NACA

RESEARCH MEMORANDUM

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S-816 BLADES IN A TURBOJET ENGINE

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SUMMARY

An investigation was conducted to study the effects of several heat treatments on the operating life of turbine blades in a J33-33 turbojet engine operated without an afterburner. As-forged blades, blades solution treated at temperatures high enough to produce germinated grains, blades given a double-aging treatment, and blades overaged by overtemperature heat treatments were evaluated. The engine was operated in a cyclic manner, 15 minutes at rated speed and 5 minutes at idle.

The as-forged group of blades and the group of blades aged without prior solution treatment performed twice as well as a group of blades given the standard AMS heat treatment for S-816 (AMS 5765A) and performed better than all other groups of heat-treated blades. The superior performance of these groups of blades was associated with a high hardness and a dense and uniform precipitation of carbides throughout the microstructures. The forging operations were concluded to be responsible for the superior performance of these groups by strain-hardening the matrix prior to engine operation and by promoting the uniform and dense precipitation of carbides during engine operation.

A double-aging treatment following the standard solution treatment for S-816, which was intended to produce randomly scattered and dense precipitation, failed to improve blade performance relative to the group given the standard heat treatment. The second aging temperature was felt to have caused too great a degree of overaging for S-816. Two groups of blades heat treated at 2300° F to germinate grains had abnormally low failure times. In one of the groups eutectic melting was observed and in the other, thick grain boundary formations. Blades which were given a low-temperature overaging treatment (1550° F for 16 hr) to simulate a low-overtemperature condition possible in engine operation failed at very low times, while blades given a high-temperature overaging treatment (1900° F for 16 hr) were found to perform almost as well as blades given the standard heat treatment. The better performance of blades overaged

at the higher temperatures may have been due to simultaneous solution treatment of some minor phases which permits subsequent strengthening of the alloy by precipitation during engine operation.

INTRODUCTION

Alloy S-816 (AMS 5765A) is a cobalt-chromium-nickel-base alloy designed for service in the temperature range from 1200° to 1500° F where high strength and corrosion resistance are required. S-816 is readily forged to form turbine blades for jet engines and is, at the present time, the most widely used alloy for this application in the United States.

The heat treatment recommended for aircraft forgings in the Aircraft Materials Specifications is a solution treatment at 2150° F for 1 hour, followed by water quenching, and then aging for 16 hours at 1400° F. Higher solution-treating temperatures than 2150° F have been shown by others to yield better high-temperature strengths than are possible from this recommended standard solution treatment. For example, the short-time tensile strength and the stress-rupture life of S-816 at elevated temperatures may be generally increased by raising the temperature of the solution treatment from 2150° to 2300° F (unpublished data). Furthermore, stress-rupture and creep data reported in reference 1 indicate that the optimum strength of S-816 is obtained by solution treating for 1 hour at 2350° F, water quenching, and aging 16 to 24 hours at 1400° or 1500° F. At test temperatures of 1500° F and higher, an aging temperature of 1350° F appeared to yield optimum results.

The increase in strength gained by raising the temperature of the solution treatment is accompanied by the germination of large grains in those areas of the turbine blades which received critical amounts of hot-cold working during the forging process. Germinated grains, particularly when bunched together next to areas of fine grains, have been considered harmful by the manufacturer, and apparently for this reason the solution treatment at 2150° F for 1 hour has been adopted.

A recent study conducted by the NACA (ref. 2) describes results for the performance of J33-9 turbojet-engine blades of wrought S-816 which had been given the standard heat treatment. The operating life of the blades varied from 181 to 539 hours at rated speed, with a 50-percent failure time at 305 hours. The large scatter in blade performance was attributed to variables in the fabrication of the turbine blades. This lot of blades was considerably superior in engine performance to a lot of S-816 blades which had been forged several years before, as discussed in reference 2.

As part of an earlier program at the NACA, tensile specimens were machined from the blade airfoil section of J47 turbine blades of wrought S-816 alloy and were given stress-rupture tests at 1500° F. Results were obtained for specimens which had received the standard heat treatment, specimens which were tested in the as-forged condition, and also specimens which had been aged for 16 hours at 1400° F without prior solution treatment. These results are given in table I and appeared to indicate that the specimens which had received no heat treatment or had been aged only without solution treatment were, at least, as good as those which had been fully heat treated. Similar results have been obtained from stress-rupture tests carried out on tensile specimens machined from one 3/4-inch-diameter rolled bar stock of S-816 (unpublished data). Results for this group of specimens indicated that the stress-rupture condition was fully as good as that for specimens which had received the standard heat treatment. While both groups of results are insufficient to permit definite conclusions to be drawn as to the effect of heat treatment or lack of heat treatment on the life of turbine blades when operated in a jet engine, they do serve to warrant an investigation of such effects.

The effects of heat treatment on the properties of wrought Haynes-Stellite 21, a cobalt-base precipitation-hardening alloy, are reported in reference 3. In contrast to S-816, this alloy showed very poor stress-rupture life at 1500° F in the as-wrought condition. Improvement was obtained when the carbide precipitates in the as-wrought structure were first dissolved into the matrix by solution treatment at 2250° F and then reprecipitated by controlled aging at lower temperatures. The highest stress-rupture lives were associated with a uniform dispersion of fine particles of carbide precipitate throughout the matrix. This microstructure was best obtained by a double-aging treatment in which the Haynes-Stellite 21 alloy was first solution treated for 16 hours at 2250° F, then air cooled, aged for 72 hours at 1200° F, and finally aged for 24 hours at 1500° F. The essential feature of this treatment is that precipitation from the solution-treated structure is nucleated at a relatively low temperature. This treatment produces a uniform scattering of sites for the subsequent growth of precipitate at a second and higher aging temperature.

Recent interest in the occurrence of overtemperature conditions in the jet engine has been concerned with the possible damaging effects on the material properties of the components, particularly the turbine blades. For this reason, two aging treatments at temperatures above those normally used were included in this investigation.

In line with the preceding remarks, the general purpose of the present investigation has been to study the effects of several heat treatments on the operating life of J33 turbine blades of forged S-816 alloy. This investigation includes an evaluation of the operating life of the blades (1) in the as-forged condition, (2) heat treated to produce germinated grains and simultaneously a better degree of solution treatment, (3) solution treated and double aged, and (4), finally, overaged.

This study is a continuation of a program being carried out by the NACA in order to understand the fundamental factors which determine the high-temperature properties of alloys and to extend the operating life of turbine blades.

EXPERIMENTAL PROCEDURE

Turbine blades. - The turbine blades used for the present investigation were J33-33 turbojet-engine blades of wrought S-816 alloy (AMS 5765A) having the following nominal percentage chemical composition:

Co	Ni	Cr	Mo	W	Cb + Ta	Fe	C
40 min.	19-21	19-21	3.5-4.5	3.5-4.5	3.5-4.5	5 max.	0.32-0.42

The blades were received from two sources. The first lot of blades was supplied by the manufacturer in the as-forged condition. All these blades were forged from bar stock from the same heat. These blades were used for heat-treatment groups 1 to 6, as described in table II. The second lot of blades was withdrawn at random from U.S. Air Force stock. These blades had already been given the standard heat treatment by the manufacturer and were used for groups 7 to 9.

The heat treatments described in table II were carried out in an atmosphere of argon gas and were selected on the basis of the preliminary work discussed in the INTRODUCTION. Specific reasons for the selection of each heat treatment are shown in the table. All the groups of blades consisted of six blades except group 7, which contained four blades.

Stress and temperature distribution in turbine blades during engine operation. - The cross-sectional areas of J33-33 turbojet-engine blades were obtained from blueprints of the blade, and the distribution of centrifugal stress along the length of the blades was calculated by the method described in reference 4. Six thermocouples were welded to the midchord of the blade airfoils at different distances from the blade base, and a trial run in a J33-33 engine was made to determine the temperature distribution in this type of blade. Previously, J33-9 engine blades had been investigated for temperature distributions, and the method used for the J33-33 engine blades is essentially the same (ref. 5). Figure 1 shows the measured temperature and the calculated centrifugal-stress distributions for these blades during engine operation at full power as well as the stress-rupture life of S-816 bar stock at the different combinations of stress and temperature in the blades. The latter curve shows that the most severe combination of constant stress and temperature (21,400 psi and 1450° F, respectively) occurs at a distance

2.4 inches from the base platform of the blade. The curve may be used to locate the zone of possible stress-rupture failure. However, as will be discussed in the section Blade failure mechanisms, there is no reason to expect that the S-816 blades will run the 860 hours corresponding to the minimum in this curve.

Engine operation. - The nine groups of blades were installed in a single turbine rotor of a J33-33 turbojet engine, and the engine was operated until all but four of the blades had failed. The test consisted of repeated 20-minute cycles of 5 minutes of idle followed by 15 minutes at the rated speed of the engine (11,750 rpm). Only the time at rated speed is discussed throughout the report, since the stresses and temperature at idle are too low to be significant. Blade stress and temperature at rated speed were held constant by controlling the engine speed and the exhaust-nozzle opening. Blade temperature was measured by thermocouples inserted in two special blades in the disk. A slip-ring system was used to connect the thermocouples to recording instruments. Further details of engine instrumentation and temperature controls are contained in references 5 and 6.

Blade elongation measurements. - One blade from each of the nine groups was scribed near the trailing edge at 1/2-inch intervals as described in reference 6. However, measurements were made over 1-inch segments. One segment was designated as zone A and extended from $1\frac{7}{8}$ to $2\frac{7}{8}$ inches above the base. The other segment, zone B, extended from 7/8 to $1\frac{7}{8}$ inches above the base. After blade failures or necessary engine shutdowns, the elongation of each scribed segment of intact blades was measured with an optical extensometer.

Macroexaminations of failed blades. - A blade was said to have failed and was removed from the engine either when actual fracture occurred or when numerous cracks in the airfoil or severe necking made it apparent that failure was imminent. Blades which failed were examined at low magnifications to determine as nearly as possible the manner by which the failures originated. In addition, blades from each group were macroetched to show grain growth and to differentiate between intercrystalline and transcrystalline cracking. The failures may be classified into the following catagories as a matter of convenience:

(1) Stress-rupture: Blade failures which occurred by cracking within the airfoil, by necking of the airfoil, or by fracturing in an irregular and jagged intercrystalline path. In addition to the main fracture, other similarly formed cracks frequently occurred near the origin of the main fracture or crack.

(2) Fatigue: Cracks which progressed from nucleation points, usually at or near either the leading or trailing edges, in straight paths. The cracks frequently were smooth, often showed progression lines or concentric rings, and appeared to be transcrystalline.

(3) Stress-rupture followed by fatigue: Blade failures which appeared to be caused by a combination of the preceding mechanisms. The fractured surfaces of blades in this group consisted of a small area which had the characteristics described for the stress-rupture catagory and a larger area with the fatigue characteristics already described. A further criterion was that other cracks also appeared immediately near the nucleation area of the main crack or fracture edge and appeared to be stress-rupture cracks.

In all cases, the blades failed finally in tension because of the progressive reduction in the load-carrying area, so that all blade failures showed a large area of rough fracture surface.

(4) Damage: Blades showing nicks or dents in the airfoil which could initiate fracture. These blades were considered apart from the preceding three catagories of blade failures since they did not give a true indication of material properties.

Metallographic examination of as-heat-treated specimens and of failed blades. - Metallographic examinations were made on sections of blade airfoils which had been heat treated along with the blades in order to show microstructural changes and grain growth during heat treatment. Similar examinations were made on specimens cut from the airfoil of the first and last blade failures in each of the heat-treatment groups so that changes in microstructure during engine operation could be observed. Care was taken that the specimens of blade failures were cut from the section of airfoil about the origin of the failure so that paths of fracture and fracture mechanisms could also be studied.

Grain-size measurements. - Grain sizes of the as-heat-treated specimens and of specimens cut from the first and last blade failures from each group of blades were measured using a metallograph and an ASTM grain-size measuring eyepiece. In addition, blades from each heat-treatment group were macroetched after the blades had been run to failure using a solution of 150 milliliters of HCl, 150 milliters of H_2O, 9 milliliters of HNO_3, and 40 grams of $FeCl_3$. Specimens were vapor blasted and then boiled in this solution for 5 minutes.

Hardness tests. - Rockwell A hardness measurements were made of all the as-heat-treated specimens used for the metallographic examinations.

In the case of failed blades, segments were cut from the first and last blade failures of each group of blades in zones as close to the fracture areas as possible. Hardness measurements were made over the entire cross section of the airfoils. Rockwell superficial readings (R-15-N) were used to obtain accurate results in the narrow portions of the airfoils. Both the Rockwell A and the R-15-N readings were converted to Rockwell C values. It should be cautioned here that hardness values reported after engine operation were obtained 1/8 inch from the fracture edges, and the values might be influenced by the elongation of the blades in these areas.

RESULTS

Engine Operating Results and Blade Performance

The engine operating results for the different groups of blades are presented in table III along with a record of the location of the blade failures and the mechanism of the failures for each blade. Engine runs were discontinued after 430 hours when the relative performance of the different blade groups was clear.

In order to facilitate the presentation of results as well as the subsequent discussion, the blade data are plotted in figure 2. The as-forged group of blades (group 1) performed at least twice as well as the group given the standard heat treatment (group 3). The blades of group 2, which were aged at 1400° F only, behaved essentially the same as the blades of group 1. The failure mechanisms of all blades in groups 1 and 2 were either stress-rupture or stress-rupture followed by fatigue, whereas the blades of group 3 failed chiefly from fatigue.

Some of the blades given the double-aging treatment (group 4) failed at lower times than those given the standard heat treatment (group 3). Also, some of the blades solution treated at 2300° F followed by aging at 1600° F for 16 hours, or by aging at 1200° F for 24 hours and 1500° F for 48 hours (groups 5 and 6), failed in unusually short times.

Air Force stock specimens given the standard heat treatment (group 7) failed at lower times than the blades forged from the selected bar stock and given the standard treatment (group 3). Two of the four blades in the former group failed by damage, and the results should be considered in this light. All the Air Force stock blades of group 8 which had been overaged (overheated) at 1550° F for 16 hours failed by fatigue at low times, while some of the blades which had been overaged at 1900° F for 16 hours (group 9) failed by stress-rupture followed by fatigue. The behavior of the latter group as a whole is essentially the same as that of the group given the standard heat treatment.

Elongation of blades during engine operation. - Elongation results are given in figure 3 for zone A. Elongation curves for zone B are not shown, but the maximum elongations measured in both zones are compared in figure 4.

Macroexaminations of failed blades. - Most of the blade failures (71 percent) occurred between $2\frac{5}{8}$ and $3\frac{1}{4}$ inches from the blade root platform as shown in table III and figure 5. This zone is farther from the base of the blade than the critical zone (defined by stress-rupture considerations), which was found to be 2.4 inches from the base. Of the total number of blade failures, 39 percent occurred in the segment for which elongation measurements were made and plotted (zone A), whereas 56 percent of the failures occurred above zone A.

Photographs of typical "failed" and "unfailed" blades are shown in figure 6. The cracked blade contains typical stress-rupture cracks in the center of the blade; the blade which failed by fatigue contains the progression lines typical of a fatigue zone; and the stress-rupture followed by fatigue failure has some cracks below the fatigue zone, the criterion for this type of failure. Of the total number of failures other than damage failures, approximately 35 percent were classified as stress-rupture, 20 percent as stress-rupture followed by fatigue, and 45 percent as fatigue failures.

Metallurgical Studies of As-Heat-Treated Specimens

Microstructure. - The microstructures of test pieces which were cut from representative blade segments and given the same heat treatment as the blade groups run in the engine are shown in figure 7. Metallographic specimens cut from a single as-forged blade were found to contain areas with elongated or severely distorted grains (fig. 7(a), left) as well as areas of equiaxed grains (fig. 7(a), right). The grain sizes observed were very small in the elongated areas (A.S.T.M. 8) and slightly larger in the equiaxed areas (A.S.T.M. 5 to 8). Precipitation in slip lines may be observed in the photomicrographs shown, but other areas appeared to have been partially solution treated. Elongated grains were most prominent near leading and trailing edges, while equiaxed grains were predominant at the center of the blade airfoil. The structure of the specimen aged at 1400° F for 16 hours (fig. 7(b)) shows an area with equiaxed grains similar to the area of equiaxed grains shown in figure 7(a), with precipitation in slip lines and twin and grain boundaries. The specimen given the standard heat treatement for S-816 contains the usual massive residual columbium carbides (probably (Cb,Ta)C) and fine precipitation in the grain boundaries. General precipitation in the matrix is not noticeable. The specimen given the standard solution treatment at 2150° F for 1 hour followed by double aging (fig. 7(d)) shows increased matrix precipitation and, of course, the residual carbides. The specimen given a solution treatment for 4 hours at 2300° F followed by the standard aging treatment (fig. 7(e)) has a structure essentially the same as that of the specimen given the standard heat treatment except that the average grain size is greater. Some of the larger carbides appear to be idiomorphic, indicating that at higher temperatures the carbide forming element (Cb or Ta) diffuses to the residual carbide sites and precipitates as a carbide upon the orginial carbide. In figure 7(f) the specimen solution treated at 2300° F followed by double aging shows very large grains with an unusually large quantity of matrix precipitation as well as a small quantity of Widmanstätten structure. Larger carbides also appear to be idiomorphic, as was the case in the preceding specimens.

The structure of a typical Air Force stock blade is shown in figure 7(g) and is almost identical to that of figure 7(c), which represents the same condition of heat treatment. Upon overaging or overheating the Air Force stock blades at 1550° F for 16 hours, the precipitation in the grain boundaries appears to increase (fig. 7(h)) and a little matrix precipitation also has occurred. The specimen overheated at 1900° F for 16 hours (fig. 7(i)) shows that some of the carbides, particularly those in the grain boundaries, have spheroidized and that the matrix is relatively clean.

Grain-size measurements of as-heat-treated microspecimens. - The grain sizes of the preceding microspecimens are listed in table IV. All the specimens are of a fine, or at least microscopic, grain size except specimens of groups 5 and 6, which were heat treated at 2300° F to deliberately coarsen the grains.

Hardness measurements of as-heat-treated specimens. - The hardness of the as-heat-treated specimens is shown in table V. The average hardness of the as-forged and aged groups (groups 1 and 2, which performed the best) was greater than the average hardness of any other group. Blades of groups 1 and 2 had an average hardness of Rockwell C-35, whereas blades given the standard heat treatments (groups 3 and 7) had an average hardness of Rockwell C-25 to C-26.

Metallurgical Studies of Failed Blades

Microstructures. - Photomicrographs cut from the first and last blade failures for each group are shown in figure 8. Typical stress-rupture and fatigue portions of fractures are shown. The stress-rupture portions of cracks are intergranular, as shown in figure 9(a), and transcrystalline in the case of the fatigue crack, figure 9(b). The microstructures of groups 1 and 2 (the blade groups which performed best) were similar prior to operation in certain locations, as noted previously, and are essentially similar after operation. The precipitation which occurred in grain boundaries, slip lines, and twins during the forging operation had been spheroidized and it still outlines some of these areas. However, it is more difficult to define grain boundaries, and therefore fracture paths, in specimens cut from blades because the spheroidized carbides obscure the fracture paths. Intergranular tears and some evidence of jaggedness at fracture edges confirm the macroexaminations made of the fracture edges, table III, and show that the specimens in these groups failed predominantly by stress-rupture. The last failure of group 2 (fig. 8(b)) is believed to have occurred by fatigue initiated near the trailing edge by intergranular penetrations of oxides. The degree of spheroidization made it difficult to note any differences between the structures of the first and last failures.

In figure 8(c), the specimens given the standard heat treatment (group 3), the first failure is shown to have a smooth transcrystalline fracture edge, indicative of fatigue; and the last failure is shown to have stress-rupture characteristics, as shown by the jagged fracture edge and the stress-rupture cracks below the fracture edge. This confirms the classification of these specimens made in table III. The precipitation in the first failure is well developed, but since it has run only 167 hours, spheroidization has not taken place to a great degree. The microstructure of the last specimen to fail (failure time, 310 hr) shows, at high magnification, that the precipitation in the immediate vicinity of the fracture origin is to a large degree in a Widmanstätten form, that the precipitation in the grain boundaries is much more nearly continuous than those of the specimens of groups 1 and 2, and that the grain size near the fracture origin is greater than in groups 1 and 2. Any of these conditions could account for the poorer performance of this group of blades relative to groups 1 and 2.

In figure 8(d), the photomicrographs represent the group of specimens given the double age (group 4). Since the specimens treated in the same manner (fig. 7(d)) contained a considerable quantity of precipitation, it is surprising to note so little matrix precipitation in the first blade to fail. The matrix precipitation is beginning to spheroidize, and the grain boundaries contain enough carbides to be almost continuous. The structure of the last blade to fail corresponds more closely with the structure shown in figure 7(d). In figure 8(e) (group 5), the first failure, which occurred in an unusually short time (36 hr), contained a considerable quantity of eutectic melting and the stress-rupture cracks occurred in areas with considerable evidence of grain boundaries melting. This group of blades was solution treated at 2300° F. In the blade which failed last the grain boundaries were very thick (although this cannot be observed in the photomicrographs). This failure contained a formation similar to "Chinese script," which may be evidence of melting. In both the first and last failures, large quantities of precipitation have occurred in slip lines and in Widmanstätten patterns and are very extensive.

Additional metallographic examinations revealed three of the remaining four specimens contained eutectic melting. In view of this melting, the results for group 5 cannot be considered as representative of the heat treatment employed.

Blades of group 6 were also solution treated at 2300° F but did not show new evidence of eutectic melting (fig. 8(f)). The photomicrograph of the first specimen to fail shows the area at or near the origin of the crack. The fracture edge is transcrystalline and smooth, indicative of fatigue. Again all the blades in this group were sectioned to determine whether any eutectic melting occurred, and nothing was found comparable with the evidence found in group 5. In one of the blades some possible Chinese script was found, but it was not extensive. In another blade, the fatigue crack was propagated across very fine grains and adjacent germinated grains without deviating from a straight path. The most deleterious formation observed in this group of blades consisted of thick carbide formations in grain boundaries.

The structures of Air Force stock blades given the standard heat treatment are shown in figure 8(g). There is a noticeable difference between the structures of the blade failures of this group and those of the blade failures of group 3, which also received the standard heat treatment. The Air Force stock blade which failed first has much less general precipitation than the specimens of group 3, and the first failure time is less. Grain boundary precipitates and residual carbides appear to be the same in both groups, but the grain size of the specimen of group 7 is finer than that of group 3. This could be the result of forging variables. An interesting but unexplainable happenstance is that all failures of group 3 originated at the trailing edges, whereas all failures of group 7 originated at the leading edges. Stress-rupture and fatigue characteristics are present in the photographs of the first and last failures, respectively.

The structures of failed blades which had been overaged (overheated) at 1550° F are shown in figure 8(h) (group 8). Carbides precipitated in the matrix, and slip lines and twin boundaries have been partially spheroidized. Matrix precipitation has not occurred to any great extent. The fracture edges of both specimens are predominantly transcrystalline, indicative of fatigue failures, and again this correlates well with the results of macroscopic examinations shown in table III.

Stress-rupture characteristics are evident in the photomicrographs of figure 8(i), which represent the blades overaged at 1900° F. The grain boundaries contain a large amount of spheroidized carbides, and in general the spheroids in these specimens are the largest of any of the heat-treated groups. Matrix precipitation (of the salt and pepper type) is not evident in either of the blade structures, but this type of precipitation is frequently difficult to see when large quantities of spheroidized particles are present.

Hardness. - The results of the hardness tests of the first and last blade failures are shown in table VI. Blades from groups 1 and 2, the best performing groups, which had a hardness of Rockwell C-35 before testing, maintained the highest hardness values (Rockwell C-32 to C-34) of any of the groups. Individual readings made for a given blade specimen varied considerably in some cases. For example, in group 6 readings ranged from Rockwell C-24 to C-30.

Grain size. - Macroetching of specimens from different groups revealed no grain germination in any blades except those of groups 5 and 6. Photographs of macroetched specimens from these groups are shown in figures 10 and 11. According to the usual concept of grain growth, the areas in which germinated growth has occurred are areas in which critical stresses were produced by forging. The photographs show that most of the blades had received critical amounts of deformation in the leading and trailing edges and in the upper-third portion of the blade airfoils. Stress-rupture cracks are shown in figure 10. Blade 3 of figure 11 has uniform grains throughout except at the extreme leading and trailing edges at the tip. This blade, which was a first failure, failed by fatigue in an area where germinated grains were adjacent to microscopic grains.

DISCUSSION

Operating life of different blade groups. - The excellent behavior of the as-forged blades (range of failure times: group 1, 357 to 430 hr; group 2, 317 to 430 hr) relative to the blades given the standard heat treatment (range of failure times: group 3, 148 to 310 hr) was most interesting. This behavior is in contrast to the stress-rupture data shown in table I, which indicates that the as-forged specimens should behave about as well as the blades given the standard heat treatment.

The superior performance of the as-forged blades and the as-forged and aged blades might be explained by two factors: (1) cold working during forging, which would be expected to strengthen the matrix; and (2) the formation of nucleation sites during forging, which could result in a desirable form and distribution of precipitate during engine operation. The occurence of cold working during forging is supported by the observation that the hardness of groups 1 and 2 was significantly higher prior to and after engine operation than the hardness of the other groups. The occurence of beneficial precipitation during engine operation is supported by metallographic examinations which showed that the microstructures of groups 1 and 2 contained larger quantities of precipitates and residual carbides after engine operation than did the other groups.

The range of failure times obtained for the blades of group 3 (167 to 310 hr) shows that this lot of blades is not significantly better than the blades selected from Air Force stock (group 7). It is recognized that the results for group 7 are not conclusive in themselves because there are only two true failures. However, it is shown in reference 2 that S-816 Air Force stock blades for the J33-9 turbojet engine (with the same airfoils as the blades used in this investigation but with different roots) failed in cyclic tests of the same type used in this investigation at times ranging from 163 hours to 394 hours.

The performance of the double-aged blades of group 4 (range of failure times: 102 to 310 hr) was not significantly better than those of group 3, which were given the standard heat treatment. As was previously mentioned, past experience with wrought Haynes Stellite 21 (ref. 3) indicated that double aging might improve the performance of S-816 blades. However, in the case of S-816, the alloy was apparently overaged in the double-aging treatment, and shorter times or lower temperatures should have been selected. Overaging was shown by the large quantity of visible precipitation and by the fact that the hardness of the as-heat-treated specimens given the double aging appeared slightly less than the hardness of the specimens given the standard heat treatment (Rockwell C-25 and C-26, respectively). Although the double-aging treatment used did not improve the properties of these blades, predistribution of numerous, well scattered nucleation sites by low-temperature aging prior to high-temperature aging or operation is believed to be a sound general principle. Such distribution of precipitates, perhaps, may also be obtained by water quenching from solution-treating temperatures or, as previously mentioned, by hot-cold working prior to testing or high-temperature operation.

The results of the heat treatment to produce germinated grains (solution heat treated at 2300° F for 4 hr, water quenched, aged 16 hr at 1400° F) and simultaneously a greater degree of solution treatment (range of failure times: group 5, 36 to 285 hr) were obscured because eutectic melting was found in this group of specimens. The poor performance of the blades of group 5 may be largely attributed to the eutectic melting observed. The previous data from unpublished information and reference 1 indicated that solution treatment at 2300° F and 2350° F yielded better high-temperature properties than were obtained by solution treating at lower temperatures. No eutectic melting was observed in a preliminary test specimen heat treated at 2300° F for 4 hours (fig. 12) nor was any found in the blade specimen examined after heat treatment (fig. 7(d)). No melting was found in the blades of group 6 (range of failure times: 59 to 401 hr), which were also solution treated for 4 hours at 2300° F. The presence of eutectic melting in the blades of group 5 after engine operation cannot be explained.

Although no certain evidence of eutectic melting was observed in any of the blades of group 6, the heat treatment was harmful as shown by the fact that two blade failures occurred by fatigue at very low times of 59 and 99 hours. Heterogeneous grain sizes were observed in blades of this group with large areas of fine grains adjacent to large areas of coarse grains (fig. 11). However, the possible harmful effects of the germinated grains are obscured by thick carbide formations in the grain boundaries, which could alone account for the poor performance of this group of blades. It is interesting to note that the path of a fatigue crack in the first blade of this group (fig. 8(f)) did not deviate in going from the fine- to the coarse-grained area. This may indicate that the heterogeneity of the grain size (per se) is not important.

The operating lives of the blades of group 8, which were standard Air Force stock blades given the low-temperature, overaging treatment at 1550° F for 16 hours, were very low (range of failure times: 55 to 131 hr). The 1550° F treatment per se is apparently not as important as the microstructure produced. A similar deleterious microstructure might be produced by overtemperature engine operation at shorter times but higher temperatures. The visible precipitation produced by the 1550° F treatment occurred mainly in grain and twin boundaries (fig. 7(h)). During engine operation further precipitation and agglomeration occurred in these sites. This type of precipitation apparently does not improve the performance but does considerably increase the hardness compared with the standard heat treatment (see table V).

The blades of group 9 (range of failure times: 86 to 213 hr), which were overheated at 1900° F for 16 hours, were overaged as shown by the hardness (Rockwell C-27) and by the spheroidization of carbides in the microstructures (table V and fig. 7(i)). The heat treatment may also have partially solution treated the alloy, dissolving smaller microconstituents into the matrix. This would permit strengthening of the blades by precipitation during engine operation, which would explain the comparable performance of this group of blades with that of the standard Air Force stock blades (range of failure times: group 7, 94 to 202 hr).

Elongation of different blade groups. - The elongation curves show that the blades which performed the best (groups 1 and 2) had the lowest creep rates. Blades of group 5 also had a low creep rate but in this case, poor blade life. This combination of low creep rate and poor blade life for group 5 may have resulted from the eutectic melting found in this group. The blades of group 9, which were given the high-temperature overaging treatment (1900° F), exhibited the greatest total elongation and creep rate, which may be associated with the spheroidization noted in the microstructure prior to testing.

Blade failure mechanisms. - Centrifugal stress and temperature are not the only service conditions that are known to limit the performance of turbine blades. Engine vibration and blade flutter may contribute to early failures by fatigue, rapid heating or cooling may set up thermal stresses within the blades, and the action of the hot combustion gases may cause surface and intergranular corrosion. In many cases, the classifications made by visual examination were supported by later microscopic examination of the microstructure. A basic difficulty in defining the failure mechanism from the appearance of the fracture surface alone is that the effect of superimposed fatigue damage is not always evident. Ferguson (ref. 7), for example, measured the high-temperature life of specimens subjected to vibratory loads superimposed upon constant loads and found that, in many cases, the fracture surface of the specimens showed no evidence of fatigue damage, although the reduction in life caused by the vibratory loads made it clear that fatigue must have been an important factor in causing early failures.

From the known distributions of temperature and centrifugal stress in the blade, and from the known material properties, the stress-rupture life of material under the same combinations of stress and temperature which exist along the length of the blade airfoil can be calculated. The results of such a calculation are shown in figure 1, where the stress-rupture life of S-816 bar stock given the standard heat treatment is shown plotted against the distance above the base for the corresponding combinations of stress and temperature. The minimum in this curve shows that the most severe conditions for the lowest stress-rupture life are located at a distance of 2.4 inches above the platform of the blades, where the stress is 21,400 psi and the temperature is 1450° F. Because of the material differences which exist between bar stock and forged blades, the stress-rupture life of the forged blades under operating conditions should not be expected to be equal to the minimum value given by this figure. Only the location of the critical zone as defined by the stress-rupture strength is considered in the following discussion.

In all but two instances (exclusive of damage failures) the turbine blades failed in operation in a zone above that of minimum stress-rupture life (fig. 5). Creep measurements (figs. 3 and 4) show that the rate of creep and the amount of creep deformation just prior to fracture were greater in zone A than in the zone closer to the base (zone B). Exclusive of damage failures and group-5 blades (which exhibited eutectic melting), 58 percent of the blades failed at or above a point $2\frac{7}{8}$ inches from the base platform (the top portion of zone A). While no elongation measurements were made above zone A, past experience (ref. 6) indicates that maximum creep would be centered about the position of minimum stress-rupture life (in this case, 2.4 in. above the platform). It therefore would appear that the majority of blades failed in a region having less than maximum creep. From these observations, the conclusion can be drawn that other factors than stress-rupture must contribute to the failure mechanism of turbine blades during operation.

The preceding conclusion is supported by the examination of the blade fractures. Exclusive of those blades which failed by damage, 65 percent of the blade failures could be classified as fatigue or a combination of fatigue and stress-rupture. This classification was based upon the appearance of fatigue progression rings on the fracture surface, the transgranular character of the failure, or the single-line path of fracture. The remaining 35 percent of the failures, which were classified as stress-rupture failures, may actually have undergone considerable damage by fatigue which could not be detected from appearance alone, as already noted. The early blade failures were more often fatigue or fatigue plus stress-rupture, while the later failures tended to have predominantly stress-rupture characteristics (fig. 5).

Implications of hardness measurements. - Previous work with S-816 (ref. 2) showed a wide range of scatter of hardness values. Since the scatter of hardness shown in tables V and VI could have been due to macrodifferences within the specimen or segregations of microconstituents, the average readings are used to represent the hardness of the alloy.

The high hardness of the as-forged blades and the as-forged and aged blades (groups 1 and 2) may be largely attributed to residual stresses introduced into the blade airfoils during forging. Precipitation alone at 1550° F for 16 hours (group 8) increased the hardness over that of the fully heat-treated condition (group 7) by as much as 4 Rockwell C units. Other precipitation treatments might increase the hardness further, but it seems unlikely that the hardness of Rockwell C-35 (groups 1 and 2) could be attained by heat treatment alone. The solution treatment at 2300° F for 4 hours with single and double aging (groups 5 and 6, respectively) permitted greater hardening than the solution treatments at 2150° F followed by single and double aging (groups 3 and 4). The lower hardness of as-heat-treated group 9 as compared with group 8 may be explained by the greater degree of overaging that occurs at a temperature of 1900° F. Some solution of minor phases may have occurred during heat treatment at 1900° F, and this would have been followed by precipitation during engine testing. This would explain the higher hardness of group 9 after engine operation as compared with the as-heat-treated hardness of this group.

Further comments regarding the hardness of S-816 can be made at this point. S-816 has been regarded by others to be a "solution-strengthened" alloy. This implies that upon increasing the thoroughness of the solution treatment the strength (or hardness) of the alloy should increase as more alloying elements enter solid solution. Observations at the NACA have shown that solution treatment of wrought S-816 bar stock produced the following hardness changes:

Solution temperature, °F	Solution time, hr	Quench	Rockwell C hardness
2150	1	Water	26
2250	1	Water	24
2250	16	Water	23
2300	4	Water	19

From these results, solution hardening does not appear to occur. Metallographic results, to be discussed subsequently, indicate that columbium precipitates even at 2300° F, increasing the size of the massive carbides in the alloy. Since Cb has a very large atomic diameter, it should contribute to hardening when in solution. Its precipitation on the massive carbides could more than offset the hardening effect of a solution of smaller carbide forming elements.

The matrix hardness (and strength), therefore, appears to be increased by hot-cold working or precipitation but not by solution treatment alone.

Analysis of metallographic results. - The massive carbides in the microstructure of S-816 have been shown to be CbC in references 8 to 11. Since some of the later S-816, including that of the present investigation, contains tantalum along with columbium, the massive carbides in the structures shown in the photomicrographs could be either CbC or TaC or solid solutions of the two carbides. Furthermore, nitrogen could also be present in these structures since the nitrides and carbides of Cb and Ta are isomorphous. Regardless of the composition of the massive carbides, it is unlikely that their effect in the alloy is great, since upon solution treatment they increase in size as a result of precipitation and become more or less idiomorphic rather than dissolve into the matrix. However, some of the smaller particles of CbC type carbides may dissolve during solution treatment and may subsequently strengthen the alloy by precipitation during use at high temperatures.

There is some X-ray evidence that the $Cr_{23}C_6$ type carbide forms in this alloy (refs. 10 and 11), and this probably is the most prevalent carbide in the grain boundaries and slip lines formed during aging or engine operation. This type of carbide would not be expected to have as great a strengthening effect upon the matrix of the alloy as the CbC type, since the atomic spacings of the metal atoms in $Cr_{23}C_6$ are very closely related to the atomic spacings of the matrix.

A previous investigation relating the microstructural characteristics of wrought Haynes-Stellite 21 to stress-rupture properties (ref. 3) has shown that drastically different carbide formations could be formed by various heat treatments and that these formations were largely responsible for the differences in the stress-rupture properties obtained. $Cr_{23}C_6$ has been found to be the principle carbide in the wrought Stellite 21 alloy (ref. 11). The as-wrought structure consisted of massive $Cr_{23}C_6$ particles in a matrix of face-centered cubic and hexagonal close-packed solid solutions. These carbides readily dissolved upon solution treatment at 2250° F for 16 hours and subsequently transformed upon heat treatment at lower temperatures. However, the CbC type carbides of S-816 do not dissolve even at 2300° F, so that subsequent aging or double-aging treatments did not improve the S-816.

SUMMARY OF RESULTS

An evaluation of S-816 turbine blades given different heat treatments along with as-forged blades has been made in a J33-33 turbojet engine. The engine was operated in a cyclic manner, 15 minutes at rated speed and 5 minutes at idle. At rated speed, blade stresses and temperatures were controlled at 21,400 pounds per square inch and 1450° F, respectively, at 2.4 inches above the base of the blade. The following is a summary of the results obtained and conclusions reached from the investigation:

1. The as-forged blades and the as-forged and aged blades performed better than blades given the standard heat treatment or any of the other heat treatments studied. The life of the as-forged blades ranged from 357 hours to over 430 hours, after which time the tests were discontinued. The failure times of the as-forged and aged blades ranged from 317 to 430 hours. In comparison, the blade group given the standard heat treatment for S-816 had failure times ranging from 148 to 310 hours.

The superior performance of the as-forged blades and the as-forged and aged blades was associated with a high hardness and a dense and uniform precipitation of carbides throughout the microstructure of the alloys. Forging was concluded to be responsible for this superior performance both by the introduction of strain-hardening and by promoting the uniform and dense precipitation of carbides during engine operation.

2. A double-aging treatment (24 hr at 1200° F, air cooled, 48 hr at 1500° F) following the standard solution treatment for S-816, which was intended to produce randomly scattered and dense precipitation, failed to improve blade performance relative to the group given the standard heat treatment. Failure times in this group of blades ranged from 102 to 310 hours. The aging treatments selected were felt to have caused too great a degree of overaging for S-816.

3. Two groups of specimens which were heat treated at 2300° F for 4 hours to produce germinated grains and simultaneously a greater degree of solution treatment than is produced by the standard solution treatment of 1 hour at 2150° F performed very poorly. Blade failures of one group (aged 16 hr at 1400° F) ranged from 36 to 285 hours, and failures of the other group (aged 24 hr at 1200° F and 48 hr at 1500° F) ranged from 59 to 401 hours. In the first of these groups, eutectic melting was observed and undoubtedly accounted for the poor performance, and in the second group, thick grain boundary formations or germinated grains or both were possible causes of the poor blade failures.

4. Blade lives of specimens given an overtemperature heat treatment at 1550° F for 16 hours (low-temperature overaging) were very low, ranging from 55 to 131 hours.

5. Blade lives of specimens given an overtemperature heat treatment at 1900° F for 16 hours (high-temperature overaging) ranged from 86 to 213 hours. For comparison, blades from the same stock given the standard S-816 heat treatment failed at times ranging from 94 to 202 hours. Therefore, high-temperature overaging was not harmful. Carbides in the as-heat-treated blades were largely spheroidized, but apparently partial solution treatment of some of these carbides occurred during the heat treatment. Partial solution treatment would permit strengthening of the blades by reprecipitation during operation in the engine.

6. The results from the overtemperature groups of blades just presented suggest that any heat treatment that would effectively overage the blades without simultaneously partially solution treating the blades could be harmful. Conversely, overaging at a high enough temperature to permit solution of some of the microconstituents could be beneficial.

7. The blade groups which performed best, the as-forged and aged groups, had the lowest creep rates and also the lowest total elongation. Total elongations of the as-forged and aged groups were $2\frac{1}{2}$ and 1 percent, respectively. Specimens given the standard S-816 heat treatment elongated 6.3 percent and specimens given the overtemperature treatments at 1550° F and 1900° F elongated 2.2 percent and 8.8 percent, respectively.

8. From known distributions of temperature and centrifugal stresses in the blades and from the known material properties, a theoretical zone of minimum strength of the blade airfoil was located at a distance of 2.4 inches from the base of the blades. In all but two cases (exclusive of damage failures), the turbine blades failed in operation above this zone. Creep measurements showed a trend which indicated that fracture probably occurred outside the zone of maximum creep. Such failure patterns suggest that factors other than centrifugal stress contribute to the failure mechanism of the blades. Vibratory conditions in the

engine could have accounted for the above failure patterns; and, in fact, exclusive of those blades which failed by damage, 65 percent of the failures were classified from macroexaminations and microexaminations as fatigue or a combination of fatigue and stress-rupture failures. The remaining 35 percent of the failures, which were classified as stress-rupture failures, may actually have undergone considerable damage by fatigue which could not have been detected from appearance alone.

9. Evidence presented in this investigation indicated that S-816 may be hardened (or strengthened) by hot-cold working or by precipitation but not by solution treatment alone.

Lewis Flight Propulsion Laboratory
National Advisory Committee for Aeronautics
Cleveland, Ohio, November 29, 1954

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TABLE I. - STRESS-RUPTURE RESULTS AT 20,000 PSI AND 1500° F FOR SPECIMENS MACHINED FROM AIRFOIL OF J47 TURBINE BLADES

Solution treated 1 hr at 2150° F, water quenched, aged 16 hr at 1400° F, air cooled		No heat treatment; tested in as-forged condition		Aged 16 hr at 1400° F without solution treatment	
Life, hr	Elongation, percent	Life, hr	Elongation, percent	Life, hr	Elongation, percent
80	30	165	20	185	15
112	25	255	12	620	--
133	35	290	15		
152	40				
450	42				
760	40				

TABLE II. - HEAT TREATMENT GIVEN DIFFERENT BLADE GROUPS

AND REASONS FOR TREATMENT

Group	Solution treatment			First age[a]		Second age[a]		Purpose of heat treatment, to determine -
	Temperature, °F	Time, hr	Quench	Temperature, °F	Time, hr	Temperature, °F	Time, hr	
1[b]	As-forged							Whether as-forged blades would behave equally as well as heat-treated blades
2[b]	None			1400	16			Whether aging alone could improve as-forged structure or properties
3[b,c]	2150	1	Water	1400	16			(Standard for comparison)
4[b]	2150	1	Air	1200	24	1500	48	Whether aging to develop nucleation sites followed by aging to cause precipitation in the sites can improve properties
5[b]	2300	4	Water	1400	16			Whether germinated grains are harmful or whether increased solution which occurs at higher temperatures improves life
6[b]	2300	4	Air	1200	24	1500	48	Whether double aging (see group 4) will compensate for germinated grains if they are shown to be harmful in group 5
7[c,d]	2150	1	Water	1400	16			(Standard for comparison between experimental groups 1 to 6 and standard stock previously run in identical types of test)
8[d]	None			1550	16			Whether low-temperature overaging (overheating) is harmful
9[d]	None			1900	16			Whether high-temperature overaging (overheating) is harmful

[a]All aging treatments followed by air cooling.

[b]Blades forged from single lot of bar stock from same heat.

[c]Standard heat treatment.

[d]Blades from U.S. Air Force stock.

TABLE III. - RESULTS OF ENGINE OPERATION OF TURBINE BLADES

Group	Blade	Order of blade failure	Failure time, hr at rated speed	Failure mechanism (a)	Location of failure origin (b)	Distance of failure from root platform, in.
1	5	1	356.63	SR	T.E.	$2\frac{13}{16}$
	1	2	356.65	SR	T.E.	$2\frac{3}{4}$
	2	3	379.25	SR	Near C	$2\frac{7}{8}$
	6	4	404.50	D	L.E.	$3\frac{3}{8}$
	3	-	>430	--	----	---
	4	-	>430	--	----	---
2	3	1	317.25	SR-F	T.E.	$2\frac{3}{4}$
	5	2	376.50	SR	C	$2\frac{7}{8}$
	1	3	421.30	F	T.E.	$2\frac{27}{32}$
	2	4	430	D	Tip	$4\frac{1}{4}$
	4	-	>430	--	----	---
	6	-	>430	--	----	---
3	4	1	147.95	D	T.E.	Tip
	6	2	166.78	F	T.E.	$3\frac{1}{2}$
	3	3	193.75	F	T.E.	$3\frac{1}{8}$
	1	4	202.37	F	T.E.	$2\frac{1}{2}$
	5	5	249.75	D	T.E.	3
	2	6	309.78	SR-F	T.E.	3
4	1	1	101.72	SR-F	T.E.	$3\frac{1}{8}$
	6	2	110.37	D	L.E.	$1\frac{1}{2}$
	3	3	116.65	F	L.E.	$2\frac{27}{32}$
	5	4	194.08	F	L.E.	$2\frac{31}{32}$
	2	5	209.25	SR(Necked)	Entire cross section	$2\frac{3}{4}$
	4	6	309.78	SR(Cracked)	C	$2\frac{5}{16}$
5	4	1	36.00	SR(Cracked)	C	$2\frac{5}{16}$
	6	2	49.25	SR(Cracked)	C	3
	2	3	116.65	SR(Cracked)	C	$3\frac{1}{16}$
	5	4	142.50	SR(Cracked)	C	$2\frac{1}{2}$
	3	5	142.50	SR(Cracked)	C	$2\frac{15}{16}$
	1	6	285.17	F(?)	T.E.	$2\frac{5}{8}$
6	3	1	59.17	F	T.E.	$3\frac{3}{16}$
	2	2	93.43	D	T.E.	$3\frac{9}{16}$
	1	3	99.28	F	L.E.	$3\frac{1}{16}$
	4	4	194.00	D	T.E.	$3\frac{7}{16}$
	5	5	209.25	D(?),SR	L.E.	$2\frac{15}{32}$
	6	6	401.00	SR	C	$2\frac{5}{8}$
7	1	1	93.72	SR-F	L.E.	$3\frac{1}{16}$
	3	2	109.57	D	L.E.	$2\frac{1}{4}$
	2	3	147.95	DD	L.E.	$2\frac{3}{4}$
	6	4	201.83	F	L.E.	$3\frac{1}{16}$
8	5	1	54.50	F(?), D	T.E.	$2\frac{13}{16}$
	6	2	60.42	F	L.E.	$2\frac{15}{16}$
	4	3	69.00	F	L.E.	$3\frac{1}{8}$
	2	4	73.25	D	L.E.	$2\frac{15}{16}$
	3	5	78.57	F	T.E.	$2\frac{3}{4}$
	1	6	131.22	F	T.E.	$3\frac{1}{8}$
9	5	1	86.08	SR-F(?)	L.E.	$3\frac{1}{4}$
	2	2	98.00	SR-F	L.E.	$3\frac{3}{16}$
	3	3	170.68	F	L.E.	$2\frac{15}{16}$
	1	4	173.03	F	L.E.	$2\frac{15}{16}$
	6	5	206.67	SR-F	L.E.	$3\frac{1}{16}$
	4	6	212.80	D	L.E.	2

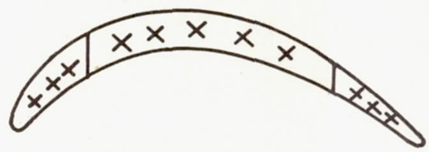
[a]SR, stress-rupture; D, damage, SR-F, stress-rupture followed by fatigue; F, fatigue.

[b]T.E., trailing edge; C, center; L.E., leading edge.

TABLE IV. - GRAIN SIZE OF AS-HEAT-TREATED SPECIMENS

Group	Largest A.S.T.M. grain size	Most prevalent A.S.T.M. grain size	Smallest A.S.T.M. grain size
1	8	8	8
2	5	7	8
3	6	8	8
4	2	6	8
5	2	5	8
6	3xl	1	5
7	4	6	8
8	1	5	8
9	4	6	8

TABLE V. - HARDNESS OF AS-HEAT-TREATED SPECIMENS

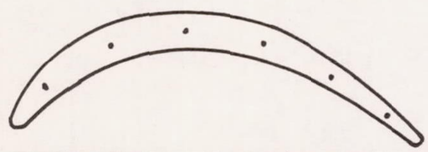
Group	Individual hardness readings,[a] Rockwell C 	Average hardness,[a] Rockwell C
1	39 36 33 37 31 35	35
2	33 34 34 36 37	35
3	25 28 26	26
4	24 26 25 26 25	25
5	25 32 29[b]	29[b]
6	29 29 30 29 26	29
7	-- -- -- -- -- -- -- -- -- -- --	25[c]
8	29 31 29 29 29	29
9	26 25 26 28 28	27

[a]Converted from Rockwell A readings.

[b]Individual readings were converted from R-15-N and are three of six readings, the average of which is Rockwell C-29.

[c]Average of readings made from seven specimens cut from Air Force stock and taken from table II, ref. 3.

TABLE VI. - HARDNESS OF FIRST AND LAST BLADE FAILURES IN CROSS SECTION APPROXIMATELY 1/8 INCH FROM FAILURE ORIGIN

Group	Failure[a]	Individual Rockwell C hardness readings[b]						Average Rockwell C hardness[b]	Location of failure origin (c)
		1st	2nd	3rd	4th	5th	6th		
									
1	First	37.0	30	33	33	35	34	34	T.E.
	Last	33	32	33	34	34	34	33	C
2	First	33.5	33	32	31	32	33	32	T.E.
	Last	34	34	33	33	34	34	34	T.E.
3	First	29	29	31	28	29	24	28	T.E.
	Last	30	30	31	31	33	29	31	T.E.
4	First	27	30	30	30	28.5	29	29	L.E.
	Last	24	26	26	24	26	23	25	C
5	First	28	31	32	31	28	27	30	C
	Last	29	32	31	32	30	32	31	T.E.
6	First	29	30	28	26	26	24	27	T.E.
	Last	27	30	30	31	31	24	29	C
7	First	30	29	29	28	29	24	28	L.E.
	Last	29	31	33	33	33	29	31	L.E.
8	First	29	29	30	29	27	26	28	T.E.
	Last	30	27	28	26	31	30	29	T.E.
9	First	29	30	30	31	31	30	30	L.E.
	Last	30	31	30	31	31	27	30	L.E.

[a]Designation of first or last failure pertains to failures exclusive of those which occurred by damage.

[b]Hardness measurements were made with Rockwell superficial hardness tester and readings were converted to Rockwell C.

[c]T.E., trailing edge; C, center.

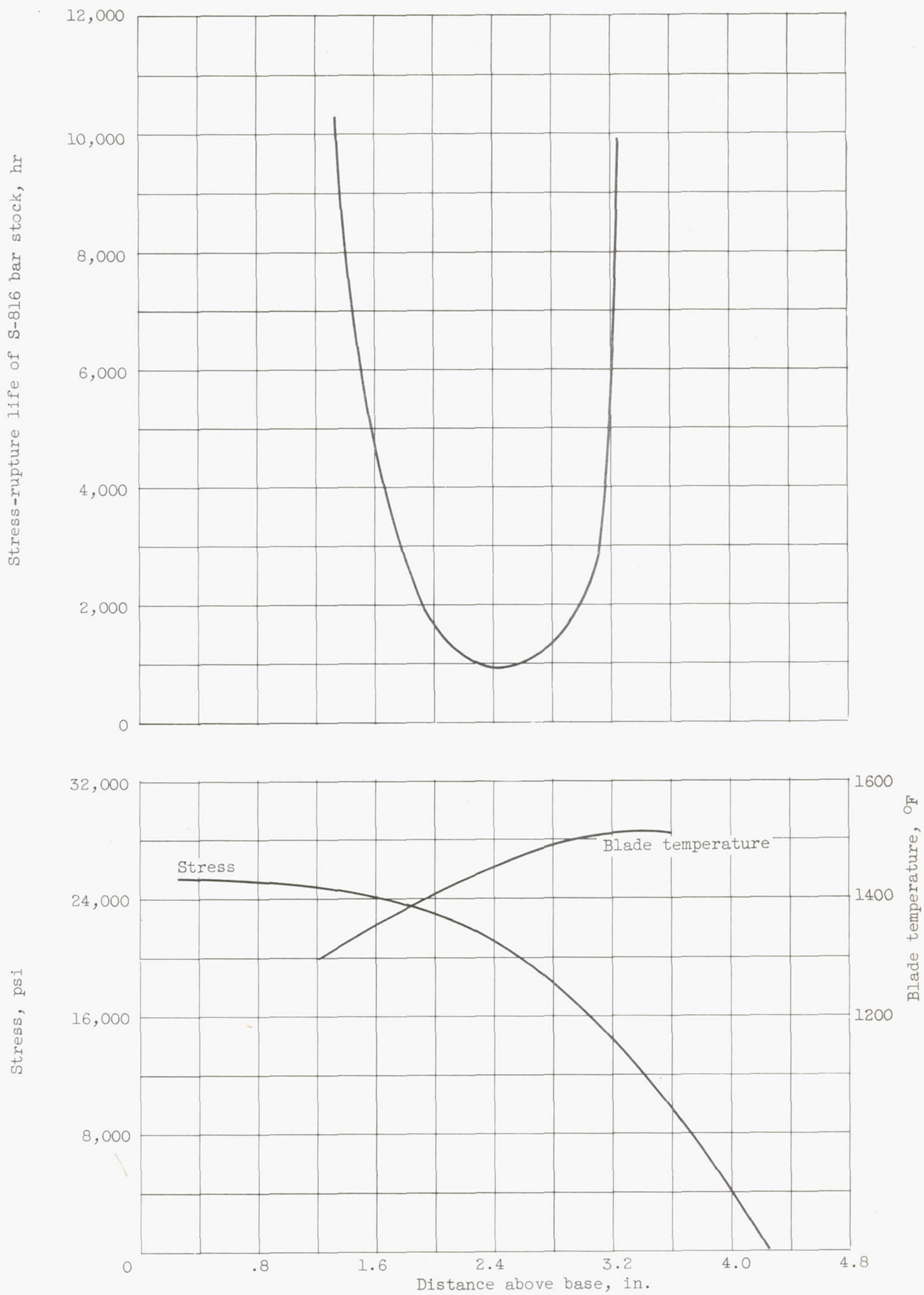
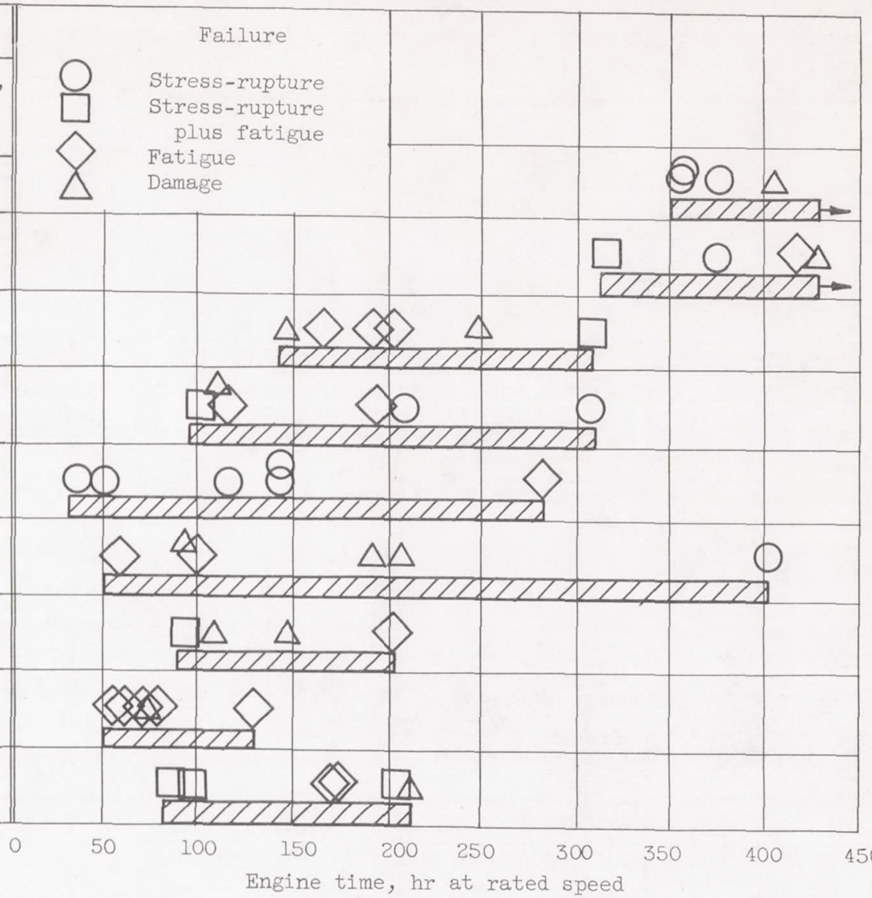


Figure 1. - Stress and temperature distribution in J33-33 turbine blades operated at full power and corresponding stress-rupture life of S-816 bar stock.

Group	Solution treatment			First age[a]		Second age[a]	
	Temperature, °F	Time, hr	Quench	Temperature, °F	Time, hr	Temperature, °F	Time, hr
1[b]	As forged						
2[b]	None			1400	16		
3[b,c]	2150	1	Water	1400	16		
4[b]	2150	1	Air	1200	24	1500	48
5[b]	2300	4	Water	1400	16		
6[b]	2300	4	Air	1200	24	1500	48
7[c,d]	2150	1	Water	1400	16		
8[d]	None			1550	16		
9[d]	None			1900	16		



[a]All aging treatments followed by air cooling.
[b]Blades forged from single lot of bar stock from same heat.
[c]Standard heat treatment.
[d]Blades from U.S. Air Force stock.

Figure 2. - Blade performance of J33-33 turbine blades of as-forged and heat-treated wrought S-816.

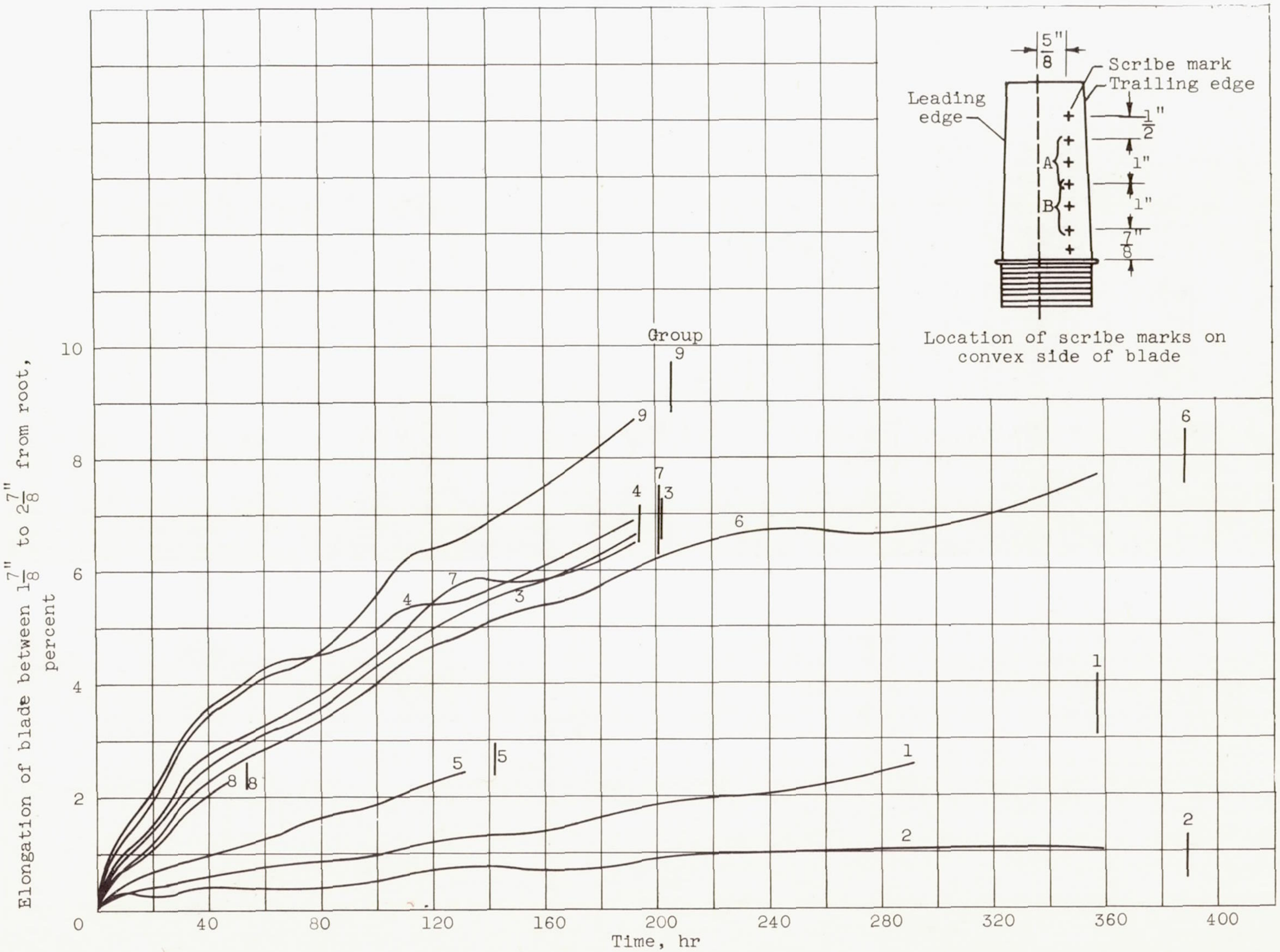


Figure 3. - Elongation of portions of blades during engine operation. (Zone A.)

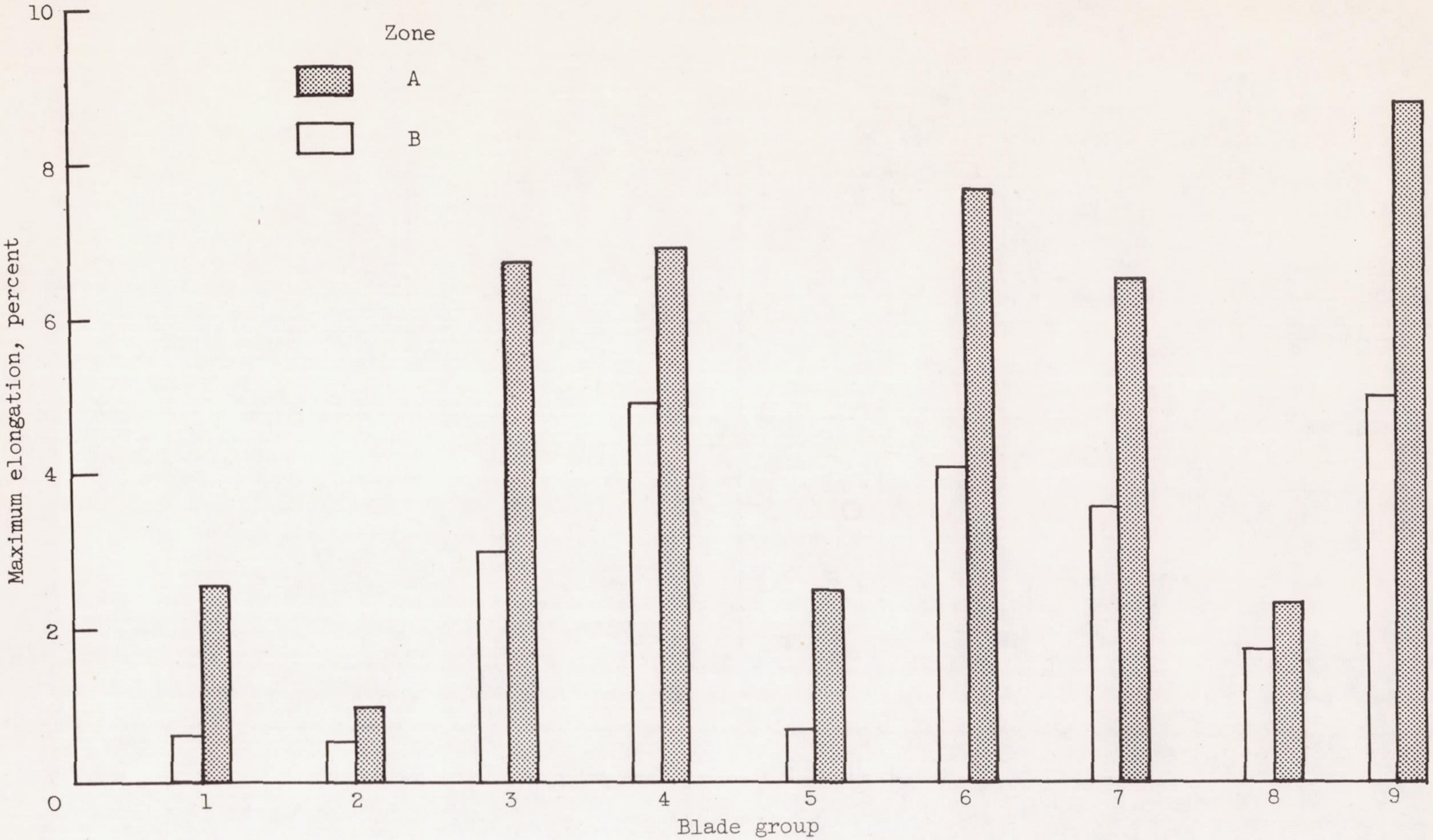


Figure 4. - Maximum percent elongation measured for segments of blades extending from $1\frac{7}{8}$ to $2\frac{7}{8}$ inches from base (zone A) and from 7/8 to $1\frac{7}{8}$ inches from base (zone B).

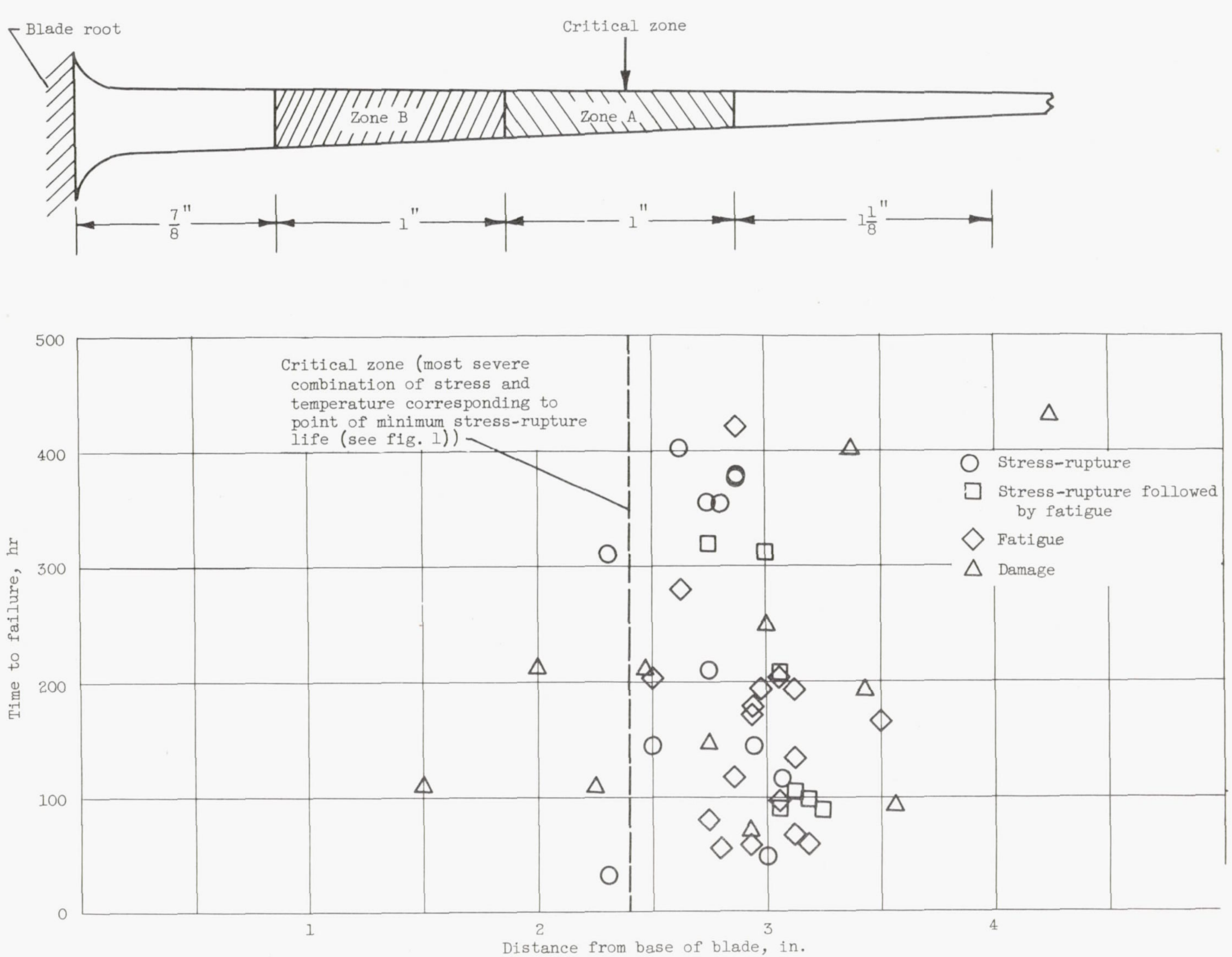
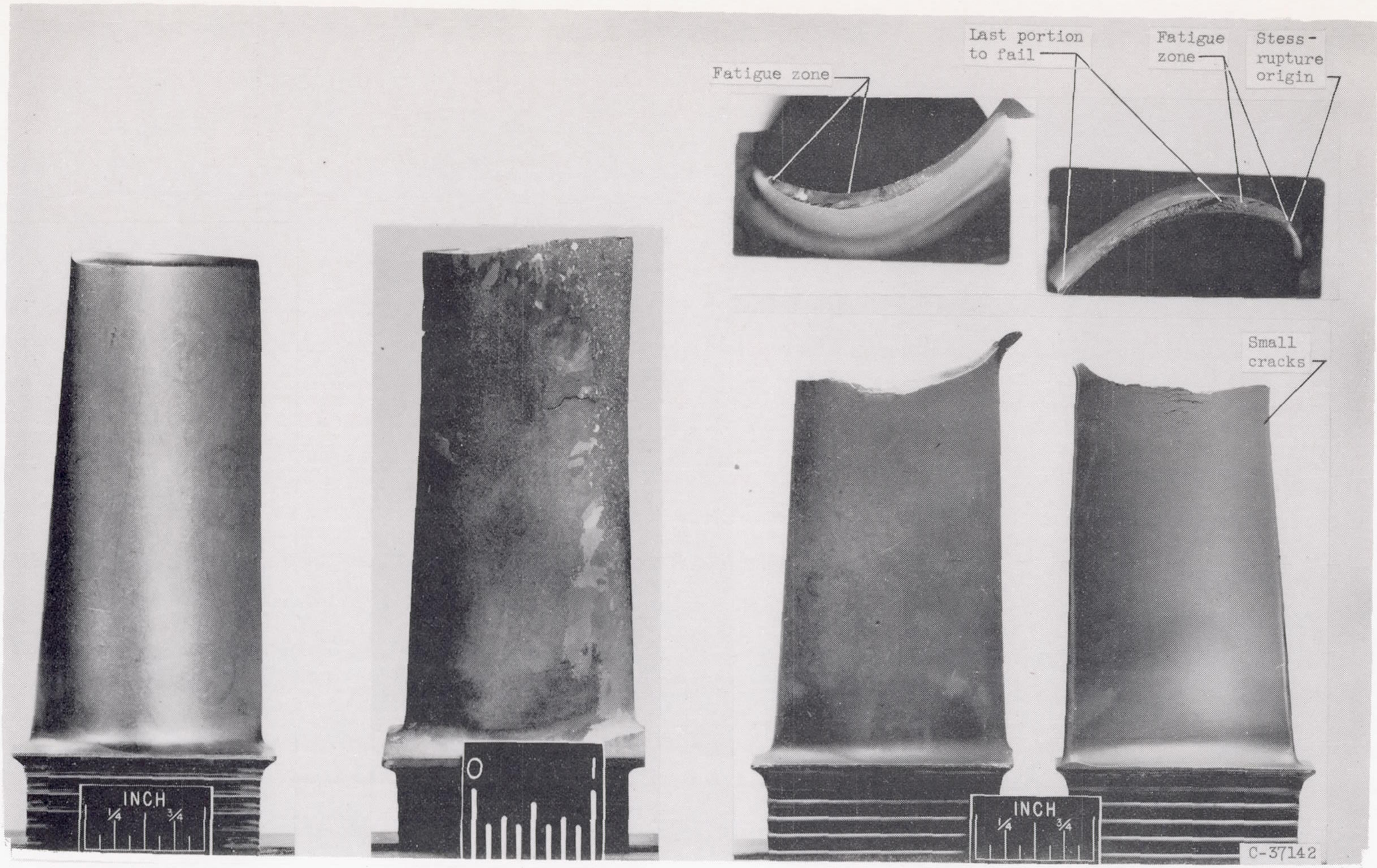


Figure 5. - Location of failure origins in blades.



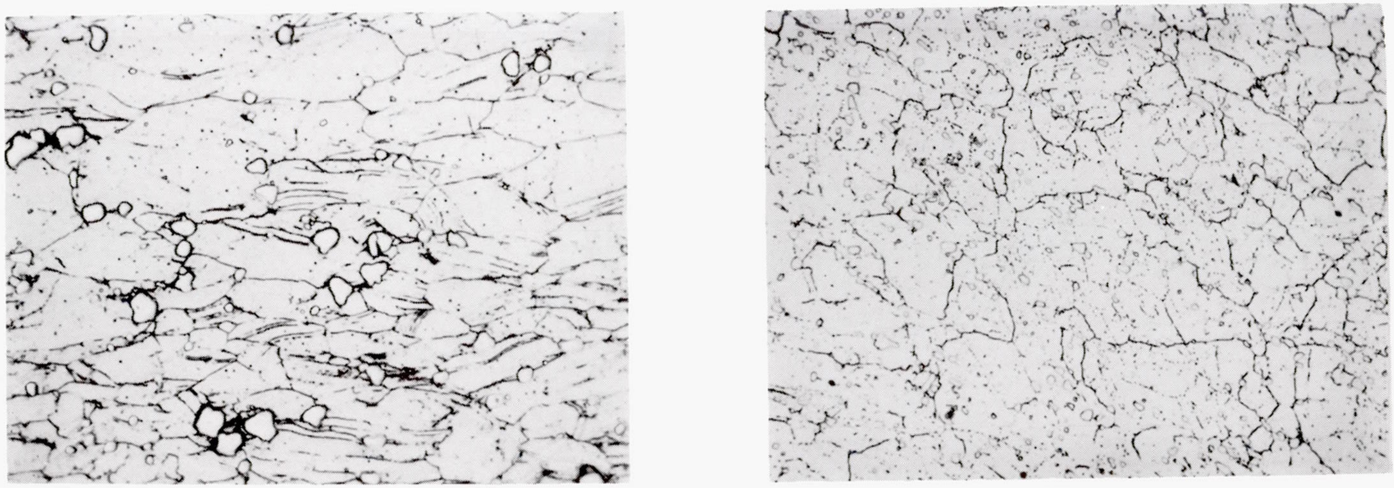
(a) As-forged.

(b) Typical stress-rupture crack.

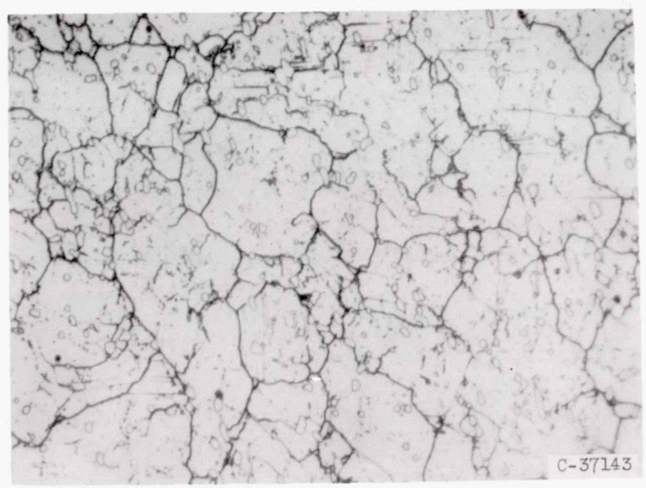
(c) Typical fatigue failure.

(d) Typical stress-rupture followed by fatigue failure.

Figure 6. - As-forged blade and typical failed blades.

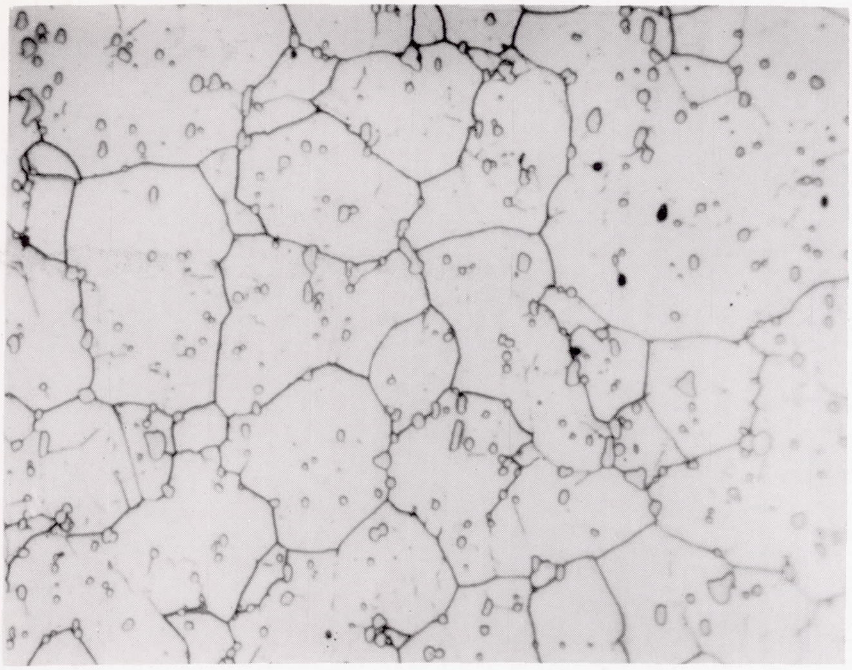


(a) Group 1 (as-forged).

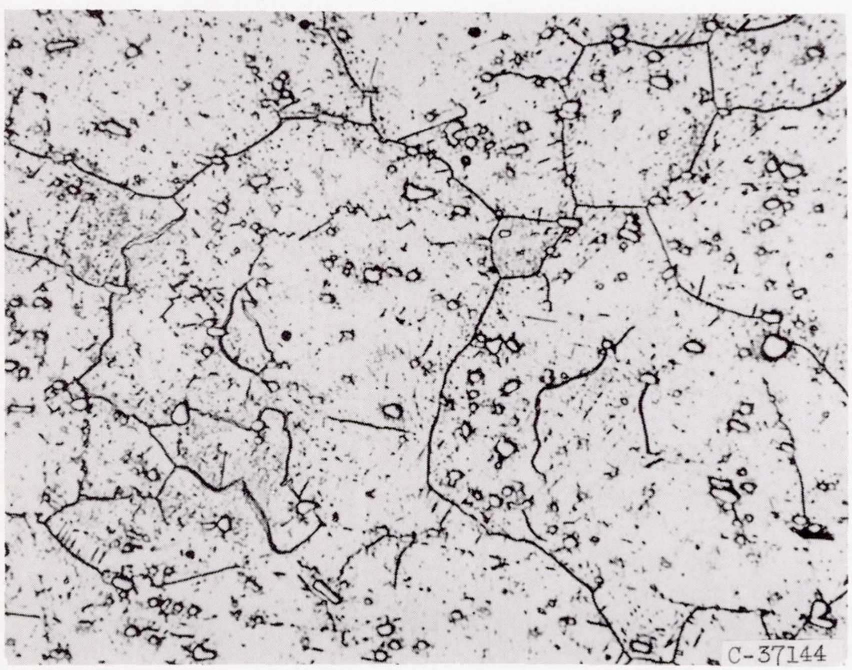


(b) Group 2 (aged 16 hr at 1400° F).

Figure 7. - Microstructures of as-heat-treated specimens cut from representative segments of blades.

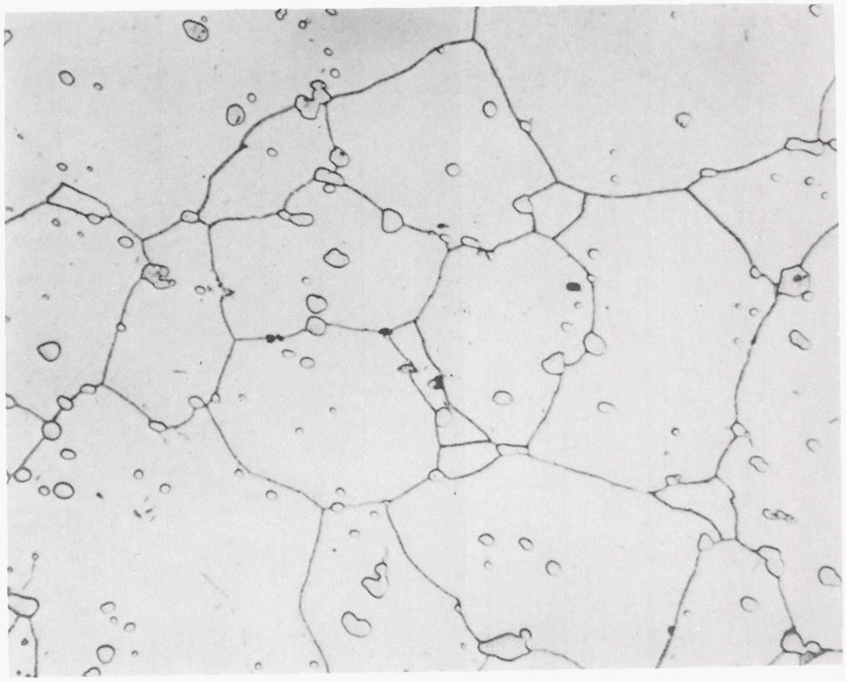


(c) Group 3 (standard heat treatment: 1 hr at 2150° F, water quenched, aged 16 hr at 1400° F, air cooled).

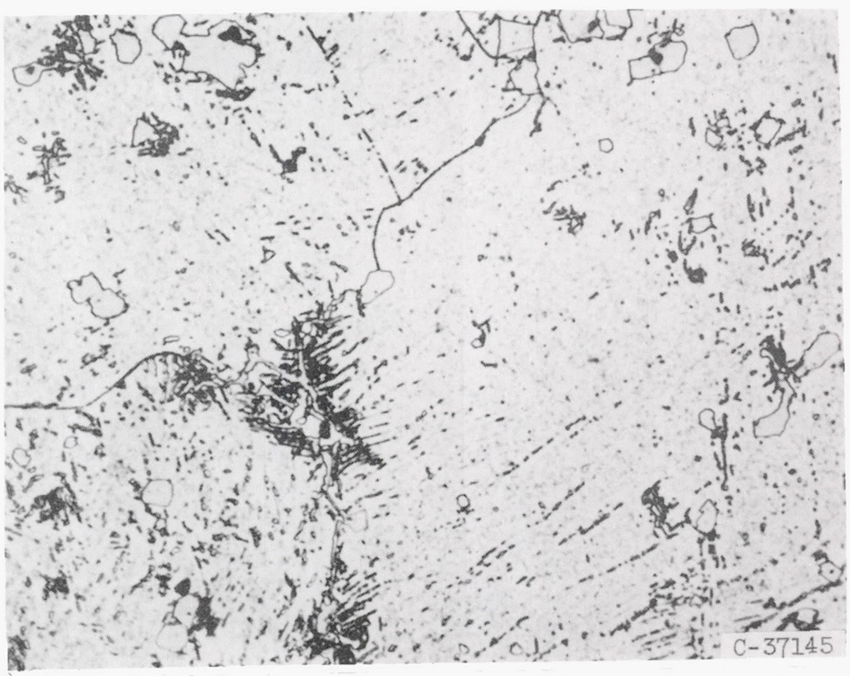


(d) Group 4 (1 hr at 2150° F, air quenched, aged 24 hr at 1200° F, air cooled, aged 48 hr at 1500° F, air cooled).

Figure 7. - Continued. Microstructures of as-heat-treated specimens cut from representative segments of blades.

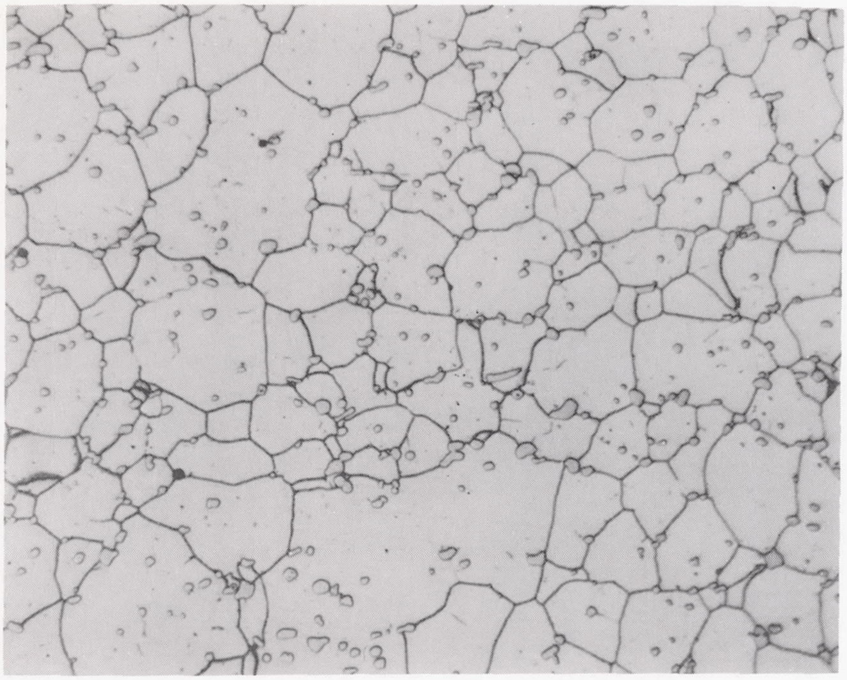


(e) Group 5 (4 hr at 2300° F, water quenched, aged 16 hr at 1400° F, air cooled).

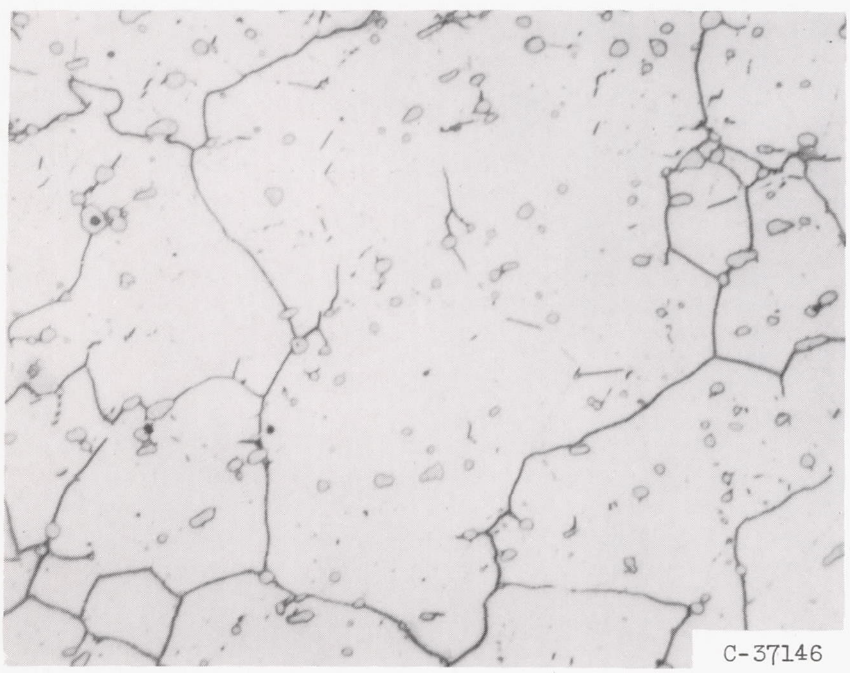


(f) Group 6 (4 hr at 2300° F, air quenched, aged 24 hr at 1200° F, air cooled, aged 48 hr at 1500° F, air cooled).

Figure 7. - Continued. Microstructures of as-heat-treated specimens cut from representative segments of blades.

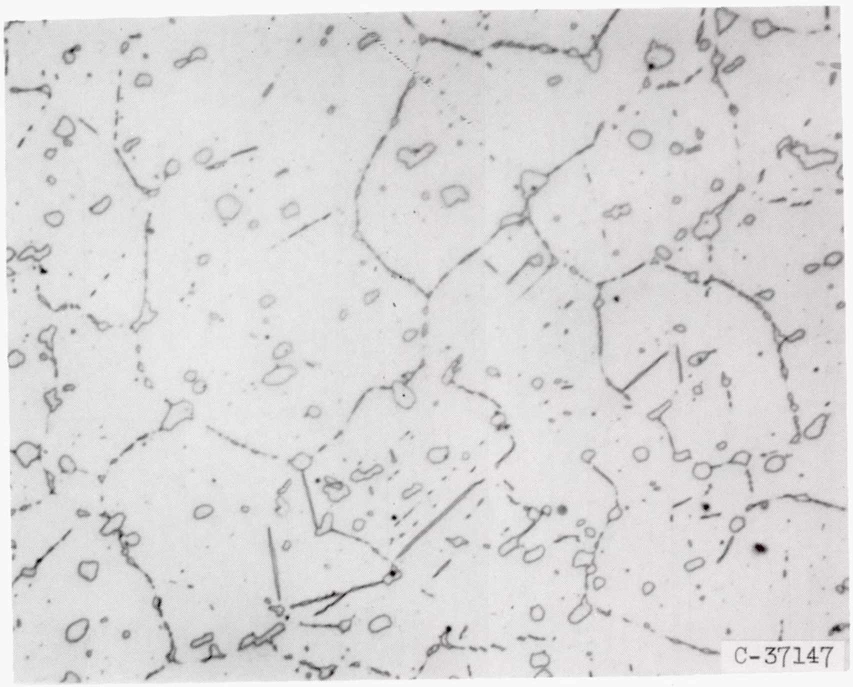


(g) Group 7 (Air Force stock; standard heat treatment: 1 hr at 2150° F, water quenched, aged 16 hr at 1400° F, air cooled).



(h) Group 8 (Air Force stock; aged 16 hr at 1550° F, air cooled).

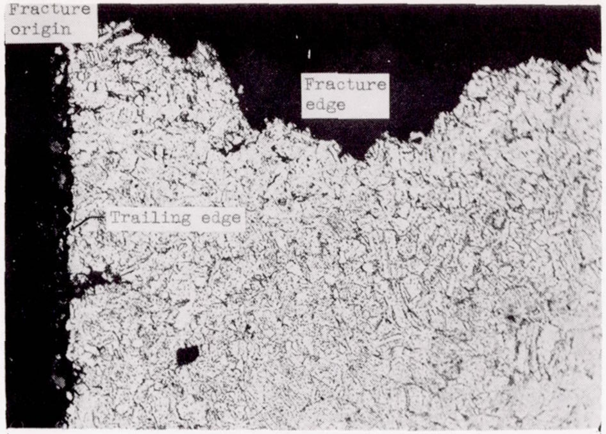
Figure 7. - Continued. Microstructures of as-heat-treated specimens cut from representative segments of blades.



(i) Group 9 (Air Force stock; aged 16 hr at 1900° F, air cooled).

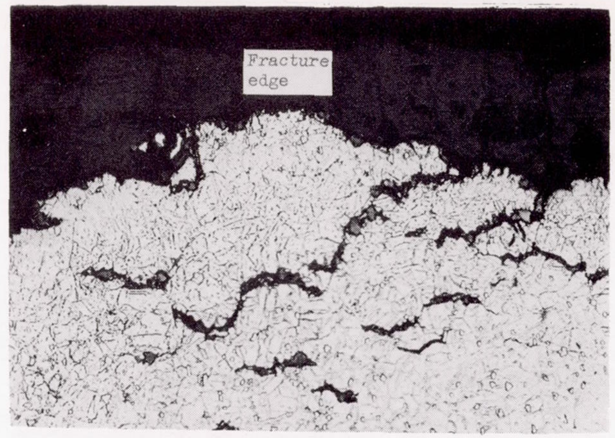
Figure 7. - Concluded. Microstructures of as-heat-treated specimens cut from representative segments of blades.

First blade to fail, 357 hr

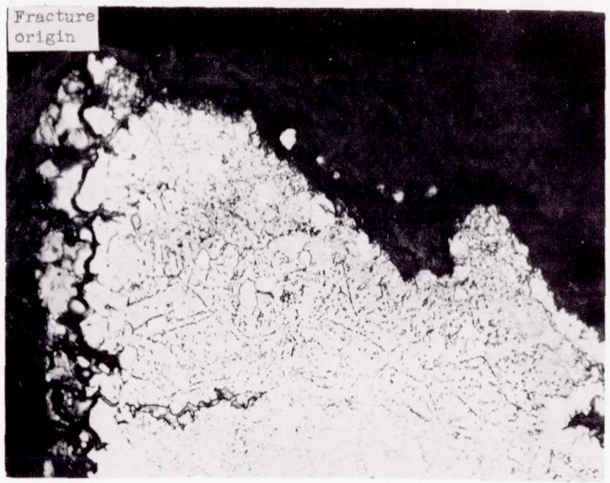


X250
Tears below fracture edge and irregular fracture edge are indicative of stress-rupture failure mechanism.

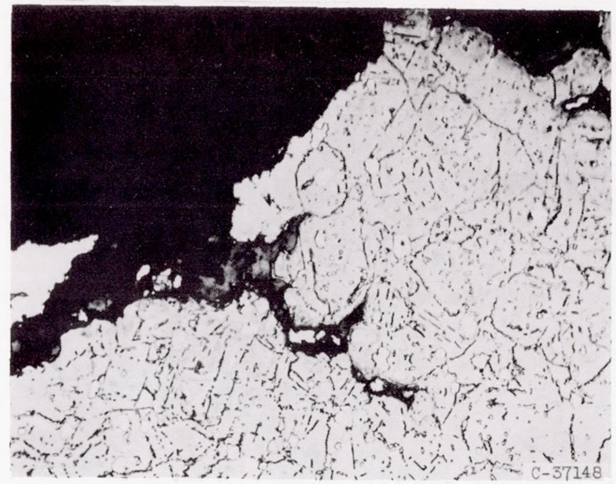
Last blade to fail (third failure), 379 hr



X250
Typical stress-rupture tears in center portion of blade where failure originated.



X750
Oxide in intergranular crack below fracture edge. Area is same as at X100 except that specimen was repolished

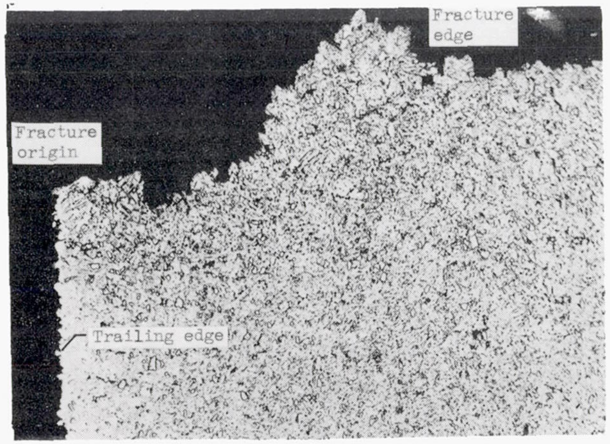


X750
Enlarged view of fracture edge.

(a) Group 1.

Figure 8. - Microstructures of failure origins or areas near failure origins in first and last blade failures.

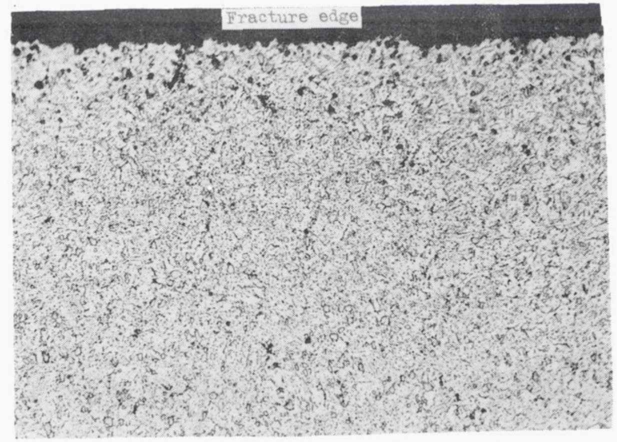
First blade to fail, 317 hr



X250

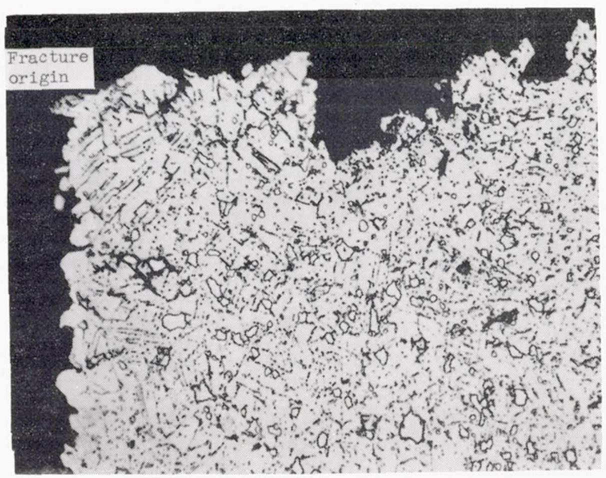
Stress-rupture characteristics in form of tears and jagged edges.

Last blade to fail, 421 hr



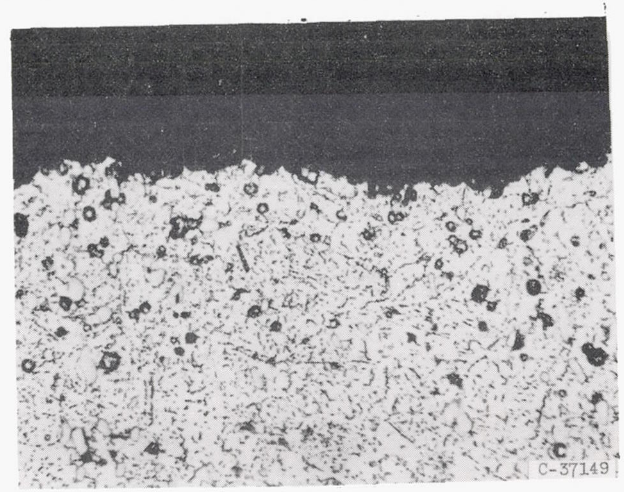
X250

Fatigue originated 1/32 in. from edge. Fracture edge shown is in area of origin and appears smooth and straight.



X250

Same area as above.



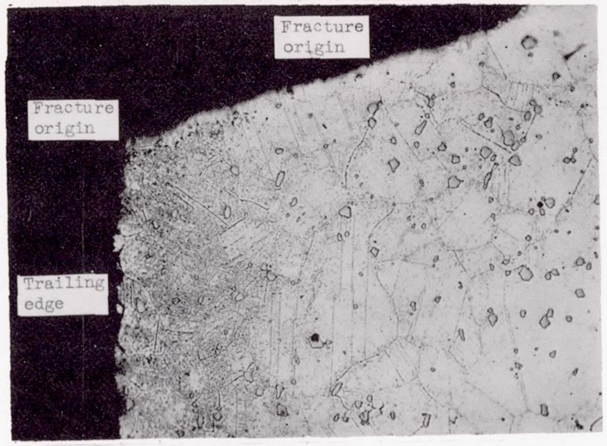
X750

Same as above area. Fracture edge appears transcrystalline.

(b) Group 2.

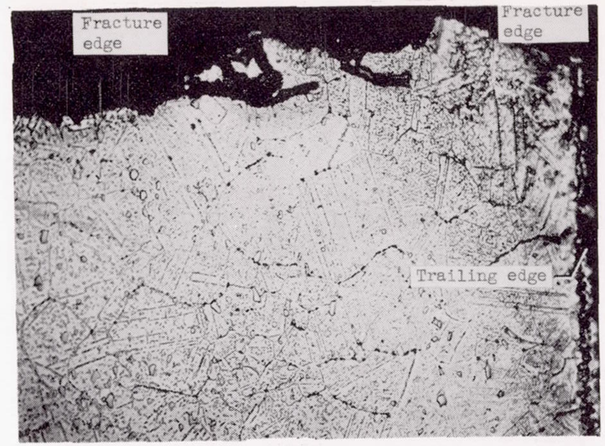
Figure 8. - Continued. Microstructures of failure origins or areas near failure origins in first and last blade failures.

First blade to fail, 167 hr

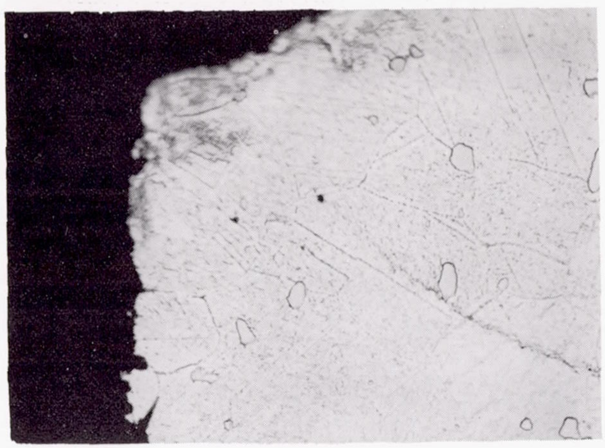


X250
Smooth transcrystalline nature of fracture edge, which is indicative of fatigue.

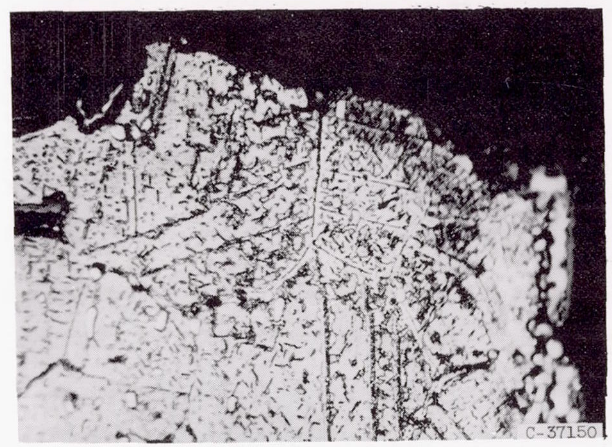
Last blade to fail, 310 hr



X250
Jagged fracture edge and stress-rupture tears.



X750
Same area as above.

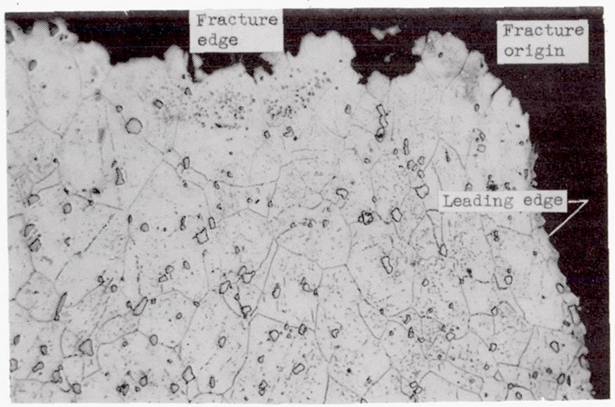


X750
Same area as above. Fine Widmanstätten structure.

(c) Group 3.

Figure 8. - Continued. Microstructures of failure origins or areas near failure origins in first and last blade failures.

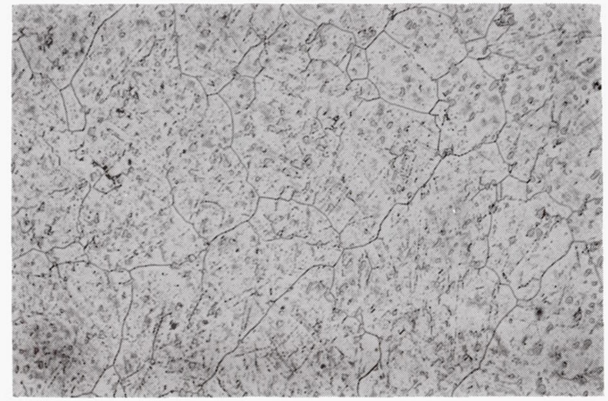
First blade to fail, 102 hr



X250

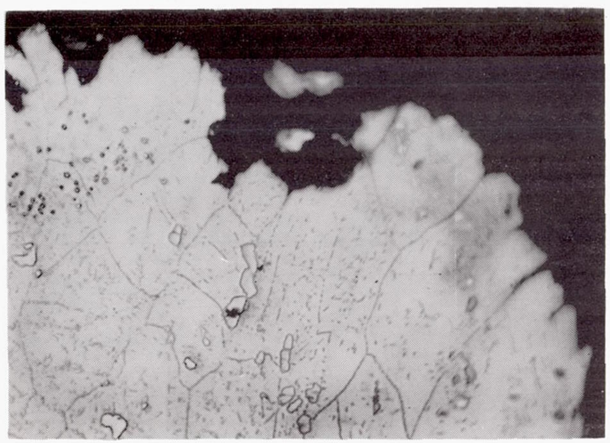
Very little general precipitation in structure, which is not consistent with as-heat-treated structure of fig. 6(d). Jagged edge is indicative of stress-rupture failure. Dent on leading edge could have resulted after blade failed.

Last blade to fail, 310 hr



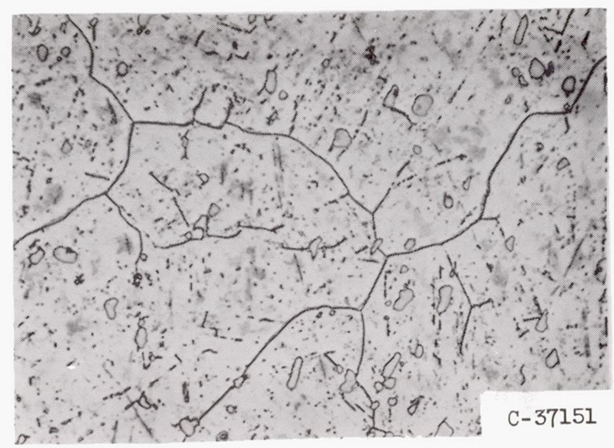
X250

Stress-rupture crack was removed by polishing. Structure shown is in immediate area of crack which was present.



X750

Same area as above.



X750

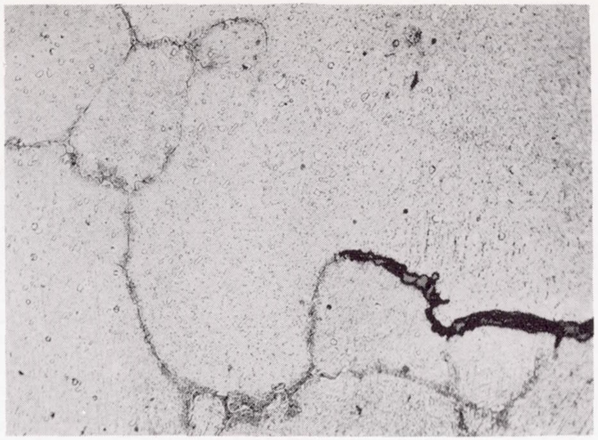
Large quantities of general matrix precipitation are evident.

(d) Group 4.

Figure 8. - Continued. Microstructures of failure origins or areas near failure origins in first and last blade failures.

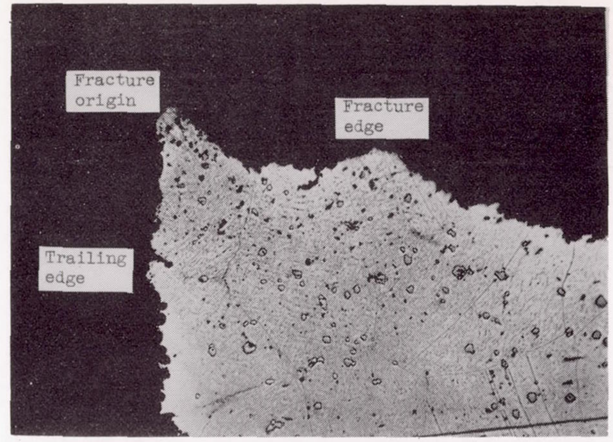
First blade to fail, 36 hr

Last flade to fail, 285 hr



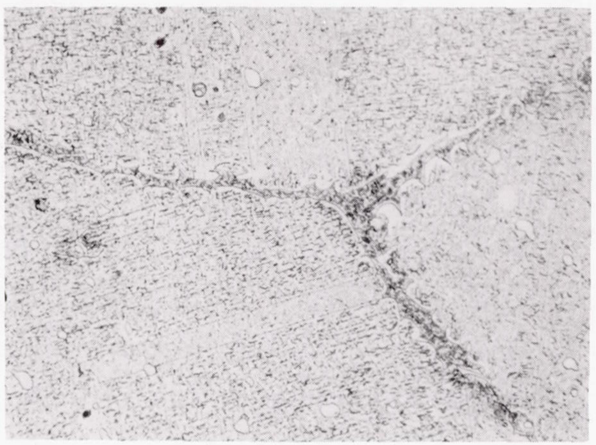
X250

Intergranular crack following grain boundary exhibiting eutectic melting.



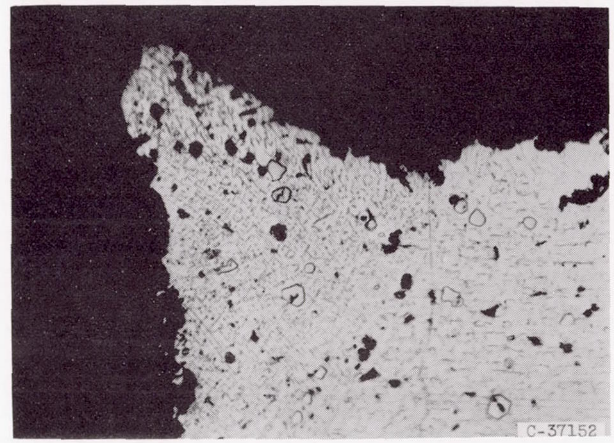
X250

Fracture edge appears transcrystalline, indicative of fatigue.



X750

Eutectic melting in grain boundaries and rosette formations are evident.



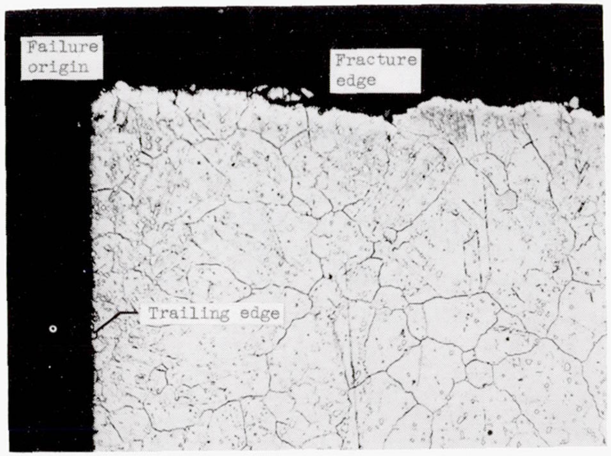
CX750

Grain boundaries contain thick precipitation. Voids could be evidence of melting or etching effect.

(e) Group 5.

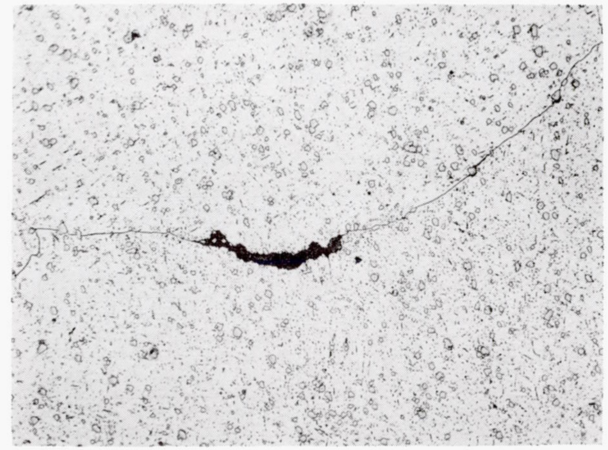
Figure 8. - Continued. Microstructures of failure origins or areas near failure origins in first and last blade failures.

First blade to fail, 59 hr

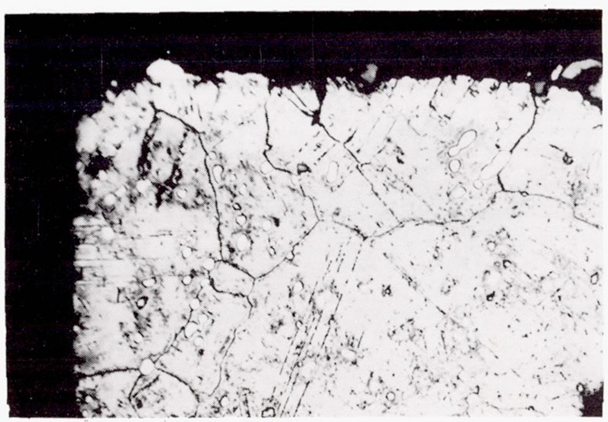


X250
Fracture edge is transcrystalline, indicative of fatigue.

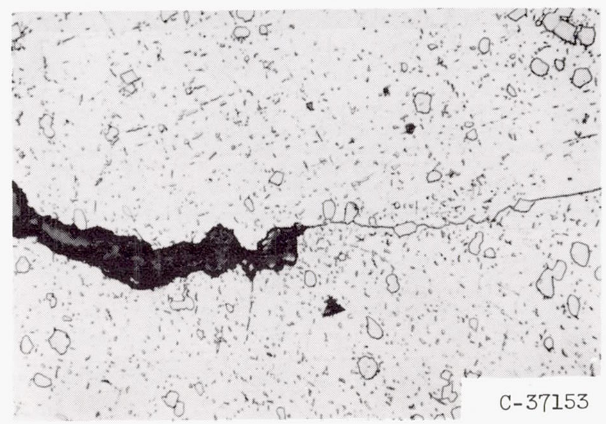
Last blade to fail, 401 hr



X250
One of several stress-rupture cracks occurring in grain boundaries in center of blade airfoil.



X750
Same area as above.

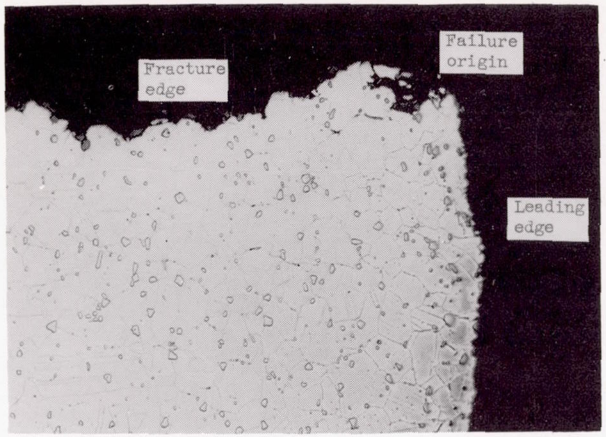


X750
Same area as above.

(f) Group 6.

Figure 8. - Continued. Microstructures of failure origins or areas near failure origins in first and last blade failures.

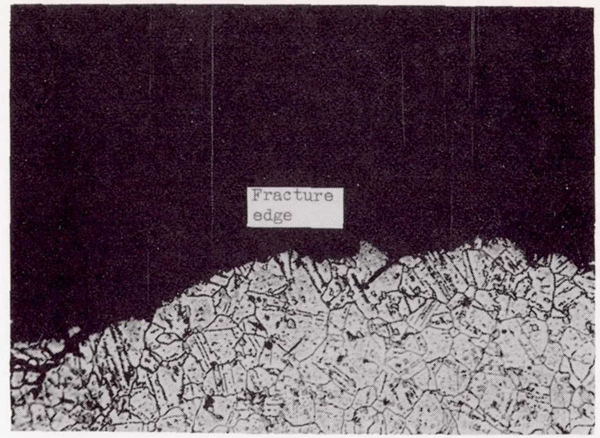
First blade to fail, 94 hr



X250

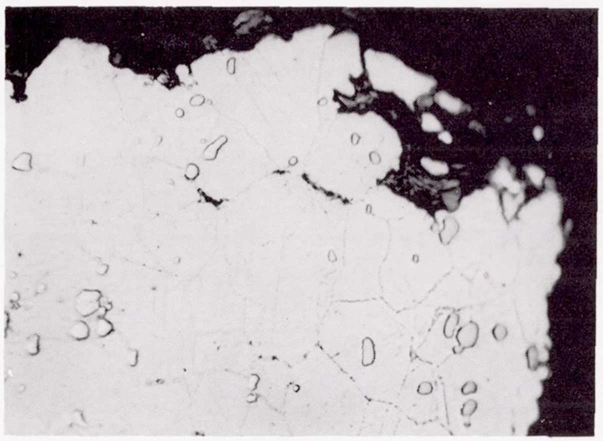
Stress-rupture characteristics of failure occur at jagged failure origin, irregular fracture edge, and intercrystalline tear.

Last blade to fail, 202 hr



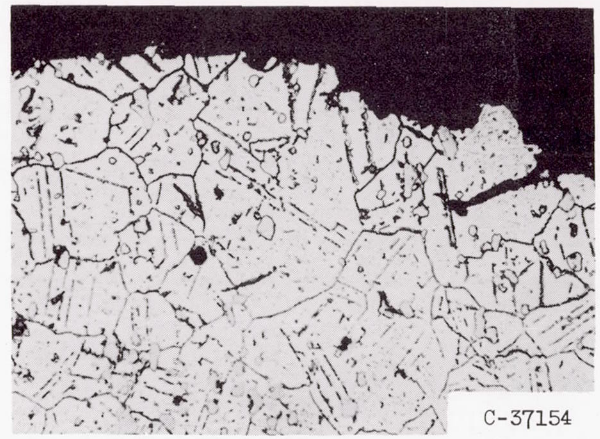
X250

Fracture originated 1/4 in. from leading edge and progressed in transcryatalline manner indicative of fatigue.



X750

Same area as above.



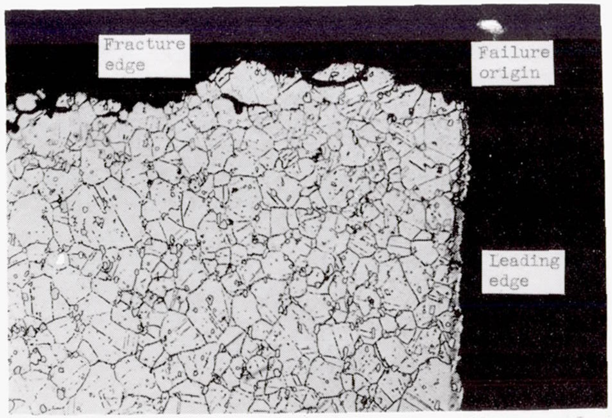
X750

Same area as above.

(g) Group 7.

Figure 8. - Continued. Microstructures of failure origins or areas near failure origins in first and last blade failures.

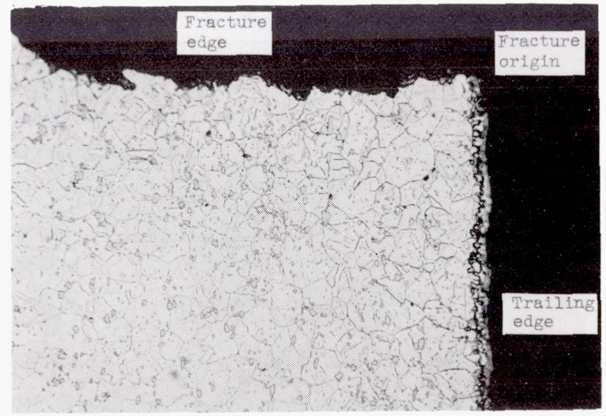
First blade to fail, 60 hr



X250

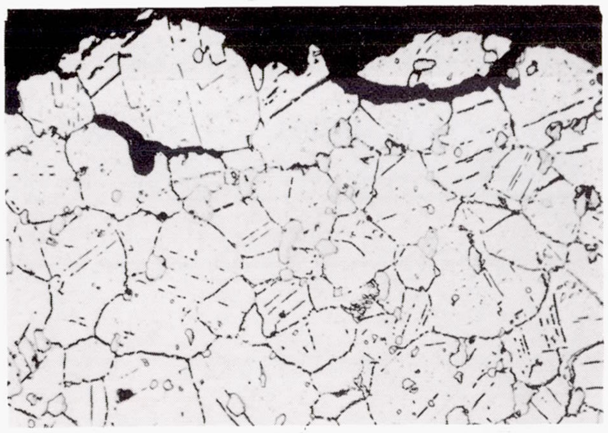
Fracture edge appears both transcrystalline and intercrystalline. An intercrystalline crack is formed at leading edge. Microexamination indicates stress-rupture followed by fatigue failure, whereas macroexamination indicates fatigue.

Last blade to fail, 151 hr



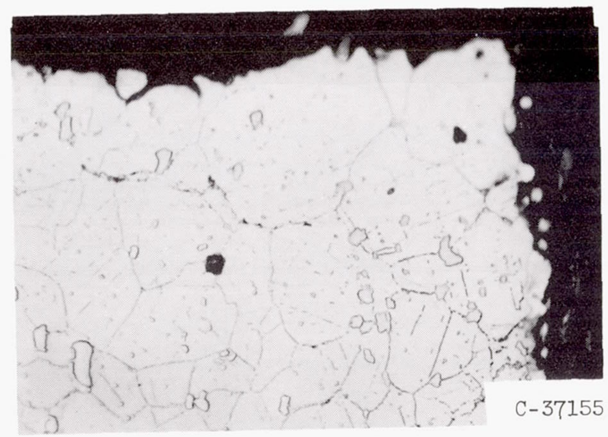
X250

Fracture edge is smooth and transcrystalline.



X750

Same area as above.



C-37155

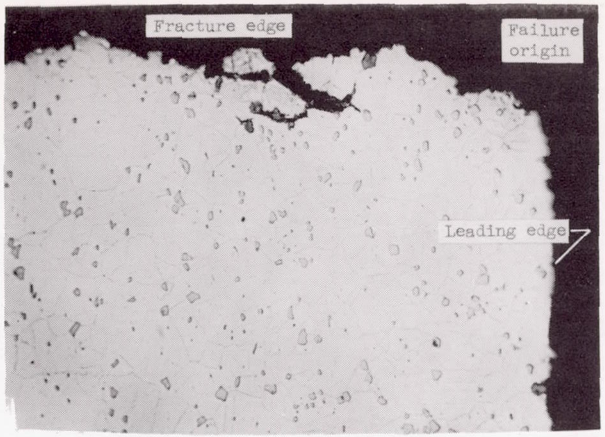
X750

Same area as above. Matrix of grains have very little precipitation compared with groups 3 and 7.

(h) Group 8.

Figure 8. - Continued. Microstructures of failure origins or areas near failure origins in first and last blade failures.

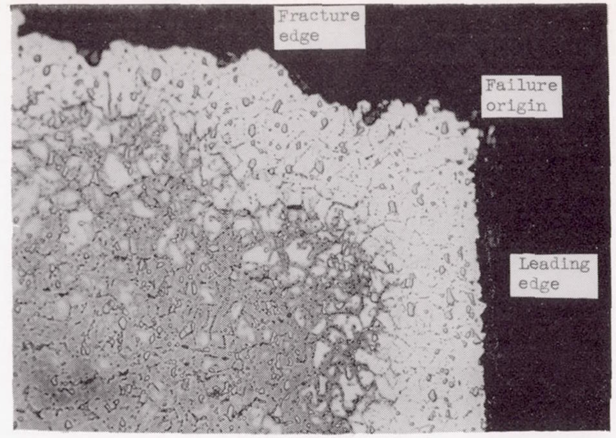
First blade to fail, 86 hr



X250

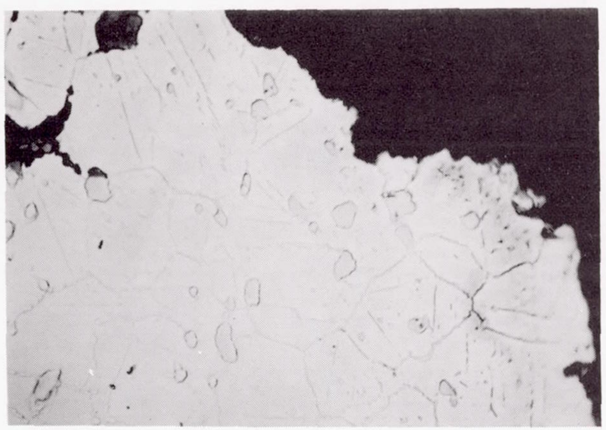
Jaggedness and intercrystalline characteristics at and near origin of fracture indicate that fracture originated by stress-rupture. However, this is a borderline case and could be a pure fatigue failure. Fatigue characteristics occur in the left-hand portion of the photograph.

Last blade to fail, 207 hr



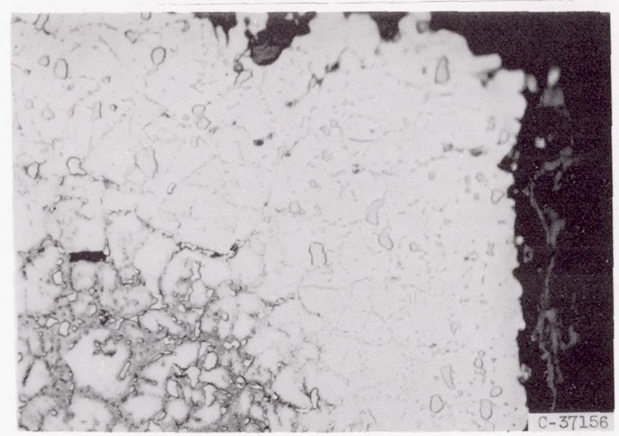
X250

Fracture edge appears transcrystalline, but small tears within specimens are indicative of stress-rupture. The structure has been stained differentially, but this does not obscure detail.



X750

Thick grain boundary precipitation and very little general or matrix precipitation.

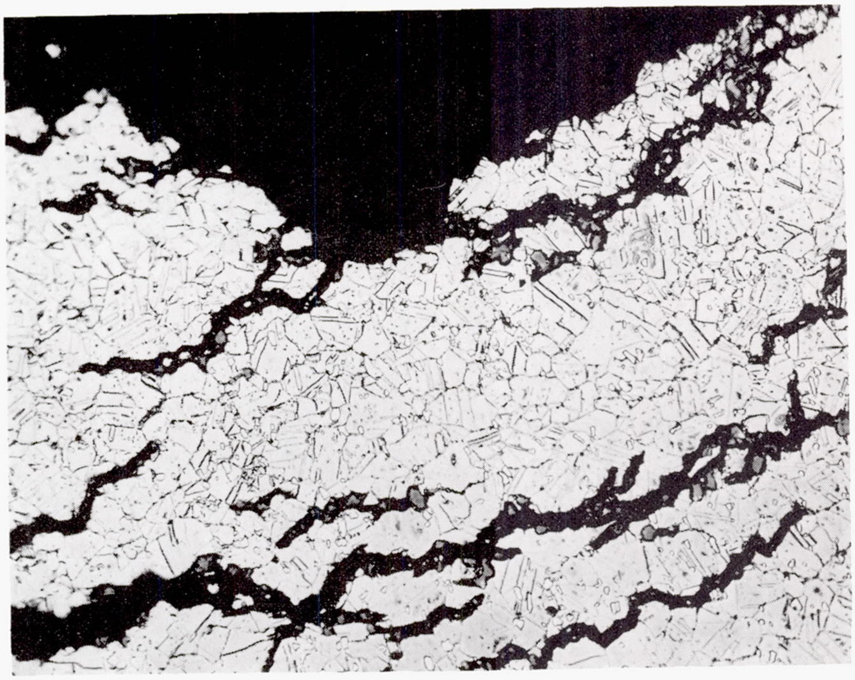


X750

Same area as above. Very thick carbides and very little matrix precipitation.

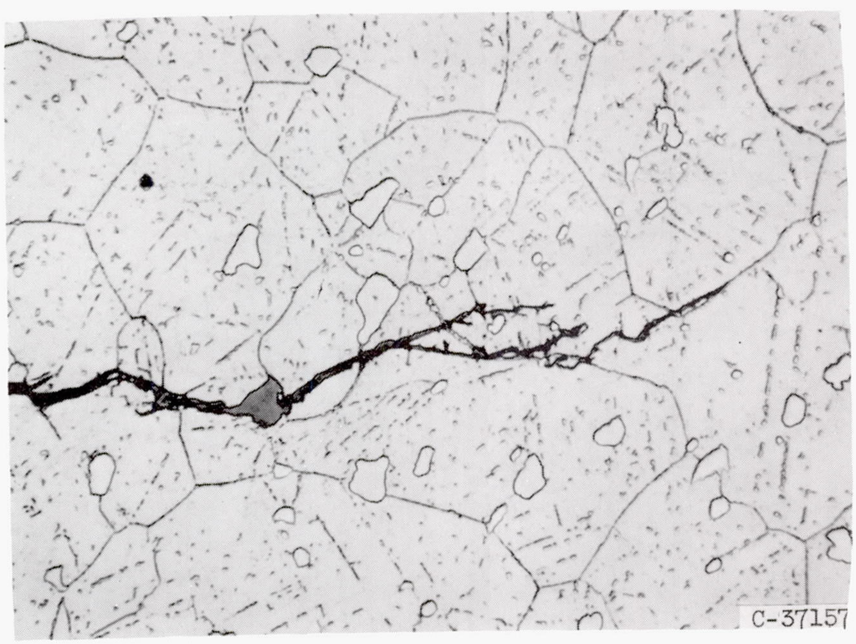
(i) Group 9.

Figure 8. - Concluded. Microstructures of failure origins or areas near failure origins in first and last blade failures.



X250

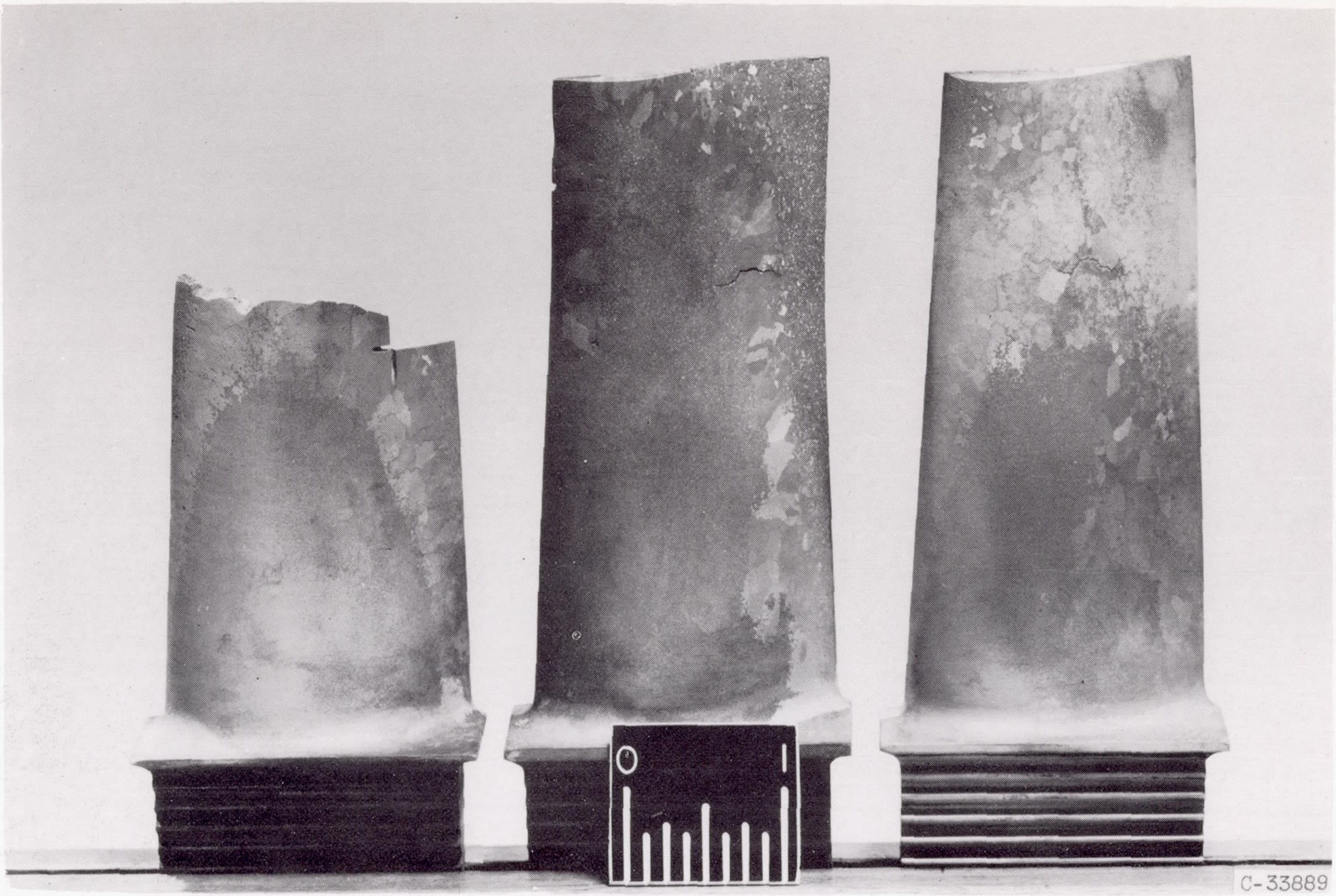
Stress-rupture portion of fracture edge and typical stress-rupture cracks. Etched in 5 percent aqua regia in water electrolytic followed by 10 percent HCl in alcohol electrolytic.



X750

Fatigue cracks which radiated from main fracture edge which failed by stress-rupture followed by fatigue. Etched in 5 percent aqua regia in water electrolytic followed by 10 percent HCl in alcohol electrolytic.

Figure 9. - Typical stress-rupture and fatigue cracks associated with main fracture.

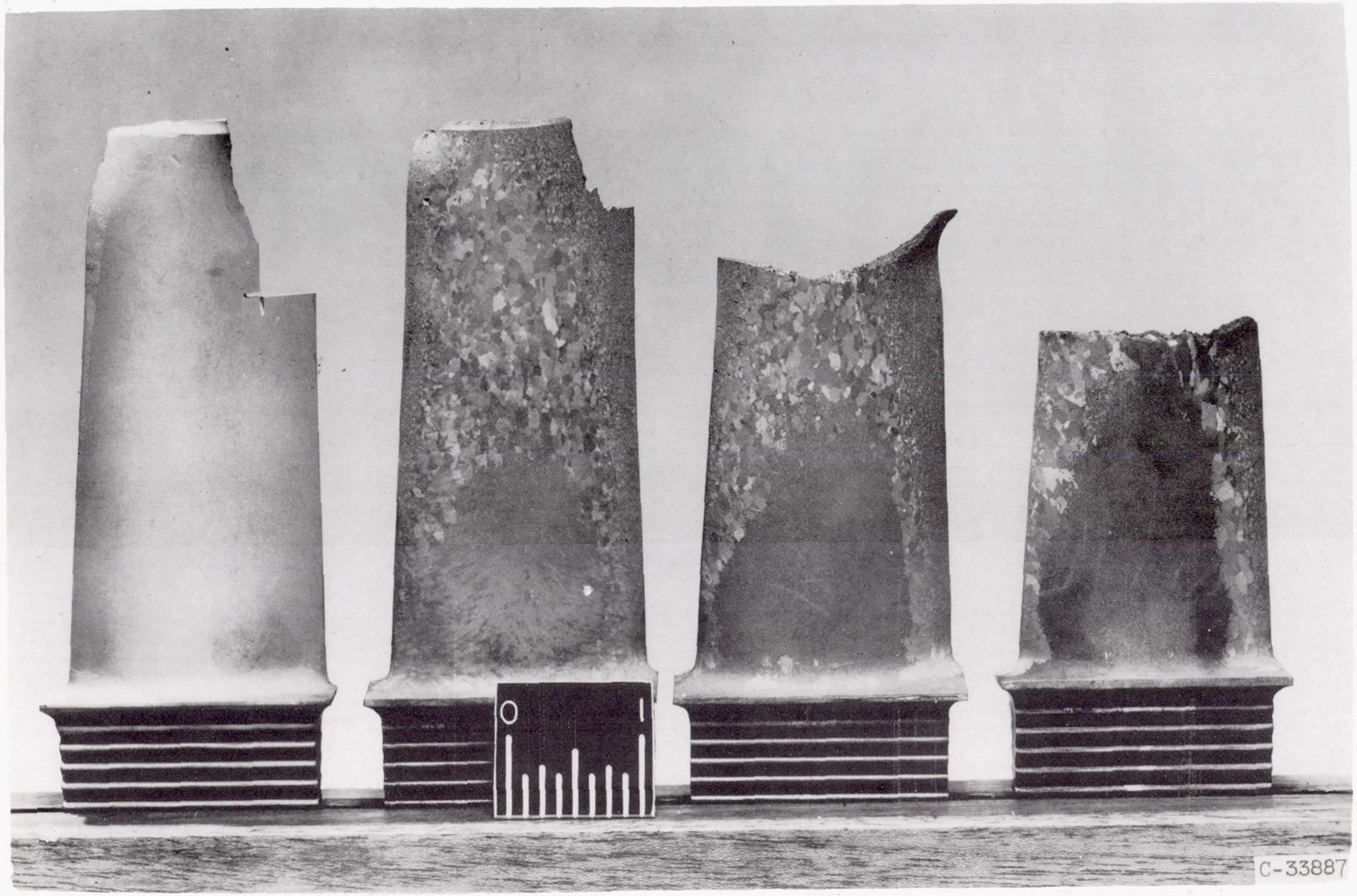


(a) Blade 1. Failed by stress-rupture in 285.17 hours.

(b) Blade 3. Failed by stress-rupture cracking in 142.50 hours.

(c) Blade 6. Failed by stress-rupture cracking in 49.25 hours.

Figure 10. - Blades of group 5 with germinated grains revealed by macroetching.



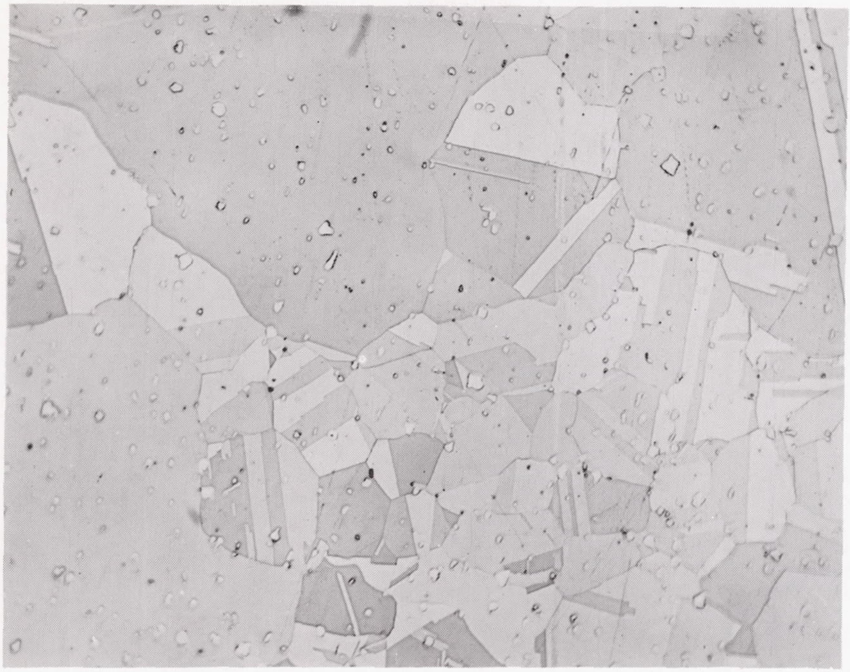
(a) Blade 3. Failed by fatigue in 59.17 hours.

(b) Blade 2. Failed by damage in 93.43 hours.

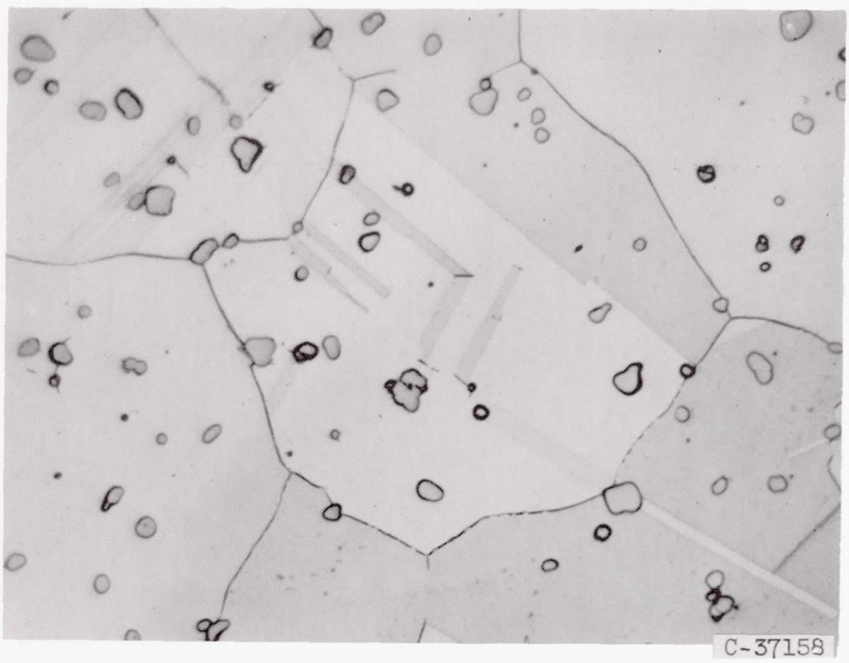
(c) Blade 1. Failed by fatigue in 99.28 hours.

(d) Blade 5. Failed by damage in 209.25 hours.

Figure 11. - Blades of group 6 with germinated grains revealed by macroetching.



X250



X750

Figure 12. - Microstructure of trial specimen solution treated at 2300° F for 4 hours and water quenched. No evidence of melting was observed. Structure appears ideal for a subsequent aging treatment.