COMPARISON OF SHOT PEENING RESIDUAL STRESS DISTRIBUTIONS IN A SELECTION OF MATERIALS

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ABSTRACT

The residual stress distributions in a variety of engineering materials which have been shot peened to different intensities have been measured using non-destructive neutron diffraction methods. The results reveal that the same characteristic shape of stress distribution is obtained in all the materials with the surface in compression and the existence of a tensile peak subsurface. Surface compression approaching yield was observed in some cases which was balanced by subsurface tension of about a third of the value. The results are discussed in terms of the mechanical properties of the materials. The implications of the tension for initiating fatigue failures from the interior of shot-peened components is also discussed.

KEYWORDS

Shot peening, residual stress, neutron diffraction.

INTRODUCTION

Shot peening is an effective method of improving the fatigue life of engineering components by introducing compressive residual surface stresses which can approach the yield stress of a material, Townsend (1) and Wang (2). The affected region is normally within 1mm of the surface. Careful selection of the peening parameters is necessary to produce the optimal residual stress distribution for maximum fatigue life improvement. Verification of the residual stresses introduced by shot peening has largely been carried out using the X-ray diffraction technique in conjunction with progressive material removal to obtain the depth profile, Fuchs (3) and Li (4). However there is some uncertainty over the accuracy of this technique since compensation is required to allow for the material removal. For reliable predictions of the fatigue performance of shot peened components, it is essential to have accurate information of the through thickness residual stress distribution. Neutron diffraction, which is non-destructive, has been used in this investigation to determine these stresses. Experimental results are presented of neutron diffraction measurements of residual stress distributions that have been produced by different intensities of peening in a steel, and also in alloys of nickel, titanium and aluminium.

NEUTRON DIFFRACTION FOR MEASURING ENGINEERING STRAINS

Background

Neutrons of sufficient intensities for stress measurements are produced in either nuclear reactors or synchotron radiation sources by the process of nuclear fission or spallation respectively. Samples must therefore be taken to a neutron source of which there are a limited number worldwide. The neutron diffraction measuring

method is identical in principle to the well known X-ray stress measuring technique which is described in detail by Noyan (5). Diffraction stress measurements can only be performed on crystalline material. Because of the limited penetration properties of X-rays only the surface can be examined with this method. Typically the maximum depths of penetration is of the order of 50 microns. Consequently, to obtain stress with depth measurements, the surface of the material must be progressively eroded by electrochemical methods so as to avoid the introduction of residual stresses by the material removal process. The stresses are then measured at the exposed depth. In contrast, because of their neutral charge, neutrons can penetrate several centimetres in some metals depending on the neutron scattering properties of the material. In steels for instance neutron total path lengths of incident and scattered neutrons can be of the order of 40 mm. As such, a direct measurement of the strain tensor in the plane of interest is measured as shown in Fig. 1. For meaningful measurements by neutron diffraction the material must have low neutron absorption properties and be a good neutron scatterer. All the materials used in this investigation satisfy this criterion, although aluminium and titanium based materials may require long neutron count times for satisfactory results.

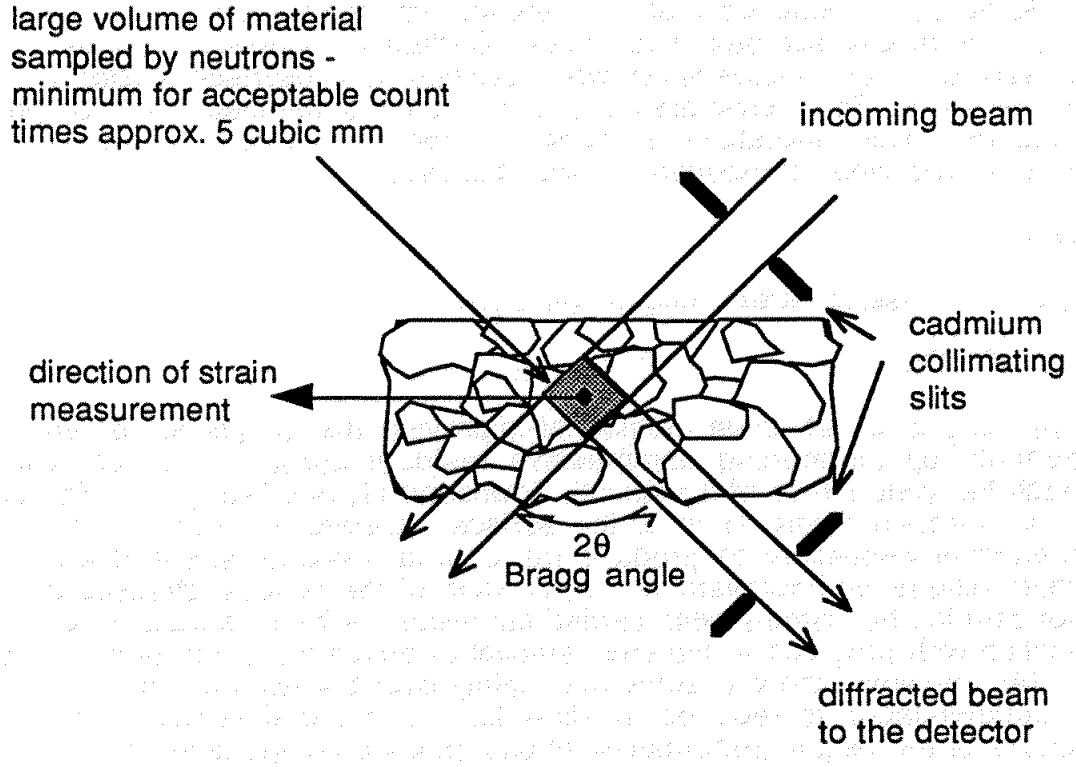

Figure 1 Principle of the Neutron Diffraction Method

Theory and Application to Strain Measurements

The neutron diffraction stress measuring technique is described by Allen (6). As with all stress measuring techniques, strains are measured and converted into stresses using appropriate elastic stress/strain relations. The interatomic spacing of a selected reflection is used as the strain gauge with neutrons diffracted according to the Bragg equation;

$$m\lambda = 2d_{(hkl)}\sin\theta_{(hkl)} \quad (1)$$

where λ is the neutron wavelength, $d_{(hkl)}$ is the interatomic spacing of the (hkl) atomic plane and $2\theta_{(hkl)}$ is the Bragg angle (see Fig. 1) Differentiation of the above equation for a fixed wavelength gives the strain tensor ε_i for a neutron beam diffracted through an angle $2\theta_i$;

$$\varepsilon_i = -(\theta_i - \theta_0)\cot\theta_0 \quad (2)$$

$2\theta_0$ is the diffraction angle for unstrained material and can be obtained by measurements in unstrained material. For a general three dimensional stress system 6 measurements of direct strain are required. This allows the principal strains to be determined as well as the shear components. If the principal directions are known only three orthogonal strains need to be measured and these can be used to obtain the principal stresses using eqs. (3) below, Smith (7).

$$\sigma_x = \frac{E}{(1+\nu)(1-2\nu)}\{(1-\nu)\varepsilon_x + \nu(\varepsilon_y + \varepsilon_z)\}$$

$$\sigma_y = \frac{E}{(1+\nu)(1-2\nu)}\{(1-\nu)\varepsilon_y + \nu(\varepsilon_x + \varepsilon_z)\} \quad (3)$$

$$\sigma_z = \frac{E}{(1+\nu)(1-2\nu)}\{(1-\nu)\varepsilon_z + \nu(\varepsilon_x + \varepsilon_y)\}$$

In these equations E is the elastic modulus and ν Poissons ratio for homogeneous material. In many engineering problems, the principal directions can be identified by observing the geometry of the component and also the nature of the loading. Measuring in the principal directions only significantly reduces the time to acquire the neutron data. For a more accurate determination of stresses and considering that most crystal systems are anisotropic, the value of the elastic properties E and ν used in eqs. (3) should be replaced by the elastic properties for the (hkl) reflection used for the diffraction measurements, $E_{(hkl)}$ and $\nu_{(hkl)}$. Values of $E_{(hkl)}$ can be obtained by calibration measurements, Allen (8) and Ezeilo (9) or by calculation, Noyan (5). For engineering purposes it is generally convenient to measure strains using a reflection which gives a good neutron intensity and has a small peak width in the diffraction profile. Resolution is also improved if the diffraction angle of the selected (hkl) reflection is close to 90° to avoid elongated sampling volume shapes and also if the (hkl) elastic modulus is close to the bulk value.

DESCRIPTION OF SHOT PEENED SAMPLES

A variety of engineering materials were examined which had received a selection of peening intensities. The materials comprise 2 nickel superalloys, a titanium alloy, a steel and an aluminium alloy. The sample thicknesses were all greater than 5 mm and were therefore considered to be thick samples. Counterbalancing bending stresses were therefore not expected to contribute significantly to the final residual stress profile. The shot peening conditions are given in table 1 below with the peening intensities given in terms of the Almen intensity strip A. An increase in the Almen intensity implies a higher peening intensity for a particular strip.

Table 1 Shot Peening Parameters

Alloy	shot size	Intensity	Coverage (%)
Udimet 720 I	MI 330 R	18 - 20 A	200
Udimet 720 II	M 110 H	6 - 8 A	125
Waspaloy	MI 230 R	14 - 16 A	200
Jethete	MI 230 R	14 - 16 A	200
IMI 834	MI 330 R	18 - 20 A	200
Al 7071	S 110	8 - 12 A	125

EXPERIMENTS PERFORMED

Neutron diffraction residual stress measurements were made on all the samples described above at different neutron reactor sources (see table 2 below).

Table 2 Diffraction Information

Alloy	base element	crystal structure	neutron spectrometer and wavelength (Å)	(hkl)	sampling volume mm^3
Udimet 720 I,II	Ni	FCC	D1A, Grenoble, France, 1.9	(311)	1 * 1 * 25
Waspaloy	Ni	FCC	HB4, Petten, Holland, 1.9	(311)	1 * 1 * 25
Jethete	Fe	BCC	G5.2,LLB,Saclay, Fr, 2.9	(211)	0.5 * 10 * 10
IMI 834	Ti	HCP	D1A, Grenoble, France, 1.9	(11-22)	1 * 1 * 25
Al 7071	Al	FCC	D1A, Grenoble, France, 1.9	(311)	1.5 * 2 * 20

Cadmium masks were inserted in both the incoming and the diffracted beams to define a matchstick shaped sampling volume as shown in Fig. 2. Only for the steel sample was a box like sampling volume used. The sample was traversed across the neutron beam, with smaller steps in the surface region and larger ones in the interior, see inset Fig. 2. Measurements were made typically in one in-plane direction (x or y in Fig. 2 making use of the fact that $\varepsilon_x=\varepsilon_y$ for peened material) and in the z direction, these representing the principal directions. Diffraction peak centres were obtained by fitting the peak profile with a Gaussian function. In general very good diffraction data were obtained with peak fitting routines giving strains to within 50 $\mu\varepsilon$. The stress free diffraction angle was in most cases obtained by either measurements on unstressed samples or by taking measurements in the interior of the sample away from the peened surface, using a wide neutron beam.

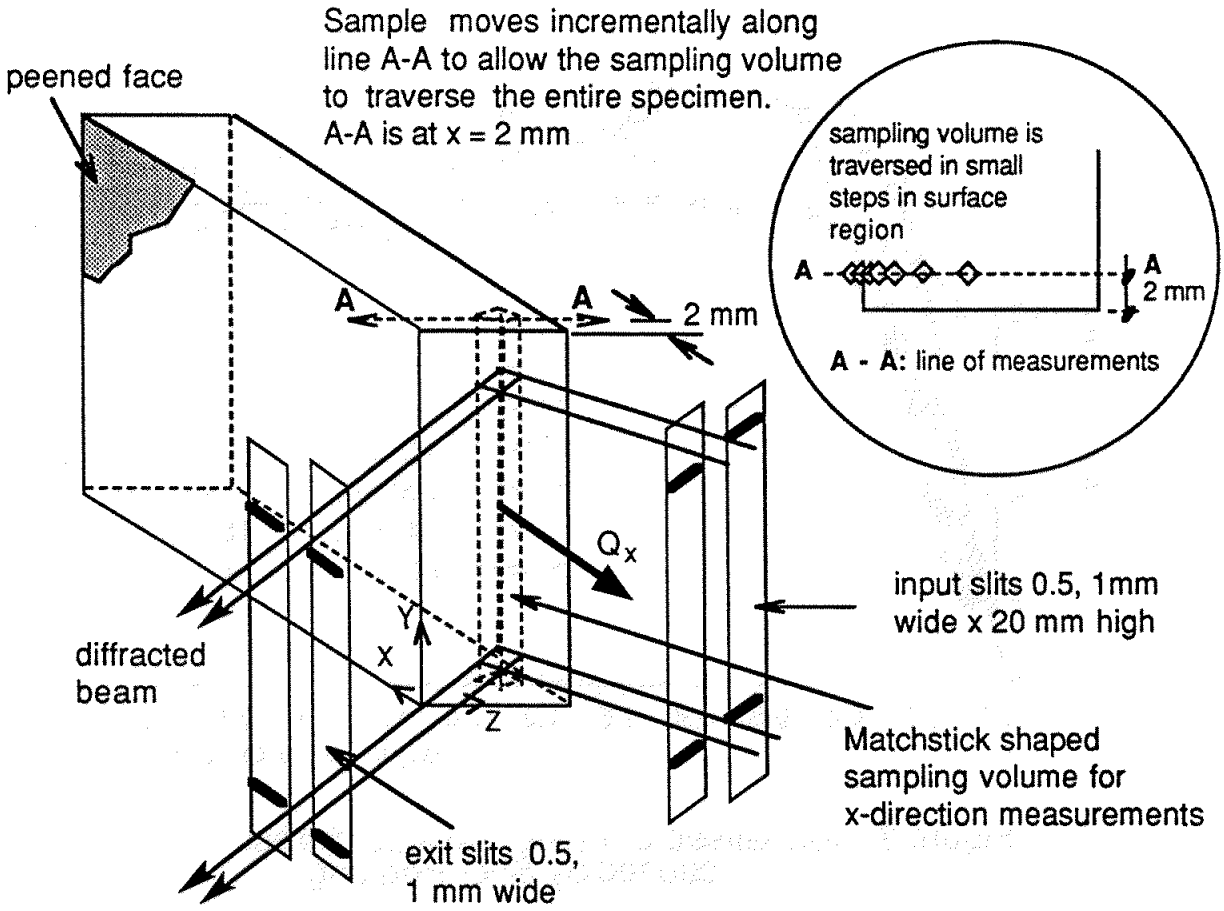

Figure 2 Sampling Volume for Shot-Peened Plate

RESULTS

The residual stress with depth profiles normalised with respect to material yield stress are shown in Fig. 3. This figure indicates that all the profiles have the same characteristic shape regardless of the material or the peening intensity used. The results, which are summarised in table 3 below, indicate surface compression balanced by subsurface tension of about a third of the maximum surface compression. It is seen that an increase in peening intensity, for the same material (Udimet 720), causes an increase in the maximum compressive and tensile residual stresses measured. The tensile peak is also pushed deeper below the surface. At the highest intensity the maximum surface compression is in the region of the material yield strength. It is also observed that the residual stress profile is dependent on the

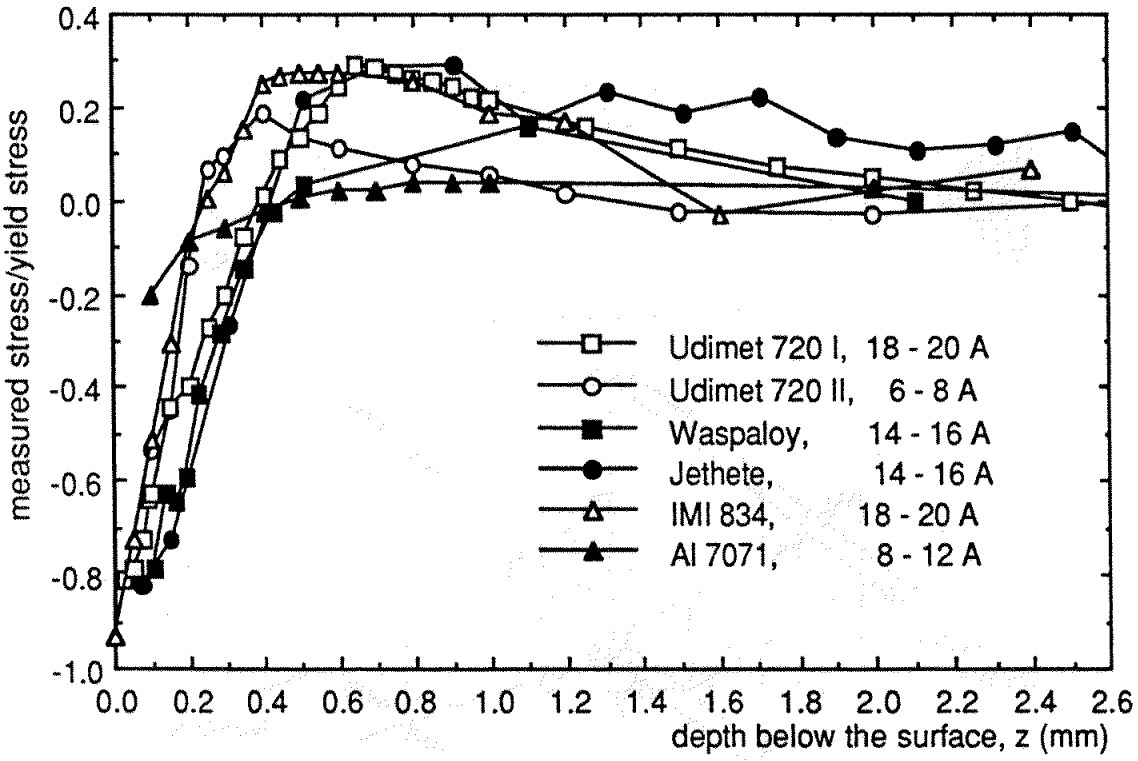

Figure 3 Normalised In-Plane Residual Stress Distributions caused by Shot Peening

material properties for the same peening parameters. For instance, for the same surface strains it is harder to yield the IMI 834, which has an elastic modulus of about half that of Udimet 720 and a yield stress almost as high as Udimet 720, than it is to yield the Udimet 720. As such, it is expected that the compressive stresses in the nickel will extend to a greater depth as observed in Fig. 3. The Waspaloy and the Jethete which have similar moduli and yield stresses show very similar stress distributions although insufficient measurements were made on the Waspaloy to accurately determine the tensile peak. Little residual stress is observed in the centre of the plate for any of the materials. Table 3 below summaries the main results from the experiments.

Table 3 Shot Peening Results

Alloy	Intensity (%)	σ_{max}/σ_{yp} compression	σ_{max}/σ_{yp} tension	% maximum tension/compression	depth of max tension (mm)
Udimet 720 I	18 - 20A	-0.81	0.293	36	0.65
Udimet 720 II	6 - 8A	-0.54	0.19	35	0.40
Waspaloy	14 - 16A	-0.79	0.16	20	1.10
Jethete	14 - 16A	-0.82	0.291	35	0.90
IMI 834	18 - 20A	-0.93	0.275	29	0.50
Al 7071	8 - 12A	-0.20	0.041	21	0.90

DISCUSSION

The magnitude of the subsurface tension has not really been appreciated in the past although it is seen here that it can be as high as 36% of the surface compression.

Comparison of the results on the two Udimet 720 samples shows that the higher the peening intensity the more compressive the surface stress and therefore the higher the subsurface tension. This is significant as subsurface crack initiation and growth has been known to occur beneath the surface of components, Almen (10) and Shin (11). The accuracy of the magnitude and distribution of the stresses measured is dependent on the size of the sampling volume used. Large sampling volumes reduce the resolutional accuracy of the stresses measured. This explains why the measured surface compression and subsurface tension in the aluminium sample, which are early results before the technique was developed, are relatively low. Also in the case of the Waspaloy sample, there were not enough data points in the tensile stress region to accurately obtain the magnitude and position of the subsurface tension. Because of the finite volume of material needed for neutron measurements, it is difficult to produce precise measurements within the first 100 μm of the surface. As a consequence the characteristic increase in compression from the surface to a point just below the surface as measured by X-rays is often not detected with neutrons.

CONCLUSIONS

Residual stresses have been obtained in a variety of engineering materials which have experienced different shot peening treatments. These stresses show surface compression balanced by subsurface tension of about a third of the surface compressive stress value. The stress distributions obtained demonstrate the capability of the neutron diffraction stress measuring technique to determine shot peening residual stresses.

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