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# MATERIAL EXAMINATION OF THE FRACTURED PROPELLER BLADES FROM THE ICEBREAKER C.M.S. CAMSELL

by

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PHYSICAL METALLURGY DIVISION

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## SUMMARY OF RESULTS

The stubs from three fractured propeller blades, which failed on one shaft of the icebreaker C.M.S. Camsell, were examined to determine whether failure resulted from defective material. The blades were cast from a nickel-vanadium steel. Two of the stubs were examined and failure of one blade was attributed to defective material. The material from this blade was found to be brittle, possibly due to a primary grain boundary precipitate, and to contain numerous surface cracks in the area where fracture occurred. The presence of notches in a brittle material would greatly reduce the impact strength of the blades. Failure of the third blade could not be attributed to defective material.

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# INTRODUCTION

The stubs from three fractured propeller blades from the icebreaker C.M.S. Camsell were submitted by the Department of Transport for metallurgical examination to determine whether failure was caused by defective material.

The request for the examination, along with some of the details of the icebreaker and the history of the failure, came in a letter from Mr. J. Rankine Strang, Director, Shipbuilding Branch, Department of Transport, Ottawa, dated September 15, 1960 (Department of Transport's File No. 8906-58). Some of the details of the letters are given below.

The Camsell is a twin-propeller ship with each propeller consisting of four cast nickel-vanadium steel blades bolted to a hub. The general appearance of the blades is shown in Figure 1. The blades under consideration were manufactured at William Kennedy and Sons Limited, Owen Sound, Ontario. The failure was reported to have occurred while the Camsell was in service in the Western Arctic. Three blades from one shaft fractured in a brittle manner at the root of each blade. The stubs from the three fractured blades are shown in Figure 2. The sequence of failure was that one blade failed first and after the propeller had operated for ten days with only three blades, two other blades fractured.

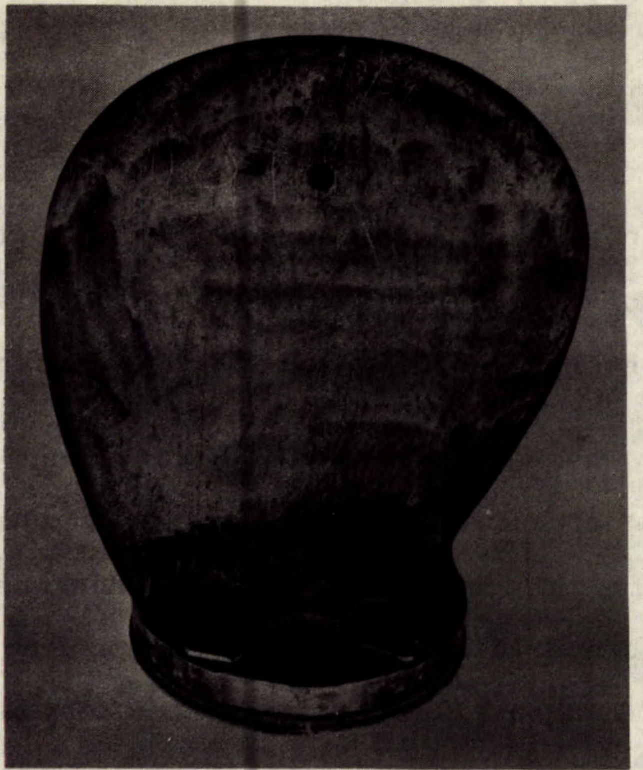


**Figure 1. General appearance of the propeller blades. Grooves on the root of the blade shown are where cracks were chipped out in preparation for repair welding.**

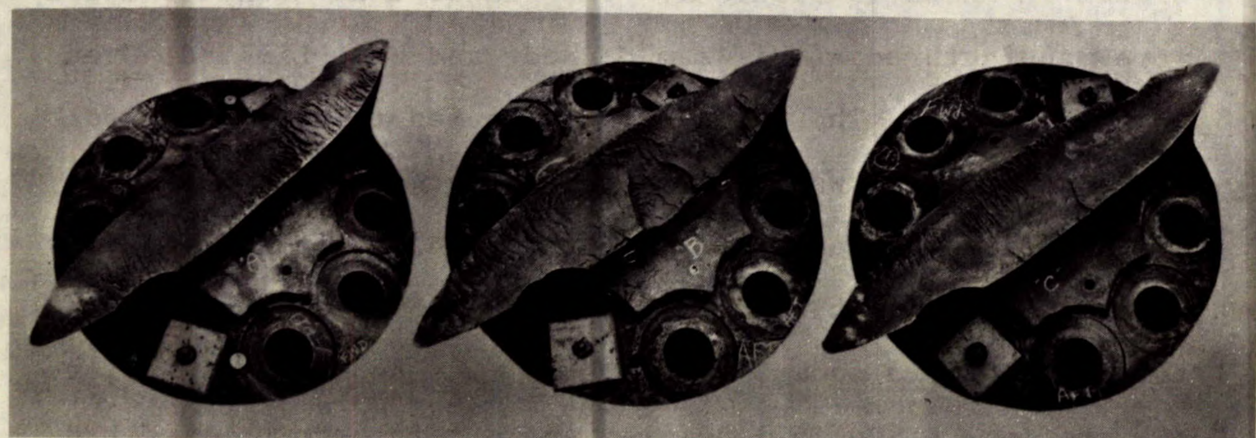


**Figure 2. Fractured stubs submitted for examination. White markings on stubs are cracks which were detected visually and by magnafluxing.**

## PROCEDURE

The three stubs were marked "A", "B" and "C". Only stubs "A" and "B" were examined.

The material examination consisted of the following:

(1) Chemical Analyses

(2) Mechanical Properties

(3) Surface Examination

(4) Metallographic Examination

  (a) Macrostructure

  (b) Microstructure

(5) Fracture Studies.

## RESULTS

### Chemical Analyses

The chemical analyses, along with the chemical specification for this type of blade, are shown in Table 1. The chemical composition was within the specification except for a slightly high vanadium content. The vanadium is added as a grain refiner and to increase the low temperature impact strength. No detrimental effect would be expected as a result of the slightly high vanadium content.

TABLE 1

Chemical Composition

| Element | Blade "A" | Blade "B" | Specification |
|---|---|---|---|
| Carbon | 0.11 | 0.17 | 0.18 |
| Manganese | 0.81 | 0.81 | |
| Silicon | 0.86 | 0.76 | |
| Sulphur | 0.021 | 0.010 | |
| Phosphorus | 0.019 | 0.018 | |
| Nickel | 1.90 | 1.94 | 1.5 (min) |
| Vanadium | 0.15 | 0.20 | 0.10-0.12 |
| Aluminum (chemical) | 0.07 | 0.05 | |
| Aluminum (spec) | 0.23 | 0.10 | |
| Nitrogen | 0.007 | 0.009 | |
| Zirconium (spec) | 0.06 | 0.05 | |
| Titanium (spec) | nil | nil | |

The aluminum content was determined by both chemical and spectrographic methods. The spectrographic analyses were taken on samples (about 1 in. x 1 in.) cut from the centre of the blades. Forty-eight readings were taken on this sample and the average of these readings is shown in the Table 1. The chemical analyses were taken from drilled samples. The spectrographic values were in both cases higher, two times in blade "B" and over three times in blade "A". The large variation could be accounted for by segregation of alumina in the spectrographic samples.

The presence of zirconium indicates deoxidation with a zirconium alloy. This was verified by the manufacturer.

## Mechanical Properties

The tensile and low temperature impact properties are shown in Table 2, along with the specification values. The mechanical test specimens were machined from the centre of the casting just below the fracture surface, with the test specimen axis parallel to the fracture surface and the blade surface. Standard 0.505" diameter tensile specimens and Charpy V-notch impact specimens were used.

The mechanical properties, except for the ultimate tensile strength of Blade "A", did not meet the required specification. The test specimens were taken from the centre of the casting, not from test coupons, and therefore some of the values would be expected to be low. Studies on the effect of mass[1] have shown that the ultimate tensile strength decreased as the mass increased, while the ductility and impact strength decreased only slightly. This could, therefore, account for the low tensile strength of blade "B" and possibly the low elongation of blade "A". However, the elongation value of blade "B" is much lower than can be accounted for by mass effect.

TABLE 2

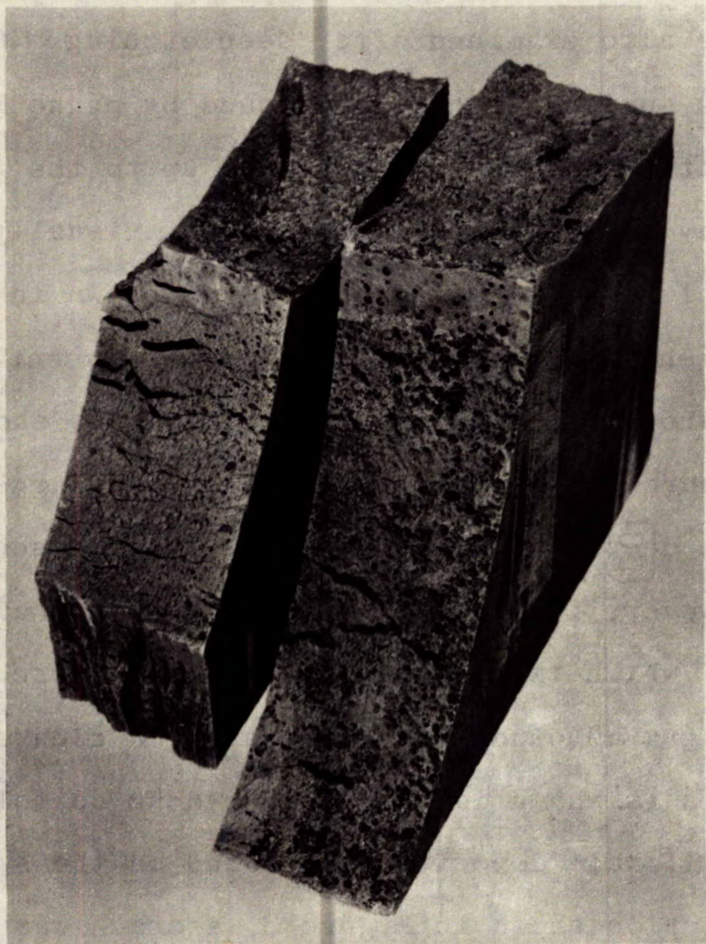
Mechanical Properties

| Property | Specification | Blade "A" | | | | Blade "B" | | | |
|---|---|---|---|---|---|---|---|---|---|
| | | | | | Average | | | | Average |
| U.T.S. (psi) | 75,000 | 77,500 | | 77,200 | 77,350 | 68,600 | | 73,600 | 71,100 |
| Yield Strength (psi) | not spec | 54,400 | | 54,700 | 54,550 | 54,900 | | 54,800 | 54,850 |
| Elongation (%) | 25 | 19.0 | | 18.5 | 18.8 | 4.0 | | 8.0 | 6.0 |
| Reduction in Area (%) | not spec. | 27.1 | | 22.6 | 24.9 | 13.9 | | 17.4 | 15.7 |
| Bend (degrees) | 160 | - | | - | - | - | | - | - |
| Impact Strength | not spec. | | | | | | | | |
| Charpy V-notch (ft-lb) room temp | | 32, | 32, | 30 | 31 | 20, | 32, | 26 | 26 |
| 32°F | | 19, | 19, | 15 | 17 | 20, | 20, | 12 | 17 |
| 0°F | | 11, | 6, | 11 | 9 | 14, | 15, | 9 | 13 |
| -25°F | | 9, | 14, | 8 | 10 | 2, | 2, | 2 | 2 |
| -50°F | | 2, | 2, | 2 | 2 | | | | |

## Surface Examination

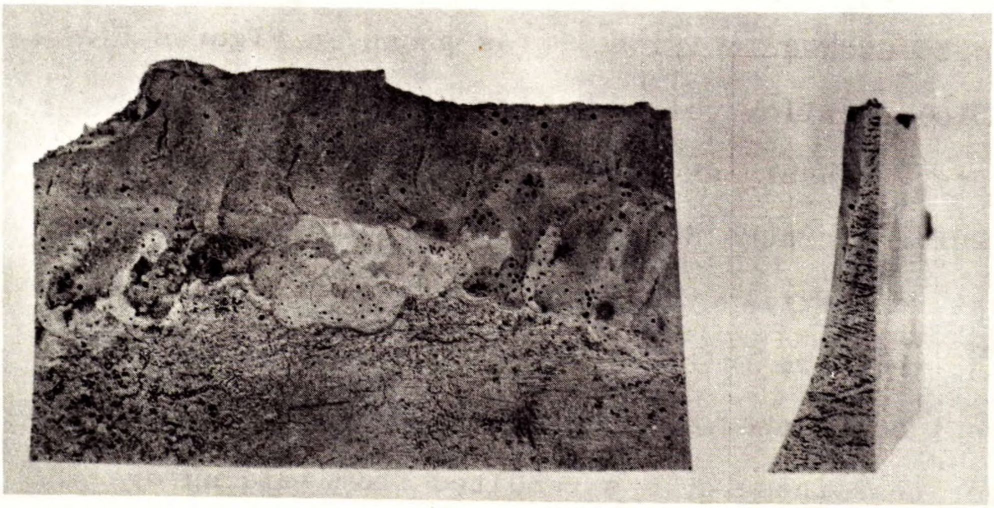
The surfaces of the two blades investigated were examined visually by magnetic particle technique. The surfaces were also examined after deep etching. On stub "B", numerous surface cracks were found by magnetic particle examination in the root area and up to the fracture surface. Many of the cracks were evident visually. A typical example of the surface cracks is shown in Figure 3. The surfaces shown are both sides of a cross-section taken at the fracture. The top edge of the sections shown are the fracture. The cracks are slightly enlarged as a result of the deep etch treatment given to the sections. The etch, however, indicated the presence of a hairline crack network which is probably the result of etching out of a primary grain boundary precipitate. A clearer indication of this is shown later. Some cracks do follow the network. Small weld deposits were found on the surface, and in the section shown in Figure 3, a crack was found in a weld deposit.

In stub "A", no cracks and only a slight indication of grain boundary network were found on the surface. A section of the surface, after etching, is shown in Figure 4. Again, a weld deposit was found at the fracture but, in this case, only one large deposit was found on each side.



Deep Etched 1:1 HCL X3/4

Figure 3. Section taken from stub "B" showing opposite surface of the casting. Top surface shown in the photograph is the fracture.



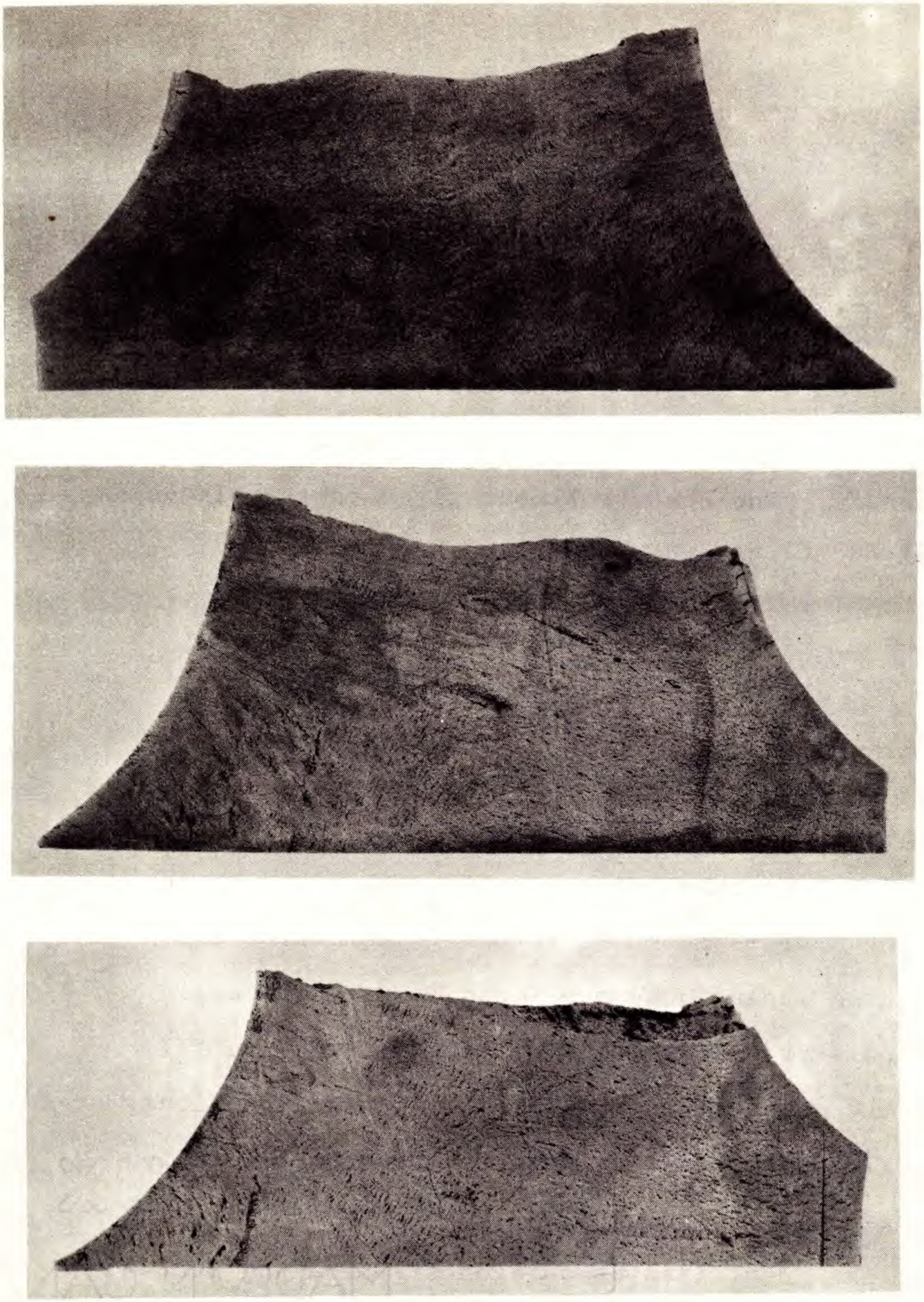
Deep Etched 1:1 HCl X3/4

Figure 4. Section taken from stub "A" showing the surface (photograph on left) of the casting at the fracture. Top edge is the fracture. Photograph on right is a view of the edge of the same section.

## Metallographic Examination

(a) Macroscopic

The macrostructures from three different cross-sections of stub "B" are shown in Figure 5. The fracture surface intersects the prepared surface at the top of each photograph. Each cross-section was deep-etched in 1:1 HCl. As shown in Figures 5(a) and 5(b), shallow weld deposits were found on the surface, probably as a result of repairing some of the surface cracks during the manufacturing. Many cracks were still evident on this stub. It was also found, as shown in Figure 5, that some cracks were present below the weld deposit which would indicate either that these cracks resulted from welding or that the cracks were not chipped out deep enough prior to repair welding. The latter is more probable, as the welding tests of the material from stub "B" resulted in no cracks being formed. A hairline crack network was found on the macrostructure which, as reported by Lorig and Elsea[3] is indicative of a primary grain boundary precipitate. This network can cause brittleness in the material and result in an intergranular (rock candy) fracture. The fracture studies described later verify the presence of some grain boundary precipitate.



Deep Etched 1:1 HCl X$\frac{1}{2}$

Figure 5. Macrostructure of cross-sections taken from stub "B". Top surface of each cross-section is the fracture surface.

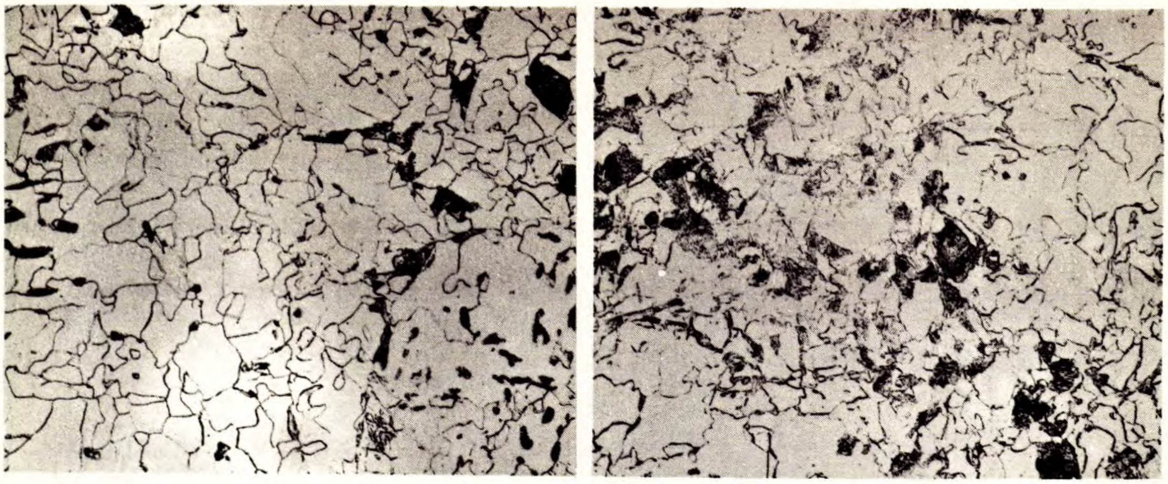
None of the above defects were found in the macrostructure of stub "A". The depth of the large weld deposit on this stub is shown in Figure 4.

(b) Microstructure

The general microstructure of the material taken from the centre of the casting is shown in Figure 6. The structure is typical of an annealed structure. The heat treatment given to the blades as supplied by the manufacture is:

> Anneal 1650°F, 5 hours and furnace cool. After grinding and repair welding, reanneal 1650°F, 8 hours, and furnace cool.

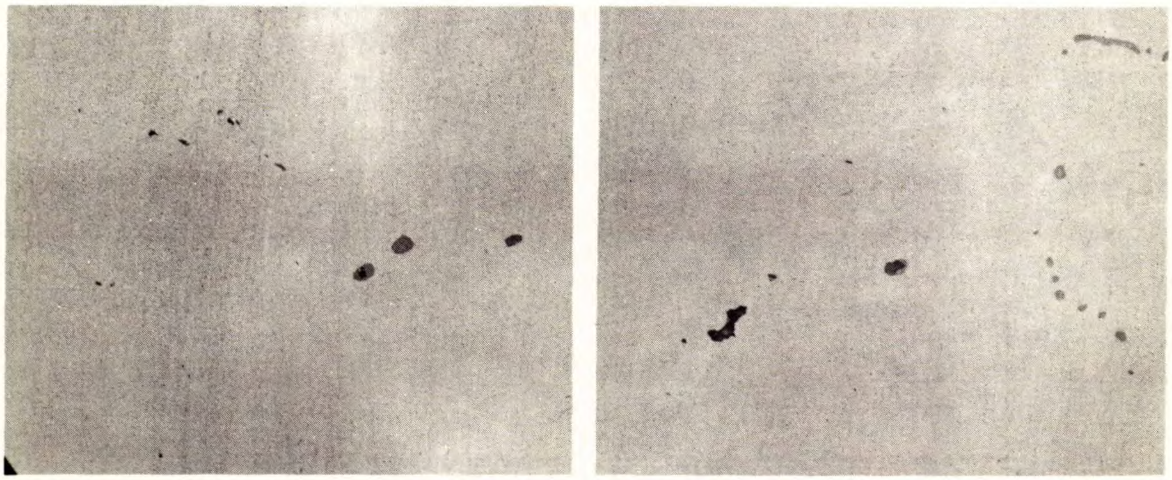
The unetched microstructure showing the inclusion type is shown in Figure 7. Sulphide and zirconium inclusions were found. The sulphide inclusions were mainly high melting point, Type III. Near the surface, some low melting point, injurious, Type II inclusions were found. The other inclusion was orange in colour and by the etch tests[2], part of it was identified as an iron sulphide and part as either zirconium sulphide or zirconium nitride.



Nital etch X250 Nital etch X250

Blade "A" Blade "B"

Figure 6. General microstructure of material.



Unetched X250 Unetched X250

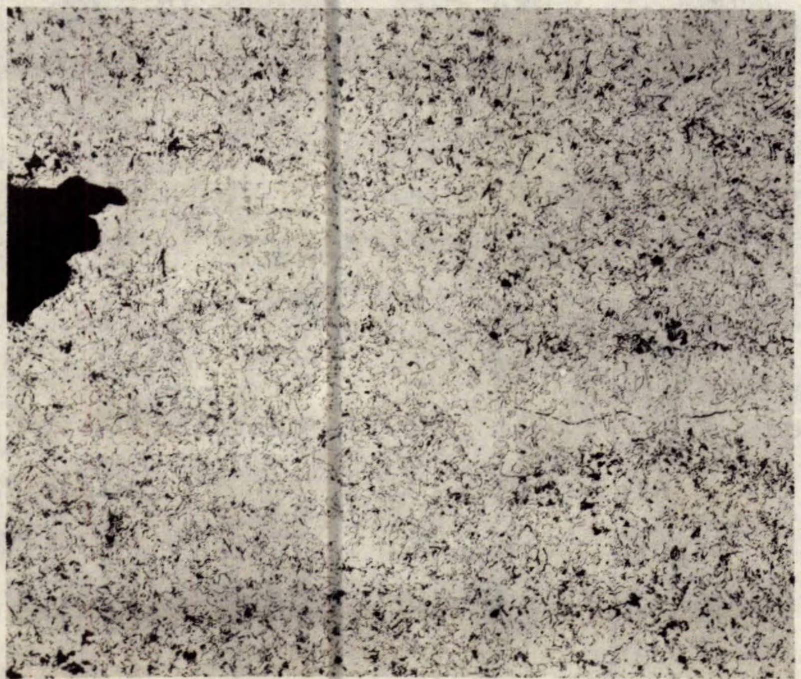
Blade "A" Blade "B"

Figure 7. General Inclusion Type.

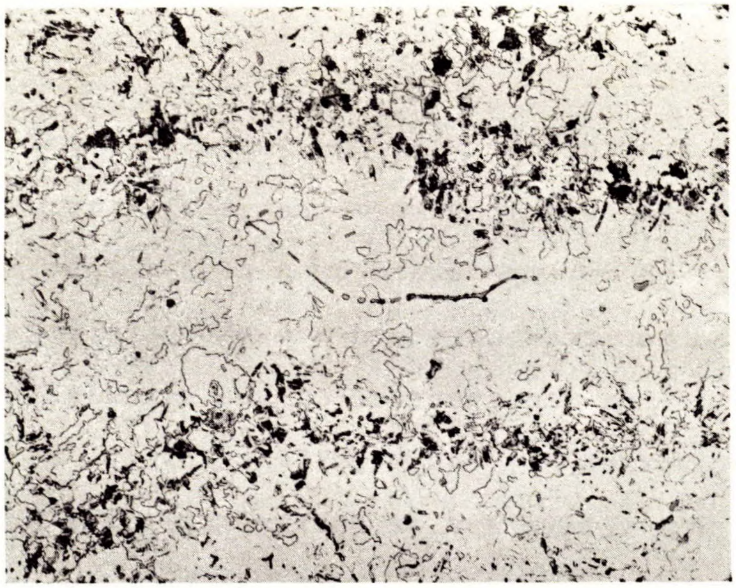
The microstructure of the material below one of the surface cracks of stub "B" is shown in Figures 8 and 9. The crack appeared to extend into a decarburized zone. In the decarburized zone, as shown in Figure 8, there were Type II sulphides.

A careful microscopic investigation was undertaken to find a grain boundary precipitate which caused the network formed on the macroetches. In Figure 10, a fine chain-like precipitate was found, possibly located at the primary grain boundary. The precipitate was too fine to identify. Lorig and Elsea[3] and Woodfine and Quarrell[4] have associated precipitates of this type with intergranular fracture.

crack

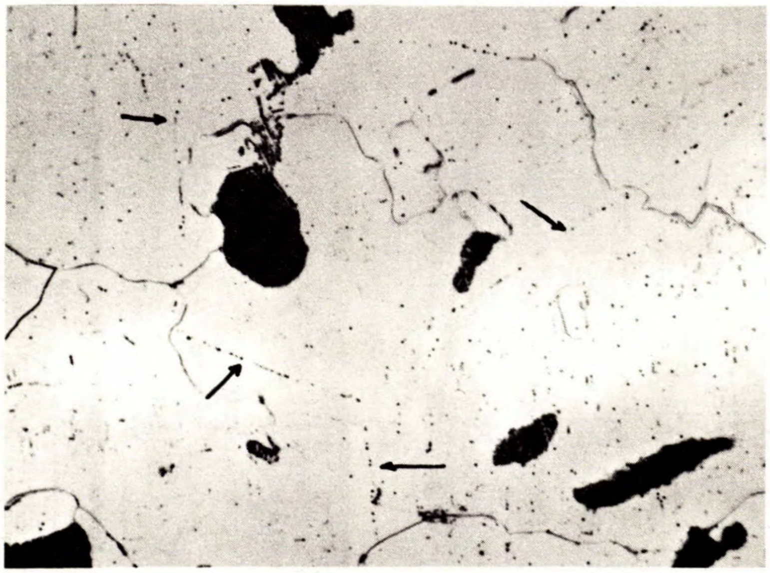


Nital etch X40
Figure 8. Microstructure at the base of one of the surface cracks.



Nital etch X100

Figure 9. Magnified section of decarburized zone extending below the crack shown in Figure 8.

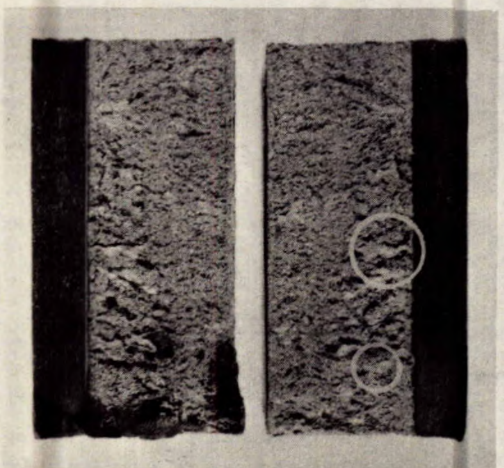


Picral etch X1000

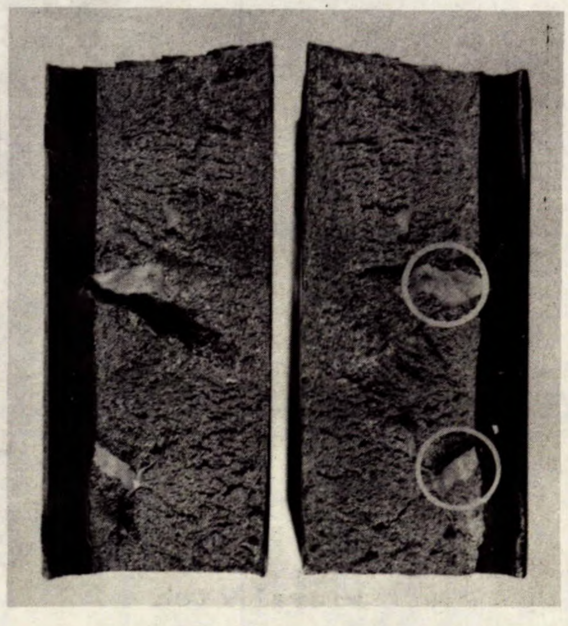
Figure 10. A fine precipitate (marked by arrows) possibly at primary grain boundaries.

## Fracture Studies

Rectangular test specimens of about $1\frac{1}{2}$" x 3" in cross-section were cut from each stub, notched and fractured at room temperature by a drop-weight apparatus. The fracture surfaces obtained are shown in Figure 11. About 5 to 10% of the fracture of the material from stub "B" was intergranular, and much less for the material from stub "A". The intergranular surfaces are the shiny parts of the fracture shown in the photographs. In stub "B" each intergranular surface was much larger in area than stub "A", which indicates a larger grain size. An intergranular fracture is often referred to as a "rock candy" fracture.



X3/4
Stub "A"



X3/4
Stub "B"

Figure 11. Fractured surfaces of the test specimens taken from stub "A" and stub "B". Specimen fractured at room temperature. The top edge in both photographs is the fracture surface which occurred in service. Examples of intergranular fracture are shown within circles.

## DISCUSSION

The low tensile ductility and low impact strength of the material from blade "B" are indications of brittleness. The cause of this brittleness is possibly due to a precipitate at the primary austenite grain boundary. Fracture in this type of material is sometimes intergranular and is then described as "rock candy".

Lorig and Elsea[3] and Woodfine and Quarrell[4] have reported that intergranular fracture is principally associated with aluminium nitride precipitated at the primary grain boundaries, although Lorig and Elsea consider any primary grain boundary constituent, such as ferrite or carbide networks or Type II sulphides, can cause intergranular failure.

The best indication of the existence of a primary grain boundary precipitate is the hairline crack network evident on the deep etches. This, as shown by Lorig and Elsea, is associated with intergranular fracture. They, however, reported that when a network is evident on the deep etches, the fracture must be at least 50% intergranular. In this material the fracture studies indicate only about 5-10% intergranular. The existence of a chain-like precipitate found on the metallographic examination also is evidence of a primary grain boundary precipitate.

The chemical composition of the blade suggests the possibility of a primary grain boundary precipitate. The aluminum content is high enough to expect an aluminium nitride precipitate. The zirconium was probably added to prevent the formation of aluminium nitride. This, however, as shown by Woodfine and Quarrell is not effective when zirconium and aluminium are added together. From thermodynamic considerations, zirconium oxide is more stable than aluminium oxide, and when zirconium and aluminium are added, the oxygen combines with the zirconium leaving the aluminium to combine with the nitrogen. Zirconium by itself is reported to reduce intergranular failure as do titanium and the combination of titanium and aluminium(4).

The precipitation of aluminium nitride is reported by Woodfine and Quarrell to occur by slow cooling below 1150°C (2100°F). The heat treatment for these blades requires a long holding period at 910°C (1670°F) and, therefore, the heat treatment itself would be expected to cause some aluminium nitride precipitation.

The amount of intergranular precipitate was felt not to be sufficient to cause failure by itself. If however, notches were present in a somewhat brittle material, the impact strength would be greatly reduced. The failure of blade "B" can, therefore, be attributed to the combination of the two effects.

The cracks in some cases were found to follow the hairline crack network. This, however, does not necessarily mean that the cracks resulted from the primary grain boundary precipitate which is believed to cause the network. The fact that decarburization was associated with several cracks suggests hot tearing which also would occur at the primary grain boundary but above the temperature at which the precipitation of aluminium nitride would occur. If the aluminium nitride precipitate caused the cracking, cracking would have to occur below 1150°C (2100°F) and less decarburization would be expected.

The material of the other blade investigated did not contain any of the above defects. The impact and tensile ductility were high and no pronounced indication of an intergranular fracture was found. Failure of this blade could not be attributed to defective material.

## CONCLUSIONS

From the investigation undertaken on two of three fractured stubs, the following conclusions regarding the material and the possible causes of failure can be made.

1. The material was within the chemical specification laid down for the production of these blades, except for slightly high vanadium

content. The high vanadium content would not be expected to have any detrimental effects.

2. The mechanical properties of the material, determined from test specimens taken from the casting, were below those permitted by specification, except for the ultimate tensile strength of the material from stub "A". The low mechanical properties of the material from stub "A" may be accounted for by mass effect (test specimens taken from a large section, rather than from small test coupons). The mechanical properties of the material from stub "B", however, cannot be accounted for by mass effect, especially the elongation. The mechanical properties indicate a brittle material.

3. Numerous surface cracks were found on stub "B" in the root area and up to the fracture surface. Some weld deposits were found in this area and cracks were found below the welds. In stub "A" no surface or sub-surface cracks were found, but in this area a large weld deposit was found.

4. The brittleness was believed to be caused by a primary grain boundary precipitate.

5. Failure of blade "B" was the result of defective material, i.e., the presence of notches (surface crack) in a brittle material. Neither defect by itself would be likely to cause failure.

6. The failure of the third blade could not be attributed to defective material.

TSP/ls

## REFERENCES

1. Steel Casting Handbook, Editor Charles W. Briggs, Steel Founders Society of America, 3rd edition, 1960.

2. Metals Handbook, 1948 edition, American Society for Metal, p 449.

3. C.H. Lorig and A.R. Elsea, Occurrence of Intergranular Fracture in Cast Steels, Trans. AFS, vol 55, 1947, p 160.

4. B.C. Woodfine, S.G. Quarrell, Effect of Al and N on the Occurrence of Intergranular Fracture in Steel Castings, Journal of the Iron and Steel Institute, August 1960, p. 409.