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The Effect of Shot Peening on the Fatigue Behavior of Alloy 7075-T6

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Shot peening has long been recognized as an excellent process for increasing the fatigue life of a metal part. Shot peening increases fatigue life by creating a residual compressive stress near the specimen surface. The effect of this residual compressive surface stress on the distribution of applied stress in the specimen depends on the loading mode. In reverse bending, the peak tensile stress in the specimen is reduced and shifted away from the specimen surface, Fig. 1(*a*). However, in axial fatigue ($R = -1$), the peak tensile stress is increased over the applied tensile stress, and is displaced inward from the specimen surface, Fig. 1(*b*). Nevertheless, in both cases the benefit of a lower tensile stress near the specimen surface may be offset by a roughened surface creating more fatigue crack initiation sites and possibly surface microcracks. As a result, the fatigue life of shot peened parts is a compromise between a lower surface tensile stress and a greater number of fatigue crack initiation sites. When conducted properly, shot peening leads to an increase in specimen fatigue life and long-life fatigue strength over unpeened specimens. However, the degree to which fatigue life and fatigue strength are affected is found to be highly dependent on the mode of fatigue. These key observations of the interaction between shot peening and fatigue mode result from a test program investigating the effect of shot peening on the fatigue behavior of aluminum alloy 7075-T6. The results of these tests shed some light on the fatigue strength of high strength aluminum alloys with and without shot peening.

Tension-compression fatigue tests with zero mean stress were performed on cylindrical axial fatigue specimens in air at room temperature (21°C). The specimens are 4.4 in. (11.2 cm) in length with 1.0 in. (2.54 cm) diam threaded ends tapering to a 0.5 in. (1.27 cm) gage diam. Specimens were shot peened with S230 steel shot to an Almen intensity of 0.01 A. The angle of incidence between the shot and work piece was 90 deg and all specimens were peened up to 200 pct coverage. Specimens were then tested in load control in an MTS servocontrolled hydraulic testing machine.

Figure 2 gives the *S/N* curves for uniaxial fatigue tests conducted at zero mean stress on peened and unpeened specimens of 7075-T6. These curves show a significant increase in fatigue life of shot peened specimens over unpeened specimens at high stress levels. However, at stresses approaching the long-life fatigue strength, there is little difference in the fatigue life of the peened and unpeened specimens. A crossover of *S/N* curves occurs at 29 ksi (200 MPa). Reverse bending fatigue of peened specimens results in a more characteristic and expected improvement in fatigue life at stresses near the long-life fatigue strength, Fig. 3.[1] The reverse bending specimen is a tapered cantilever beam 0.13 in. (0.330 cm) thick, 3.38 in. (8.59 cm) long, and tapering in width from 2.0 in. (5.08 cm) to 0.75 in. (1.91 cm).

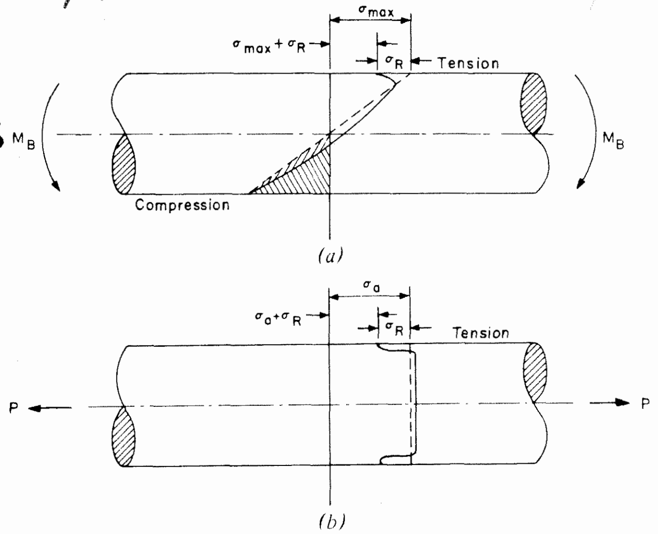

Fig. 1—(*a*) Stress distribution in a specimen loaded in reverse bending, in the unpeened (----), and peened (———) conditions, (*b*) Stress distribution in a specimen loaded in tension in the unpeened (----) and peened (———) conditions.

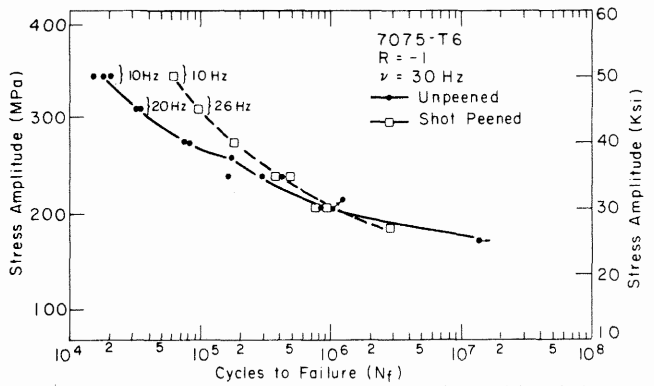

Fig. 2—*S/N* behavior of aluminum alloy 7075-T6 cylindrical fatigue specimen tested in uniaxial fatigue with zero mean stress ($R = -1$), in the unpeened (———) and shot peened (----) conditions.

A detailed fractographic investigation of the specimens tested in tension-compression gives some useful information regarding the initiation and propagation of cracks in the shot peened and unpeened specimens. In the unpeened condition, failures initiated at the specimen surface at all levels of stress amplitude, and fatigue crack growth is typical of stage II growth, *i.e.* perpendicular to the tensile axis. Stage I growth is limited to the order of a grain size (25 to 50 μm) and stage II crack growth is present up to fast final fracture, producing an essentially flat fracture surface except for narrow shear lips.

On the other hand, in the peened condition, the fracture path is chaotic and is more typical of shear failure on different planes or facets. A detailed visual and SEM examination showed that 1) initiation is subsurface at all stress levels at a depth of approximately 400 μm, and 2) from initiation to the breakthrough of

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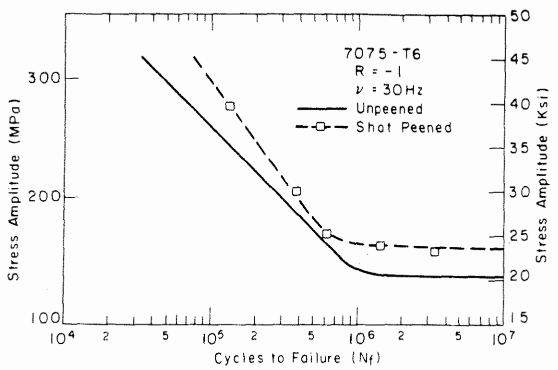

Fig. 3—S/N behavior of aluminum alloy 7075-T6 sheet specimens tested in reverse bending in the unpeened (———) and shot peened (- - - -) conditions.[1]

the crack front to the free surface, crack propagation is characterized by well defined stage I fracture. Figures 4 and 5 show the stage I shear facets, each facet corresponding to cracking through one grain. Although all the stage I crack growth occurs in a vacuum, it appears from the S/N data that the fatigue strength, that is, the critical cyclic shear stress for crack initiation, is the same in vacuum or in air.

These test results are in agreement with the well-known influence of humidity and environment on the fracture path in aluminum alloys, and the results of recent experiments by Schijve[2] on the relative importance of surface or subsurface initiation sites. Vogelsang[3,4] has indicated that there is a significant difference between fatigue mode failures in humid air and in vacuum. Water vapor in the humid air enhances the crack growth rate in a direction perpendicular to the principal stress. In vacuum, however, the absence of air and water vapor results in slow crack growth along planes of maximum shear. These results provide strong evidence that fatigue cracks in aluminum prefer to grow macroscopically in the tensile mode in air and in the shear mode in vacuum.[2]

Since environment affects crack propagation rather than crack initiation,[5] crack growth in vacuum leads to an increase in fatigue life. This accounts for the longer fatigue life of shot peened specimens at high stresses where crack nucleation is rapid and most of the specimen's life is spent in crack propagation. However, at low stresses, 90 pct of the life is spent in crack nucleation, while the fraction of the life spent in crack propagation is very small. The nearly equal fatigue lives of peened and unpeened specimens at low stresses indicate that the critical stress for crack initiation is not changed by shot peening, although initiation is below the surface.

Schijve has demonstrated the effect of residual stresses on the location of the fatigue crack initiation sites in 7075-T6 specimens tested in axial fatigue at zero mean stress.[2] Residual compressive stresses at the specimen surface, and residual tensile stresses at the specimen centerline, were produced by a rapid cold water quench at 5°C following a solution treatment at 480°C for 60 min. The rapid quenching resulted in an estimated residual tensile stress of 29 ksi (200 MPa) at the specimen centerline. Figure 6 shows Schijve's results of uniaxial fatigue tests of specimens with residual stresses and plastically prestrained specimens with no residual stresses. The data com-

Fig. 4—SEM fractograph of the crack initiation region of a 7075-T6 specimen fatigued in tension-compression ($R = -1$) showing stage I crack growth. Crack propagates from bottom to top.

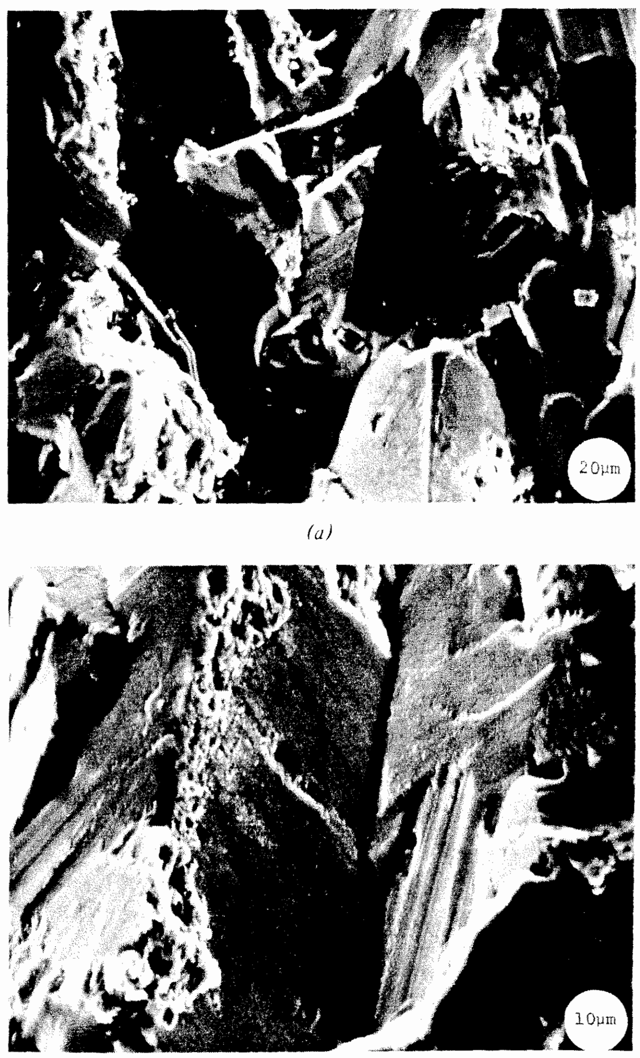

Fig. 5—SEM fractograph of shear facets produced by stage I crack growth under vacuum conditions in 7075-T6 cylindrical specimens fatigued in tension-compression ($R = -1$).

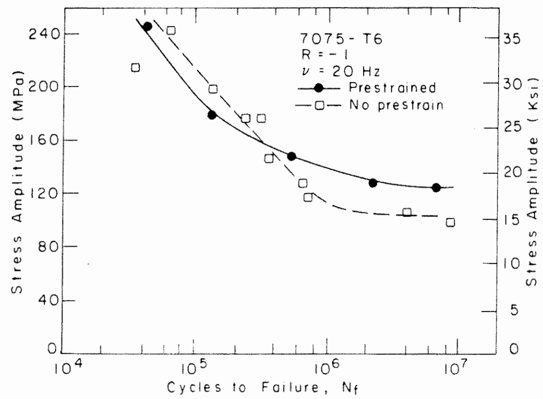

Fig. 6—S/N behavior of 7075-T6 cylindrical fatigue specimens tested in uniaxial fatigue with zero mean stress in the prestrained (———) and unstrained (- - - -) conditions.[2]

pare quite closely with the results of this investigation. Perfect shear mode failures with initiation at the centerline were observed in all specimens with residual stresses, while tensile mode failures were found in all prestrained specimens. Schijve attributes the subsurface initiation to the residual tensile stresses at the center of the specimen.

The fatigue life of specimens tested in tension-compression at zero mean stress, at stress levels above the fatigue limit, is increased by shot peening because crack propagation is slowed down by the residual compressive surface stress and the vacuum environment. However, shot peening does not affect the long life fatigue strength in tension-compression fatigue. Crack initiation is probably caused by the residual tensile stresses below the surface layer of the specimen. However, these stresses are not nearly of the same magnitude as those produced by rapid quenching. Based on measurements of residual stresses in shot peened 7075-T6, residual compressive stresses are typically on the order of 20 ksi (138 MPa).[1] A stress of this magnitude extending 0.02 in. (0.51 mm) in from the specimen's surface will result in a residual tensile stress of 3.63 ksi (25.0 MPa) if distributed uniformly across the tensile region of the specimen. This small residual tensile stress added to the applied stress may very well be sufficient to favor subsurface initiation.

A controversial point is whether the tensile stresses resulting from a shot peening operation are distributed uniformly through the remainder of the section or are concentrated in peak form immediately below the compressive surface layer. Measurements of the residual stress distribution support both points.[6,7] Nevertheless, fatigue crack initiation consistently occurs just beneath the specimen surface lending support to the conclusion that residual tensile stress, as opposed to preexisting flaws or material defects, is the cause of fatigue failure.

The S/N curve for specimens tested in the unpeened condition exhibit a "hump" or type "B" discontinuity at about 35 ksi (241 MPa) and 20,000 cycles, which is observed by others[8] and is in agreement with observations of Mori, reported by Finney,[8] who tested 7075-T6 in reverse bending. The hump is a well documented phenomenon in certain aluminum alloys including 7075-T6 and several theories have been proposed to explain the occurrence of this discontinuity.[8] However, the hump is absent from the shot peened curve.

Our observations support the contention that the discontinuity is the result of environment enhanced crack growth. Figure 2 shows the location of the discontinuity to be within the crack growth controlled regime as opposed to the crack initiation controlled regime. Extrapolation of the lower segment of the unpeened S/N curve to high stresses indicates that the environment enhances crack growth above a critical stress level of 35 ksi (241 MPa). However, only the S/N curve for specimens in the unpeened condition shows a hump since crack growth in specimens in the peened condition occurs in a vacuum. Hence, a strong possibility exists that the environmental effect on crack propagation is responsible for the occurrence of the well known hump discontinuity in the S/N curve of aluminum alloys.

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106187

The Application of a Photogrammetric Technique to the Determination of the Orientation of Stress-Corrosion Fractures

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Transgranular stress-corrosion fracture surfaces in a variety of materials consist of a series of flat, parallel facets separated by steps which are also crystallographic in nature;[1,2] typical examples are illustrated in Fig. 1(*a*) and (*b*). Because of their small size, determination of the orientation of the facets is not possible by standard two surface techniques, and attempts to use electron channelling have so far been unsuccessful, due presumably to the presence of oxides or excessive plastic deformation. In recent studies,[3,4] the orientation of the facets and the traces of the surface steps was determined by a photogrammetric technique used in conjunction with the scanning electron microscope (SEM). While the photogrammetric

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662