ARMY RESEARCH LABORATORY

Metallurgical Examination of a Failed Blade Lag Shock Absorber (Part No. 114H6802) From a CH-47D Chinook Cargo Helicopter

by Marc S. Pepi, Scott M. Grendahl, and Victor K. Champagne

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Abstract

A metallurgical examination was performed on a failed blade lag shock absorber from the aft red rotor blade of an Army cargo helicopter. The U.S. Army Research Laboratory (ARL) and the primary contractor (Boeing Helicopters, Philadelphia, PA) performed a visual examination of the failed part, fluorescent penetrant inspection, fractographic evaluation, metallography, hardness testing, conductivity testing, and chemical analysis. It was concluded that the part failed due to fatigue from an area exhibiting intergranular attack. The corrosive attack was most likely caused by the processing fluids used during the rework process. In addition, the parts may not have been properly aged, as evidenced by the higher-than-nominal yield strength values. An improper aging treatment could have facilitated the intergranular attack.

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1. Background

The U.S. Army Research Laboratory (ARL) and Boeing Helicopters, Philadelphia, PA, performed a metallurgical examination of a failed blade lag shock absorber from a CH-47D Chinook cargo helicopter. The helicopter experienced an abnormal vibration during flight and landed as a precautionary measure. Upon shutdown, the part failed and an aft rotor blade struck the fuselage, causing the damage shown in Figures 1 and 2 [1]. Figure 3 shows the location of this component on the rotor system of a CH-47D Chinook. The part was first analyzed by Corpus Christi Army Depot (CCAD), Corpus Christi, TX (see copy of report in Appendix A), and subsequently forwarded to Boeing. ARL witnessed and/or contributed to the Boeing investigation, and their report is included in Appendix B. After the joint analysis was performed at Boeing, the components were brought back to ARL for further analysis at the request of the U.S. Army Aviation and Troop Command. Figures 4 through 7 show the pieces ARL retrieved from Boeing. The following analyses were conducted in order to fully characterize the failed component: visual examination, fractographic evaluation/energy dispersive spectroscopy (EDS) analysis, metallography, hardness testing, conductivity testing, tensile testing, and chemical analysis. ARL concurred with Boeing that the intergranular attack was a contributing factor to the failure and that the attack was most likely caused by cleaning fluids from the rework process.

- Applicable Specifications (specifications referenced during the investigation):

  - BAC 5946, "Temper Inspection of Aluminum Alloys" [2].
  - MIL-A-22771, "Aluminum Alloy Forgings, Heat Treated" [3].
  - ASTM E8, "Tension Testing of Metallic Materials" [4].
  - MIL-H-6088, "Heat Treatment of Aluminum Alloys" [5].

- Service Hours of Part: unknown.

- Prior Depot Overhauls: 2.

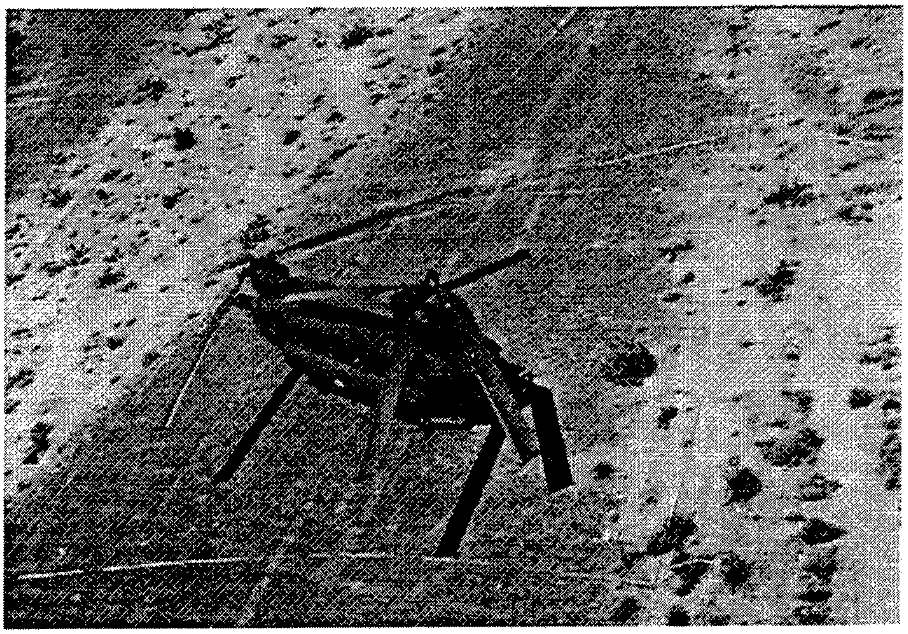

Figure 1. Overhead View Showing the Damage Encountered by the Chinook Helicopter Subsequent to the Failure of the Blade Lag Shock Absorber (Photo Courtesy of CCAD).

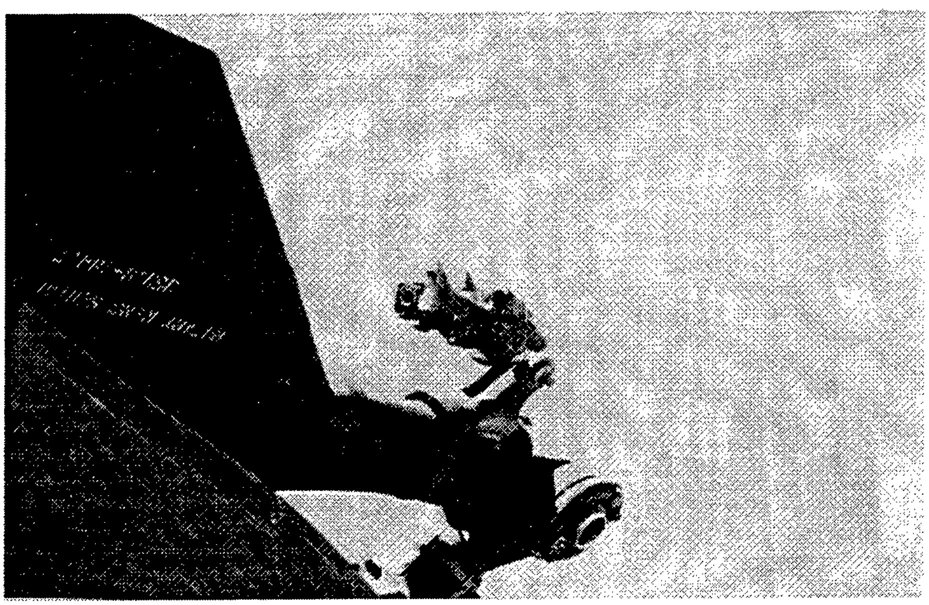

Figure 2. View Showing the Extensive Damage at the Aft Rotor, Caused by the Blade Lag Failure (Photo Courtesy of CCAD).

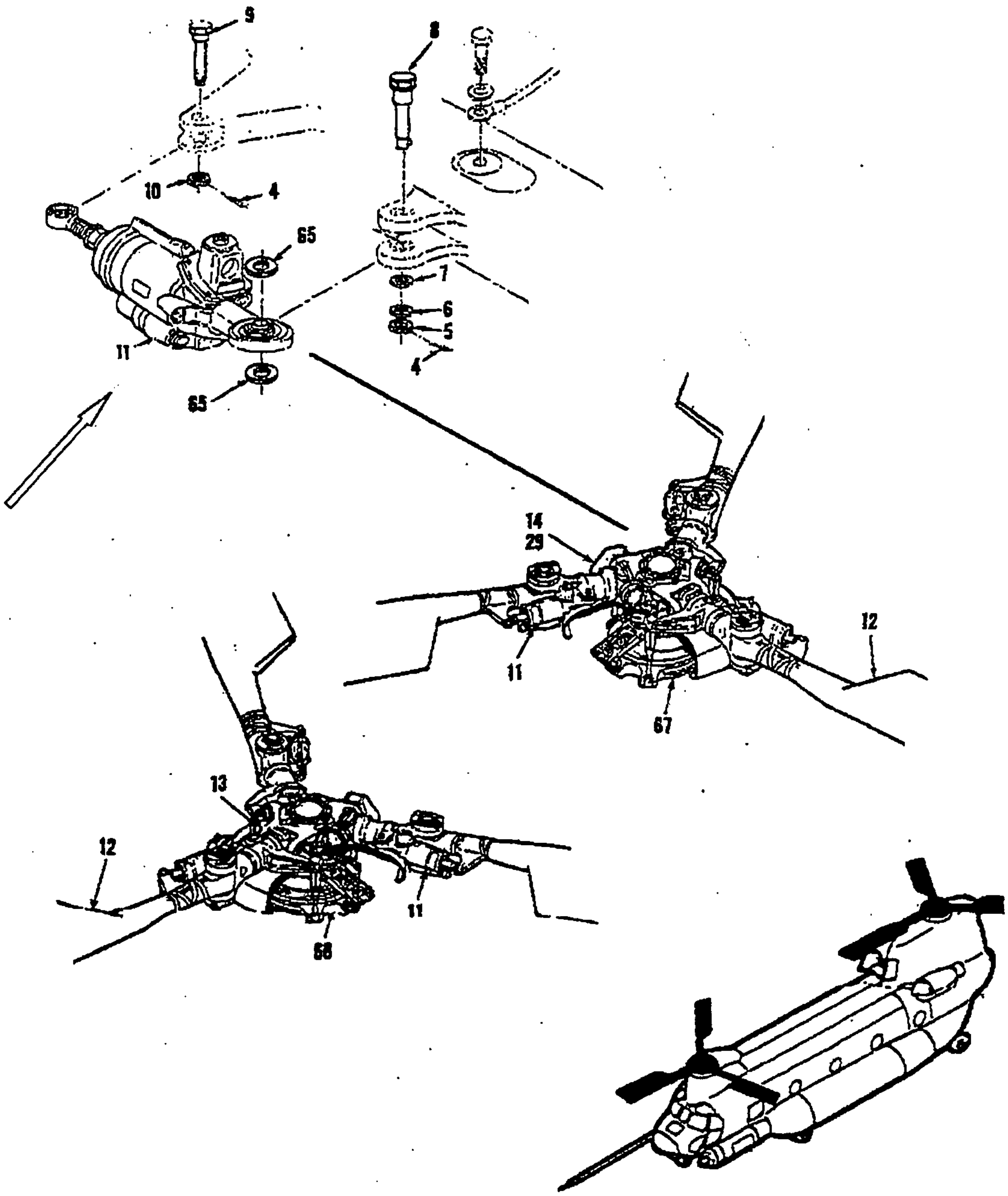

Figure 3. Schematic Showing the Location of the Blade Lag Shock Absorber on the CH-47D Chinook Helicopter.

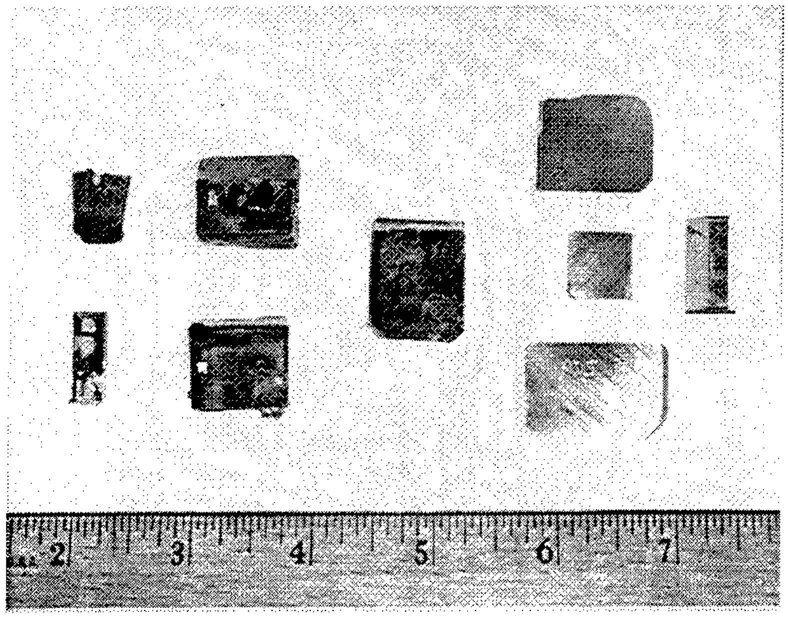

Figure 4. Sections of the Failed Blade Lag Shock Absorber as Received From Boeing; Reduced 31%.

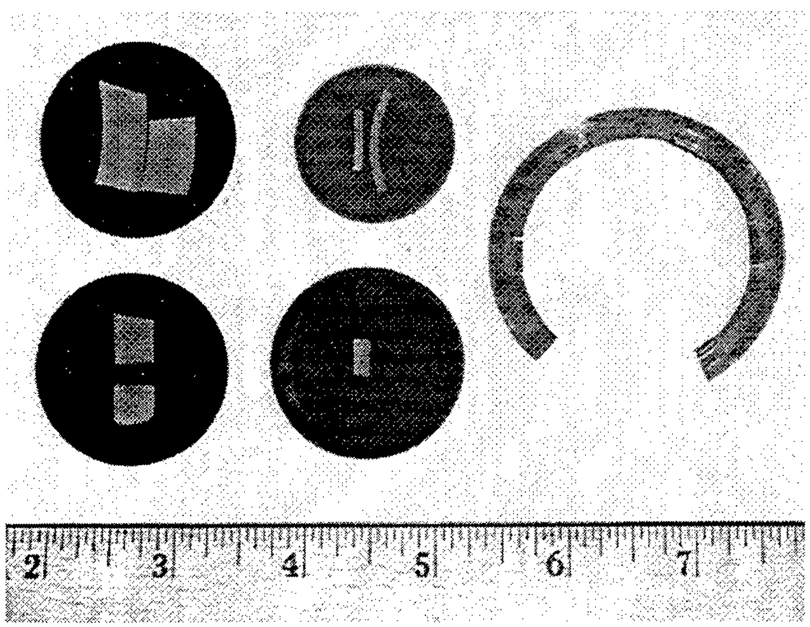

Figure 5. Mounted Samples of the Failed Blade Lag Shock Absorber and a Sectioned Collar as Received From Boeing; Reduced 28%.

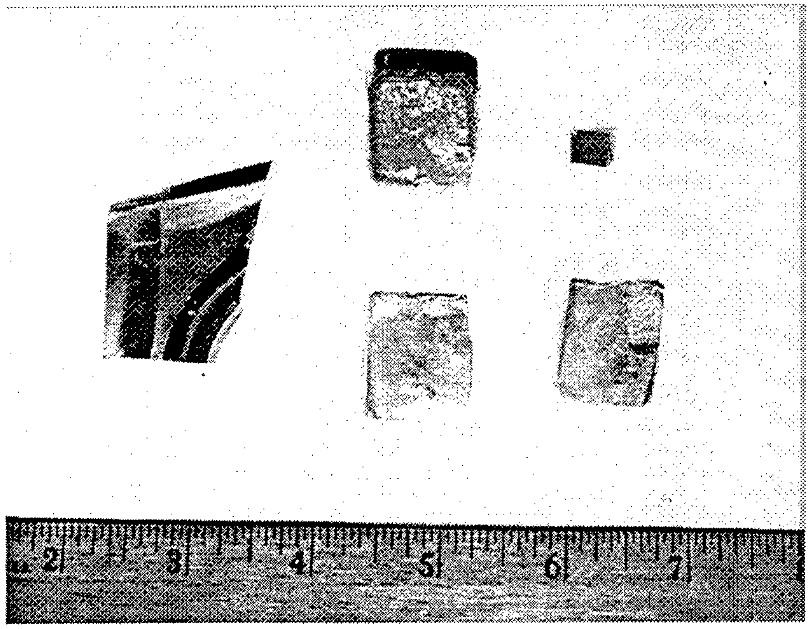

Figure 6. Sections of the Failed Blade Lag Shock Absorber Including Fracture Surfaces as Received From Boeing; Reduced 31%.

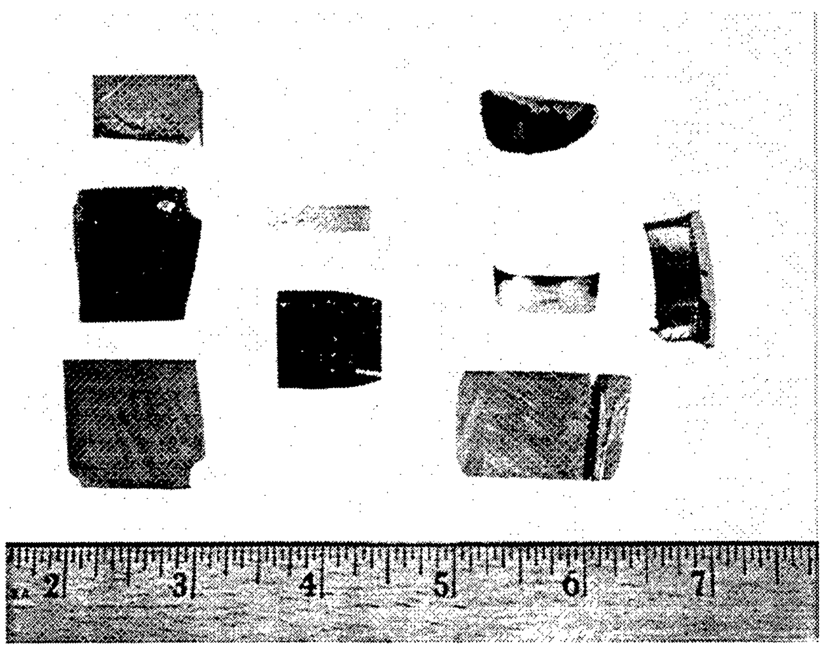

Figure 7. Additional Sections of the Failed Blade Lag Shock Absorber as Received From Boeing; Reduced 31%.

- Heat Treatment: The forging was required to be heat-treated in accordance with MIL-H-6088 [5] (for 7075-T73 forgings):

  - solution heat-treating at 860–890° F,
  - age-hardening at 215–235° F for 6–8 hr, followed by
  - age-hardening at 340–360° F for 8–10 hr.

2. Visual Examination/Light Optical Microscopy

The failed part was visually examined at Boeing. The inboard lug had two radial through-fractures at the spherical bearing bore, as shown in Figure 8 [1]. The through-fractures were at the approximate positions shown in Figure 9 of the inboard lug of the blade lag shock absorber:

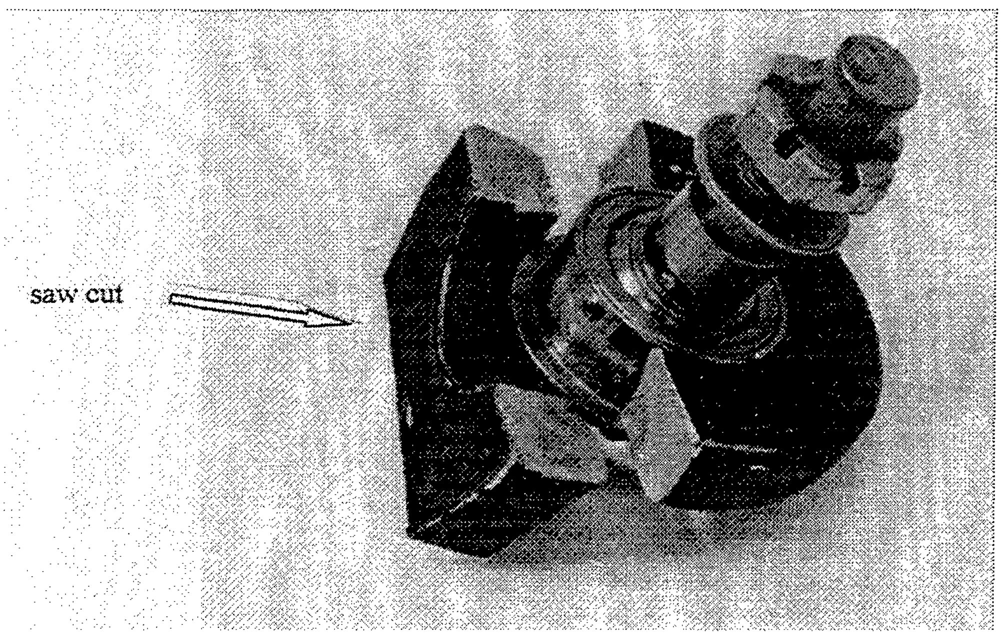

Figure 8. Failed Component Taken From the Lag Blade Shock Absorber (Photo Courtesy of CCAD).

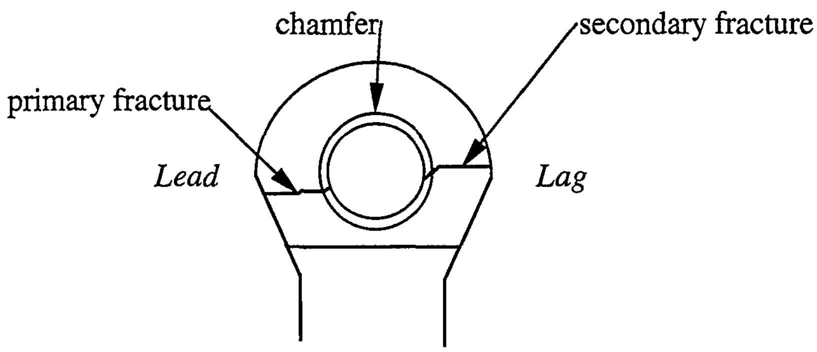

Figure 9. Schematic Showing Positions of Through-Fractures.

Fluorescent penetrant inspection performed by Boeing revealed no additional cracks. Figures 10 and 11 [1] show the primary and secondary fracture surfaces, respectively. The primary fracture (labeled as Fracture A within the context of this report) exhibited features consistent with a fatigue failure; the topography was relatively flat and smooth, with evidence of beach marks. The fatigue crack traversed approximately 95% of the fracture surface through the lead portion of the lug, while ductile shear lips comprised the remaining fracture surface. This was considered the primary fracture surface due to the extent of fatigue. The origin was determined to be located at the top of the 0.040-in, 45° chamfer. The secondary fracture (Fracture B) contained only approximately 20% fatigue through the lag region of the part, with the remainder exhibiting a ductile dimpled morphology. The origin of this fracture was determined to be located along the bore, approximately 0.05 in from the spot face of the lug. In addition, a dark-gold coating was observed along the exterior of the part. This coating was the typical color of a part that had been chromic-acid-anodized (as required in accordance with the engineering drawing). A blackish-green coating was observed on the bore surface, which was atypical of the chromic acid anodize. Finally, many axial score marks were noted on the bore surface. Some markings were shiny, while others contained the blackish-green coating. These markings were determined to be consistent with the prior removal of the bushing. The blackish-green coating and the axial score marks were evidence that the bore had been reworked.

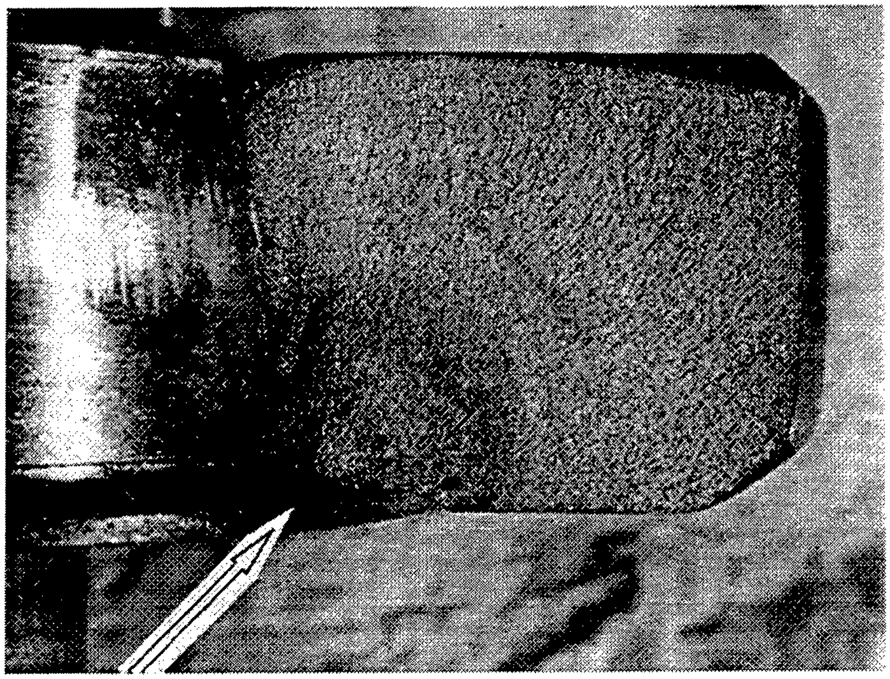

Figure 10. Optical Macrograph of the Primary Fracture (Photo Courtesy of CCAD).

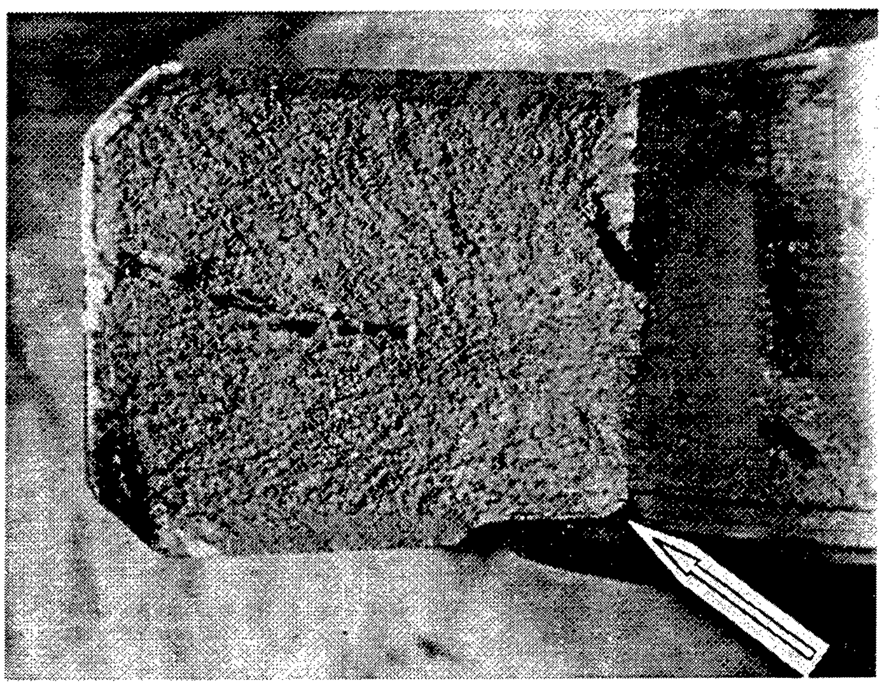

Figure 11. Optical Macrograph of the Secondary Fracture (Photo Courtesy of CCAD).

3. Metallography

Boeing fabricated three transverse metallographic samples from the inboard lug. Evidence of intergranular attack was prevalent on each sample, extending along the bore wall and spot face of the lug. Figures 12 and 13 show examples of this corrosive attack in the as-polished condition, while Figure 14 shows another region of attack that had been etched with Keller's reagent. These areas of attack were located adjacent to the primary fracture. The average depth of the intergranular attack was approximately 0.010 in. The microstructure was examined in an effort to confirm that the part was properly aged to the -T73 condition. The major microstructural difference resulting from variations in time and temperature of -T73 tempers is the size and type of precipitate. However, these differences are subtle, and impossible to detect metallographically [6]. Moreover, transmission electron microscopy (TEM) is not adequate in distinguishing between these features. It is suggested within Davis [6] that the differences within a given temper (and between tempers) be established through hardness, conductivity and, in some cases, tensile yield results. The -T73 temper contains a mixture of η' and the stable, incoherent η ($MgZn_2$) phase [7]. Figure 15 shows the precipitates of a sample sectioned through the chamfer in the as-polished condition. The light-gray particles were actually copper in color and represented the copper in the alloy. Figure 16 shows this structure etched with Keller's reagent at 200$\times$ magnification. Note the directionality of the grains, indicative of the prior forming operation. Figure 17 shows this structure at higher magnification.

4. Fractography

Boeing analyzed the fracture surfaces utilizing scanning electron microscopy (SEM). The results confirmed the origin locations, the fatigue failure mode, and the degree of fatigue crack propagation. Subsequently, ARL examined the fracture surfaces to more fully document the findings. The origin of the primary fracture is shown in Figure 18. The radial lines converge at the origin on the surface. Figure 19 shows the origin at an angle such that the bore can be seen,

Figure 12. Intergranular Attack Noted Along the Bore From a Sample Taken Adjacent to the Primary Fracture Surface; as Polished, Magnification 400×.

Figure 13. Another Region of Intergranular Attack Noted Along the Bore From a Sample Taken Adjacent to the Primary Fracture Surface; as Polished, Magnification 200×.

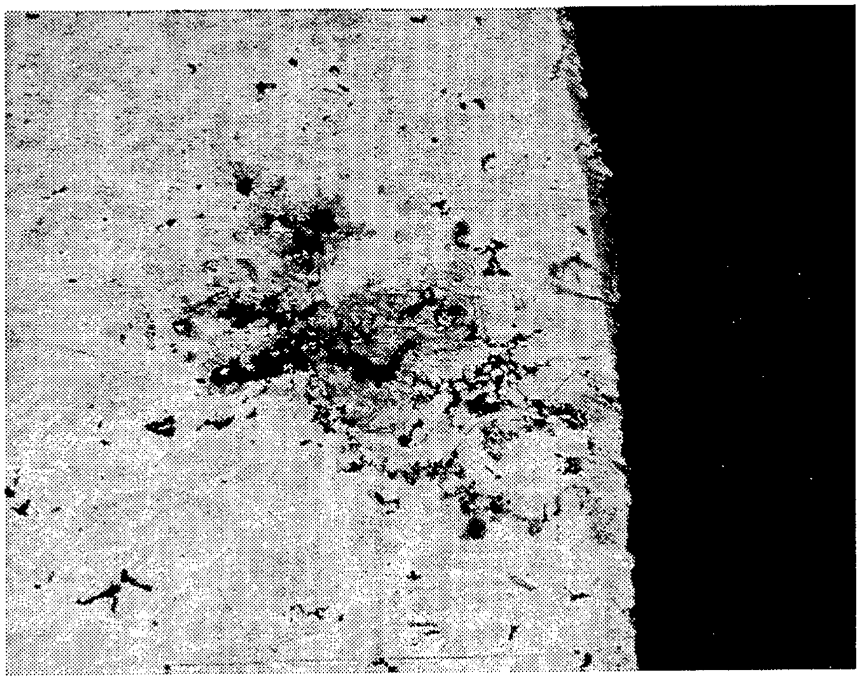

Figure 14. Intergranular Attack Noted Along the Bore of a Sample Taken Adjacent to the Primary Fracture Surface; Keller's Etch, Magnification 200×.

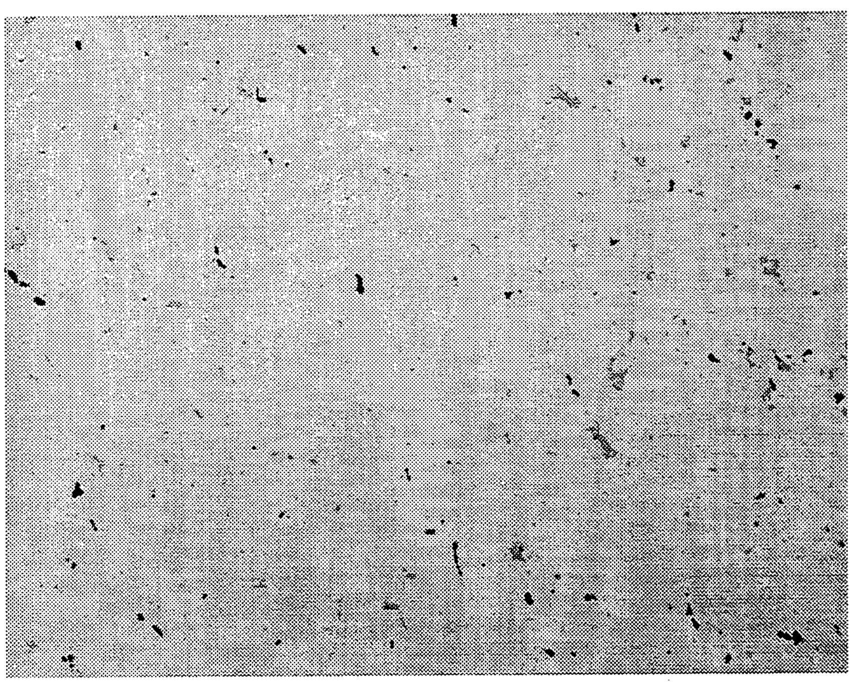

Figure 15. Micrograph of the As-Polished Microstructure of the 7075 Alloy; Magnification 200×.

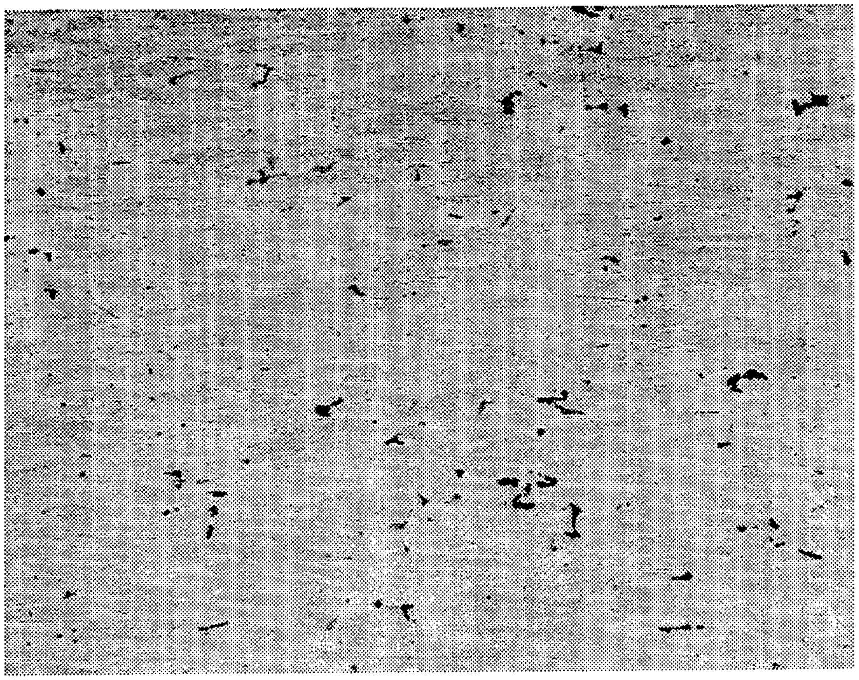

Figure 16. Micrograph of the Etched Microstructure of the 7075 Alloy; Keller's Etch, Magnification 200×.

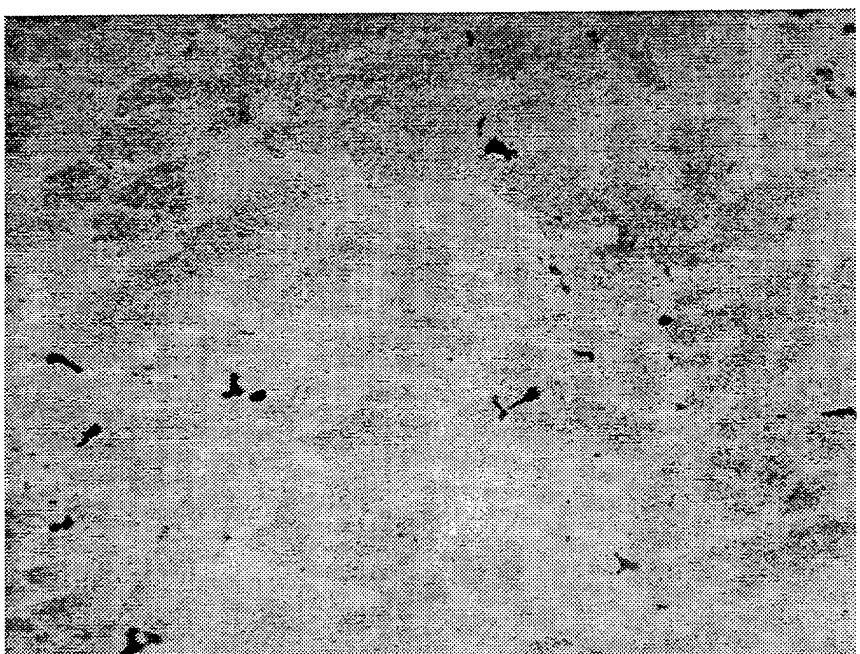

Figure 17. Micrograph of the Microstructure Shown in Figure 16, at Higher Magnification; Keller's Etch, Magnification 500×.

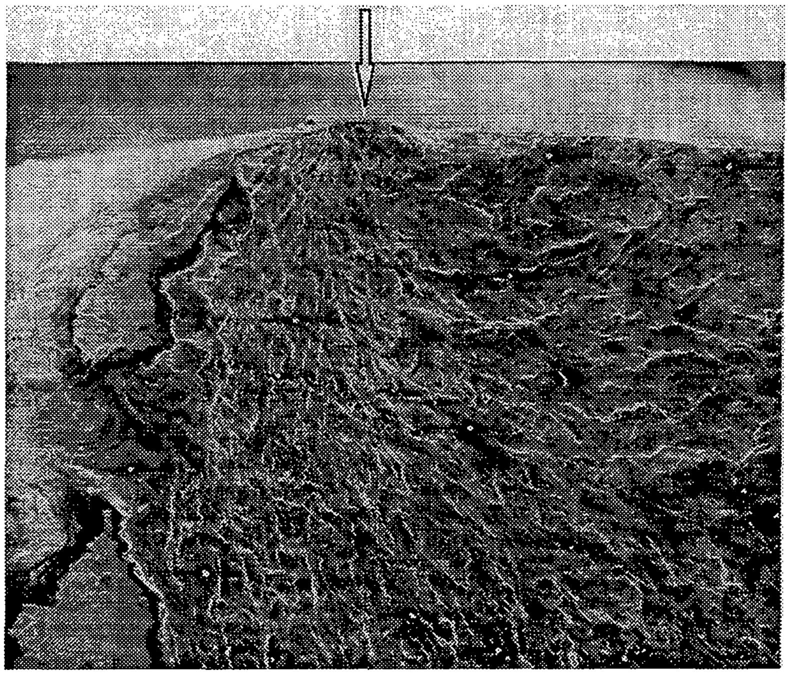

Figure 18. SEM Macrograph of the Primary Fracture Origin (Indicated by Arrow); Magnification 50×.

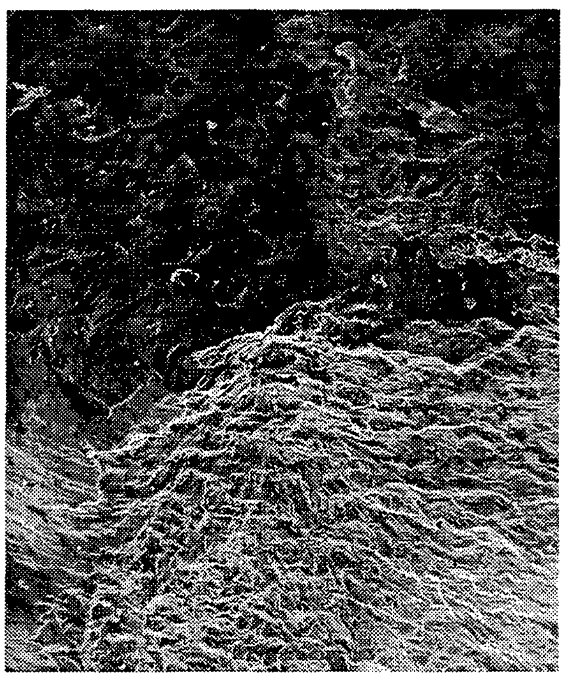

Figure 19. SEM of the Fracture Origin of the Primary Fracture, Tilted at an Angle to Reveal the Surface of the Bore; Magnification 50×.

while Figure 20 is a magnification of this area showing intergranular surface attack adjacent to the origin. Beyond the fracture region, a small shear lip consisting of ductile dimples was observed, which represented final fracture. No inherent defects or gross inclusions were noted at the origin of the primary fracture. The topography of the secondary fracture was characterized by a smaller region of fatigue, with the remainder of the surface typical of overload. The fatigue morphology was characterized by a shiny appearance, in contrast to the overload region that displayed a dull texture. A shear lip encompassed the outer periphery of the fracture surface, except for the bushing wall. An intergranular morphology was evident at the fracture origin. Figure 21 shows the origin of the secondary fracture. The origin at an angle is shown in Figure 22. Figure 23 shows intergranular surface attack adjacent to the origin. EDS examination of the dark golden colored coating at Boeing revealed a level of chromium consistent with a chromic acid anodize.

Figure 20. Magnified SEM of Figure 19, Showing Intergranular Attack on the Surface of the Bore Adjacent to the Primary Fracture Origin; Magnification 200×.

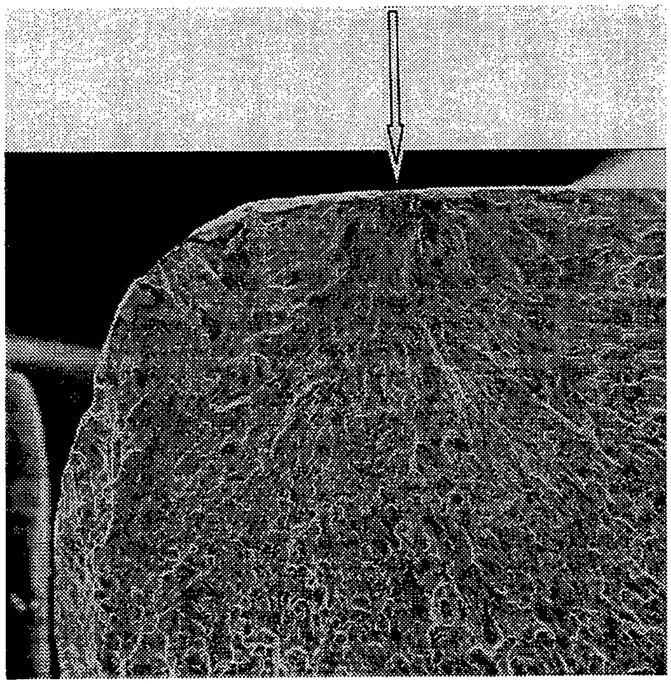

Figure 21. SEM of the Secondary Fracture Origin (Indicated by Arrow); Magnification 50×.

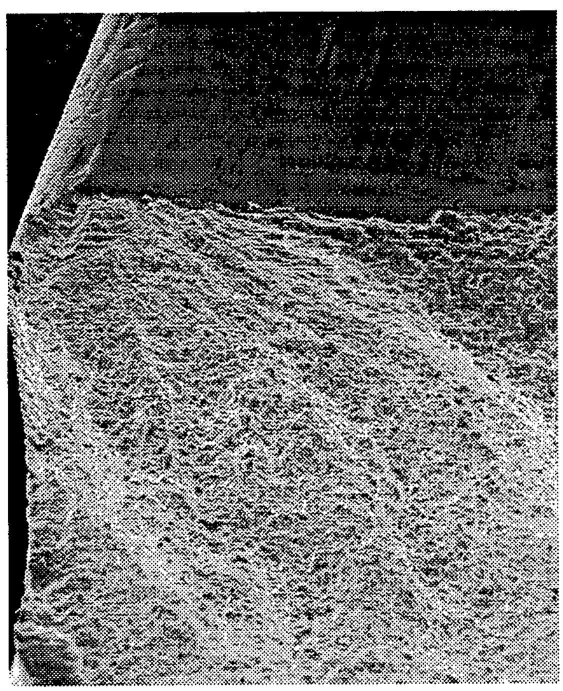

Figure 22. SEM of the Secondary Fracture Origin, Tilted at an Angle to Reveal the Surface of the Bore; Magnification 50×.

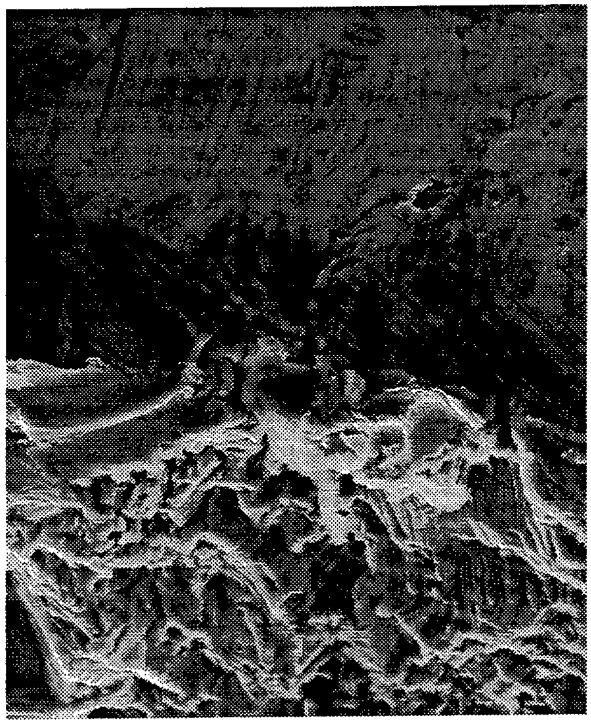

Figure 23. Magnified SEM of Figure 22 Showing Intergranular Attack on the Surface of the Bore Adjacent to the Secondary Fracture Origin; Magnification 500×.

5. Hardness Testing

Hardness testing was performed by CCAD, Boeing, and ARL. ARL performed tests on three specimens representing a thick section, a region adjacent to the primary fracture, and a region adjacent to the secondary fracture. The range of hardness results obtained by CCAD was higher than the ranges obtained by Boeing and ARL, as shown in Table 1. The CCAD results did not conform to the governing requirements, whereas, the results from Boeing and ARL did conform (1 of the 20 ARL readings did not conform, 89.5 Rockwell Hardness "B" Scale [HRB]). It was concluded that the hardness met the governing requirements.

Table 1. Summary of Hardness Results

Testing Activity	Results (HRB)
CCAD (7 readings)	91–93
Boeing (6 readings)	87.7–88.5
ARL (20 readings)	86.8–89.5
Boeing Specification BAC 5946 [2] for -T73	79.5–89.0

6. Conductivity Testing

Conductivity testing was also performed by ARL to provide a third set of data to compare to the results of CCAD and Boeing. The CCAD results averaged 38.3% International Annealed Copper Standard (IACS), which met the acceptance limit of 38.0% to 41.9% IACS as listed within the governing specification of MIL-A-22771 [3]. The results of conductivity testing conducted by Boeing fell within the range of 37.5% to 38.5% IACS, which was on the low end of the acceptable limit. Conductivity testing was also performed by ARL, and the results obtained closely resembled those measured by both CCAD and Boeing, and conformed to the governing requirements. A summary of the conductivity results is provided in Table 2.

Table 2. Summary of Conductivity Results

Testing Activity	Results (% IACS)
CCAD Average	38.3
Boeing Range	37.5–38.5
ARL Range	38.1–39.7
Boeing Specification BAC 5946 [2] for -T73	38.0–41.9

7. Tensile Testing

Boeing specification BAC 5946 [2] requires a hardness of 79.5 to 89.0 HRB. The hardness results as measured by Boeing were between 87.7 to 88.5 HRB, which met this requirement. The specification also requires a conductivity of 38.0 to 41.9% IACS. The conductivity measured by Boeing was from 37.5% to 38.5% IACS. This range was on the low end of the acceptable limits. The values of hardness and conductivity measured by ARL closely resembled those measured by Boeing. Boeing performed tensile tests on three samples, in accordance with paragraph 3.4.2.3 of MIL-A-22771 [3]:

> If the electrical conductivity is 38.0% to 39.9% IACS, inclusive, if tensile strength properties meet the minimum limits specified herein, and the longitudinal yield strength does not exceed the specified minimum by more than 11.9 ksi, the forgings shall be considered satisfactory.

and

> If conductivity is below 40.0% IACS and the longitudinal yield strength exceeds the specified minimum value by 12.00 ksi or more, the lot is unacceptable.

The results of these tensile tests revealed a yield strength of 69.6 ksi for one of the specimens, which was in violation of paragraph 3.4.2.3 of MIL-A-22771 [3], since this yield strength was greater than the required yield strength (56 ksi) by 13.6 ksi (11.9 ksi was the maximum allowance).

ARL believed the results of tensile testing were not conclusive in determining whether the proper aging sequence was performed on the failed part. It was felt that more specimens needed to be tested to provide a larger sample size. Therefore, ARL also performed five tensile tests to determine the acceptability of the material, using the acceptance criteria of paragraph 3.4.2.3. of

MIL-A-22771 [3]. Although the results of hardness testing and conductivity testing met the requirements of the more conservative Boeing specification, ARL believed that this testing was necessary to help determine whether the prior temper was indeed -T73. This was critical, since a different temper could detrimentally influence the intergranular corrosion resistance of the material.

A total of five threaded specimens (Figure 24) was fabricated in accordance with the requirements of ASTM E8, "Tension Testing of Metallic Materials" [4]. It was the initial intent of ARL to fabricate subsize specimens from the failed "C" section of the lag damper itself adjacent to the fracture surfaces, but size constraints would have made the measurement of yield strength from these specimens impossible. Since yield strength was the primary mechanical property that ARL was to use as the acceptance criteria for this material, standard specimens were fabricated from the housing, adjacent to the location in which Boeing sectioned their three tensile specimens.

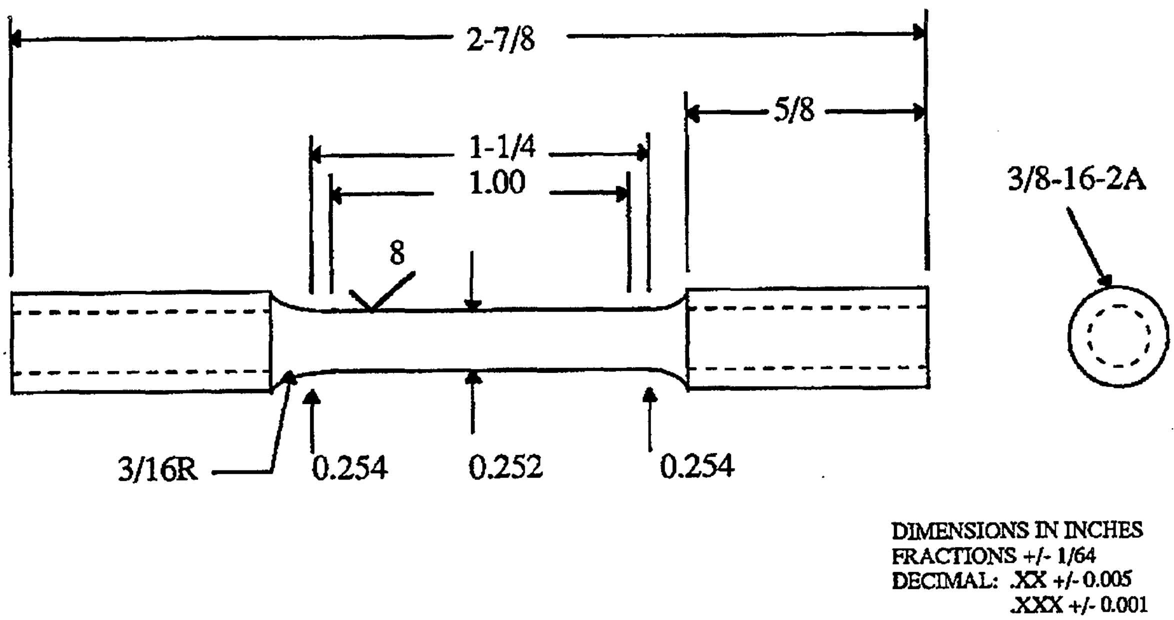

Figure 24. Schematic Showing Dimensions of Tensile Specimens Sectioned From the Blade Lag Shock Absorber Housing and Tested by ARL; Specimen Conformed to ASTM E8 [4].

The results of tensile testing are listed in Table 3. The "L" designation for each of the five samples indicates the samples were fabricated in a direction longitudinal to the length of the housing. Table 3 also contains the results of tensile testing performed by Boeing and the governing requirements. The UTS and elongation of each sample met these requirements. However, the 0.2% yield strength of four of the five specimens was greater than the maximum allowed. Including the results of Boeing, a total of five of the eight specimens failed to meet the required yield strength range.

Table 3. Summary of Tensile Testing Results

Sample	0.2% Yield Strength (ksi)	UTS (ksi)	Elongation (%)
ARL L1	71.6	80.2	10
ARL L2	70.9	79.4	11
ARL L3	67.2	75.7	9
ARL L4	70.6	76.9	10
ARL L5	68.8	78.2	10
Boeing - 1	67.4	76.7	13
Boeing - 2	69.6	78.8	10
Boeing - 3	63.5	74.9	11
MIL-A-22771 [3] (-T73)	56–67.9	66 (minimum)	7 (minimum)

Note: UTS = ultimate tensile strength.

8. Chemical Analysis

Chemical analysis was performed by ARL to verify the elemental composition of the alloy. The chemical composition is governed by MIL-A-22771 [3]. CCAD and Boeing both reported that the composition conformed to the governing requirements. The CCAD report stated that a Kevex x-ray fluorescence instrument was utilized to determine quantitative chemical analysis. Boeing and ARL utilized the wet chemistry method. ARL used inductively coupled plasma/atomic emission spectroscopy to determine the weight-percent of each element. Table 4 lists the results obtained by both ARL and Boeing (CCAD did not include the results of testing in

Table 4. **Summary of Chemical Analysis Results (Weight-Percent)**

Element	ARL Analysis	Boeing Analysis	MIL-A-22771 [3]
Copper	1.8	1.6	1.2 - 2.0
Silicon	0.19	0.13	0.40 (maximum)
Iron	0.18	0.18	0.50 (maximum)
Manganese	0.04	0.04	0.30 (maximum).
Magnesium	2.65	2.71	2.1–2.9
Zinc	5.92	5.91	5.1–6.1
Chromium	0.22	0.20	0.18–0.28
Titanium	0.04	0.02	0.20 (maximum)
Other Elements (each)	<0.01	N/A	0.05 (maximum)
Other Elements (total)	<0.05	N/A	0.15 (maximum)
Aluminum	Remainder	Remainder	Remainder
Bismuth	<0.01	—	—
Tin	<0.01	—	—
Beryllium	<0.01	—	—
Nickel	<0.01	—	—
Sodium	<0.01	—	—

their final report), as well as the requirements of MIL-A-22771 [3]. The results of both ARL and Boeing were similar, and each conformed to the governing requirement. Although small amounts of additional elements were detected by ARL, these findings were well within acceptable limits.

9. Prior History

Conversation with William Alvarez, an Engineer with the U.S. Army Aviation and Missile Command (AMCOM) revealed that, at the time of rework of this component, an aqueous alkaline immersion cleaner was being utilized at CCAD, rather than the required (and already approved) solvent cleaner [8]. This alkaline cleaner was named Daraclean 282 (data sheets are included in Appendix C) and was being used in the parts washer equipment. Apparently, CCAD personnel noted that an adherent white substance had enveloped the components processed with Daraclean 282, so a product named Daracoat G15 was used as a rinse-agent additive. In addition, tap water was being used (which was shown to contain calcium carbonate) instead of deionized water. After cleaning, the parts were sent to be nondestructively inspected (NDI). Daracoat purportedly left a greasy-like film on the components, which may have masked cracks that were already present. It was also conveyed that parts may have been stored for up to 2 days with the Daraclean cleaner on them before the nondestructive inspection took place. This may have posed the worst threat to the components since past formulations of this cleaner have been shown to have a history of failures associated with it when utilized as a replacement for degreasing operations. Northrop Report 3882-91-61 [9], dated 14 October 1991, states,

> Daraclean 282 and Oakemclean were eliminated [as viable replacements for 1,1,1 Trichloroethane for cleaning prior to NDI] because they failed the soil removal test badly at the maximum desirable operating temperatures and maximum concentration recommended by the manufacturers. Daraclean 282 was also found by General Dynamics to fail intergranular attack, sandwich corrosion test and storage stability test.

10. Discussion

The findings included in the Boeing report [10] from the examination of the inboard lug bore wall and spot face suggested that the bore had been reworked. The bore surface was the only region of the failed component that exhibited intergranular corrosion. Therefore, it is reasonable to assume that the rework process was responsible for the intergranular attack. It is likely that this attack occurred as a result of rework processing solutions. As recommended by Boeing, the fluids utilized to rework these components should be contaminant free and controlled in accordance with the governing specifications. However, the exact processing fluid needs to be identified, since the component is exposed to a variety of processing steps, including cleaning and machining. The problem can only be alleviated if the contaminated fluids are identified and corrected.

Type 7075 aluminum in the -T73 condition offers a superior combination of high strength and resistance to stress corrosion and exfoliation corrosion. The strength is less than that of the -T6 temper, but the corrosion resistance is greater (as strength increases, corrosion resistance tends to decrease for this alloy). However, the properties obtained by this condition are extremely sensitive to small variations in aging times and aging temperatures (approximately five times greater sensitivity than that of the -T6 condition, for comparison). Consequently, control of both temperature and time to achieve the mechanical properties and corrosion resistance specified for these tempers is more critical than the control required in producing the -T6 temper [5]. The -T6 aging temperature can range from ~225° F to ~275° F with a variation in UTS of only 4 ksi. However, it has been shown that variations of 50° F during the -T73 aging treatment affects the UTS of -T73 by as much as 16 ksi. (Figure 25 [11]). In addition, the rate of heating from the first to the second aging step has to be taken into account since precipitation also occurs during this period. In order to determine whether the alloy had been properly aged, hardness and conductivity testing was performed. The material exhibited hardness and conductivity measurements within the respective limits but at the low end of the conductivity range and the high end of the hardness range. Tensile testing was conducted as the deciding factor to determine whether the proper aging sequence was performed, since the conductivity measured between 38.0 and 39.0% IACS (per MIL-A-22771 [3]). Testing tensile performed by ARL revealed yield strengths greater than the maximum requirement for four of the five specimens. Taking into account the tensile testing performed by Boeing, a total of five of eight specimens failed to meet this requirement. This showed that the material may not have been properly aged per the requirements of MIL-A-22771 [3]. It is possible that this would have facilitated intergranular attack, but it is also possible that the rework solutions may have had the same effect on properly aged 7075-T73 material.

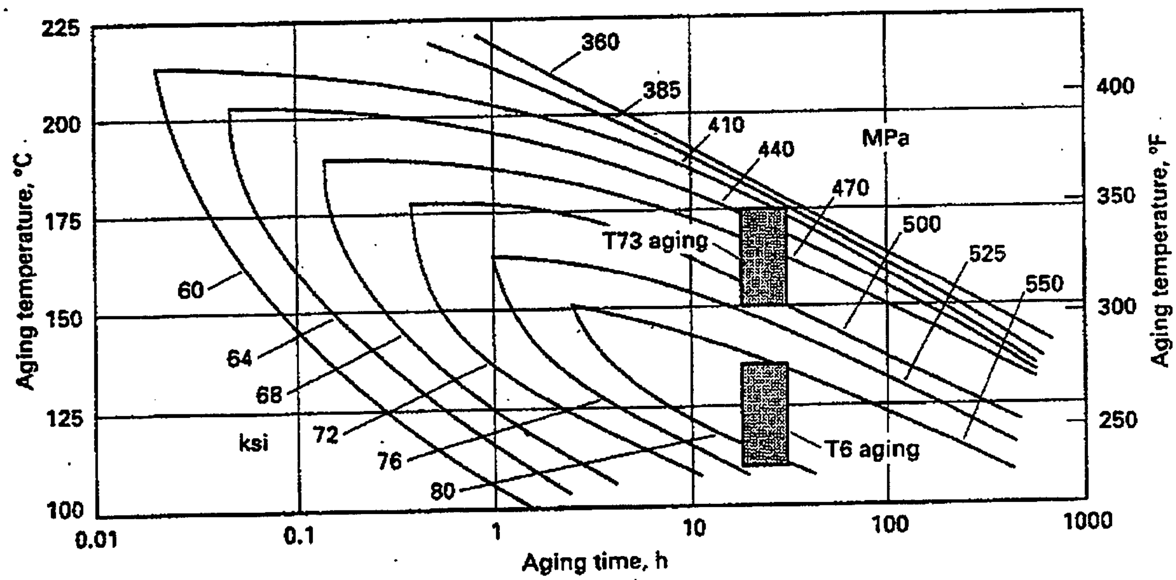

Figure 25. Iso-Yield Strength Curves for Aluminum Alloy 7075 [11].

11. Conclusions

(1) The component under investigation failed in fatigue that initiated from intergranular attack at the bore region. The corrosive attack was most likely caused by processing fluids utilized during the rework process. It was determined that the part was reworked as evidenced by axial score marks and a blackish-green coating within the bore.

(2) The part failed in fatigue producing two through-fractures. The primary fatigue crack traversed approximately 95% of the entire fracture surface. The fracture surface of the secondary fatigue contained approximately 20% fatigue. The origins of both fractures were intergranular in morphology.

(3) Hardness measurements were consistently at the high end of the acceptable range, while conductivity measurements were consistently at the low end of the acceptable range.

(4) Metallographic examination through the inboard lug showed evidence of intergranular attack only on the reworked bore region. Intergranular attack was noted on a sample taken adjacent to the primary and secondary fracture, as well as a sample taken through the chamfer.

(5) Tensile testing was conducted on five specimens to help determine whether the proper aging treatment was performed. The 0.2% yield strength results of four of the five ARL specimens failed to meet the requirements of MIL-A-22771 [3], indicating that the component may not have been properly aged. Including the results obtained by Boeing, a total of five of eight specimens failed to meet this requirement. Improper aging would have most likely impaired the intergranular corrosion resistance of this material.

(6) The results of chemical analysis conformed to the requirements of MIL-A-22771 [3] for 7075-T73 aluminum.

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12. References

1. Corpus Christi Army Depot. "Failed Housing, Shock Absorber." USASC 95-309, Corpus Christi, TX, September 1995.

2. Boeing. *Temper Inspection of Aluminum Alloys.* BAC 5946, rev. T, Philadelphia, PA, 15 December 1993.

3. U.S. Department of the Army. *Aluminum Alloy Forging, Heat Treated.* MIL-A-22771, rev. D, 30 January 1994.

4. American Society for Testing of Materials. *Tension Testing of Metallic Materials.* ASTM E8, Philadelphia, PA, 1990.

5. U.S. Department of the Army. *Heat Treatment of Aluminum Alloys.* MIL-H-6088, note 8, p. 53, 1 April 1991.

6. Davis, J. R. "Aluminum and Aluminum Alloys." *ASM Specialty Handbook*, p. 591, Materials Park, OH, 1994.

7. Kennedy, J. R. "Acoustic Emission During Deformation of Heat Treated 7075 Aluminum." *Metal Science*, vol. 18, p. 286, June 1984.

8. Alvarez, W. Personal communication. U.S. Army Aviation and Missile Command, Huntsville, AL, 26 January 1998.

9. Northrop. "Replacements for 1,1,1 Trichloroethane." Report 3882-91-61, Los Angeles, CA, 14 October 1991.

10. Boeing Materials Engineering Laboratory. "CH-47D, Blade Lag Shock Absorber, P/N 114H6800-5, S/N VY1313, Metallurgical Evaluation of Through Fractured Inboard Lug." Report 95-169, Philadelphia, PA, 10 November 1995.

11. Davis, J. R. "Aluminum and Aluminum Alloys." *ASM Specialty Handbook*, p. 315, Materials Park, OH, 1994.

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Appendix A:

Corpus Christi Army Depot Investigation Reports USASC 95-309, 95MX11, and 95MX113, Dated 12 September 1995

The contents of this appendix appear in their original form, without editorial change.

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CORPUS CHRISTI ARMY DEPOT
ANALYTICAL INVESTIGATION BRANCH
SDSCC-QLA
308 CRECY ST
CORPUS CHRISTI, TX 78419-5260

OPERATING ACTIVITY:
COMMANDER
CO B, 2/158TH AVN REGT
FORT HOOD, TX 76544-5056

INVESTIGATOR:

Javier Benavides — 11 SEP 95
Javier Benavides, Equipment Specialist (ACFT), DSN 861-2902/2903 — Date

APPROVAL:

Gary S. Tindall — 12 Sept 95
Gary S. Tindall, C, AIB, DSN 861-2598/4681/2902 — Date

FOR THE COMMANDER:

E. Charles Wilson — 12 September 95
E. Charles Wilson, C, TAD, DSN 861-3198 — Date

END ITEM: (AIRCRAFT TYPE)	HELICOPTER, CH-47D
END ITEM SERIAL NUMBER (S/N):	86-01661

EXHIBIT

NOMENCLATURE:	SHOCK ABSORBER, BLADE LAG
NATIONAL STOCK NUMBER:	1650-01-106-9510
SERIAL NUMBER (S/N):	VY-1313
PART NUMBER (P/N):	114H6800-5
REMOVAL CODE:	070
REMOVAL DATE:	26 JUL 95
REASON FOR REMOVAL:	BROKE IN FLIGHT
PRIOR OVERHAULS:	2
USAGE SINCE NEW (HOURS):	UNKNOWN
USAGE SINCE OVERHAUL (HOURS):	UNKNOWN
DATE LAST OVERHAUL:	UNKNOWN
LAST OVERHAUL/REPAIR ACTIVITY:	CCAD
DA2410 CONTROL NUMBER:	NOT REQUIRED

PRIMARY FAILED PART(S):

NOMENCLATURE:	**HOUSING, SHOCK ABSORBER**
PART NUMBER:	114H6802-3
SERIAL NUMBER:	UNKNOWN
MANUFACTURER'S CODE:	77272
FAILURE CODE:	070

BACKGROUND/REQUEST:

1. The flight crew reported feeling a "one-per-rev" vibration during forward flight. A precautionary landing was accomplished. The aft red blade impacted the fuselage during coast down. This caused damage to the driveshaft covers and the red blade.

2. The blade lag shock absorber was submitted to the Analytical Investigation Branch (AIB) for analysis after it failed in flight.

CONCLUSIONS:

1. Analysis of the fractured blade lag shock absorber revealed that the failure originated and propagated due to corrosion fatigue. The fracture occurred through the inboard lug (bushing bore). Corrosion found on the fracture surfaces indicates that the fracture originated at some time prior to the final overstress fracture when the remaining material could not support the load.

2. The remainder of the shock absorber and its parts failed to reveal any significant defects.

DETAILS:

1. The lug fracture occurred essentially perpendicular to the longitudinal axis of the blade lag shock absorber (Exhibit 1, Arrows A-A).

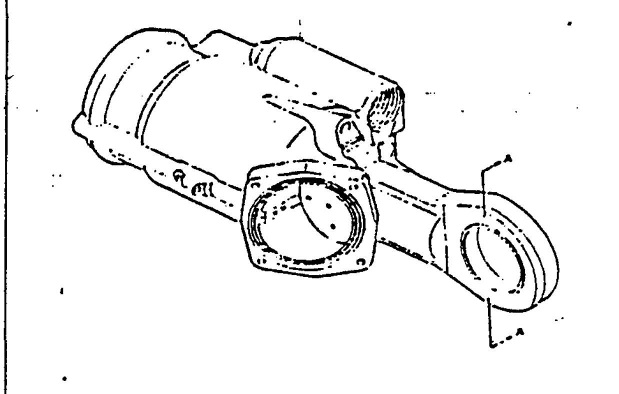


Exhibit 1. Location of lug fracture.

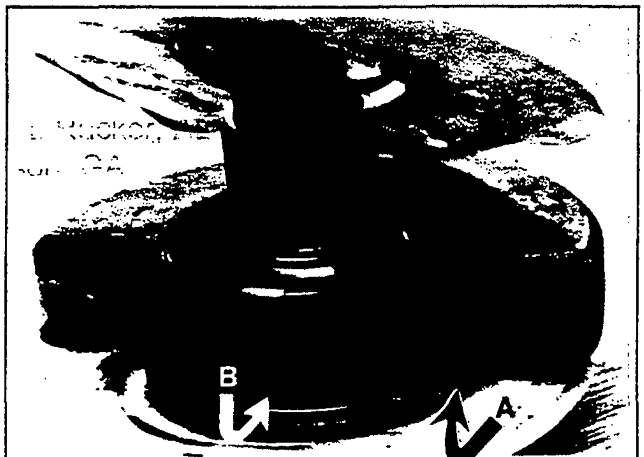


Exhibit 2. Outboard portion of lug, bushing, and bearing.

2. The bushing, P/N 114H6803-2, and the bearing, P/N 114H6S681-2, remained attached to the outboard portion of the lug (rotor head end).

3. The fracture surface of the blade lag shock absorber showed evidence of surface corrosion (Exhibit 2, Arrow A).

4. The topography of the fracture surface near the corrosion product appeared to exhibit beach marks (fatigue striations).

5. The bushing, P/N 114H6803-2, displayed a rust colored residue and some shiny bare metal areas on its outer diameter (Exhibit 2, Arrow B).

6. Removal of the bushing from the fractured lug revealed a dark colored residue, rust colored residue, shiny bare metal areas, and vertical score marks on the bushing bore. The score marks were made during installation or removal of the bushing.

Exhibit 3. Outboard end of lug (blade end).

7. The opposite end (blade end) of the lug fracture showed similar fracture features, as expected (Exhibit 3). The bushing bore showed a dark colored residue, a rust colored residue, and score marks similar to those found on the rotor head end of the lug (Exhibit 3).

8. The oil noted on the sight glass appeared amber in color.

9. The exterior surfaces of the remainder of the shock absorber failed to reveal any other significant defects.

FINDINGS:

1. Metallurgical examination (Reference: Corpus Christi Army Depot Analytical Investigation Branch Laboratory Report Number 95MX111, Enclosure 1) of the shock absorber revealed the following conditions:

- The fracture originated at the bushing bore chamfer.
- The exact cause (pits, mechanical notch, etc.) could not be determined due to the damaged condition.
- The failure was due primarily to corrosion fatigue.
- The secondary failure was due to overstress. The remaining solid metal could not support the flight stresses.
- The housing was of the specified alloy.
- The housing hardness was slightly higher than the accepted limits.

- The conductivity was lower than required for 7075-T73 material.

2. Radiographic examination of the shock absorber, as received, failed to reveal any obvious defects (Exhibits 4 and 5).

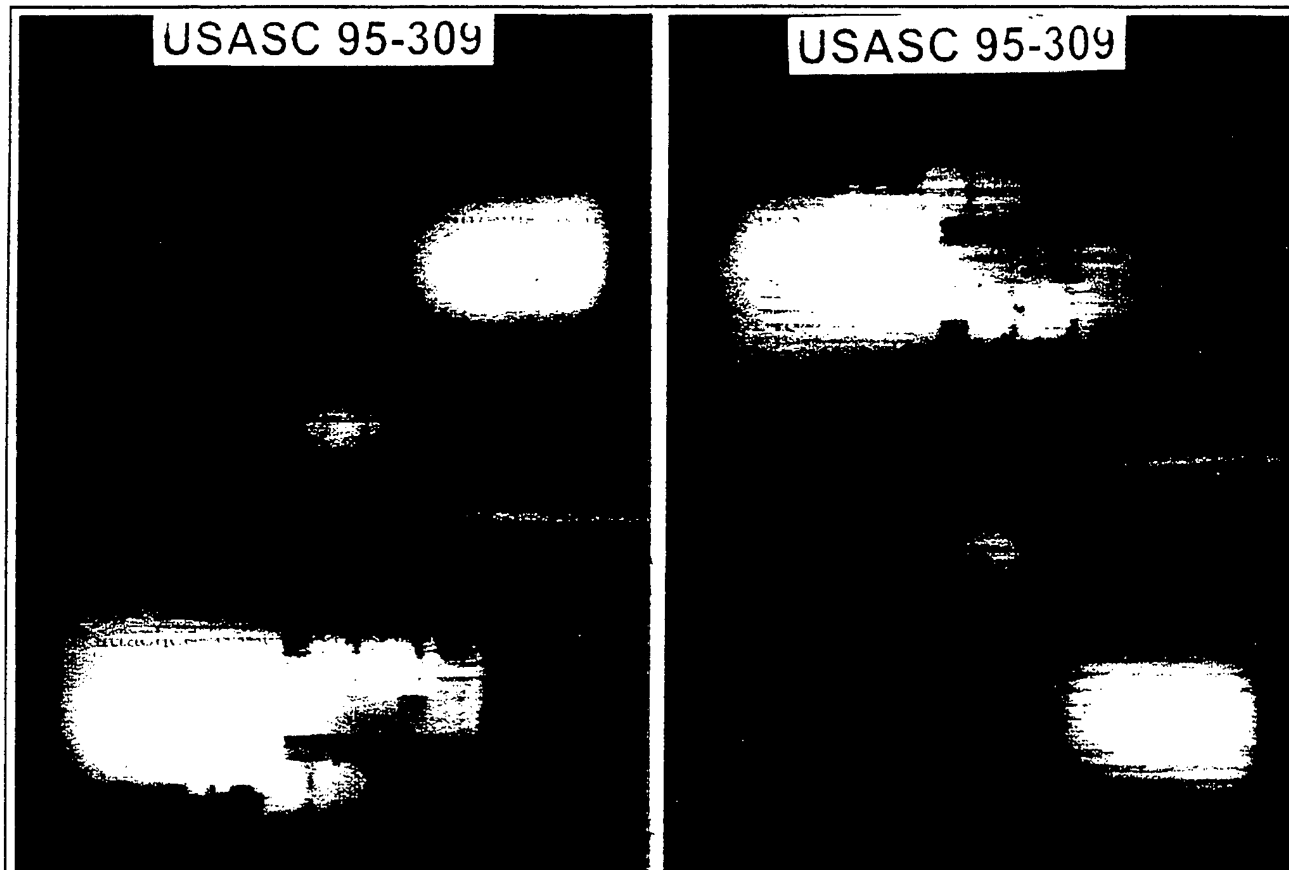


Exhibit 4 Exhibit 5

3. Functional testing of the component was not accomplished due to the condition of the shock absorber housing. The housing could not be safely installed on the test fixture.

4. Chemical analysis (Reference: Corpus Christi Army Depot Analytical Investigation Branch Laboratory Report Number 95MX113, Enclosure 2) of the residue found on the bushing bore disclosed the following:

- The dark residue was determined to be anodic coating (anodizing).
- The adhesive tape smears taken from the fracture surface did not provide information of value to the analysis.

9. Chemical analysis (Reference: Corpus Christi Army Depot Analytical Investigation Branch Laboratory Report Number 95 MX112, Enclosure 3) of the hydraulic oil sample disclosed the following:

- The fluid met the chemical requirements of MIL-L-5606, HYDRAULIC FLUID, PETROLEUM BASE, AIRCRAFT, MISSILE, AND ORDINANCE.
- MIL-L-5606 Hydraulic Fluid is red in color. The color of the oil submitted for analysis was dark amber.
- Spectrographic and ferrographic analysis revealed wear particles in the heavy use category.

10. Disassembly of the component and examination of its parts revealed heavy deposits throughout its internal components (Exhibit 6, Arrow). The internal component showed heavy wear typical of extended usage. The internal components did not reveal evidence of a failure or an impending failure (Exhibit 7). Several other shock absorbers were disassembled and examined. They appeared to be in the same condition (wear and debris). The CCAD shop mechanics at the CCAD Hydraulic Shop claim that the majority of the shock absorbers they process are in the same condition as the failed shock absorber (worn and contaminated). The discoloration of the hydraulic fluid is commonly seen by these mechanics.

Exhibit 6. Debris (typical) found inside shock absorber.

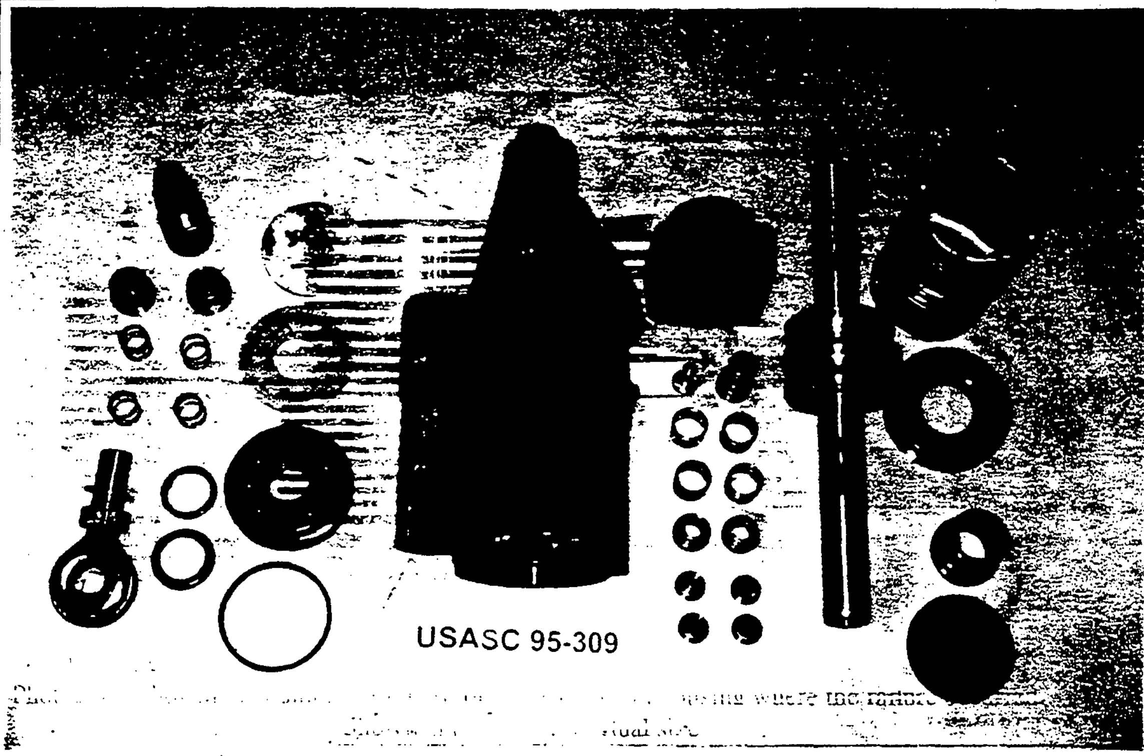


Exhibit 7. Disassembled blade lag shock absorber.

11. Dimensional inspection of the bushing bore on the housing was not possible due to the fracture of the lug. The outside diameter of the cylindrical portion of the bearing bushing (P/N 114H6803-2) was 0.0023 inch oversized. The effect the oversized bushing would have on the housing lug is not known at this time. All other dimensions are not considered a factor at this time.

12. Two unserviceable shock absorber housings were located at this facility. The housings were dimensionally inspected (lug area) and examined. The two housings were within specifications.

RECOMMENDATIONS:

1. Nondestructive test the inboard end lug area of all used blade lag shock absorbers.

2. Hermetically seal the inboard end bushings on the blade lag shock absorber housing during bushing replacement.

3. Incorporate a hermetic type seal on the lug/elastomeric bearing interface of the latest configuration blade lag shock absorbers (P/N 114H6800-11).

DISPOSITION:

The exhibit was shipped to Boeing Defense & Space Group (at ATCOM's request) for further analysis.

DISTRIBUTION:

COMMANDER
U.S. Army Aviation Troop Command
ATTN: AMSAT-R-E
AMSAT-I-ME
AMSAT-MED (CCAD)
AMSAT-C-X
AMSAT-A-TBB
U.S. Army Safety Center, ATTN: CSSC-O-RP, Ft. Rucker, AL
Forces Command, ATTN: AFPI-S0, Ft. McPherson, GA

CORPUS CHRISTI ARMY DEPOT

ANALYTICAL INVESTIGATION BRANCH
LABORATORY REPORT

FOR: AIB Exhibit # USASC 95-309 REQUESTER: J. Benavides

PREPARED BY: Dwight C. Deck
DWIGHT C. DECK, Materials Engineer DSN 861-4681

ASSISTED BY: Enrique J. Sosa
ENRIQUE J. SOSA, Materials Engineering Technician

RELEASED BY: Gary S. Tindall 17 August 1995
GARY S. TINDALL, Chief, Analytical Investigation Branch Date

SUBJECT

Part Shock Absorber, P/N 114H6802-3, S/N unknown

*Note: The failure location (portion of the housing containing the uniball which attaches to the Rotary Wing Head) was removed with the aid of a power band saw. Only this portion of the housing was brought to the Metallurgical Laboratory for failure analysis.

Aircraft CH-47D, S/N, 86-01661

Major Assembly Shock Absorber Blade Lag, P/N 114H6800-5, S/N VY-1313

Material Aluminum

Keywords Corrosion, fatigue, overstress

OBJECTIVES

To determine the mode of failure of the portion of the Blade Lag Shock Absorber that contained the uniball.

CONCLUSIONS

Scanning electron microscopic and metallographic examinations defined that fracture of the Blade Lag Shock Absorber **originated and propagated due to corrosion fatigue**. The failure originated in a bore of the housing at the uniball location near (within 0.1") the chamfer located at the top position. Corrosion (pitting and intergranular) provided sites for initiation of cracking, which most likely began on one side of the housing, and was propagated by cyclic stress encountered in service. Final fracture occurred when the remaining sound metal could no longer support this stress.

The housing was specified to be of 7075-T73 aluminum. Test results confirmed the part to be of the specified alloy, however, the hardness (90 to 93 Rockwell "B") was slightly above Boeing's acceptance limits[1] (79.5 - 89 Rockwell "B") and above the AMS 2658A-91 requirement (86 Rockwell "B"). In addition, the conductivity of the part was below the AMS 2658A-91 requirements for the 7075-T73 material.

DETAILS OF ANALYSIS

Visual and Stereomicroscopic Examinations

The portion of the Blade Lag Shock Absorber housing where fracture occurred was removed from the remainder of the part and brought to the laboratory for study. Fracture occurred approximately one eight inch from the center line of the bore containing the uniball bearing (where the Shock Absorber is attached to the Rotary Head) and was approximately 90° to the length of the housing. The bearing, with attached hardware (bushing, bolt, etc.) remained attached to the larger section which fractured from the housing. See Figure 1.

Figure 1 -- Photograph of section of Blade Lag Shock Absorber brought to the laboratory for study. Note the location of the fracture. Shown approximately 0.6X actual size.

[1]per Jim Kachelries, Staff Metallurgist & Corrosion Control Specialist, Materials Engineering Helicopters Division, Boeing Defense & Space Group.

Inspection of the fracture surfaces (to both sides of the bearing) revealed yellow to brown coloration near the chamfered edge located on the top side (positioned under the shoulder of the bushing) of the bore. The coloration was much more pronounced on one fracture suggesting this fracture occurred first (hereafter referred to as the primary fracture). Beach markings, indicative of fatigue damage, appeared to have originated at, or near, the chamfered edge of this fracture. The beach markings were visible on approximately 75% of the fracture surface. Inspection of the second fracture (hereafter referred to as the secondary fracture) revealed a coarse texture suggesting this failure was primarily due to overstress. See Figure 2.

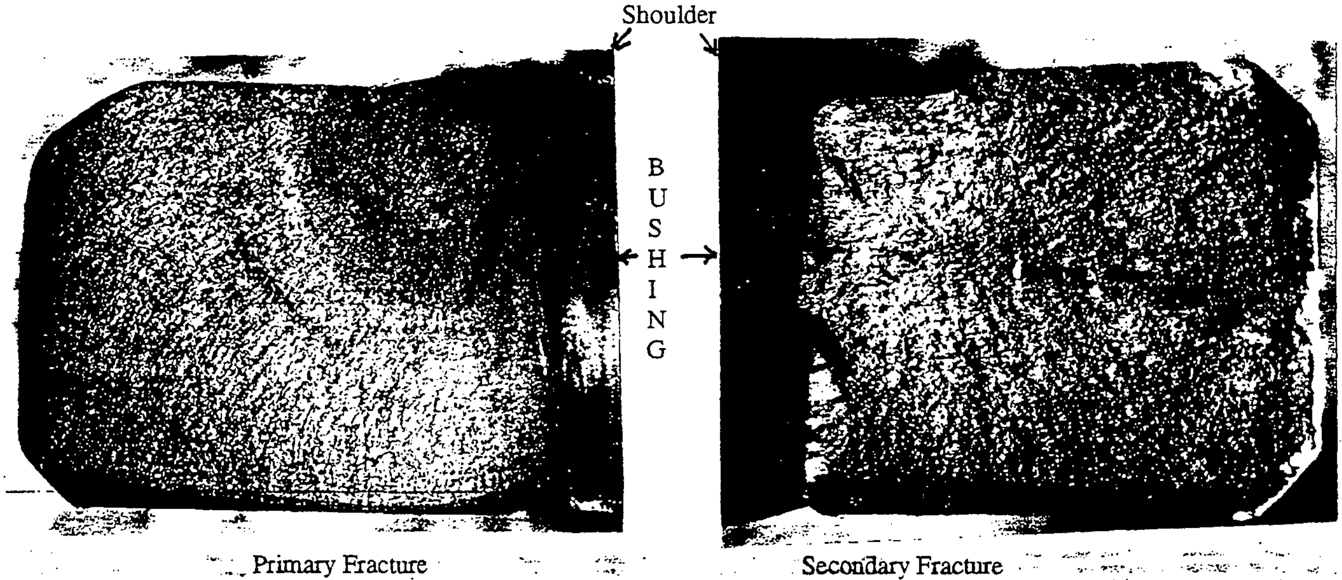


Figure 2 -- Photographs showing appearance of fracture surfaces. Note the visible beach marks. Note darkened area which appeared to be origin of cracking.
Shown approximately 2.8X actual size.

A single light tap with 8 oz. ball peen hammer was used to remove the portion of housing which remained attached to the bushing/uniball bearing. Inspection, at magnifications to 30X, of both halves of the unibearing bore in the housing revealed the following and are shown in Figure 3.

*** The surfaces were covered with an oily residue.

*** Longitudinal score markings, apparently resulting during installation, on the bore surface .

*** Bright metallic areas to opposite sides of the width of bore surface. These areas were not completely around the circumference of the bore but were located to both sides of the primary fracture surface and appeared to have resulted due to applied pressure (as could occur due to high areas as the bushing was installed).

*** Arc shaped darker colored areas were noticeable on both sides of the bearing width, located 90° to the fractures, and on the portion of the bore which stayed on the remainder of the housing.

*** There appeared to be slight pitting corrosion on the bore surface and on the side of the housing located under the busing.

*** Areas other than those described above were mostly black in color.

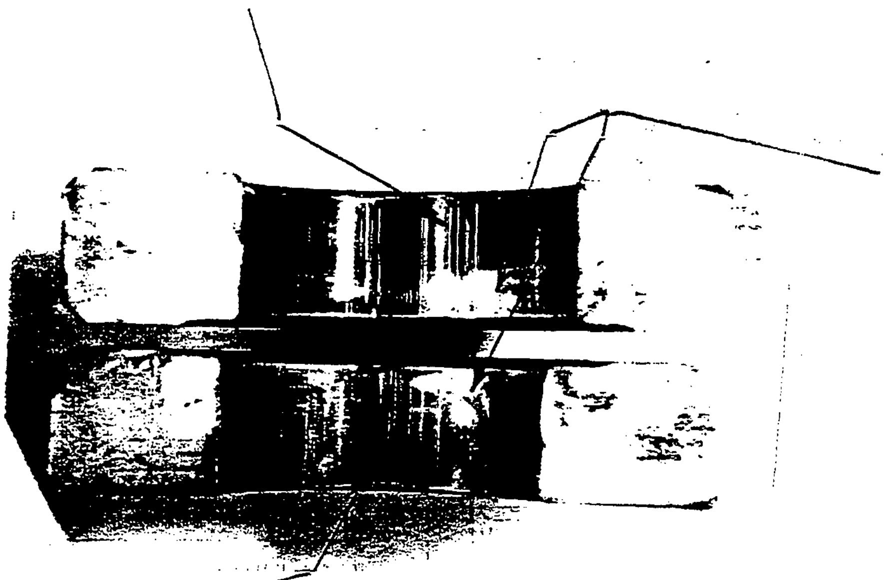


Figure 3 -- Photograph showing conditions observed on uniball bore in housing where the failure occurred. Shown approximately actual size.

The outer surface of the bushing (which mates with the bore in the housing) was also inspected at magnifications to 30X. See Figure 4. These inspections revealed the following.

*** Deep, short score marks located near the center of the surface which contact the bore surface. The score marks were present under the portion of the housing which fractured from the remainder of the part.

*** Evidence of fretting and slight pitting corrosion.

*** A pronounced scored area, located near the center of bushing width, adjacent to where the housing fractured. This area was located under the piece of housing which was still attached to the remainder of the housing.

*** A black band, likely of corrosion products, was present across the width of the bushing. One edge of the band followed the contour of the fractured edge of the bore of the housing.

*** Evidence of corrosion to the other associated hardware (bolt, nut, etc.).

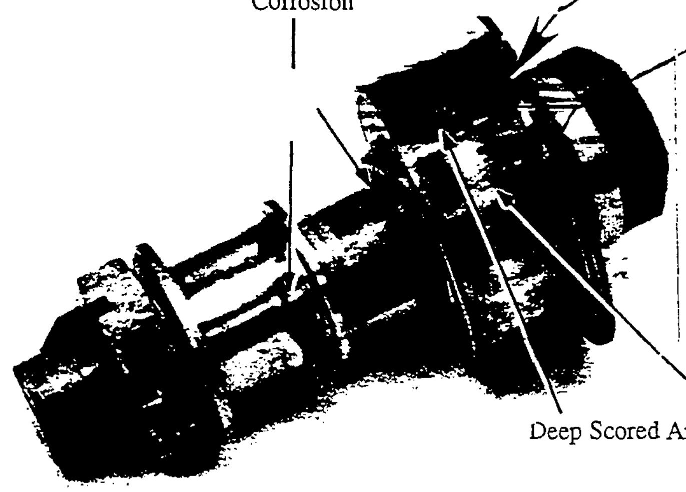


Figure 4 -- Photograph of Bushing and other hardware brought to laboratory with the fracture section of the housing.

Scanning Electron Microscopic (SEM) Examinations

The fracture surfaces were cut from the bore section using an abrasive cut-off, dried and rinsed with acetone, then mounted in a micro-vise and inserted in the vacuum chamber of the scanning electron microscope.

Magnifications to 5000X were utilized during the SEM examinations. Above 2000X, resolution of the fracture topography limited these examinations.

Initial studies revealed "dried river bed" appearing areas on both fracture surfaces. (Note: This appearance most often results from a film of deposits left after a contaminated liquid present on a surface evaporates.) After cleaning the fracture surfaces using acetone and a solution of Branson's Cleaning Concentrate, the fracture surfaces were re-examined. The observed conditions area shown in Figure 5 and 6 and described as follows..

1. Secondary fracture (See Figure 5):

    *** Well defined striations, visible at 500X, circulated around the area of the chamfer (positioned on flared side of the bushing, in the corner). This defined that the failure likely originated due to fatigue and the origin was at or near the chamfered edge.

    *** Small areas of intergranular fracture were visible near the surface at the bore.

*** The fracture surface, in the area where the striations hadindicated to be the origin, experienced post cracking damage, that is, was flattened and appeared to have been corroded. It thus was impossible to pinpoint the exact point of the fracture origin but appeared to be from the bore surface and within one-sixteenth inch of the chamfered edge.

*** Striations were visible to approximately one-fourth inch from the chamfered edge defining that the cracking was propagated due to cyclic stress. The remainder of the fracture topography was indicative of that resulting from overstress.

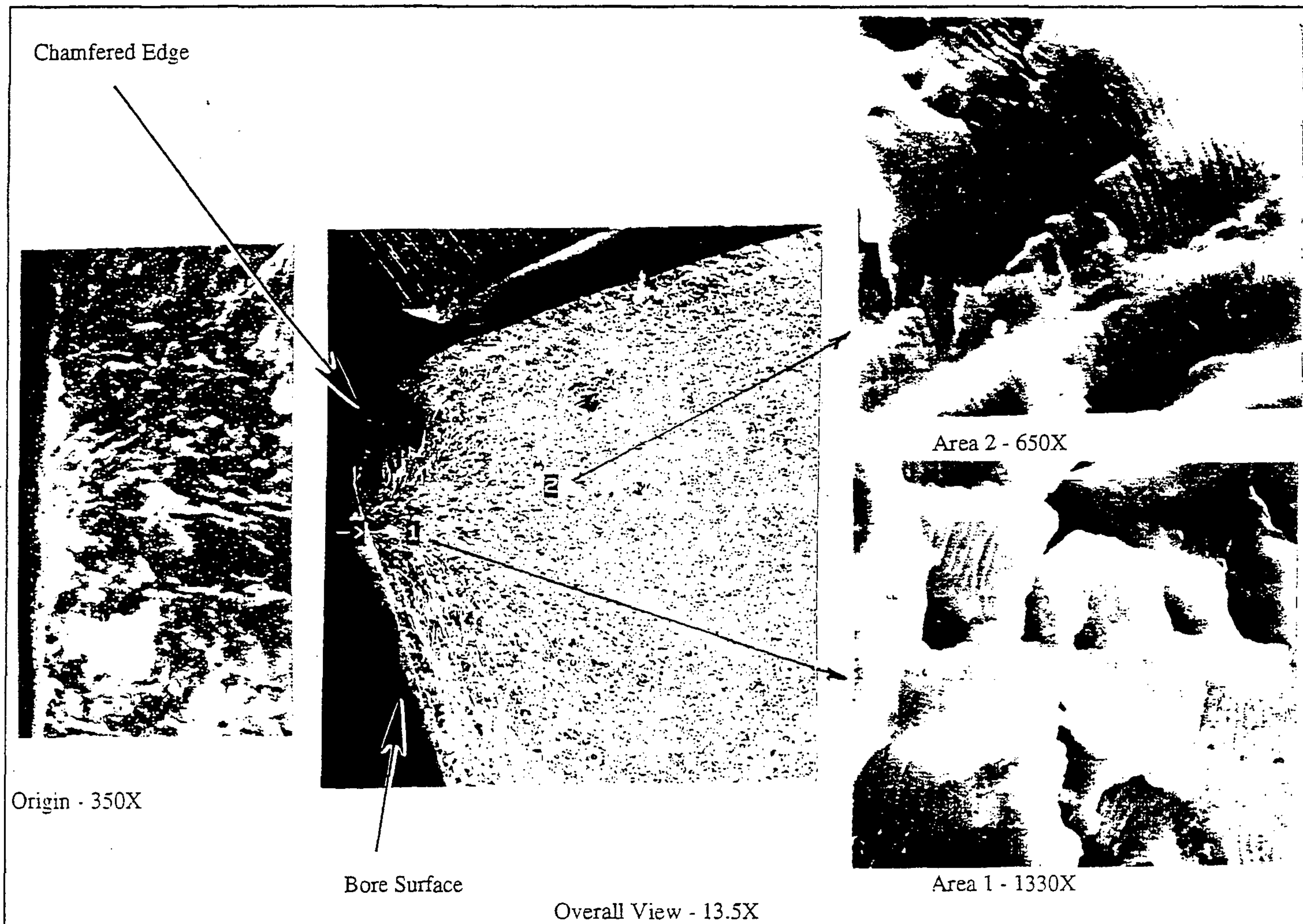


Figure 5 -- SEM views of fracture surface of secondary fracture. Note clear definition of striations (fatigue damage). Magnifications as shown.

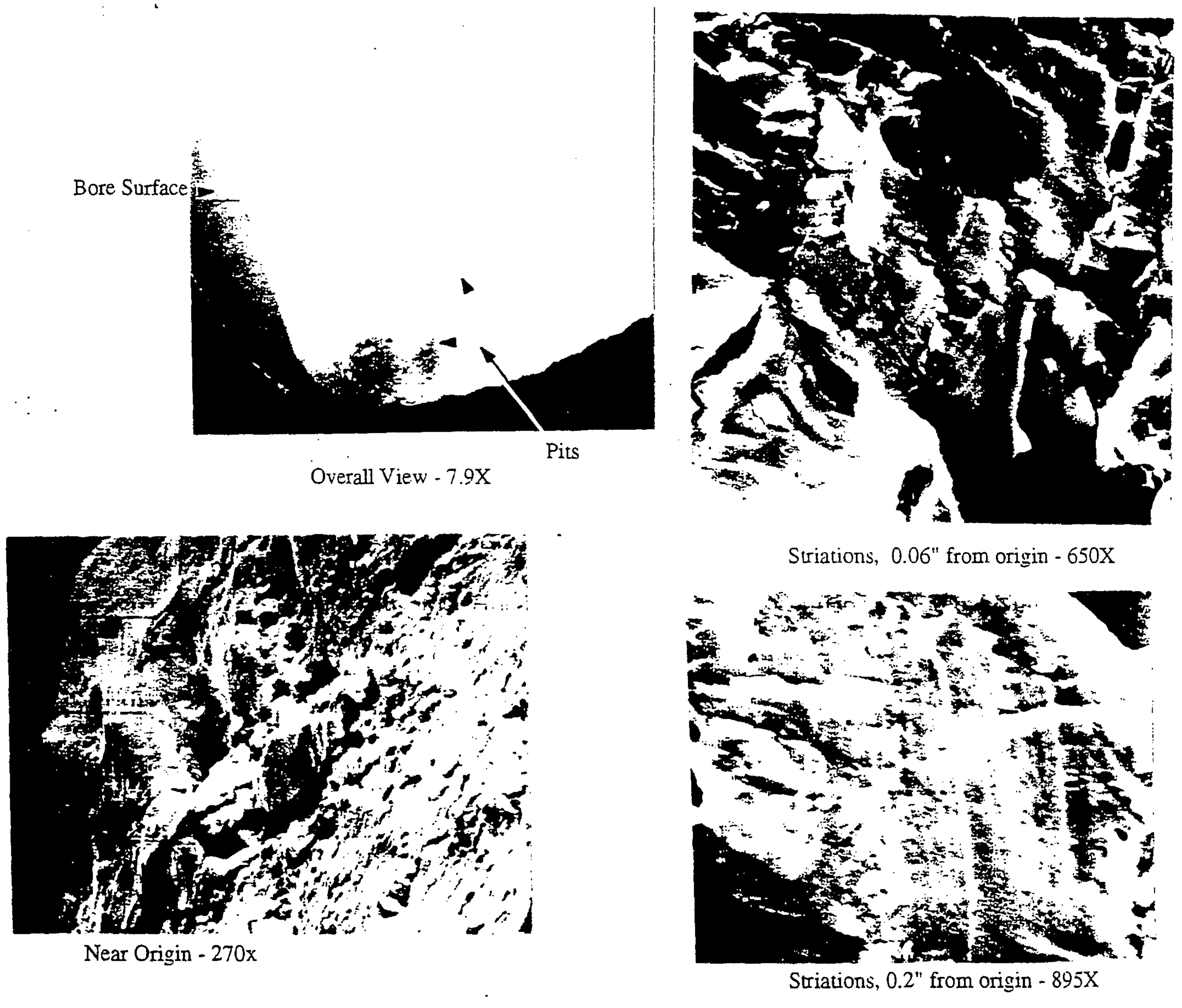
2. Primary fracture (See Figure 6):

*** Like the secondary fracture, striations circulated around the area of the chamfer (positioned on the shouldered side of the bushing, in the corner). This defined that the failure likely originated due to fatigue and the origin was at, or near the chamfered edge.

Encl 1

*** Some pitting and intergranular corrosion was noted on the fracture surface where the deposits were observed during the visual examination. This, as well as mechanical damage (flattening from contact of mated surface) prevented definition of the exact point of the fracture origin.

*** The striations, clearly visible at magnifications to 1000X, were present on 80-90% of the fracture surface. The final fracture, located on the diametrical opposite corner of the fracture surface, had a topography indicative of overstress.



Overall View - 7.9X

Striations, 0.06" from origin - 650X

Near Origin - 270x

Striations, 0.2" from origin - 895X

Figure 6 -- SEM photomicrographs showing topographies of primary fracture surface. Note areas of corrosion and striations. Magnifications as shown.

Metallographic Examinations

The two fractures were mounted separately in Lucite mounting material with the side facing the shoulder of the bushing facing down. This technique was used as an attempt to grind/polish to the point of the fracture origin.

After the first grind/polish, and at the depth of contact with the chamfered edge , the following conditions were observed and are shown in Figures 7 thru 9.

*** Several very fine fatigue cracks, many of which originated in pits in the surface, were observed. The cracks were transgranular, that is, ran across the grains of the microstructure. All the cracks originated from the bore of the housing. The cracks in the secondary fracture were more numerous, but were not as deep, than those observed in the primary fracture.

*** Intergranular corrosion (following the grain boundaries) was observed on the polished surface, that is, on the surface facing the shoulder.

*** Some secondary cracks from the fracture surfaces. These cracks were small and transgranular.

320X 400X 80X

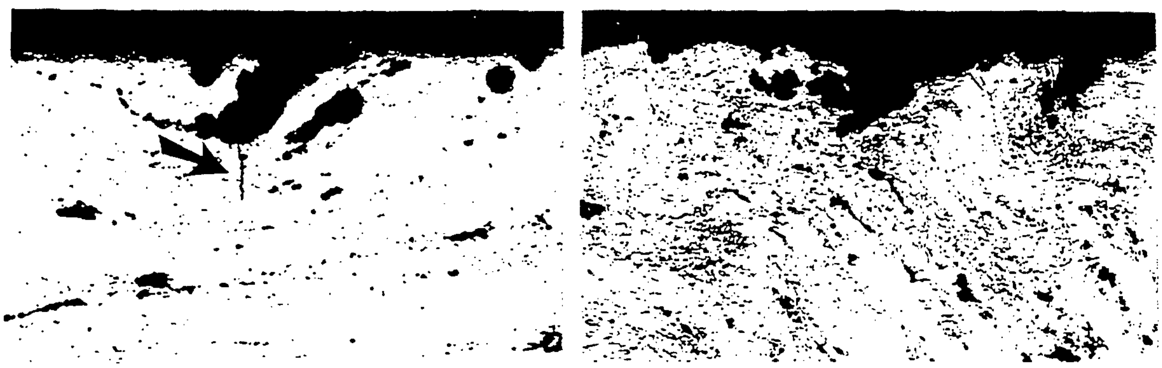
Figure 7 -- Metallographic views of corrosion fatigue cracks, typical of those which originated from corroded areas (pits and/or intergranular attack) on the surface of the chamfer near both the primary and secondary fractures. Note that the cracking is transgranular.

Magnifications: As shown Etchant: $HF+HCl+HNO_3$

Figure 8 -- View showing transgranular cracking from bore surface (top) and secondary cracking from fracture surface (left). See arrows
Magnification: 32X Etchant: $HF+HCl+HNO_3$

Figure 9 -- View of intergranular corrosion. This area of attack originated from side of housing which contacted the shouldered side of the bushing.
Magnification: 320X Etchant: $HF+HCl+HNO_3$

Sections were also removed approximately 10° and 90° (around the circumference of the bore) to the primary fracture. Pitting corrosion was present at both locations. A few (3), very shallow, transgranular cracking was present in the section remove one-half inch from the primary fracture but none were present in the section removed 90° from the fracture (See Figures 10a and 10b). The grain orientation 90° from the fracture was nearly longitudinal to the part length thus should have favored intergranular cracking if the hoop stress resulting from installation of the uniball bearing were the stress which caused the failure; however, no cracking was observed. It appeared very probable that cyclic tensile stress imparted at the location of the failure during operation provided the stress necessary to cause the fracture.



10a - 1/2" from fracture, 400X 10b - 90° from fracture, 200X

Figures 10a and 10b -- Views of chamfer located ~ one-half inch (left) and 90° (right) from fracture. Note small crack (see arrow) and grain orientation. Magnifications as shown.

The bushing was sectioned through the scored areas. Although oxides were present in the area of the deep score marks, there was no evidence of overheating ("burning" as could result from use of a grinding wheel) in the microstructure below the scored areas. See Figure 11.

Figure 11 -- Cross section through scored area on bushing. Note that oxides are present on surface (see arrow) but that there is no evidence of overheating of the microstructure.
Magnification: 400X Etchant: Chrome Regia

Note: In addition to the examination of the housing at uniball location of failed shock absorber, two other shock absorber housings were sectioned for study at the location of the uniball. The bore of one of these housings (uniball location) was black with, what appeared to be paint, but free of deep scoring. The bore of the second shock absorber was metallic, that is, free of deposits. This bore had a deep circumferential groove worn in one side. Sections were removed from the bore of both of these shock absorbers where the one housing had fractured and 90° to this location. There was no cracking observed in either of these housings.

Composition, Hardness, and Conductivity Test Results

One side of the housing, near the fracture, was lightly ground on 60 grit silicon carbide paper to remove paint from the surface. This ground surface was used for the following tests.

a. Composition

Test results, obtained using the Kevex x-ray fluorescence instrument, identified the housing to be of the specified alloy, 7075 aluminum.

b. Hardness

Rockwell "B" hardness results were 91 to 93 (approximately 7 determinations). These values are above the Boeing's Acceptance Limit of Rockwell "B" 79.5 to 89.0 and above the AMS 2658A-91 requirement of less than Rockwell "B" 86 for 7075-T73 of thickness 0.5" to 2.0".

c. Conductivity

Test Result -- 38.3 % IACS

Boeing's Acceptance Limit -- 38 to 41.9 % IACS AMS 2658A-91 Specification - 40 to 43 (Note: 38 to 39.9 acceptable if Rockwell "B" is <86 for material of thickness 0.5" to 2.0".)

Military Specification MIL-A-22771 for heat treated 7075-T73 forgings requires tensile and yield strength properties be used as acceptence criteria when conductiviey is 38.0 to 39.9 % IACS. Standard specimens for tensile properties could not be prepared from the absorber housing because of its geometry.

Note: The test specimen was then reground on silicon carbide paper to assure that all effects resulting from cold work (shot peening and/or dents imparted during service or handling) were removed. The results of these "re-tests" and the results of test performed on ground surfaces of the two additional housings described above (See Metallographic Examination), are shown in the following table.

Shock Absorber Housing (Uniball Area)	Hardness* Rockwell "B"	Conductivity (% IACS)
a. Fractured	90	39.8
b. Comparison 1	87	41.1
c. Comparison 2	87.5	40.5

* Average of four determinations.

CORPUS CHRISTI ARMY DEPOT

ANALYTICAL INVESTIGATION BRANCH

LABORATORY REPORT

FOR: AIB Exhibit # USASC 95-309 REQUESTER: J. Benavides

PREPARED AND RELEASED BY: Gary S. Tindall 8/11/95

GARY S. TINDALL, Chief, Analytical Investigation Branch Date

ASSISTED BY: Corando Gallegos

CORANDO GALLEGOS, Physical Science Technician

SUBJECT

Part Shock Absorber, Blade Lag, P/N 114H6800-5, Serial Number VY-1313

Sub-Part Shock Absorber Housing, Part Number (P/N) 114H6802-3 Serial Number Unknown

Aircraft CH-47D, S/N 86-01661

Material Aluminum

Keywords anodize

OBJECTIVES

To characterize the surface coating integrity of the rotary wing head, blade lag shock absorber, uniball housing.

CONCLUSIONS

1. The surface of the blade lag shock absorber was covered with a black colored paint system. Directly beneath the paint was an anodic coating (anodize) such as is required in Military Specification MIL-A-8625, "ANODIC COATINGS, FOR ALUMINUM AND ALUMINUM ALLOYS".

2. The uniball housing surface area was anodized. However there were numerous score lines that had penetrated the anodize coating into its aluminum alloy parent metal.

3. The adhesive tape believed to have evidence of material removed from the uniball housing fracture surface was inspected under magnification and no material associated with the blade lag shock absorber housing was found.

DETAILS OF ANALYSIS

Anodize Coating Test

The blade lag shock absorber (Figure 1) fractured uniball housing bore areas (Figures 2 & 3) were examined for electrical conductivity using a volt/ohm meter. It was obvious that the surface of the bore was not electrically conductive thus anodized. Contained within the bore were score lines that must have penetrated the anodize coating as they were electrically conductive.

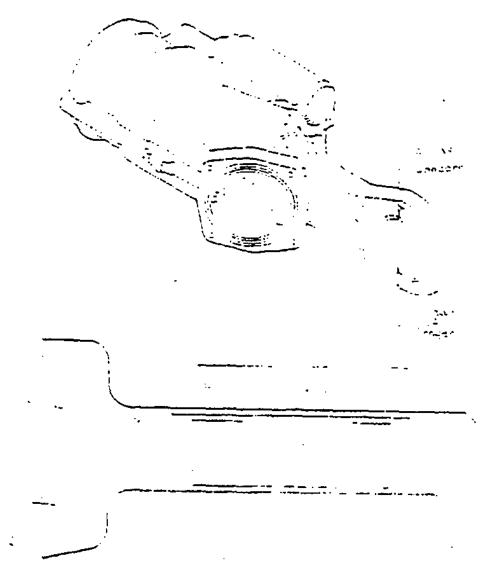


Figure 1. The blade lag shock absorber.

Figure 2. The uniball bore with the indicated score lines. Shown approximately 1.6x.

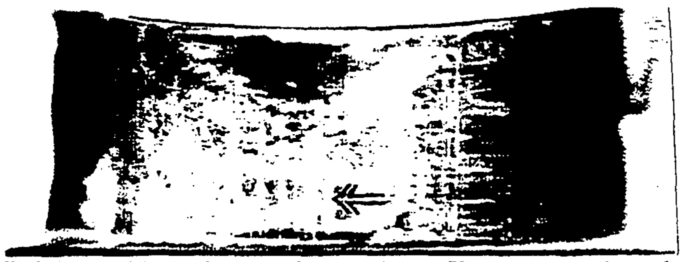


Figure 3. The uniball bore with the indicated score lines. Shown approximately 1.6x.

CORPUS CHRISTI ARMY DEPOT

ANALYTICAL INVESTIGATION BRANCH

LABORATORY REPORT

FOR: AIB Exhibit # USASC 95-309 REQUESTER: J. Benavides

PREPARED AND RELEASED BY: Gary S. Tindall 11 August 1995

GARY S. TINDALL, Chief, Analytical Investigation Branch Date

ASSISTED BY: Corando Gallegos

CORANDO GALLEGOS, Physical Science Technician

SUBJECT

Part Shock Absorber, Blade Lag, Part Number (P/N) 114H6800-5 Serial Number (S/N) VY-1313

Aircraft CH-47D, S/N 86-01661

Material Hydraulic Fluid

Keywords MIL-H-5606

OBJECTIVE

To chemically characterize the fluid submitted for analysis with consideration of in use hydraulic fluids.

CONCLUSIONS

1. The fluid was chemically characteristic of a hydraulic fluid meeting the chemical requirements of MIL-H-5606, "HYDRAULIC FLUID, PETROLEUM BASE; AIRCRAFT, MISSILE, AND ORDNANCE", with the exception of the color of the in use fluid.

2. The color of the submitted hydraulic fluid was a dark amber. The Military Specification requires that new hydraulic fluid be red in color.

3. The spectrographic and ferrographic analytical results indicate that the hydraulic fluid contains wear particles in the heavy use category.

DETAILS OF ANALYSIS

Infrared Analysis of the Hydraulic Fluid

The hydraulic fluid sample was analyzed using a polystyrene film standardized Fourier Transform Infrared (FTIR) spectrometer. The resulting 64 infrared (IR) spectral scan average was compared to the IR spectra of other known fluids and fluid mixtures. The IR spectrum of the hydraulic fluid sample indicated that it was chemically equivalent to a hydraulic fluid meeting the requirements of Military Specification MIL-H-5606.

Gas Chromatograph Analysis

Using a temperature programmed gas chromatograph, equipped with a column designed to separate hydrocarbon compounds, a 0.02 microliter sample of the fluid was injected and analyzed using a flame ionization detector assembly. The resulting chromatogram was compared with the chromatograms of known fluids, and fluid mixtures. The fluid sample chromatogram was essentially that of a petroleum based hydraulic fluid having the chemical characteristics of Military Specification MIL-H-5606. There was no indication of any gasoline or kerosene based fuel contamination.

Emission Spectrograph Fluid Analysis

A sample of the submitted hydraulic fluid was submitted to the routine Army Oil Analysis Program (AOAP) emission spectrograph test. The results indicated that there was an elevated amount of the elements iron, chromium and copper present.

Ferrographic Analysis

A sample of the fluid was analyzed using standard methodology employed for ferrographic analysis. The results indicated that there was a heavy amount of overall wear exhibiting ferrous and non-ferrous debris.

Appendix B:

Boeing Materials Engineering Laboratory Report 95-169, Dated 10 November 1995

The contents of this appendix appear in their original form, without editorial change.

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Distribution List For

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MATERIALS ENGINEERING LABORATORY REPORT

MELR NO.	95-169
DATE	11/10/95

SUBJECT: CH-47D, Blade Lag Shock Absorber, P/N 114H6800-5, S/N VY1313, Metallurgical Evaluation of Through Fractured Inboard Lug

<u>Materials</u>: Housing - 7075-T73 per MIL-A-22771, Anodized Type I, Class I
Bushing - 410 Stainless Steel per QQ-S-763 Cond. T (110-140 ksi)

REFERENCE: A. BH Materials Engineering Memo 8-7525-5-0257, Fractographic Evaluation of Failed Lag Lug conducted at Corpus Christi Army Depot, Texas, dated 8/04/95

ENCLOSURES: I Illustration of the location of the Through Fractures in the Inboard Lug
II-VII Photographic Documentation of the Subject Shock Absorber

I OBJECTIVE

The objectives of this investigation were to: (1) determine the fracture mode, origin location(s) and any factors that may have contributed to the initiation of the through fractured Shock Absorber inboard lug and (2) perform a TEM striation analysis of the Shock Absorber inboard lug fracture surface.

II BACKGROUND

It was reported that on July 26, 1995, a Fort Hood CH-47D Aircraft (86-01661) experienced a "one-per-rev" vibration during forward flight at Fort Irwin, California. The aircraft made a precautionary landing; and during shutdown, the aft red rotor blade struck the fuselage. Subsequent examination of the aft rotor head assembly, S/N A5-421, disclosed a through-fracture in the inboard lug of the red blade lag shock absorber, P/N 114H6800-5, S/N VY1313. The failed shock absorber, along with some of the attaching hardware, were removed from the aircraft and forwarded to the Corpus Christi Army Depot (CCAD), Texas, for metallographic evaluation (Reference A). After evaluation at the CCAD metallurgical laboratory the subject Blade Lag Shock Absorber, P/N 114H6800-5, S/N VY1313, was submitted to the BH Materials Engineering Laboratory for metallurgical evaluation.

III TEST RESULTS

An illustration representing the location of the subject components on the CH47D aircraft is shown in Figure 1. The as-received subject shock absorber is shown in Figure 2.

A. Inboard Lug Fracture Surfaces

Visual Examination and Fluorescent Penetrant Inspection

Visual examination of the as-received portions of the inboard lug disclosed two radial through-fractures at the 1.8750 inch diameter spherical bearing bore (Figure 3). The through fractures, labeled A and B, were located at approximately the 3:00 (lag side) and 8:00 (lead side) positions of the bore, respectively. For reference purposes, the 12:00 position is located at the inboard end of the shock absorber, looking down, (Figure 1). Fluorescent penetrant inspection of the segments of the inboard lug revealed no evidence of additional cracks.

Fractographic Evaluation of Fracture A

Examination of fracture A disclosed a fatigue mode of crack propagation, as characterized by a flat smooth topography with beach marks evident. Fatigue characteristics were observed to extend through approximately 95% of the lug. Due to the extent of fatigue, this fracture was considered to be primary. Visual and scanning electron microscope (SEM) examination revealed the fatigue origin to be located at the top of the 0.040 inch (45°) chamfer between the 1.8750 inch diameter bore and the shouldered bushing spot face of the lug (Figures 5, 6, and 7). No anomalies were observed in the origin area.

> NOTE: Only one half of the primary fatigue origin area was evaluated by BH Materials Engineering. The other half of the origin area was metallographically polished through at CCAD.

Fractographic Evaluation of Fracture B

Examination of through fracture B disclosed a fatigue mode of crack propagation (Figure 3). The extent of fatigue propagation was approximately 20% through the lag portion of the lug. Visual examination revealed the origin to be located on the 1.8750 inch diameter bore approximately 0.05 inch from the spot face of the lug (Figures 4 and 8). SEM examination revealed evidence of an intergranular fracture topography at the origin area. The intergranular area was approximately 0.004 inch deep by 0.0006 inch wide (Figure 9).

Further examination of the secondary fracture revealed additional cracks (area 1 Figure 4) oriented parallel to the secondary fracture. Fractographic evaluation of these cracks, which were approximately 0.01 to 0.02 inch in length, revealed a fatigue topography. These fractures also initiated in intergranular areas along the bore wall (Figures 10 and 11).

B. Metallographic Evaluation of the Inboard Lug

Three transverse metallographic cross-sections were made through the inboard lug. Examination of all three cross-sections revealed evidence of an intergranular corrosive attack extending along the bore wall and spot face of the lug (i.e., the spot face is the surface which mates with the underside of the hat bushing). One of these cross-sections was made approximately 0.01 inch from the primary origin. The maximum depth of intergranular attack observed was approximately 0.010 inch. Figures 12 and 13 show typical areas of intergranular corrosion.

C. Examination of the Inboard Lug Bore Wall and Spot Face

The following additional observations were made during visual examination of the subject Shock Absorber inboard lug:

1. A dark golden color coating was observed on the spot face and chamfer areas. EDXA of the dark golden colored coating on the spot face and the chamfer area revealed a level of chromium that is comparable with a chromic acid anodize. The color of the dark golden coating is typical for a chromic acid anodize as per the Engineering Drawing requirement.

2. A blackish-green coating was observed on the bore surface only. This color coating is not typical for a chromic acid anodize.

3. The inboard lug spot face and chamfer disclosed complete and uniform shot peen coverage, as evidenced by the dimpled topography. X-ray diffraction analysis was performed on the inboard lug spot face and the bore wall revealing residual stress profiles which were typical for a shot peened 7075-T73 aluminum alloy.

4. Numerous axial score marks were observed on the bore surface. Some of these marks were shiny in appearance, while others contained the blackish-green coating. Examination of one of the score marks revealed the direction of scoring to be from the bottom to the top of the bore which is associated with the removal of the bushing.

The above observations, i.e., axial score marks, blackish-green coating, indicates that the bore had been reworked.

D. TEM Striation Analysis on the Primary Fracture Surface

In an effort to determine the fatigue striation spacing, multiple chrome/carbon replicas were obtained from the primary fracture surface. These replicas were subsequently examined using the Transmission Electron Microscope (TEM). Fatigue striation spacings were observed at three (3) locations; labeled #1, #2 and #3 as shown in Figure 14. The fatigue striation data is contained in Table I.

Table I

Location	Distance from Primary Origin (in)	Average Striation Spacing (in/striation)	Figure
#1	0.05	2.2×10^{-6}	
#2	0.16	6.7×10^{-6}	15
#3	0.35	8.3×10^{-6}	

Based on the data in Table I, the estimated number of cycles to propagate the primary fracture from area #1 to the end of the fatigue propagation was approximately 160,000.

This estimate is considered to be a minimum number of cycles since fatigue cracks propagate much slower in the early stages of crack growth and the above estimate does not reflect this behavior. Furthermore, the striation spacing near the origin could not be accurately determined because the size of the spacings, typically in the 10^{-9} inch range, is beyond the resolution of the replication tape and microscope.

E. Inboard Lug Bushing P/N 114H6803-1

Visual and magnetic particle examination of the 410 stainless steel shouldered bushing, P/N 114H6803-1, revealed circumferential cracks on the underside of the hat portion of the bushing in the 0.062 inch wide groove and in the top surface in the areas that were staked. These cracks were not fractographically evaluated. It should also be noted that the outside diameter of the bushing was determined to be 1.8793 inch (see Reference A). The requirement, per the Engineering Drawing, is 1.8770 (+0.0000 / -0.0005)

F. Hardness, Conductivity, Tensile and Chemistry Data

The hardness and conductivity of the subject 7075-T73 aluminum Shock Absorber were found to be 88-89 HRB and 37.5-38.5% IACS, respectively. The hardness and conductivity acceptance limits per revision T of Boeing specification BAC 5946 for 7075-T73 are 79.5-89.0 HRB and 38.0-42.5% IACS respectively.

Tensile specimens, which conformed to the requirements of ASTM E8, were machined from the remaining portions of the lag damper housing. See Table II below for tensile data.

Table II

	Ultimate Tensile Strength	*Yield Strength*	*% Elongation*
T1	76.7 ksi	67.4 ksi	13
T2	78.8 ksi	69.6 ksi*	10
T3	74.9 ksi	63.5 ksi	11
Average	76.8 ksi	66.8 ksi	11
Required	66 ksi min	56 ksi min	7% min

* Per MIL-A-22771, if the conductivity is 38.0-39.9% IACS inclusive the longitudinal yield strength must not exceed 67.9 ksi.

The chemistry of the lag damper conformed to the requirements of 7075-T73 per MIL-A-22771.

IV DISCUSSION

Intergranular corrosion is a selective corrosive attack of the grain boundaries or closely adjacent regions without appreciable attack of the grains themselves. Intergranular corrosion is characterized by a localized attack at the surface (i.e., no significant evidence of pitting) with a widespread intergranular attack below the surface. The depth of this attack is generally about 0.005 to 0.010 inch below the surface. The copper bearing aluminum alloys (i.e., 2014, 7075, etc.) are susceptible to intergranular corrosion. This attack is generally attributed to contaminated processing solutions.

Evidence of intergranular corrosion was observed on the bore and spot face surfaces of the inboard lug. In addition, evidence of an intergranular topography was observed in the origin areas of the secondary fracture as well as the multiple fatigue fractures which initiated along the bore wall. Intergranular corrosion is considered to be a contributing factor to the initiation of the secondary crack and the adjacent multiple fatigue fractures.

Visual examination of the bore revealed evidence, as detailed herein, that it had been reworked. The presence of intergranular corrosion was only observed in the bore and spot face surfaces of the lug. Similar evaluation of other locations away from the bore area revealed no evidence of intergranular corrosion. This would indicate that the intergranular corrosion was most likely associated with the rework accomplished in the bore area.

Although not observed in the primary origin area, intergranular corrosion was observed within 0.1 inch of the primary fatigue initiation site. Rework of the bore may have altered the topographical features of this area, i.e., anomalies associated with the primary origin area may have been removed during rework.

V. CONCLUSIONS

A. The subject Blade Lag Shock Absorber, P/N 114H6800-5, S/N VY1313, contained two (2) fatigue through-fractures in the inboard lug located at approximately the 3:00 (lag side) and 8:00 (lead side) positions of the bore, respectively. The primary fracture initiated in fatigue at the top of the 0.040 inch (45°) chamfer between the 1.8750 inch diameter bore and the shouldered bushing spot face of the lug. The extent of the fatigue propagation was approximately 95% through the lug cross-section. No anomalies were observed in the origin area.

B. The secondary fracture initiated in fatigue on the 1.8750 inch diameter bore, approximately 0.05 inch from the spot face of the lug, at an intergranular area approximately 0.004 inch deep by 0.0006 inch wide. The extent of fatigue propagation was approximately 20% of the lug cross-section. Intergranular corrosion is considered to be a contributing factor towards the initiation of the secondary crack.

C. Multiple fatigue fractures were observed along the bore wall of the secondary fracture surface. These fractures initiated on the bore wall in area that contained an intergranular topography. Intergranular corrosion is considered a contributing factor towards the initiation of these multiple fatigue fractures.

D. Metallographic examination of three cross-sections, through the inboard lug, revealed evidence of an intergranular corrosive attack extending from the bore and spot face that mates with the underside of the hat bushing. Examination of areas outside the reworked bore revealed no evidence of this attack.

E. Examination of the inboard lug revealed that the bore had been reworked as evidenced by the axial score marks and the blackish-green coating.

F. The hardness, conductivity, and chemistry data of the subject 7075-T73 Shock Absorber conformed to the requirements of the Engineering Drawing and related specifications.

VI. RECOMMENDATION

It is recommended that fluids utilized during rework processing of similar lag damper components be free of contamination. Review of the solutions/fluids should be accomplished to assure they are controlled in accordance with applicable specifications.

Prepared by: S. Holder   Reviewed by: R. Walsh   Approved by: R. Cunningham

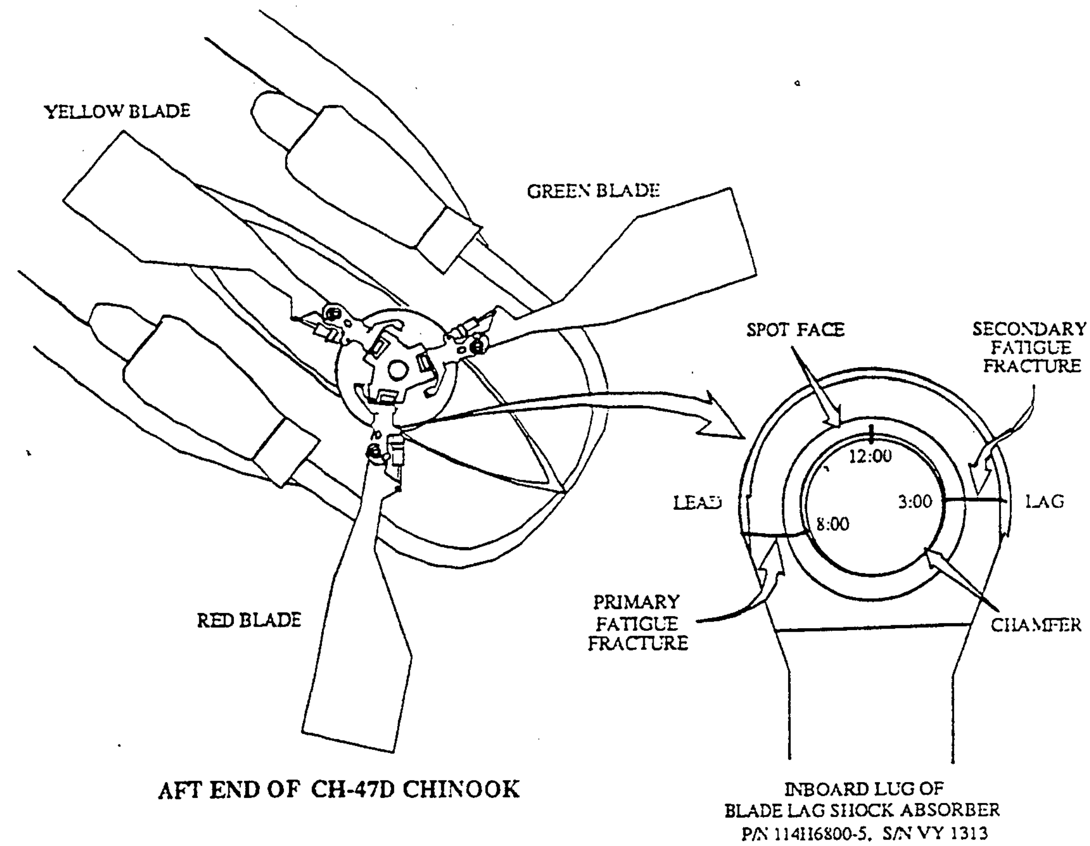

Figure 1 No Scale
Sketch illustrating the location of the through fractures in the inboard lug of the aft red shock absorber

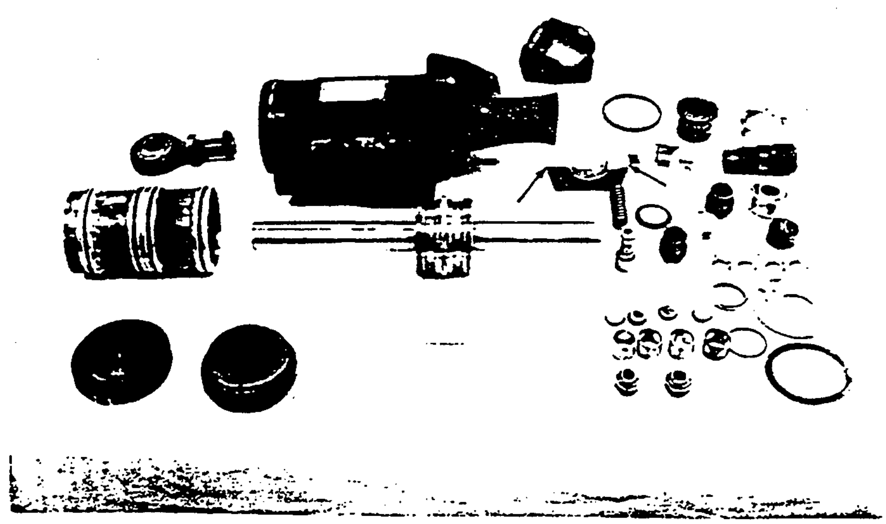

Figure 2 0.1X
As-received photograph of the Blade Lag Shock Absorber, P/N 114H6800-5, S/N VY1313. Arrows denote inboard lug fracture surfaces.

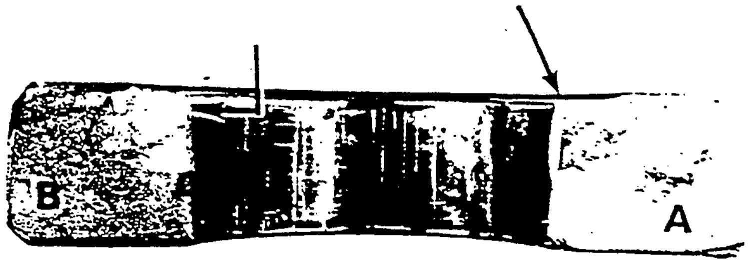

Figure 3 1X
Photograph of the through fractured inboard lug displaying fatigue fracture surfaces A (primary) and B (secondary). Arrows indicate approximate origin locations.

Figure 4 3.5X
Magnified view of fracture B which originated on the 1.8750 inch diameter bore approximately 0.05 inch from the spot face of the lug (arrow). Note: See Enclosure V to view area 1.

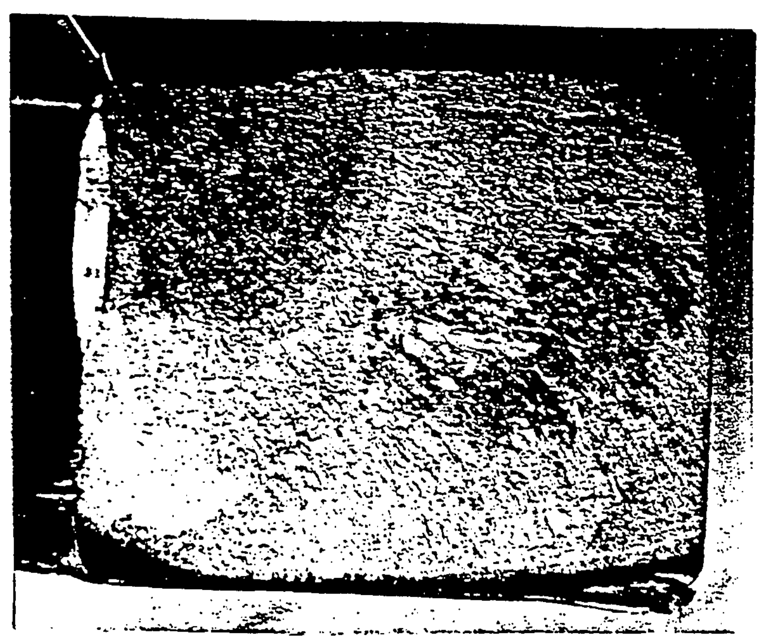

Figure 5 3.5X
Magnified view of fracture A which originated at the top of the 0.040 inch (45°) chamfer between the 1.8750 inch diameter bore and the shouldered bushing spot face (arrow).

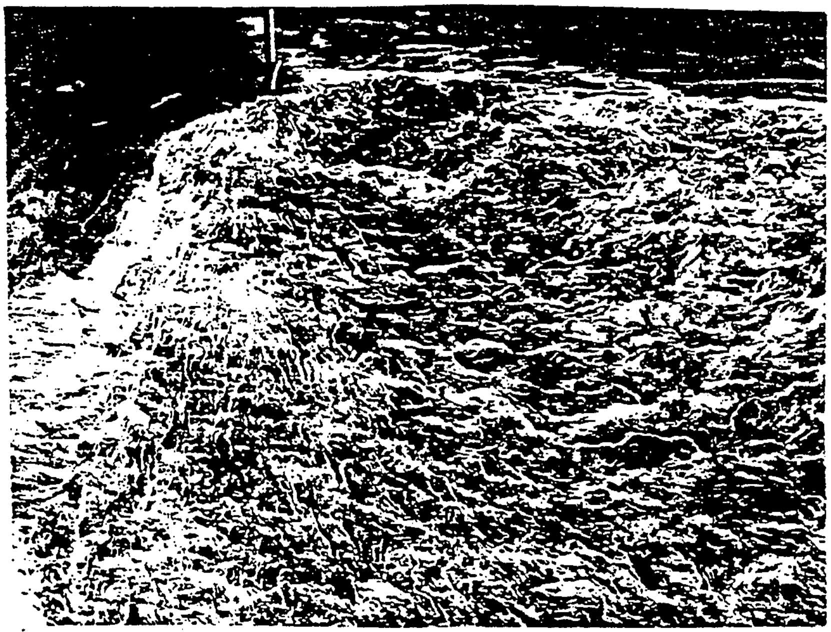

Figure 6 50X
SEM photograph of the primary fracture surface A. Arrow denotes the approximate origin location.

Figure 7 500X
Magnified SEM photograph of the primary fracture origin area (arrow). Note no anamolies are present.

Figure 8 50X
SEM photograph of the secondary fracture surface B . Arrow denotes the origin location.

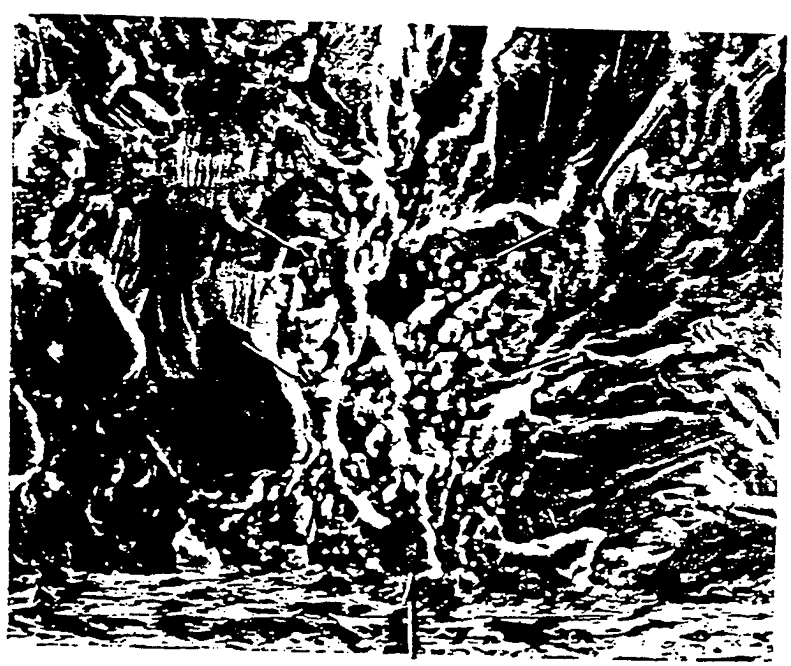

Figure 9 500X
Magnified SEM photograph of the secondary fracture origin area (large arrow) revealing an intergranular topography (small arrows) approximately 0.004 inch deep by 0.0006 inch wide.

Figure 10 50X
SEM photograph of the secondary fracture surface revealing fatigue fractures which initiated in the bore wall.

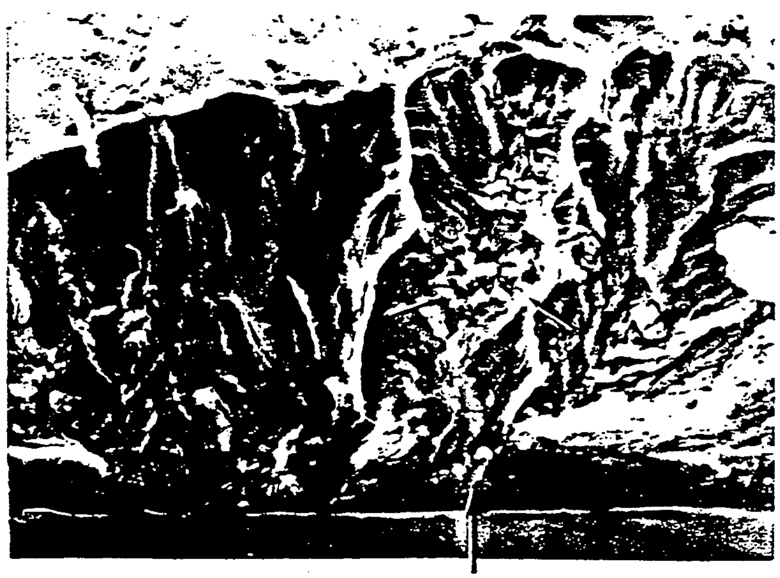

Figure 11 550X
Magnified SEM photograph of one of the fatigue cracks in the bore wall. Evidence of an intergranular topography (small arrows) was observed in the origin area (large arrow).

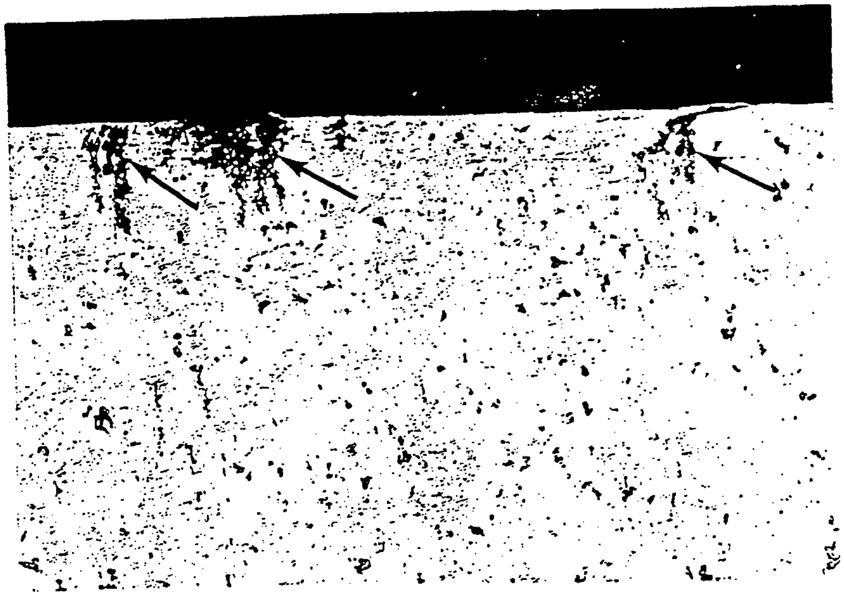

Figure 12 100X
Photomicrograph of a cross-section through the lug approximately 90° from the primary origin, revealing intergranular corrosive attack (arrows) extending from the bore surface.

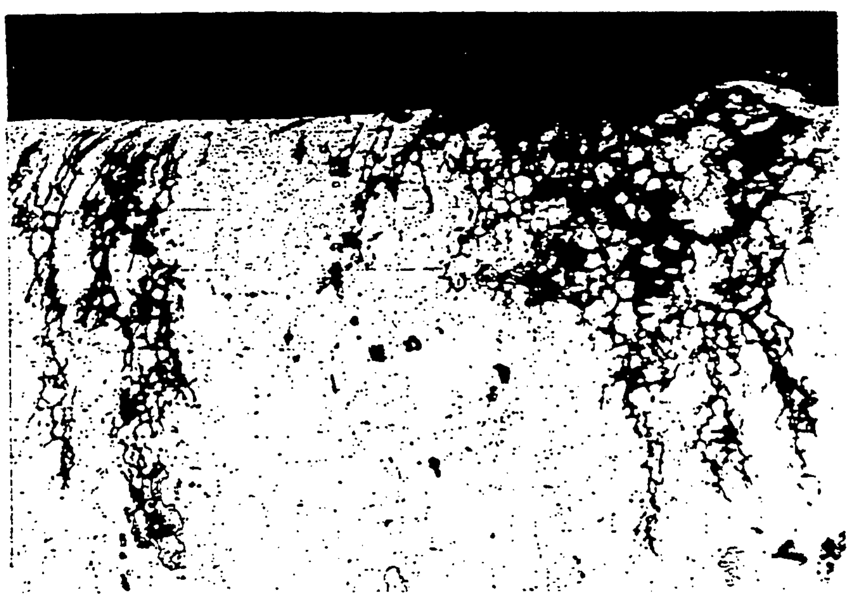

Figure 13 400X
Magnified view of the cross-section through the lug bore revealing intergranular attack extending to an approximate depth of 0.006 inch.

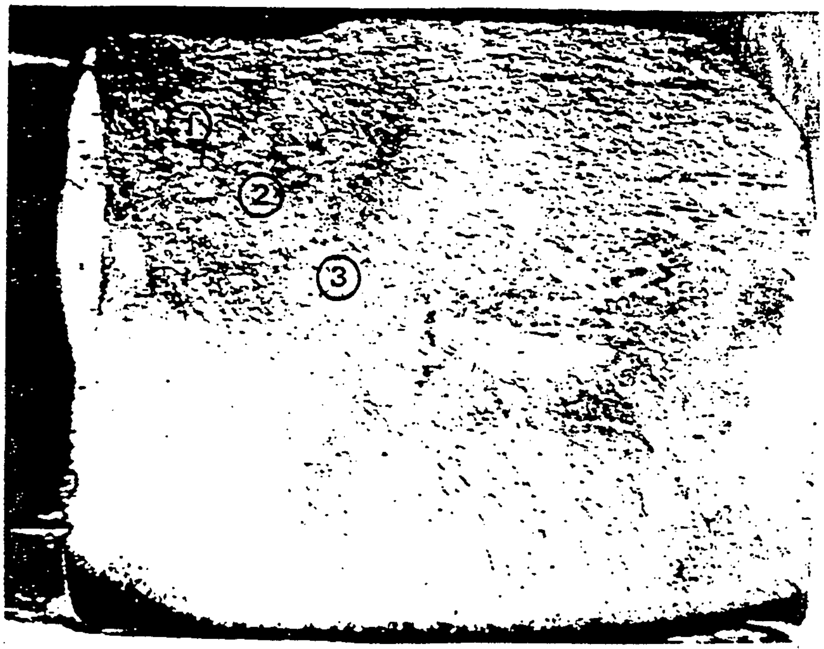

Figure 14 4X
Photograph of the primary fracture surface. Areas 1, 2 and 3 denote locations where measurements were obtained for fatigue striation spacing determination.

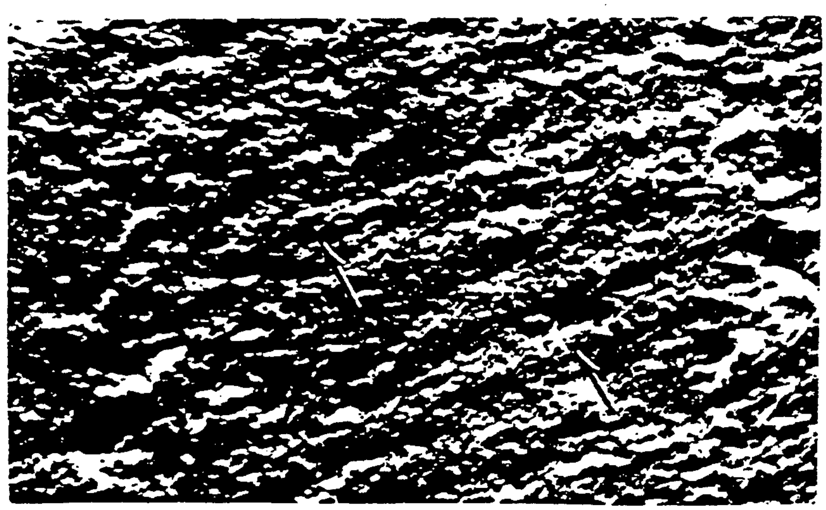

Figure 15 50,000X
TEM photograph of fatigue striations (arrows) taken from area #2 Figure 14, revealing an average propagation rate of 6.7×10^{-6} inch/cycle.

Appendix C:

Daraclean 282 Data Sheets, Grace Metalworking Fluids

The contents of this appendix appear in their original form, without editorial change.

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GRACE Metalworking Fluids

Data Sheet

Daraclean® 282: Aqueous Alkaline Cleaner

Daraclean® 282 is a low-foam alkaline liquid all-purpose cleaner. It is multi-metal safe and effective at temperatures ranging from 80 - 200°F. Daraclean® 282 can be used in soak, agitation, and spray applications. This product is especially formulated to be non-aggressive toward aluminum and zinc alloys. Designed specifically for aerospace and electronics applications, Daraclean® 282 has a unique self-cleaning ability which provides for long fluid life and vastly reduced disposal requirements.

· Aerospace qualified
· Low foam · Multi-metal safe
Biodegradable · Provides in-process rust protection
· Effective at room temperature · Recyclable for long life
· Formulated without chlorine, sulfur, phosphorous, or nitrites
· Excellent hard water tolerance

Applications/Starting Dilutions

Method	Concentration	Temperature	Typical Duration
Agitation	3 - 25%	80 - 180°F	2 - 30 Minutes
Hand Wipe	10 - 100%	Ambient	As Required
Soak	3 - 25%	80 - 180°F	2 - 30 Minutes
Spray	1 - 12% (2-5%*)	130 - 180°F	1/4 - 3 Minutes
Steam	1 - 12%	150 - 200°F	1 - 5 Minutes
Ultrasonic	3 - 25%	80 - 180°F	2 - 30 Minutes

* Preferred range

Concentration, temperature and cleaning time may be adjusted for optimum performance.

Concentration Check Procedure

<u>Test Kit Titration Method</u>

Sample Size:	10 mls	or	10 mls
Titrant:	0.5 N Acid		1.0N Acid
Indicator:	Phenolphthalein		Phenolphthalein
Concentration (%):	Drops titrant x 1.0		Drops titrant x 2.0

<u>Conductivity Method:</u> Conductivity reading (µS) ÷ 155 = % Daraclean 282

DCN282PI 0.050

W.R. Grace & Co.-Conn., Metalworking Fluids Group

5225 Phillip Lee Drive, S.W. Atlanta, Georgia 30336 (404) 691-8646
6000 West 51st Street, Chicago, Illinois 60638 (708) 458-0340
55 Hayden Avenue, Lexington, Massachusetts 02173 (617) 861-6600
2140 Davis Street, San Leandro, California 94577 (510) 568-3427

Typical Physical Properties

Appearance	Clear, yellow liquid
Odor	Citrus
Weight/U.S. Gallon	8.5 lbs
Specific Gravity	1.02
Refractive Index	13.6 - 14.8 (undiluted, Brix scale @ 20°C)
pH	12.4 - 13.0 (undiluted, 25°C)
	11.0 - 12.0 (10% diluted inDI water, 25°C)
Alkaline Equivalent	0.29 - 0.39 (MEQ to pH 8.3)
	0.43 - 0.53 (MEQ to pH 4.0)
Foam	≥10 ml (1% in tap water @ 40 - 60°F)
	≤5 ml (1% in tap water @ 105 - 120°F)
VOC (EPA Method 24)	0.5#/US Gallon (24.0gm/l)
Vapor Pressure	29 mmHG @ 23°C

Performance Properties

Cast Iron Corrosion	
at 2%	Excellent
at 5%	Excellent
Steel Wool Corrosion Test: 2%, 5%, 10%	Excellent
Film Residue	Dry, non-tacky
Foam	
20 ml of 1% at 40-60°F (cold tap water)	5 ml foam
20 ml of 1% at 100-120°F (hot tap)	Flat, no foam

Availability, Storage, And Handling

Daraclean® 282 is available in 5 gallon pails, 55 gallon drums, and tank truck quantities. It is recommended that Daraclean® 282 be stored in well ventilated areas at temperatures between 40°F and 100°F. The recommended shelf life of this product is one year.

National Stock Numbers have been issued for Daraclean® 282

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Metallurgical Examination of a Failed Blade Lag Shock Absorber (Part No. 114H6802) From a CH-47D Chinook Cargo Helicopter

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13. ABSTRACT *(Maximum 200 words)*
A metallurgical examination was performed on a failed blade lag shock absorber from the aft red rotor blade of an Army cargo helicopter. The U.S. Army Research Laboratory (ARL) and the primary contractor (Boeing Helicopters, Philadelphia, PA) performed a visual examination of the failed part, fluorescent penetrant inspection, fractographic evaluation, metallography, hardness testing, conductivity testing, and chemical analysis. It was concluded that the part failed due to fatigue from an area exhibiting intergranular attack. The corrosive attack was most likely caused by the processing fluids used during the rework process. In addition, the parts may not have been properly aged, as evidenced by the higher-than-nominal yield strength values. An improper aging treatment could have facilitated the intergranular attack.

14. SUBJECT TERMS
metallurgical examination, failure analysis, intergranular attack, alloy 7075-T73, aluminum

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