressure vessel steels by x-ray diffraction and mechanical hole drilling techniques.

The results presented here consist of the x-ray diffraction portion of the first phase of the comparison of measurement methods. The data obtained for the determination of the macroscopic residual stress have been further reduced to determine the microscopic stress distribution in order to quantify the depth of the cold worked layer produced by shot peening, and its uniformity between identically prepared specimens.

Shot peening residual stress distributions were chosen for the comparison of x-ray diffraction and mechanical residual stress measurement methods because of the anticipated uniformity both with direction on the specimen surface and between specimens similarly prepared. Also, the shot peening residual stress distribution should vary gradually and predictably as a function of depth, minimizing the complications possible in applying and comparing the two different methods of residual stress measurement.

SAMPLE PREPARATION

Four samples were manufactured from ASTM type SA 508 Class 2 pressure vessel steel with overall dimensions of 2.5 x 2.5 x 0.75 in. Any residual stresses present in the samples were eliminated by stress relieving for four hours at 1125° F. The mechanical properties and hardness of the stress relieved material were determined from five repeat tests. Details of the nominal composition, heat treatment, and mechanical properties after sample preparation are presented in Table I.

Residual Stress for Designers & Metallurgists,
American Society for Metals, Metals Park, OH 1981, pp. 151-168

TABLE I
SAMPLE MATERIAL AND PROPERTIES

ASTM SA 508 Class 2 Pressure Vessel Steel

Nominal Composition (%):

C	Mn	P	S	Si	Ni	Cr	Mo	V
.27*	.7	.025*	.025*	.25	.75	.35	.6	.05*

*Max. Amounts

Heat Treatment

I. Heat to 1125° F. at 100° F./Hr.
II. Hold at 1125° F. for 4 hrs.
III. Cool to 600° F. at 50° F./Hr.
IV. Furnace cool from 600° F.

Mechanical Properties:
(Average of Five Tests)

Ultimate Tensile Strength	86.2 ksi
Yield Strength	63.8 ksi
Hardness	86. Rb
Elastic Modulus	29.0 x 10^6 psi
Poisson's Ratio	0.29 – 0.30

After stress relieving, the four plate samples were shot peened on one square face to a 14-16A Almen intensity for 200% coverage with steel shot. During the shot peening process, the samples were held on a horizontal rotating table under vertical oscillating shot streams in an attempt to produce a near-surface compressive residual stress distribution which would be as uniform with direction and as consistent between specimens as possible. Details of the peening parameters are shown in Table II.

TABLE II
SHOT PEENING PARAMETERS

Intensity	14-16A
Coverage	200%
Shot	330 Steel
Pressure	70 psi

Two of the four samples were retained for later measurement with the mechanical hole drilling technique, and two were employed in this investigation. A region approximately 1.25 in. square in the center of the two test specimens described here was electropolished to obtain the subsurface x-ray diffraction data. The remaining surface area of the specimens was preserved in its original condition for subsequent hole drilling.

TECHNIQUE

Determination of Single Crystal Elastic Constants

The macroscopic residual stress in a body can be determined employing x-ray diffraction techniques (assuming a condition of plane stress exists in the surface being examined) by measuring the strain in a specific crystallographic direction for groups of crystals which are aligned at two or more known angles with respect to the sample surface. The strain in the crystal lattice is calculated from the lattice spacing of a specific (hkl) set of atomic planes, the spacing of which depends upon the angle of tilt, ψ, and the stress being measured, $\sigma\phi$ as,

$$d(\psi) = \left(\frac{E}{1+\nu}\right)_{(hkl)} d_o \sigma\phi \sin^2\psi - \left(\frac{\nu}{E}\right)_{(hkl)} d_o(\sigma_1 + \sigma_2) + d_o \quad (1)$$

where d_o is the unstressed lattice spacing for the (hkl) planes being examined, and $(\sigma_1 + \sigma_2)$ is the sum of the principal stresses in the surface at the point of measurement. The stress measured, $\sigma\phi$, is oriented at some angle ϕ to the maximum normal principal stress vector σ_1. The quantities ν and E are Poisson's ratio and the modulus of the elasticity, respectively, for the (hkl) direction in the crystal lattice. These single crystal elastic properties may differ significantly from the bulk values determined in a mechanical test. Because the last two terms in Equation 1 are constants depending upon the material properties, lattice spacing, and principal stresses at the measurement site, the lattice spacing is a linear function of the quantity $\sin^2\psi$.

Although the quantity d_o is unknown, the sum of the last two terms in the equations (the value of d measured at $\psi = 0$) differs by only approximately 0.1% from d_o. Making this approximate substitution, it is possible to solve for the stress present if the lattice spacing d is measured at $\psi = 0$ and for some other known value of ψ, typically on the order of 45 deg. The stress is then given by,

$$\sigma\phi = \frac{d(\psi) - d_\perp}{d_\perp}\left(\frac{E}{1+\nu}\right)_{(hkl)} \frac{1}{\sin^2\psi}$$

where d (ψ) and $d_\perp$ are the (hkl) lattice spacings measured at some known angle ψ and perpendicular to the surface of the specimen.

The bulk elastic constants obtained from mechanical test data represent averages over randomly oriented crystallographic directions, and have been found to differ by as much as 40%, as a result of a elastic anisotropy in the crystal lattice, from the corresponding values in a specific (hkl) direction.(1) A variety of methods have been developed for determining single crystal elastic constants employing various sample geometries producing regions of uniform stress on a surface, which when loaded on a diffractometer can be used to measure the change in lattice spacing of the (hkl) planes to be employed for stress measurement as a funtion of applied stress. The slope of the linear dependence of the change in lattice spacing as a function of applied stress is proportional to $E/(1 + \nu)$ in the (hkl) direction. In this investigation, a simple rectangular beam with nominal dimensions of 4.0 x 0.75 x 0.100 in., instrumented with a full bridge of strain gages, the output of which provides the average stress on the surface of the beam in the area irradiated during measurement, was loaded in four-point bending to provide the necessary data. Details of the technique used, and tables of data for specific alloys illustrating the effects of elastic anisotropy have been described previously.(1)

A rectangular beam manufactured from SA 508 Class 2 steel, as described above, was instrumented with strain gages, calibrated by dead weight loading, and placed in four-point bending on the diffractometer to applied stress levels of 5.0, 27.5, and 50.0 ksi. The change in the lattice spacing measured between the ψ angles 0 and 45 deg. was determined six times at the minimum and maximum applied stress levels, and twice at the mid-level to check linearity. The data were collected in this fashion to minimize the uncertainty in the slope of the linear relationship between applied stress and Δd (211). The resulting set of data are presented in Figure 1. Employing Equation 2 for known changes in stress, the value of $E/(1 + \nu)$ in the (211) direction for SA 508 Class 2 steel in the stress relieved condition was determined to be $27.4 \pm 0.2 \times 10^6$ psi, where the error shown is one standard deviation based upon the least-squares fit to the data. Because of elastic anisotropy in the crystal lattice, the value of $E/(1 + \nu)$ obtained by direct measurement in the (211) direction was found to be approximately 22% higher than the value obtained from the bulk elastic constants shown in Table I.

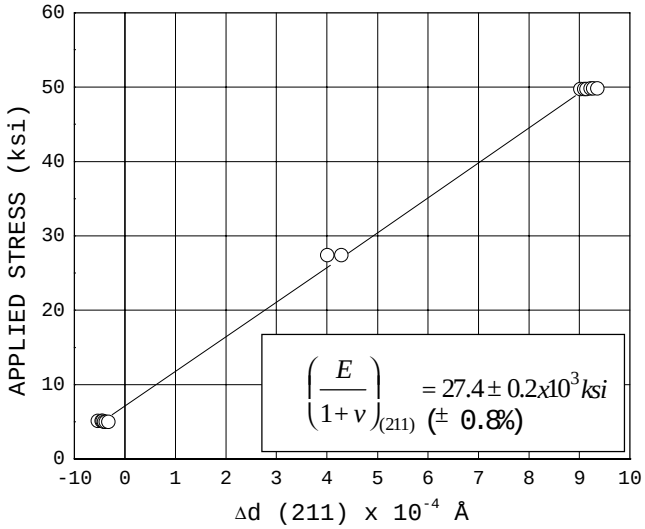


Fig. 1 – Determination of $E/(1+\nu)$ (211) Δd (211) vs. applied stress, SA 508 Class 2 Steel.

Because the difference between the elastic properties in (211) direction and the bulk values was higher than generally anticipated for steels, effort was made to verify the experimental determination of the elastic constants. A "stiff-back" specimen which could be loaded directly in tension was prepared from the same material, and instrumented with strain gages so that the change in stress on the front face of the rectangular specimen could be measured both mechanically and by x-ray diffraction. The sample used is shown in the top of Figure 2, with strain gages on both sides of the region irradiated during x-ray diffraction applied stress measurement. The stiff-back tensile specimen was placed in its loading fixture on the diffractometer, and the residual stress present on the surface of the ground sample was measured with only sufficient load to hold the sample in position. Tensile stresses of 12.3, 21.8, and 34.4 ksi were then applied to the specimen, and the sum of the residual plus applied stress was measured by x-ray diffraction.

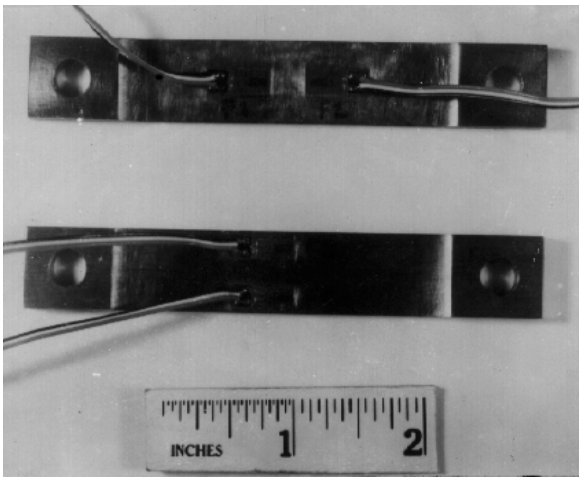
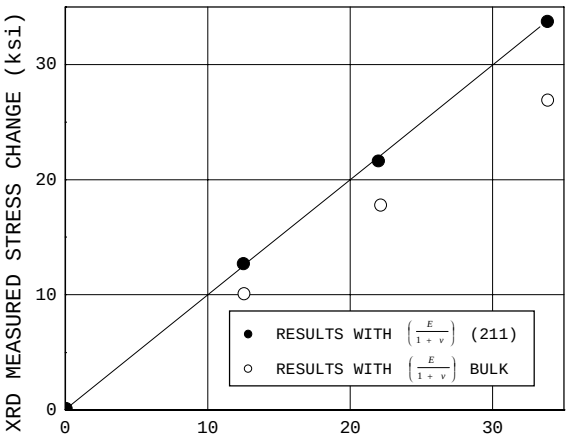


Fig. 2



Mechanical Applied Stress Change (ksi)

Measured Stress (ksi)		Change in Stress (ksi)	
Mech.	XRD	Mech.	XRD
0.1	-64.2	0	0
12.4	-51.6	12.3	12.6
21.9	-42.7	21.8	21.5
34.5	-29.8	34.4	34.4

Fig. 3 – Comparison of mechanical and x-ray diffraction measured applied stress, Type SA 508 Class 2 Steel.

The applied stress measured mechanically, the sum of residual plus applied stress measured by x-ray diffraction, and the change from the initial values are shown both in tabular and graphic form in Figure 3. The results which would have been obtained had the bulk values of the elastic constants been employed in the comparison of mechanical and x-ray diffraction applied stress measurement are shown as open circles with respect to a unit slope line which would

represent perfect agreement between the mechanical and x-ray diffraction measurements. The x-ray measured change in stress was found to be within 0.5 ksi of the mechanically measured values, verifing the single crystal elastic constants obtained in four-point bending.

Macroscopic Residual Stress Measurement

In order to determine the principal residual stresses in planes parallel to the surface as a function of depth through the compressively stressed layer produced by shot peening, x-ray diffraction residual stress measurements were made in directions parallel, at 45 deg., and at 90 deg. to a reference edge of the square steel specimens. Three stress measurements were made in this rectangular configuration on both specimens at the surface and at nominal subsurface depths spaced at approximately 0.002 in. increments to a depth of 0.012 in., and at increments of nominally 0.005 in. from 0.015 to 0.025 in. beneath the surface. The stress measurement depths were chosen to best define the anticipated -compressive residual stress distribution produced by shot peening.

X-ray diffraction residual stress measurements were made by the Two-Inclined Angle technique as recommended by the Society of Automotive Engineers[2]. No attempt was made to determine the linearity of the dependence of lattice spacing upon $\sin^2\psi$, because earlier investigations of shot peened steel and aluminum have indicated that a nearly ideal linear relationship exists on surfaces deformed by shot peening.[3] Measurements were made employing the diffraction of chromium K-alpha radiation from the (211) planes of the BCC structure of the type SA 508 Class 2 steel. The diffraction peak angular positions were determined for ψ angles of 0 and 45 deg. employing a five-point parabolic regression procedure after correction of the raw intensity data for the effects of Lorentz-polarization, absorption, and for a linearly sloping background intensity. The apparatus used was a modified G.E. horizontal goniometer fixtured with residual stress measurement apparatus designed by the author suitable for performing stress measurements in a parafocusing geometry. Details of the diffractometer fixturing are outlined in Table III.

TABLE III
DIFFRACTOMETER FIXTURING

Incident Beam Divergence	3.0 deg.
Receiving Slit	0.5 deg.
Detector System	Si(Li)
Counts per Point	10^5
Psi Rotation	0.0-45.0 deg.
Irradiated Area	0.25 x 0.25 in.

The incident beam slit height was adjusted to provide a nominally 0.25 x 0.25 irradiated area on the sample surface. The radiation was detected employing a solid state Si(Li) detector system set for 90% acceptance of the chromium K-alpha energy.

Material was removed for subsurface measurement by electropolishing in a sulfuric-phosphoric-chromic acid electrolyte minimizing the possible alteration of the subsurface residual stress distribution as a result of material removal. All macroscopic residual stress data obtained as a function of depth in this investigation were corrected for the effects of penetration of the radiation employed for residual stress measurement into the subsurface stress gradient, and for stress relaxation which occurred as a result of material removal after the method of Moore & Evans, assuming that the specimen behaved as an infinite flat plate.[4]

Systematic error due to instrument misalignment was monitored during the course of this investigation employing a powdered iron zero stress reference sample, and found to be less than ± 2 ksi.

The macroscopic residual stresses measured in the three directions with respect to the reference edge of the specimen in each plane parallel to the surface were combined employing Mohr's circle for stress to calculate the minimum and maximum normal principal stresses, the maximum shear stress and their orientation at each level beneath the surface. The resulting minimum and maximum normal residual stresses represent bounds upon the possible residual stress for any orientation at a given level beneath the surface of the sample.

Microstress Determination

The microscopic residual stress at each level beneath the surface of the two samples was determined by measuring the full width at half maximum intensity (FWHM) of the (211) diffraction peak in the $\psi = 0$ orientation. Previous investigations of the microstress behavior in Rene 95 and Inconel 600, face-centered-cubic nickel base alloys, have shown a linear dependence of the FWHM of the (420) diffraction peak as a function of known amounts of plastic strain.[5] No calibration curves were obtained to define the dependence of the (211) diffraction peak width upon percent plastic strain for the SA 508 Class 2 steel investigated here.

Because the material was in an annealed condition, it was assumed that no carbon gradient existed in the surface layers investigated, and that line broadening due to straining of the crystal lattice as a result of martensite formation was eliminated. The high back reflection angle of 156 deg. at which the (211) diffraction peak occurs with chromium K-alpha radiation should result in diffraction peak broadening due to lattice microstrain only, and not as a result of a small crystallite size. It was, therefore, assumed that the normalized (211) diffraction peak width would vary linearly with plastic strain, and would itself be an adequate indication of the degree of plastic deformation occurring in the surface layers investigated.

The (211) FWHM was determined by calculating the width of the parabola fitted to the top 15% of the diffraction peak in the analysis to determine the diffraction peak positions, at half of the diffraction peak height above the existing background intensity. These peak widths were then normalized to produce relative plastic deformation data.

RESULTS AND DISCUSSION

The results of the independent x-ray diffraction residual stress measurements made parallel, and at 45, and 90 deg. to the reference edge of the specimens are presented for sample 1 and sample 2 in Figures 4 and 5, respectively. Although the absolute accuracy of the results shown here, including the systematic errors (such as uncertainty in the single crystal elastic constants and instrument alignment) and random errors (due to uncertainty in the diffraction peak positions and in positioning of the sample itself) is typically on the order of ± 5 ksi the repeatablity, considering random errors only, has been demonstrated to be approximately ± 2 ksi on steel samples of this type.[6]

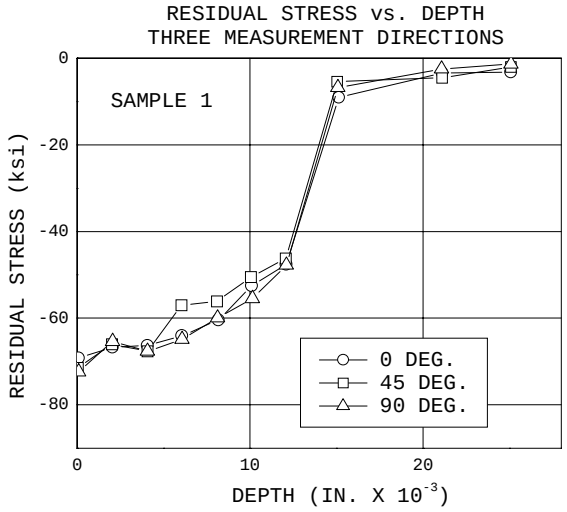


Fig 4 – Shot peened (14-16A) SA 508 Class 2 Steel, Sample 1.

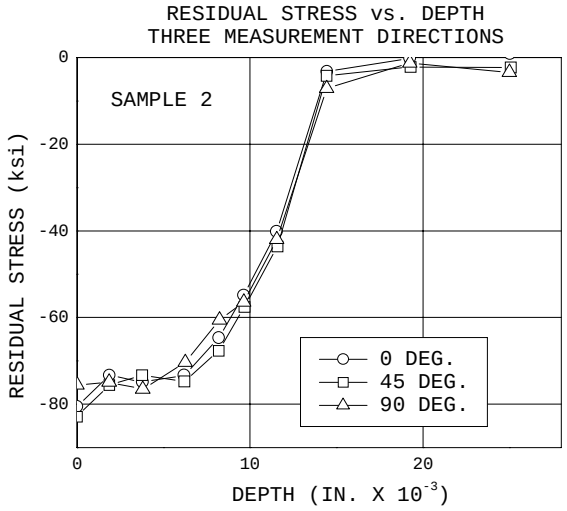


Fig. 5 – Shot peened (14-16A) SA 508 Class 2 Steel, Sample 2

Examination of the data obtained from sample No. 1, and shown in Figure 4, indicates variation of the residual stress with direction from the surface to a depth of 0.004 in. and from a depth of 0.012 to 0.025 in. which is within the anticipated range of repeatability. The results from depths of 0.006 to 0.010 in., notably in the direction perpendicular to the reference edge of the specimen, indicate a variation in subsurface residual stress magnitude with direction of as much as 10 ksi, in excess of the anticipated repeatability of the technique.

The results for the three directions of measurement shown in Figure 5 for sample 2, indicate a similar pattern with less dependence upon direction. At the surface of the specimen and at depths of nominally 0.006 and 0.008 in., the range of stress over the three directions measured appears to exceed slightly the anticipated repeatability. At the remaining levels investigated, the magnitude of the subsurface compressive stress measured on sample 2 appears to be independent of direction.

The data in Figures 4 and 5 indicate virtually identical patterns of residual stress obtained by simultaneously shot peening the two samples investigated. An exception occurs at a depth of 0.012 in., at which sample No. 1 appears to be approximately 10 ksi more compressive than sample 2. The results indicate surface compressive stresses for both samples on the order of -80 ksi, approximately 16 ksi above the bulk yield strength measured for this material in the stress relieved condition prior to peening. The magnitude of the compressively stressed layer equals or exceeds the bulk yield strength to a depth of 0.008 in. in both samples. This phenomenon is commonly observed in samples which have been shot peened, ground, machined or otherwise plastically deformed on the surface, leaving near surface layers which have been work hardened sufficiently to increase the yield strength, in some cases considerably beyond the bulk yield strength measured for the base metal.

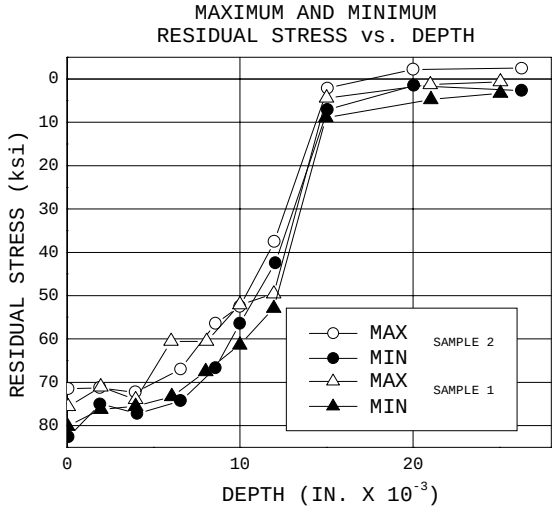


Fig. 6 – Shot peened (14-16A) SA 508 Class 2 Steel

The data shown in Figures 4 and 5 for the three independent stress measurement directions were combined to calculate the minimum and maximum normal principal stresses at each level which are shown overlayed for the two samples in Figure 6. These data indicate the bounds upon the residual stress values at each level for any possible orientation in a plane parallel to the surface. The bounds imposed by the principal stress values were found to overlap at

all depths except at 0.012 in. beneath the surfaces of the two samples where the difference on the order of 10 ksi, previously mentioned, was observed. The greatest possible variation of subsurface stress with angular orientation on the surface appears to be approximately 12 ksi at a depth of 0.006 in. on sample No. 1. For all practical purposes, the two samples could be said to be identical and free from dependence of the subsurface residual stress distribution upon direction within bounds of ± 8 ksi in the worst case.

Microstress Distribution

The results of the analysis of the relative amount of plastic deformation as a function of depth into the surface of samples No. 1 and 2 based upon the normalized width of the (211) diffraction peak at half height (FWHM) are shown in Figures 7 and 8, respectively. The values of the width of the (211) diffraction peak measured for $\psi = 0$ in each of the three orientations with respect to the reference edge of the specimen, are shown at each depth. According to the model assumed to hold (Equation 1) for the lattice spacing as a function of ψ and ϕ, the lattice spacing itself, and therefore, the mean position of the (211) diffraction peak should be independent of ϕ at $\psi = 0$. It would seem reasonable to assume that the distribution of d for various crystallites in that orientation around the mean value, and therefore, the width of the (211) diffraction peak, would also be independent of the angle ϕ for the $\psi = 0$ orientation. The agreement obtained for the three measurement directions at depths equal to, or greater than, 0.012 in., below the plastically deformed layer, indeed support this conclusion. Whether the variation seen at lesser depths is due to experimental error or a real physical phenomenon is not clear.

The results shown in Figures 7 and 8 indicate a nearly linear reduction in the relative amount of plastic deformation from the surface of the specimen to a depth of approximately 0.012 in. This depth corresponds to the maximum depth of the layer containing significant compressive macroscopic residual stress. The magnitude of the relative plastic deformation at each depth was found to be nearly equivalent for the two specimens within the observed scatter. At depths greater than 0.012 in., in the annealed material, the results are virtually identical.

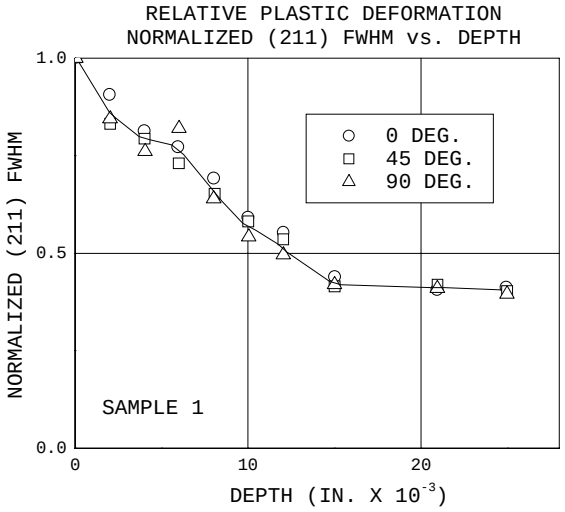


Fig. 7 – Shot peened (14-16A) SA 508 Class 2 Steel, Sample 1

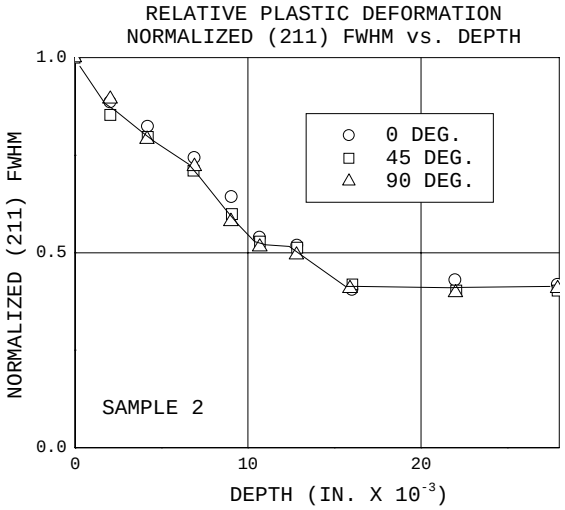


Fig. 8 – Shot peened (14-16A) SA 508 Class 2 Steel, Sample 2.

CONCLUSIONS

The results of this investigation into the nature and uniformity of the macroscopic and microscopic subsurface residual stress distributions produced by shot peening type SA 508 Class 2 pressure vessel steel indicate a dependence of the magnitude of the compressive residual stress upon orientation in a plane parallel to the surface of ± 5 ksi in the worst case. Comparison of the minimum and maximum normal principal residual stresses as a function of

depth in two identically prepared samples indicates uniformity between samples of the shot peening induced mean residual stress magnitude within ± 8 ksi. Repeatability of the measurement method used has been demonstrated in samples of this type to be on the order of ± 3 ksi. The uniformity between samples and the dependence of the magnitude of the macroscopic residual stress upon orientation are both therefore, only slightly greater than the experimental error inherent in the measurement method used.

The microscopic residual stress distribution, expressed in terms of the relative plastic deformation indicated by diffraction peak broadening, shows a linear dependence of the degree of plastic deformation upon depth extending to a maximum depth equal to the depth of the compressively stressed layer produced by shot peening. Comparison of the microscopic residual stress results obtained on two identically prepared samples indicates identical microscopic residual stress distributions within the error in the experimental technique used.

The uniformity of the shot peening process, and the lack of dependence of the magnitude of the compressive stress upon orientation, indicate that samples prepared in this manner are suitable for comparison of mechanical and x-ray diffraction methods of residual stress measurement as a function of depth in the near surface layers of type SA 508 Class 2 steel.

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