Fatigue of a Glass Bead Blasted Nickel-Base Superalloy

L. H. BURCK, C. P. SULLIVAN, AND C. H. WELLS

Room temperature fatigue crack initiation and propagation in the wrought nickel-base superalloy Udimet 700 were investigated with electropolished and glass bead blasted material. Cracks were found to initiate at the surface along coherent annealing twin boundaries oriented for maximum in-plane shear stress in both the electropolished and glass bead blasted conditions even though bead blasting more than doubled the fatigue strength. This increase was found to result from an enhanced crack initiation resistance, but even more importantly from a very pronounced retardation of early Stage I crack propagation by the residual compressive surface stress induced in glass bead blasting.

THIS investigation was undertaken to evaluate the relative and combined effects of glass bead blast induced residual stress and surface cold work on the high cycle (10^6 to 10^7 cycle) fatigue strength of the wrought nickel-base superalloy Udimet 700. Room temperature fatigue crack initiation and propagation were thus studied in electropolished and in glass bead blasted material.

EXPERIMENTAL PROCEDURE

All test specimens, except the cold swaged samples to be described below, were machined from wrought Udimet 700* in either the fully heat treated† or solutionized* condition. The grain size after heat treatment averaged 0.009 in. diam. Test specimens were of three types: flat cantilever reversed bending, axial tension, and constant moment rotating beam, Fig. 1. In addition, one specimen with gage section dimensions of 0.375 in. diam by 1.00 in. long was tested in axial compression in order to determine the effect of a compressive mean stress on fatigue crack initiation and propagation. All testing was conducted at room temperature.

*Nominal composition: Co, 18.5 pct; Cr, 15.0 pct; Mo, 5.0 pct; Al, 4.3 pct; Ti, 3.3 pct; C, 0.07 pct, B, 0.03 pct; Ni, balance.

†Full heat treat: 2140°F/4 hr + 1975°F/4 hr + 1550°F/24 hr + 1400°F/16 hr, forced air cool after each step.

*Solutionize: 2140°F/4 hr, oil quench.

In order to obtain a cold worked structure without the high residual stress which accompanies glass bead blasting, a bar of Udimet 700 was swaged at room temperature to 47 pct reduction in area after being solutionized and aged at 1975°F for 4 hr, as this was the condition considered best suited for room temperature swaging. Rotating beam fatigue specimens and one tensile specimen were then machined from this material.

All specimens were mechanically polished through 600 grit paper, electropolished, and either tested in that condition or glass bead blasted, Fig. 2. To record microstructural changes in surface detail during each test, cellulose acetate surface replicas were taken at 5×10^5 cycle intervals during testing; if no fatigue damage was noted after several million cycles, the stress level was increased. When an active fatigue microcrack could be positively identified, usually by noting a change in length when compared with preceding or subsequent surface replicas, failure was considered to have occurred, and the specimen was usually run to complete fracture. In several cases the average crack propagation rates were noted. Fatigue strength of *electropolished* material was taken as the stress required to initiate a propagating crack at approximately two million cycles. In addition to determining fatigue strengths, some samples were cycled at higher stress levels to determine crack propagation rate as a function of stress amplitude

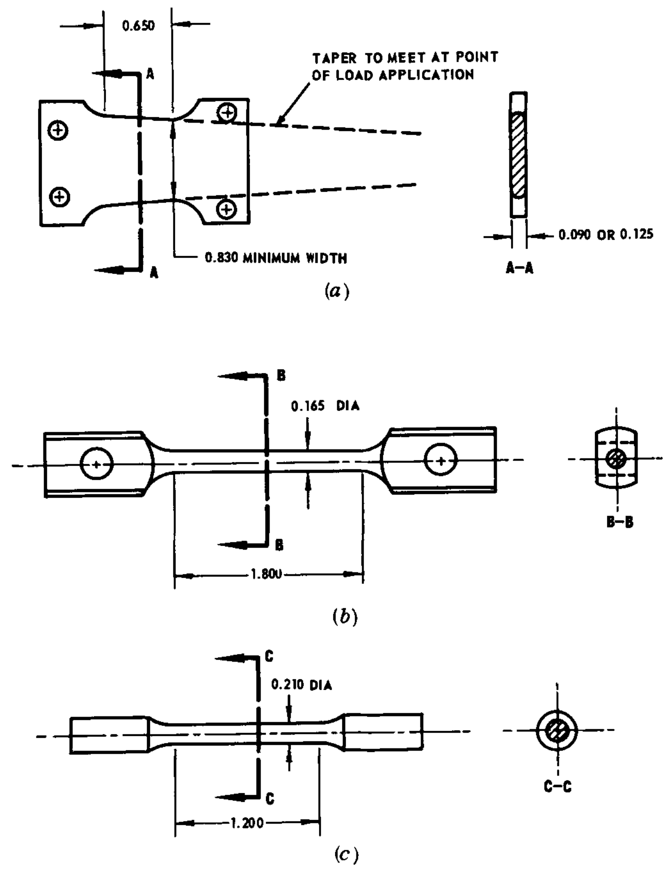
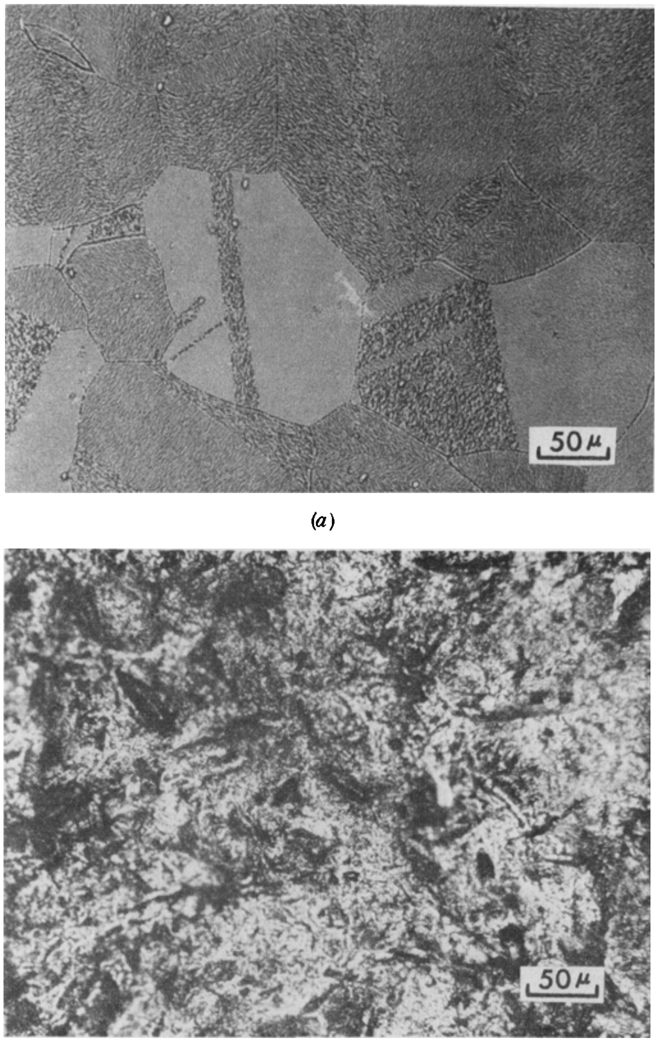


Fig. 1—Designs of (*a*) cantilever bending, (*b*) axial stress, and (*c*) rotating beam bending fatigue specimens.

L. H. BURCK, C. P. SULLIVAN, and C. H. WELLS are Research Assistant and Assistant Directors, respectively, Advanced Materials Research and Development Laboratory, Pratt & Whitney Aircraft, Division of United Aircraft Corporation, Middletown, Conn.

Manuscript submitted August 4, 1969.



(*a*)

(*b*)

Fig. 2—Surface appearance of (*a*) fully heat treated wrought Udimet 700 and (*b*) glass bead blasted wrought Udimet 700.

As will be discussed later, fatigue failure in glass bead blasted material occurred in two distinct phases: 1) crack initiation and early propagation, and 2) much faster propagation as the cracks grew out of the influence of the surface residual stress field produced by the blasting. Although the transition from slow to fast propagation occurred at surface crack lengths of less than 0.010 in., the acceleration in propagation rate was so great that this transition marked 90 to 99 pct of the total specimen life. The fatigue strengths of glass bead blasted material reported here represent this transition in propagation rate since this criterion is most analogous to that taken for nonblasted material.

The glass bead blasted samples were wet blasted with a slurry of glass beads (0.007 to 0.011 in. diam). Blasting intensities[1] ranged from 15 to 22 N Almen intensity with the standard tolerance in this range being $\binom{+5}{-0}$N. Specimens were tested both as bead blasted and bead blasted with a light (~0.0001 in.) electropolish to facilitate metallographic examination during the test. There was no apparent effect of this electropolishing on the fatigue behavior.

Slip band etching[2] and cellular recrystallization[3,4] were both employed to determine the extent of plastic deformation associated with the glass bead blasted layer. These methods yielded consistent results which gave an average depth of about 0.002 in. for an intensity of 15 N, Fig. 3.

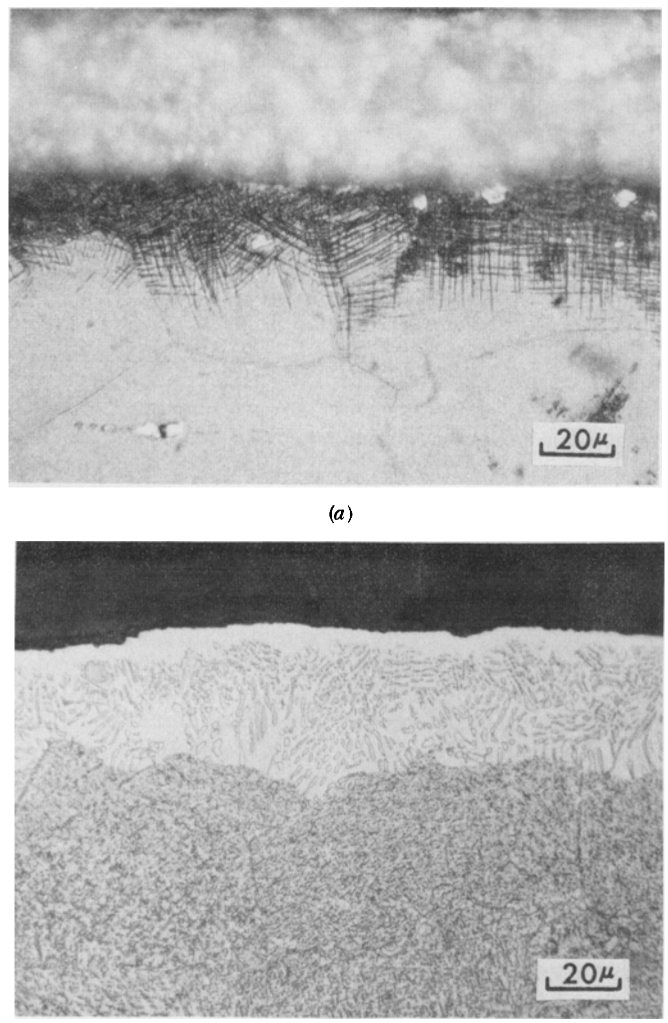
The residual stress distribution below the surface of a glass bead blasted strip (15 N intensity) of Udimet 700 was determined by electrolytically removing successive layers of strained material and noting the change in curvature of the strip,[5] Fig. 4.

EXPERIMENTAL RESULTS

Electropolished

The electropolished specimens displayed a fatigue strength of from ±30 to ±35 ksi in bending. The one axial specimen tested showed complete fracture after 0.8×10^6 cycles at 5 to 65 ksi tension which is consistent with the other results. No significant difference in fatigue strength was found between fully heat treated and solutionized material, indicating that in these tests fatigue strength was independent of heat treatment.

In many cases, microcracks were observed to initiate at the surface along coherent annealing twin bound-



(*a*)

(*b*)

Fig. 3—Cross sections (normal to surface) of glass bead blasted samples (15 N intensity) showing plastic deformation revealed by: (*a*) slip band etching and (*b*) cellular recrystallization.

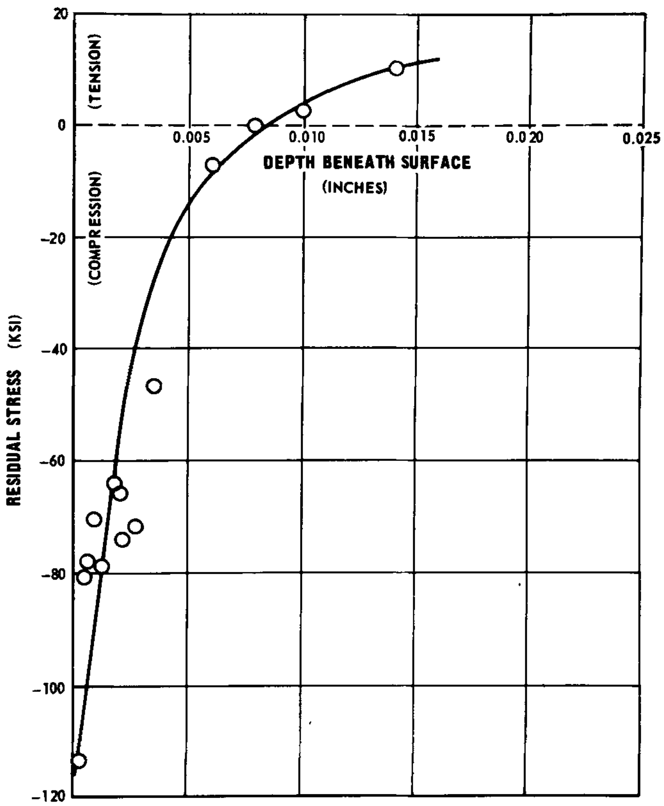


Fig. 4—Residual stress distribution resulting from glass bead blasting to 15 N intensity.

aries inherent in the wrought material. In each instance, the cracked twin boundary trace was oriented for maximum shear stress in the surface plane, Fig. 5. In general, no unusual microstructural features detectable by light microscopy were associated with these boundaries.

It was observed that fatigue cracks could not be traced back to unique points of initiation. Surface replicas taken prior to the appearance of finite length microcracks showed no indication of cracking even though extrapolations of the subsequent propagation data indicated that the cracks should also have appeared on these replicas at lengths easily resolvable in the optical microscope. In addition, extrapolations of the linear crack length vs number of cycles curves, in some instances, indicated positive (and resolvable) crack lengths at zero cycles, thus implying that the cracks either initiated in finite lengths or that they initially grew at a rate much faster than in subsequent propagation. As these cracks always initiated along twin boundaries with no discernible defects, preexisting flaws are not believed responsible for this phenomenon. With specimens cycled in bending at ±40 ksi, for example, the smallest detectable cracks were about 0.002 in. long.

The one axial specimen cycled completely in compression was found to contain a microcrack along a twin boundary after 1.0×10^6 cycles at -35 ± 30 ksi, or approximately the number of cycles needed to initiate a Stage I microcrack at $+35 \pm 30$ ksi.

As has been observed in other nickel-base superalloys,[7,8] room temperature high-cycle crack propagation in electropolished Udimet 700 occurred largely in the Stage I mode. Stage I cracks generally grew along the entire length of the initiating twin boundaries until encountering a grain boundary at which point they grew into adjacent grains in a sawtooth manner, propagating on several planes within each grain, Fig. 5. In all cases the cracking was completely transgranular.

Surface crack length was a linear function of the number of cycles during the Stage I portions of propagation. A representative plot of crack length vs number of cycles for a cantilever bending specimen is given as Fig. 6, and Stage I propagation rate as a function of alternating stress (zero mean stress) is presented in Fig. 7. The sample tested in compression showed a constant half crack propagation rate of 3.8×10^{-10} in. per cycle when cycled at -40 ksi mean stress (-40 ± 35 ksi) but when the mean stress on this same sample was increased to zero, the propagation rate increased to 1.0×10^{-8} in. per cycle.

The fatigue zones of the fracture surfaces were quite crystallographic with Stage I facets readily observed. An unusually large facet often marked the site of crack initiation and was usually oriented for maximum in-plane shear stress over its face. The facets

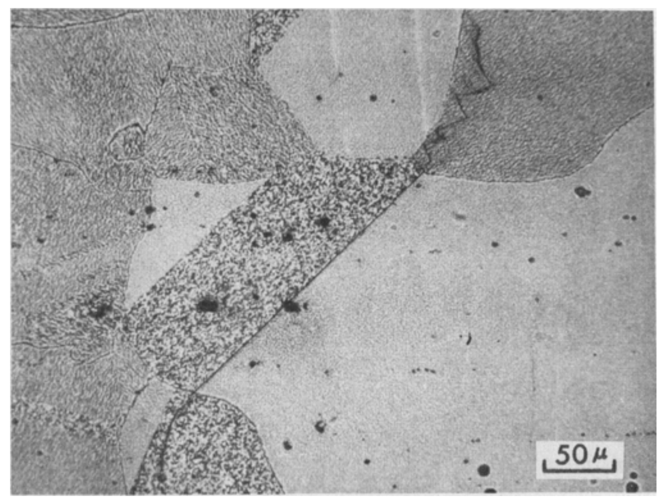


Fig. 5—Example of a twin boundary fatigue crack on the surface of electropolished wrought Udimet 700. Stress axis horizontal.

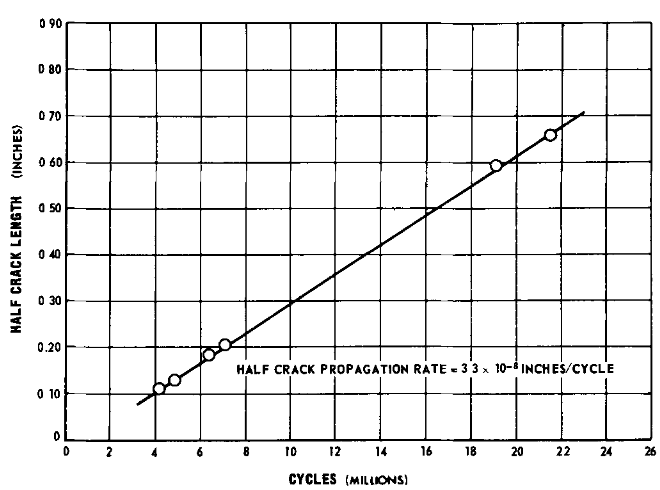


Fig. 6—Propagation of a Stage I fatigue crack in electropolished material cycled at ±35 ksi in cantilever bending. Crack propagation rate is independent of crack length.

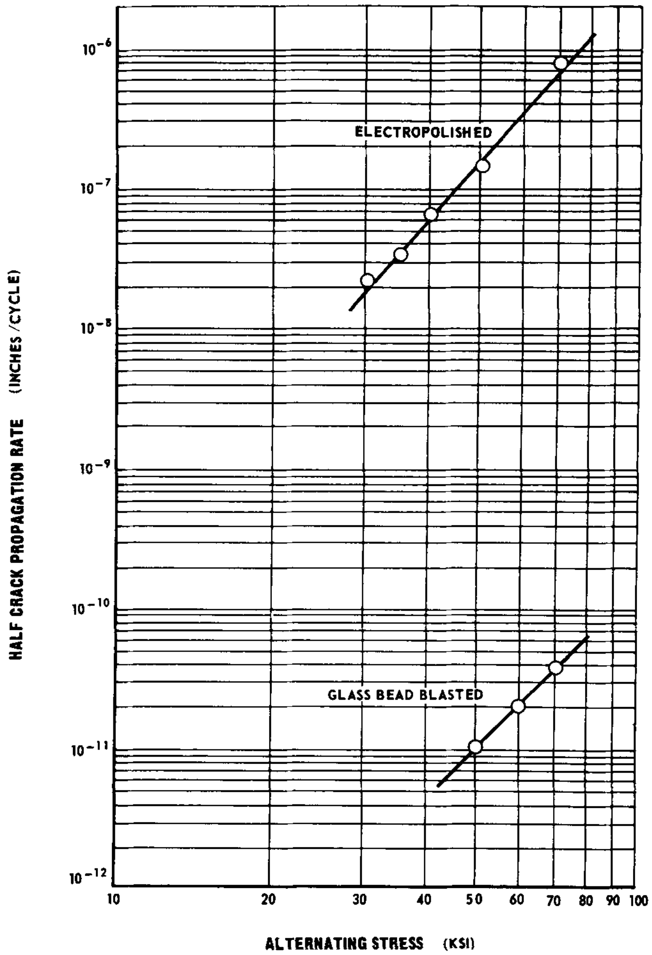


Fig. 7—Stage I half crack fatigue propagation rates at various alternating stresses in electropolished and glass bead blasted material tested in bending. Note the large decrease in crack propagation rate due to glass bead blasting.

often contained river lines and "cleavage" steps similar to those reported for another nickel-base alloy[7] and, in addition, striation markings were occasionally detected on facets somewhat removed from the initiation site.

Glass Bead Blasted

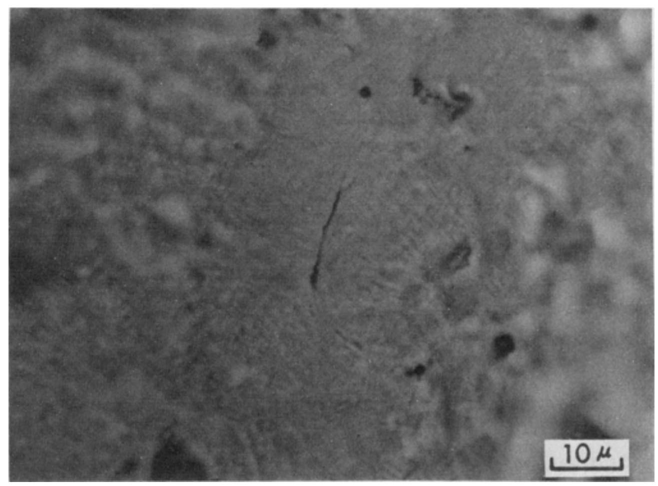
The maximum 2×10^6 cycle fatigue strength obtained with a glass bead blasted specimen was ±75 ksi in reversed bending. The minimum strength varied greatly with blasting coverage and will be discussed in more detail later. An axial specimen bead blasted to 22 N withstood 1.0×10^6 cycles at 50 ± 45 ksi without damage and failed after 4.0×10^5 cycles at 52.5 ± 47.5 ksi.

When viewed at 200X magnification, the surfaces of the glass bead blasted specimens appeared quite irregular containing many gouges as well as the expected hemispherical depressions, Fig. 2(*b*). Many extremely small microcracks induced by glass bead blasting were also noted, often occurring along the edges of these gouge marks, Fig. 8(*a*). It is important to note that none of the fatigue cracks appeared to be associated with any of these preexisting microcracks, hence they are not of concern in determining the life of a glass bead blasted specimen.

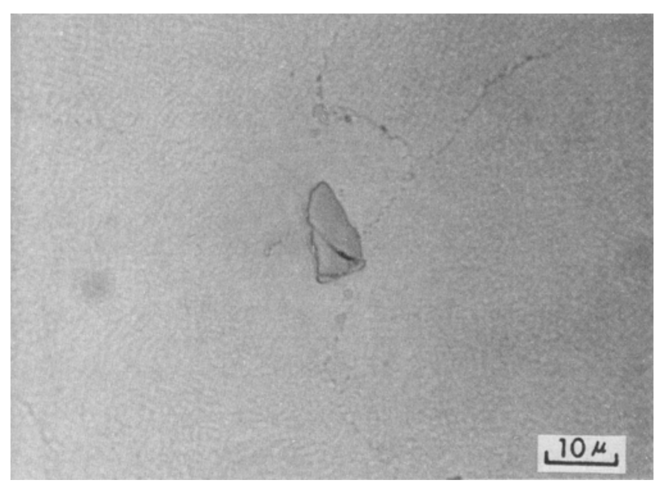
Careful polishing of the specimen surfaces and cross sections revealed the presence of fractured carbides in the bead blasted layer, Fig. 8(*b*). However, as in the case of cracks associated with gouge marks, there was no evidence that matrix fatigue cracks initiated at these previously fractured carbides.

Crack initiation in the glass bead blasted material, although deferred to somewhat higher stress levels than in the electropolished condition, was quite similar to that described for the electropolished case, *i.e.*, cracks first appeared as finite sized segments along twin boundaries oriented for maximum in-plane shear stress. Once present, however, these small cracks propagated at constant rates which were extremely slow when compared to similar cracks in electropolished material, Fig. 7. As these cracks grew larger, however, the propagation rates quickly approached those observed in electropolished material, Fig. 9.

The specimens machined from the cold swaged material showed no significant improvement in fatigue strength or crack propagation behavior over electropolished material even though the 0.2 pct yield stress was increased from about 130 ksi before swaging to



(*a*)



(*b*)

Fig. 8—(*a*) Polished surface of bead blasted sample showing a blasting induced microcrack, (*b*) cross section of sample showing a carbide particle fractured by glass bead blasting.

260 ksi afterwards and the hardness was increased from R_C 35 to R_C 48.

DISCUSSION

Electropolished

The initiation of fatigue cracks along twin boundaries has been reported for other fcc metals,[8] and a similar effect has been observed in wrought Udimet 700 subjected to low-cycle fatigue at temperatures up to 1000°F.[9] It has been suggested that noncoherent steps in the twin boundaries of OFHC copper act as dislocation sources during fatigue cycling and that when twin boundaries are parallel to the operating slip planes, these steps produce sufficient stress concentration to initiate fatigue cracks.[8] In the case of nickel-base superalloys, a lack of precipitate along the coherent twin boundaries may result in a lower critical resolved shear stress at these locations.[10]

As previously noted, all specimens showed constant crack growth rates at a given stress range during the Stage I portions of propagation. Since cracks always initiated preferentially on planes of maximum in-plane shear stress, it is reasoned that they propagate more readily in mode II (in-plane shear) than in mode III (antiplane shear). Cracks oriented for maximum shear stress in the surface plane would then grow faster along the surface than into the specimen and would assume semielliptical profiles, as reported for aluminum single crystals which also showed constant Stage I propagation rates.[11] The propagation rates of such cracks, as measured on the specimen surface, is then dependent on the range of stress intensity factor ΔK_{II} which, in turn, is a function mainly of crack depth, not length, for elliptical cracks.[12] Since the propagation rates of through thickness cracks in this material have been shown[13] to be proportional to $\Delta K_{II}^{3.4}$, the constant propagation rates observed here imply that ΔK_{II} at the crack tip (major semiaxis) remains constant as the surface length of the crack increases. One would then expect, as observed, that the rate of crack propagation would be approximately proportional to the 3.4 power of the stress range, Fig. 7.

It should be noted that, in some cases, average crack propagation rates are less than one Burgers vector ($\mathbf{b} = 1.2 \times 10^{-8}$ in.) per cycle. This fact, together with the sudden appearance of crack segments at initiation, and the fourth power dependence of propagation rate on stress range tend to support a model of cumulative damage.[14,15]

Glass Bead Blasted

Increases in fatigue strength with glass bead blasting are generally attributed to the presence of: 1) a residual compressive stress at the surface[16-19] and 2) a cold worked surface layer.[18-20] It has been demonstrated that a tensile mean stress lowers the fatigue strength of most materials and that a compressive mean stress raises it,[21,22] while cold working may either lower[23] or raise[18] the strength depending on the material, the degree of working, and the fracture mode. The relative effects of residual stress and surface working and the resulting net effect of glass bead blasting then depend on the material, the blasting parameters, and the test conditions.

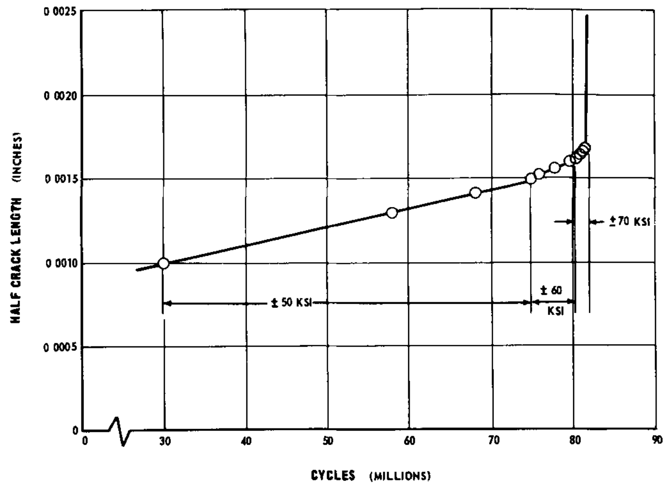


Fig. 9—Propagation of a small crack in glass bead blasted material (15 N) cycled in bending at ±50, ±60, and ±70 ksi. The nearly vertical segment in the ±70 ksi range indicates a propagation rate which approaches that of electropolished material.

The results presented here show that while cold working of Udimet 700 may have a large effect on its tensile properties, it has little effect on either fatigue crack initiation or the early stages of propagation at room temperature.

Mean stress, however, does have a significant effect of Stage I propagation. A compressive mean stress increases the friction force between the crack faces, and the resulting inhibition of relative shear displacement may thereby affect Stage I propagation. Preliminary calculations[24] have substantiated this point.

It appears then that much of the beneficial effect of glass bead blasting on the fatigue properties of Udimet 700 is derived from a very low rate of early crack propagation under the high compressive residual stress in the surface layer. A very important parameter in the case of Udimet 700 was the degree of blasting coverage. Although all specimens were bead blasted to Almen saturation, *i.e.*, until additional blasting time had little effect on measured Almen intensity, samples which were allowed additional blasting time apparently attained a more uniform stress distribution and showed a markedly improved fatigue strength. It should be noted, however, that excessive bead blasting may cause the fatigue strengths of some materials to decrease[18,25] and thus each material should be evaluated individually.

The residual compressive stress apparently also prevented bead blast microcracks and carbide cracks in this material from propagating in fatigue.

CONCLUSIONS

1) High-cycle fatigue cracks initiated at the surface along coherent annealing twin boundaries which were oriented for maximum in-plane shear stress in both electropolished and glass bead blasted wrought Udimet 700.

2) Stage I crack propagation was independent of crack length but was approximately proportional to the fourth power of the stress range.

3) A fatigue test run with a compressive mean stress showed that the Stage I crack propagation rate was

greatly retarded compared to tests run at zero or tensile mean stress.

4) The observed increase in fatigue strength with glass bead blasting was due in large part to an inhibition of early Stage I crack propagation by the large compressive residual stress at the surface. Cold working due to bead blasting did not seem to be an important factor.

5) The major factor in obtaining the maximum fatigue improvement through glass bead blasting in the present fatigue tests was thorough coverage. Almen intensity saturation did not necessarily indicate that the proper coverage for maximum blasting benefit had been achieved.

ACKNOWLEDGMENTS

The authors are grateful to S. W. Hopkins and J. J. Nolan for their assistance in the mechanical testing, and to T. T. Field who assisted in the metallography. They would also like to thank Dr. M. Gell for helpful discussions and suggestions.

REFERENCES

1. H. O. Fuchs: Am. Soc. Testing Mater., Spec. Tech. Publ. No. 196, 1962.
2. M. Gell and T. T. Field· *Trans. TMS-AIME,* 1969, vol. 245, p. 1666.
3. K. B. Valantine: *Trans. ASM,* 1948, vol. 40, p. 240.
4. J. M. Oblak and W. A. Owczarski: *Trans. TMS-AIME,* 1968, vol. 242, p. 1563.
5. W. Leaf: *Soc. Exp. Stress Analysis Proc.,* 1952, vol 9, p. 133.
6. R. B. Slack: M.S. Thesis, Hartford Graduate Center, Rensselaer Polytechnic Institute of Connecticut, 1967.
7. M. Gell and G. R. Leverant. *Acta Met.,* 1968, vol 16, p. 553.
8. R. C. Boettner, A. J. McEvily, Jr., and Y. C. Liu: *Phil. Mag.,* 1964, vol. 10, p. 95.
9. C. H. Wells and C. P. Sullivan: *ASM Trans. Q.,* 1967, vol. 60, p. 217.
10. C. H. Wells and C. P. Sullivan: *ASM Trans. Q.,* 1964, vol. 57, p. 841.
11. A J. McEvily and R. C. Boettner: *Acta Met.,* 1963, vol. 11, p. 725.
12. M. K. Kassir and G. C. Sih: *J. Appl. Mech., Trans. ASME,* 1966, vol 33, p 601
13. C. H. Wells and S W. Hopkins: Unpublished Research, 1967, Pratt & Whitney Aircraft, Middletown, Conn.
14. J. Weertman: *Int. Conf. on Fracture Sendai,* 1965, p 153.
15. A. K. Head: *Phil Mag.,* 1953, vol. 44, p. 925.
16. R. P. Felger: ASME Paper No 58-SA-46, 1958.
17. M. Ya. Shashin: *Met. i Term. Obrabutka Metallov,* 1959, p. 42 (Brutcher Translation # 4481).
18 F. Sherratt: S & T Memo 1/66, Ministry of Aviation, U. K., 1966.
19. F. Sherratt: S & T Memo 2/66, Ministry of Aviation, U. K., 1966.
20. J. M. Lessells and W. M. Murray, *Proc. ASTM,* 1941, vol. 41, p. 659.
21. J. Morrow and J. F. Millan, ed: SAE Handbook Supplement TR-198 (SAEJ 783), 1962.
22. W. N. Findley: *Proc. ASTM,* 1954, vol. 54, p. 836.
23. F. H. Vitovec: *Proc. ASTM,* 1958, vol. 58, p. 552.
24. D O. Swenson: Unpublished Research, 1969, Pratt & Whitney Aircraft, Middletown, Conn.
25. M. Uchiyama and K. Kamishohara: *Tetsu to Hagane,* 1959, vol. 45, p. 105