AD-A061 397 NAVAL RESEARCH LAB WASHINGTON D C F/G 11/6
A STUDY OF FABRICATION CHARACTERISTICS OF 3 PER CENT ALUMINUM-5--ETC(U)
SEP 50 K B LLOYD, E J CHAPIN

UNCLASSIFIED NRL-MR-41 SBIE-AD-E000 253 NL

| OF |
ADA
061397

END
DATE
FILMED
2-79
DDC

LEVEL II



# NRL MEMORANDUM REPORT No. 41

## A STUDY OF FABRICATION CHARACTERISTICS OF 3% ALUMINUM - 5% CHROMIUM LOW CARBON TITANIUM ALLOY

Kenneth B. Lloyd
Edward J. Chapin

METALLURGY DIVISION



NAVAL RESEARCH LABORATORY, WASHINGTON, D.C.

# A STUDY OF FABRICATION CHARACTERISTICS OF 3% ALUMINUM- 5% CHROMIUM LOW CARBON TITANIUM ALLOY

by

KENNETH B. LLOYD

EDWARD J. CHAPIN



Nonferrous Alloys Branch
Metallurgy Division
Naval Research Laboratory
Washington 25, D.C.

CONTENTS

## LIST OF FIGURES

## DISTRIBUTION

| | |
|---|---|
| BuShips (Code 692) | (1) |
| BuAer | |
| Attn: Harry Boertzel | (10) |
| BuOrd | (1) |
| ONR | (1) |
| Dir., USNEES, Annapolis, Md. | (1) |
| Cdr., Naval Proving Ground, Dahlgren | |
| A & T Lab. | (1) |
| CG, Watertown Arsenal, Watertown 72, Mass. | |
| Attn: ORDBE - L | (1) |

## ABSTRACT

A 3% aluminum - 5% chromium - low carbon - titanium alloy was studied to determine effects of hot and cold rolling and annealing on microstructure, on directionality, and on recrystallization and grain growth. Results of this work indicate that this alloy has unusual recrystallization characteristics which prevent reduction of grain size by the usual methods of deformation and heat treatment. The desired improvement in ductility did not materialize from the treatments considered. Heating this alloy in air to the temperatures necessary for true hot working produces serious contamination and affects the properties.


## PROBLEM STATUS

This is a final report on this alloy study; no further work will be done on this alloy unless the Laboratory is otherwise notified by the Bureau of Aeronautics.

## AUTHORIZATION

NRL Problem M01-30
TED NRL AE411007
RDB Form 1A Index No. NA581029
Reference BuAer ltr AER-AE-41 serial 74263 dated 21 Sept 1950

## I. Introduction

This report presents the data obtained from an investigation made for the Bureau of Aeronautics on a 3% aluminum, 5% chromium, low carbon titanium alloy. An interim letter report serial No. 3500-241A/51 ehb (3510) dated 7 May 1951 was prepared on preliminary work. The work covered in the interim report is also included in this report.

The alloy was developed by P.R. Mallory & Co., under a BuAer contract, and this company's continued study of its fabrication and properties has been reported in their bi-monthly reports.[1] The development work indicated that it has desirable physical properties and offers attractive possibilities for use as sheet or plate. The ductility of the alloy, however, is somewhat below that considered necessary for satisfactory formability. Since all of the work done on this alloy by P.R. Mallory & Co. had been with carbon bearing material, it was considered desirable to investigate the possibilities of obtaining further improvement in ductility on relatively carbon free material.

The New York University Research Division is studying the low aluminum - low chromium corner of the Al-Cr-Ti ternary system.[2] The recrystallization characteristics of the 3% aluminum - 5% chromium -titanium alloy, however, have not been discussed by either Mallory or the NYU group.

The purpose of this investigation is to determine the effects of hot and cold rolling and annealing on: 1) microstructure, 2) variation in directionality of properties, 3) recrystallization and grain growth.

## II. Preparation of Material

Material for this investigation was supplied by the Bureau of Aeronautics in the form of eight sheets 8" x 15" x .080". These had been prepared from a double arc melted ingot made at Battelle Memorial Institute.

According to information supplied by P.R. Mallory & Co., of Indianapolis, the sheet was prepared in the following manner: the ingot was forged in the temperature range of 925°--970°C into a 3/8" x 3" x 59" plate with a total reduction in area of 90%. The 3-3/4" diameter ingot was first forged to a 2" square and then finished to the plate. Four reheatings on each end of the ingot were required for forging. The plate was cut into 8" lengths and hot rolled at 850°C in a transverse direction to the forging direction. Reductions of .050" per pass were taken to reduce the plate to .200" thick. Reductions of .030--.035" per pass were taken to reduce the .200" thick sheet to .100" thick. The final passes were .010" reductions to the finished thickness. Approximately ten reheatings were required with one reduction per reheat. The sheet was air cooled after the final reduction. Surface scale was removed by sandblasting followed by pickling in a nitric-hydrochloric acid solution.

Metallographic examination of the sheet material as received showed that a surface layer about .003" thick was still present on both sides of each sheet. These layers were not removed from test specimens except from some used for rolling at room temperature and at 200°C.

In order to determine the extent of oxygen, nitrogen, and hydrogen contamination existing in the sheet, a strip 3/4" wide was ground down in thickness in steps of .005" each. Each section thus reduced was analyzed by vacuum fusion methods[3] developed at the Naval Research Laboratory.

Tensile test specimens were made according to ASTM Standard Method E-8-46 and measured 8" overall x 1/2" wide along the 2" gauge length. Duplicate tensile specimens, transverse and longitudinal to the rolling direction, and corresponding bend test specimens, in quadruplicate, were cut from the as received material. Bend test specimens measured 1" x 4" and the cold rolling experiments were also conducted with this size sample. For the hot rolling experiments, pieces 2" x 8" were used, from which one tensile and two bend test specimens were later cut.

## III. Experimental Work

The sheets as received were each examined radiographically to detect possible alloy segregation and tungsten inclusions. A metallographic examination was made on specimens cut from each sheet to determine the microstructure and the grain size.

Reduction of the grain size was attempted by cold rolling followed by heat treatment, the heat treated microstructures being retained by quenching the specimens into silicone diffusion pump oil from the heating zone in a high vacuum quenching furnace (Fig. 1). In the further attempt to produce a grain refinement sheet samples were rolled, some of which had been quenched a second time from 1000°C and some of which had been fully annealed by slow furnace cooling from 1000°C. Metallographic examination was made of the worked and heat treated specimens to determine changes in microstructures.

Sheet samples in the as received condition and those whose surfaces had been sandblasted to fully remove surface layers were reduced 10 and 20% in thickness at room temperature and at 200°C to test a possibility of greater plasticity under "warm" rolling conditions. Samples were heated in air to temperatures of 700, 800, 900 and 1000°C in an electric muffle furnace and reduced 10 and 20% in single passes and allowed to air cool. Another series of samples were similarly rolled at 600, 650, 675 and 700°C. Tensile and bend test specimens were machined from the rolled strips and a metallographic examination was made of cross sectional areas of the strips.

Bend tests were made in the 'constant moment bend jig', (Fig. 2), to determine changes in bendability, and tensile tests were made on a 60,000# Baldwin Southwark testing machine. Rockwell A hardness tests were made on all rolled samples.

To assist in the study of the recrystallization characteristics, a series of 8 binary alloy buttons, each consisting of 1, 2 and 3% aluminum content, and 1, 2, 3, 4 and 5% chromium content; a commercial titanium sponge button; and 2 buttons of 3% aluminum-5% chromium-titanium alloy were prepared by arc-melting Dupont titanium sponge and commercial high purity alloying metals in a water-cooled copper crucible, using a tungsten electrode in a helium atmosphere. The buttons were turned up on edge and each remelted several times to assure thorough mixing. 1/2" square slices about 1-1/4" long sawed from the buttons, were cold rolled to .4" square in grooved rolls; and after annealing at 800°C in vacuum, these pieces were further cold rolled to .3" square bars. Sections of these 30% cold reduced bars were heat treated and quenched in the vacuum furnace, and a metallographic examination made to determine the occurrence and extent of recrystallization. Rockwell A hardness tests were also made at the various stages of preparation and heat treatment.

Manual polishing techniques were used in the preparation of the metallographic specimens. Specimens exhibiting the acicular transformed quenched structure and the 'veining' network were electropolished as well; the examination of these specimens verified that no structure variations had been introduced by the manual polishing.

The measurement of low order ductility in sheet materials has not been formalized. To determine values of elongations up to 5% in the outer fibers, Pinto[4] recommends bending around a template of varying radius. The ASTM Standard Method[5] outlines a bending test for ductility, measuring fiber elongation, which gives reproducibility within 15% on plate of 1/4" and over; this method emphasizes that local stress or strain be avoided in the bending section. The use of a "Constant Moment Bend Jig" developed by the Chance Vought Aircraft Company[6] which produces a free bend in the specimen, was suggested by the Bureau of Aeronautics. The design was modified for this problem and adapted for use on sheet materials, and a jig constructed, (Fig. 2). The radius of curvature of the bent specimen can be computed from measurement of the height of the bend on a measuring device, (Fig. 3), and the strain in the outer fiber computed from the radius of curvature and the sheet thickness, using formulas derived by Chance Vought[6] and by Lubahn and Sachs.[7] The measuring and strain evaluation calculation methods are given in the Appendix.

Elongation values determined by this bend test were compared to the elongation measured on the stress-strain recorder charts by the corresponding tensile test specimens prepared from the titanium alloy sheet because of the inability to obtain other low ductility materials. The elongation values correlate well and the use of the bend test jig allows in addition to evaluating bendability, a more rapid and a more economical method of determining elongation than provided by tensile testing.

IV. Results

1. No detectable alloy segregation or tungsten inclusions were found in the radiographic examination.

2. An abnormally large grain size of about 15 grains per square millimeter was found in the as received sheets (Fig. 4). An oxygen-rich layer of about .003" thick remained on the surfaces of the sheets, (Fig. 5), which tends to diffuse inward on any heating to 800°C or higher (Fig. 6). The alloy has a Tukon Knoop hardness of 370, (Fig. 5), which increases to 400 upon quenching from 1000°C (Fig. 7).

3. The sheets have a total average content of .32% Oxygen, .023% Nitrogen and .018% Hydrogen; but with .010" surface thickness removed from each side, the Oxygen content is .115%, Nitrogen .013% and Hydrogen .018%. On this basis, an oxygen-rich layer .003" thick would have an Oxygen content of over 2%.

4. The normal structure of the hot rolled titanium alloy sheet is large grained; it is a two phase structure, with tiny particles of alpha titanium solid solution oriented on crystallographic planes, in the matrix of transformed beta titanium solid solution (Fig. 8). On heating, the particles enlarge to a maximum size and amount at about 875°C (Figs. 9 thru 12); on further heating they decrease in size, (Fig. 13). In quenching from 950°C or higher, an acicular transformed structure is produced, (Figs. 14 & 15). The same structure changes were found also in heating previously quenched samples, (Figs. 16 thru 21), and in heating 20% cold reduced sheet samples, (Figs. 22 thru 27). Sheet reduced 20% by cold rolling showed no recrystallization on heating to temperatures up to 900°C. On heating to 950°C or over, grain growth occurred, (Fig. 6).

5. Sheet first quenched from 1000°C, reheated to 950°C or higher, and again quenched, exhibited a fine network within the large grains, (Fig. 15). X-ray back reflection photographs indicated this network to be a 'veining' within the grains, rather than any re-orientation or grain break-up. Alloy sheet in this condition was too brittle to allow any reduction by rolling.

6. Sheet fully annealed by furnace cooling at the rate of 1°C per minute from 1000°C to 700°C, showed a laminar structure of massive alpha titanium layers in the remaining beta matrix (Fig. 28). It developed surface and edge cracks when cold rolled, and fractured in the bend test. Heat treatment of the annealed sheet, with and without cold deformation, produced a transformation of the alpha solid solution back to beta (Fig. 29), in an amount corresponding to the temperature attained. No recrystallization was found. A dark etching phase, (Fig. 30), developed at 500°C in the beta matrix; this phase persisted up to 700°C, but was no longer evident at 730°C.

7. The outer fiber strain measured in the bend test correlated well with the plastic strain shown at fracture on the stress-strain recorder charts by the corresponding tensile test samples. As received alloy sheet showed an average of 2.3% strain in both test methods. Duplicate sets of tensile specimens, cut transverse and longitudinal to the rolling direction, indicated strengths of 152,000 and 160,000 psi respectively; the corresponding bend test samples did not show this directionality. A slight increase in ductility from 2.3% to 2.8% strain was indicated for sheet rolled to 10% reduction at 700°C. Sheet rolled at 800°C increased in strength from 160,000 psi to 190,000 psi, but the ductility decreased to zero. The sheet rolled at 900°C showed no change in ductility over that as received but the sample rolled at 1000°C was quite brittle, fracturing in the bend test.

8. A sheet sample reduced 10% by cold rolling showed no change in ductility. Cold rolling to 20% reduction decreased the strain measurement from 2.3% to 1%. Sheet 'warm rolled' at 200°C to 10% and 20% reductions also showed a distinct loss of ductility and fractured in the bend test (Fig. 34). In testing this series of samples, noise similar to "tin cry" was heard during the bending.

9. Rockwell A hardness was found to be relatively insensitive to the thermal and mechanical treatments given the sheet. The hardness readings averaged 71; with multiple readings on some samples extending over a 2 point range. The variation attributable to treatment would also cover this range.

10. Metallographic examination of the aluminum and chromium binary alloy series revealed that the presence of alpha titanium solid solution inhibited recrystallization and grain growth of the beta matrix (Fig. 31). When the metal transformed entirely to the beta phase, recrystallization and accelerated grain growth took place (Fig. 32). The Rockwell A hardness readings, tabulated with the microstructures in Fig. 35, indicated hardening by cold working of the alpha phase unalloyed titanium and aluminum-titanium binary alloys. These alpha phase alloys softened on heating to 800°C or higher and recrystallized. The chromium-titanium binary alloys were hardened by quenching of the beta solid solution. The ternary 3% Al - 5% Cr - Ti alloy prepared showed no response to cold working or heat treatment with a Rockwell A hardness of 68-69 for all readings.

11. No oxidation of the exposed edges of the sheet was found in the pieces for rolling that were heated to 700°C in air for ten minute periods (Fig. 33). Heating to higher temperatures however, did produce an oxidized layer in the ten minute exposure.

## V. Discussion of Results

The sheet material supplied was not flat, each piece being curled up at the ends. The surface treatment of the sheets by sandblasting and acid pickling was not uniform; each sheet varied in thickness from .075"

to .085", and the amount of the oxidation layer removed was erratic as shown by the metallographic evidence. When specimens were heated in the vacuum quenching furnace, an especially large evolution of gas was noted at about 700°C. It is suspected that this gas is hydrogen, which resulted from the acid pickling treatment.

The original aim of this problem had been primarily to correlate the effects of hot and cold rolling procedures and annealing treatments with the crystal orientations produced; however, the work became that of attempting to reduce the as received grain size of the sheet to one sufficiently small so that preferred orientation studies could be made before and after rolling and annealing. (The as received grain size would require the construction of an integrating x-ray camera which uses large area transmission samples and photographic recording. The available equipment uses reflection samples and Geiger counter recording and is limited to material of very small grain size.) Although an apparent effect of grain size refinement by veining, see Fig. 15, was found in sheet that was reheated to the beta region and quenched after a first quenching from the beta phase, no recrystallization or grain refinement was found from any combination of cold deformation and heat treatment.

The veining effect has been found in several other metals and alloys, and has been discussed by many writers.[8-14] In this study, the back reflection x-ray photographs of the veined structure did not indicate any reorientation within the large grains. Samples in the veined condition slowly cooled from 1000°C through the transformation range produced microstructures identical with those similarly cooled but not previously quenched. The requirement for veining in this alloy appears to be dependent on internal strains induced by double quenching from above the transformation range.

In studying the quenched specimens from various heat treatments, the mechanism of the changes in the microstructures became apparent. The as received structure (Fig. 8), although it has the appearance of a single phase, is actually a two phase alloy with tiny particles of alpha titanium solid solution oriented on crystallographic planes in the transformed beta titanium solid solution matrix in each large grain. The transformed beta matrix (sometimes called alpha prime in the literature) has the needle like appearance associated with titanium quenched from the beta phase, but in this alloy sheet, the needles appear very minute, as they only extend from one tiny alpha particle to another. The phase relations of the 3% aluminum - 5% chromium - titanium alloy tend to follow those of the titanium - chromium system (Fig. 36) rather than the titanium - aluminum system (Fig. 37); the lower limit of the transformation range is about 700°C for this alloy (Fig. 38). Specimens heated to 700°C (Fig. 9) and 750°C (Fig. 10) show an enlarging of the alpha particles oriented within each grain. On further heating (Fig. 11) the alpha particles continue to enlarge to a maximum size at about 875°C (Fig. 12); specimens quenched from 920°C (Fig. 13) show a much smaller alpha particle size with a network of small beta grain boundaries connecting the alpha particles. This alpha growth and solution process was found not

only in heat treating specimens of as received sheet, as discussed above, but also in heat treating sheet specimens previously quenched from 1000°C (Figs. 16 thru 21), and those cold rolled to 20% reduction (Figs. 22 thru 27). Upon reheating the samples previously quenched from 1000°C, the alpha particles developed along the martensitic type needles (Fig. 19). The smaller size of the alpha particles found in the 20% cold rolled specimens (Fig. 25) indicates the possible fragmentation of the crystallographic planes produced by the cold deformation.

The back reflection x-ray pattern (Fig. 39), of the 20% cold reduced sheet also indicates fragmentation; the microstructure of the x-rayed area is shown in Fig. 40. The back reflection x-ray pattern (Fig. 41), and the corresponding microstructure (Fig. 42), of the same sample after quenching from 920°C shows that recrystallization has taken place.

An equilibrium amount and size of alpha particles are obtained when the specimens of this alloy are heated for a sufficient length of time in the alpha plus beta temperature range. Martens[15] has recently shown that it required 16 hours at 815°C for a similar titanium alloy to reach a maximum hardness which he states is evidence of progress to an equilibrium structure. The cooling rate will further determine the alpha particle size as illustrated in Fig. 28. This sample was cooled from 1000°C to 700°C at a rate of 1°C per minute and the microstructure is 50% alpha with large laminar layers. This can be compared to Fig. 14, which shows the microstructure of a sample quenched from 980°C.

The dark etching phase (Fig. 30) found in the beta matrix of massive laminar alpha fully annealed specimens reheated in the range of 500°C to 700°C is considered to be $TiCr_2$ or a similar titanium-chromium-aluminum compound. This phase has been discussed and tentatively identified by the NYU research group,[2] and would be that expected to occur in this range from consideration of the phase relations of the ternary diagram (Fig. 38).

Cold rolling this sheet, a 10% reduction in thickness could be obtained without loss of ductility; sheet reduced 20% fractured in bend testing (Fig. 34). 'Warm rolling' at 200°C to 10% and 20% reductions produced very brittle samples (Fig. 34). An increase in plasticity with increasing working temperatures, analogous to magnesium, had been forecast for titanium by Bounds and Cooper;[16] their prediction was based on the addition of slip systems at 200°C being a property of the hexagonal lattice, rather than a characteristic of the metal. The hardening obtained in the samples 'warm rolled' does not support that hypothesis.

The slight improvement in ductility, from 2.3% strain to 2.8% (Fig. 43) found in sheet rolled at 700°C, is not sufficient improvement to reduce fabricating problems as the sheet is still too brittle to allow sufficient bending or forming. The increase in strength of sheet rolled at 800°C exceeds the hardening and strengthening found by P.R. Mallory & Co., in forging these alloys at 800°C. They had attributed the increases

to strain hardening from cold working.[1] It must be considered that any forging or rolling below 925°C will be cold working for these alloys as no recrystallization appears below that temperature. The sheet rolled at 900°C showed no change in ductility or microstructure; this approximated the temperature range in which the alloy had originally been reduced to sheet. The microstructure of sheet rolled at 1000°C was unchanged from that of the as received material, except for a saw tooth effect in the grain boundaries. Either the chilling effect of the rolls and subsequent air cooling supplied a cooling rate that recreated the original microstructure, or possibly the transformation is so sluggish that the alloy was not at temperature long enough for any new microstructure to form. The heating period of ten minutes was sufficient time for the exposed edge surface to develo an oxide coating, and for the previous surface oxide layers to diffuse inward a small amount.

The microstructures of the as received sheets, when compared with the microstructures obtained in heat treating this alloy, are not compatible with the thermal history furnished with the material. The hot rolling temperature was given as 850°C; a structure similar to Fig. 11, should have been attained at this temperature rather than that of Fig. 8. The nearest approach to the structure shown in Fig. 8, was found in specimens quenched from 750°C (Fig. 10).

The metallographic examination of the prepared binary alloy series of titanium - chromium and titanium - aluminum revealed the process of transformation and recrystallization of this group of titanium alloys. Cold-worked, arc-melted, unalloyed sponge titanium will recrystallize at 600°C; it is all in the alpha phase at this temperature. With an increase in temperature to 900°C, the alpha phase starts to transform to beta. The beta phase is nucleated both within the grains and at the grain boundaries, but more generally at the boundaries; the beta within the grains is oriented on crystallographic planes. The transformation proceeds with increasing temperatures and is virtually completed at 950°C. The same recrystallization and phase changes take place in binary titanium alloys containing 1, 2 and 3% Al, and 1% Cr; these are also single phase alpha titanium solid solutions below the transformation range. The beta grains grow rapidly after all the alpha has transformed, and all of the specimens quenched from 1000°C had a very large acicular transformed beta structure. The titanium - chromium binary alloys over 1% chromium, and the 3% aluminum - 5% chromium - titanium alloy are mixtures of alpha and beta phases at lower temperatures. When annealed at 800°C, 30% cold reduced specimens of these alloys show no evidence of recrystallization. On heating further, the alpha particles in these samples decrease in size as they transform to beta (Fig. 31), and at about 950°C the microstructure consists of small recrystallized beta grains, with the last vestiges of the alpha particles in the beta grain boundary network (Fig. 32). When quenched from 1000°C, these samples have very large acicular beta grains, accelerated grain growth having started at the final transformation of the alpha phase. Recrystallization and grain growth of these titanium alloys is thus inhibited while transformation is incomplete. This

behavior is in contrast to that found in duplex brass[17,18] in which either of the two phases, alpha and beta, may be induced to recrystallize by appropriate treatments. Titanium behavior also contrasts with that of steel, in which a grain refinement from the allotropic change takes place in passing through the gamma to alpha iron transformation.

The lack of hardening response of the alloy sheet and of the specially prepared alloy buttons of the same composition to both thermal and mechanical treatments seems to be a characteristic of this alpha-beta type of titanium alloy. P.R. Mallory & Co., had reported a similar finding with this alloy containing carbon.[1]

The recrystallization and grain growth characteristics of this alloy suggest that changes are required in the procedures for ingot breakdown and fabrication. While short time heating cycles in air in the temperature range of 700°C to 900°C will render the alloy more amenable to deformation at these temperatures without materially affecting the microstructure or grain size, heating into the all beta region can only result in an increased grain size due to the impractibility of heating just to the end of the transformation. While the majority of the forging and hot working operations that have been reported in the literature on titanium alloys have been performed at 900°C, starting forging at 1050°C with finishing not below 950°C would result in a more positive breaking up of the as-cast grain structure. In heating into this temperature range surface oxidation and oxygen diffusion, however, becomes a serious problem.

## VI. Conclusions

1. The 3% aluminum - 5% chromium - titanium alloy sheet as received had too large a grain size to permit study of the preferred orientations produced by varying rolling and annealing procedures.

2. The forging and hot working of the as received sheets were performed under conditions which promoted the formation of oxidized surface layers, and allowed diffusion of the layers into the sheet.

3. The grain size could not be reduced by the usual methods of deformation and annealing.

4. The recrystallization of this low carbon alloy is inhibited when transformation is incomplete and a second phase is present in the microstructure. When the alloy is heated into the all beta range, 925°C or over, to complete the transformation, an accelerated grain growth immediately follows, nullifying this as a method of grain refinement.

5. A slight improvement in ductility of the sheet alloy was attained by heating to 700°C for rolling; this increase however, is not sufficient to alleviate the forming difficulties of this material.

## REFERENCES


1. P.R. Mallory & Co., Progress Reports 8000-M-464 on Navy Contract [illegible] NO[illegible](s)9919 Oct 1948-Aug 1949; 8000-M-474 on Navy Contract No. NO[illegible] (s)[illegible]0683 Feb 7 1950 to Aug 1950; 8000-M-484 on Navy Contract N[illegible] NOa(s[illegible]51-006-C Nov 1950-Dec 1951

2. [illegible], Mar[illegible]n and Nielsen, N.Y.U. Reports on Navy Contrac[illegible] NO[illegible]) 41-[illegible]C Jan 1951-Nov 1951

3. Walt[illegible], Bean [illegible] Anal. Chem., 22, 297-303 (1950); al[illegible] RL Report [illegible]77

4. P[illegible] N[illegible], [illegible]l of Metals, 188, 1444, (195[illegible]

5. A.S.T.[illegible] Designa[illegible]n E[illegible]16-39, ASTM Standards 19[illegible]rt 1, 1259

6. Chanc[illegible] V[illegible] Aird[illegible]t[illegible]e., Report No. 2105[illegible]ed 5 Aug 1948

7. Lubahn, J.D[illegible] Sac[illegible] Trans. A.S.M[illegible]2, 201, (1950)

8. Northcott, Jnl I[illegible] [illegible]. 126, [illegible] (1932)

9. Northcott, Jnl Inst. [illegible], (1936)

10. Goss, N.P., Metals Tech[illegible]2)

11. Greninger, Trans. AIME[illegible]

12. Hultgren, Trans. AI[illegible]12, 493, [illegible]

13. LaCombe, Co[illegible]trength Solid[illegible] Soc. Lon. 91, (1948)

14. LaCombe, [illegible], 15, (1-2), 161, [illegible]

15. Martens[illegible]. Tech. Report on Army Cont[illegible]e. DA-04-495-Ord 18 (Rept [illegible]8)

16. Bo[illegible] A.M. and Cooper, H.W., Working of T[illegible], Superior Tube C[illegible] May 1950

17. [illegible]heycombe and Boas, Nature, 159, 847 (1947)

ACKNOWLEDGMENTS

The authors appreciate the interest and the suggestions of Dr. O.T. Marzke during the course of the work. The preparation of the samples and the microphotographs by Mrs. Betty Friske is gratefully acknowledged.

Kenneth B. Lloyd

Mr. Kenneth B. Lloyd
Research Metallurgist

E.J. Chapin

Mr. E.J. Chapin, Head
Nonferrous Alloys Branch

O.T. Marzke

Dr. O.T. Marzke
Superintendent,
Metallurgy Division

## REFERENCES


1. P.R. Mallory & Co., Progress Reports 8000-M-464 on Navy Contract No. NOa(s)9919 Oct 1948-Aug 1949; 8000-M-474 on Navy Contract No. NOa (s)-10683 Feb 7 1950 to Aug 1950; 8000-M-484 on Navy Contract No. NOa(s) 51-006-C Nov 1950-Dec 1951

2. Ence, Margolin and Nielsen, N.Y.U. Reports on Navy Contract No. NOa(s) 41-331-C Jan 1951-Nov 1951

3. Walter, Dean I., Anal. Chem., 22, 297-303 (1950); also as NRL Report 3577

4. Pinto, Norman P., Jnl of Metals, 188, 1444, (1950)

5. A.S.T.M. Designation E 16-39, ASTM Standards 1949, Part 1, 1259

6. Chance Vought Aircraft Co., Report No. 21056 dated 5 Aug 1948

7. Lubahn, J.D., and Sachs, G. Trans. A.S.M.E. 72, 201, (1950)

8. Northcott, Jnl Iron Steel Inst. 126, 267, (1932)

9. Northcott, Jnl Inst. Metals, 59, 225, (1936)

10. Goss, N.P., Metals Tech. 272, (1940)

11. Greninger, Trans. AIME 145, 262, (1941)

12. Hultgren, Trans. AIME 172, 493, (1947)

13. LaCombe, Conf. on Strength Solids, Phys. Soc. Lon. 91, (1948)

14. LaCombe, Physica, 15, (1-2), 161, Apr 1949

15. Martens, Cal. Tech. Report on Army Contract No. DA-04-495-Ord 18 (Rept 20-158)

16. Bounds, A.M. and Cooper, H.W., Working of Titanium, Superior Tube Co., May 1950

17. Honeycombe and Boas, Nature, 159, 847 (1947)

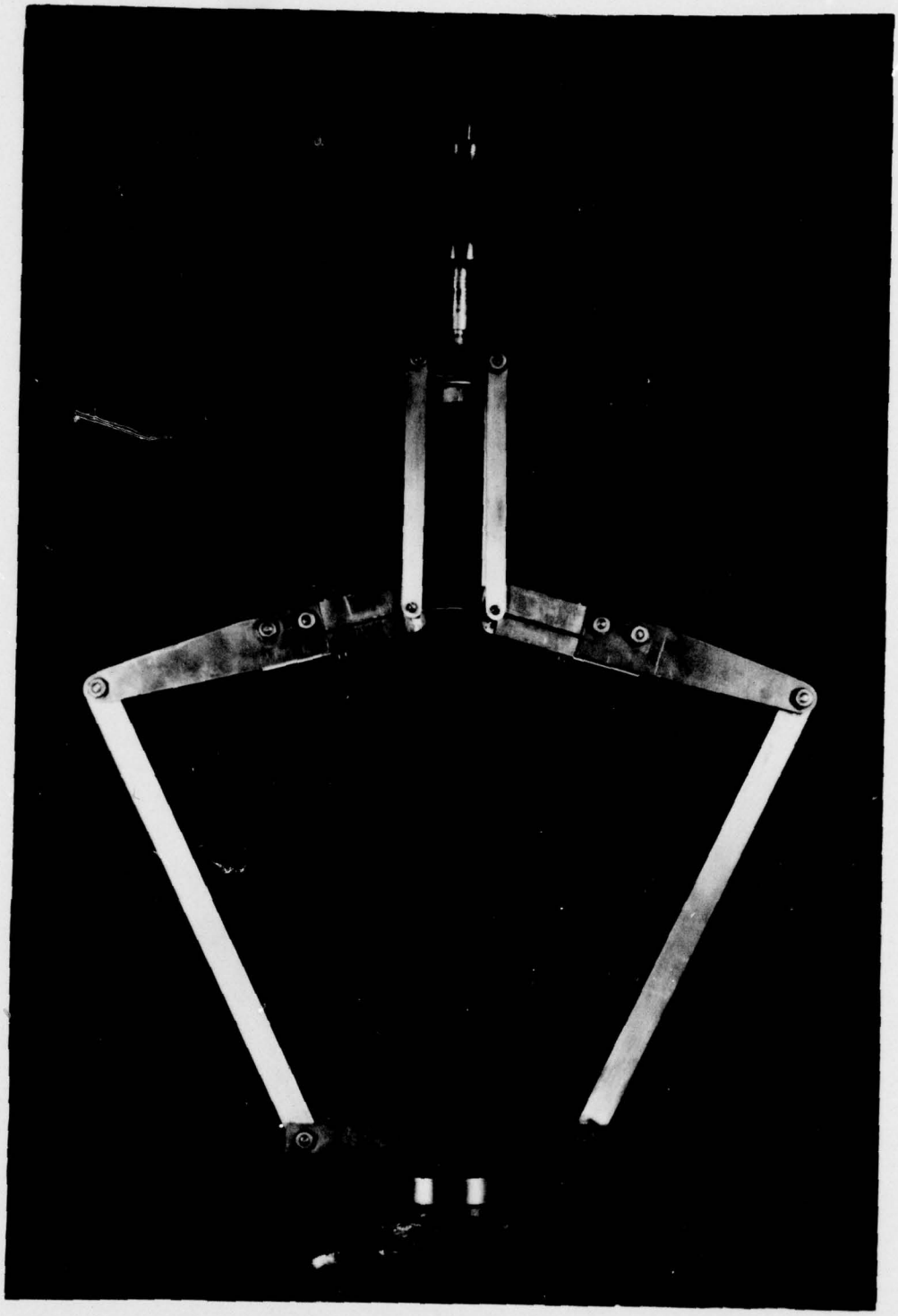
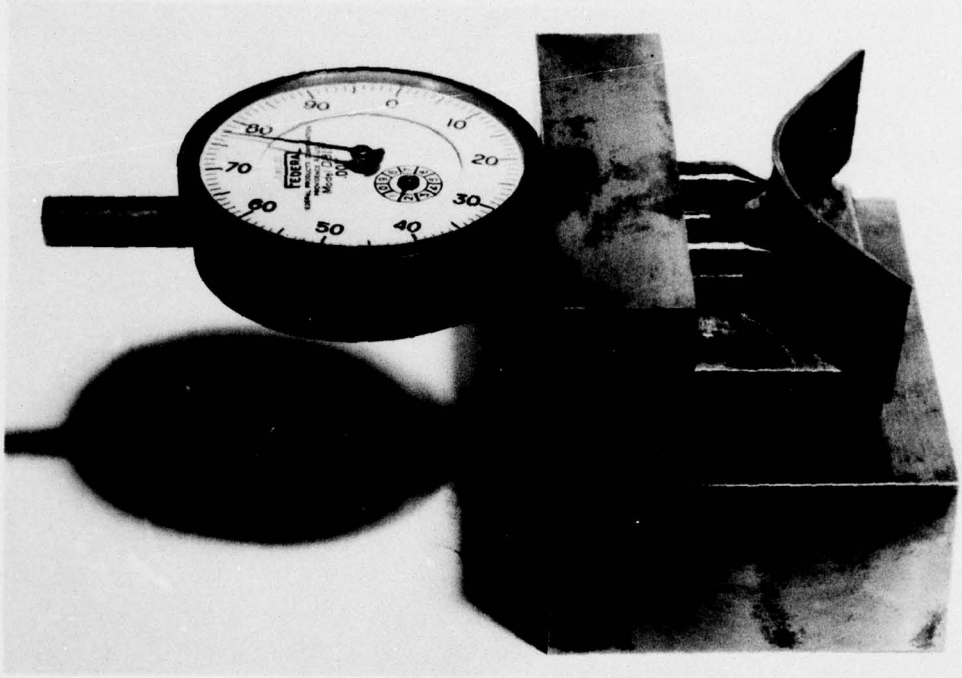


**Figure 2 - Constant moment bend jig**




CURVATURE MEASURING FIXTURE

FIGURE 3

Mallory 3% Al - 5% Cr - Ti Alloy Sheet

Fig. 4 As Received 50X

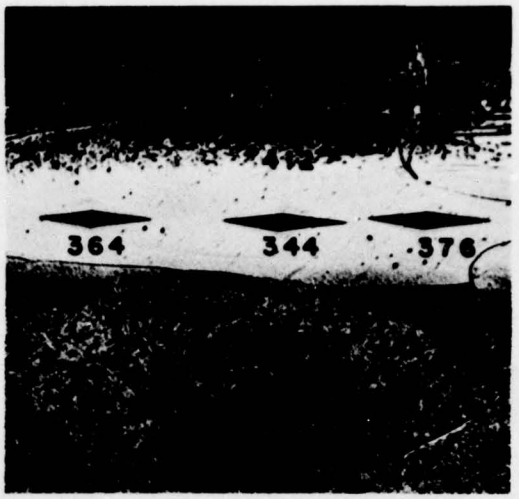



Fig. 5 As Received 200X Knoop Hard.

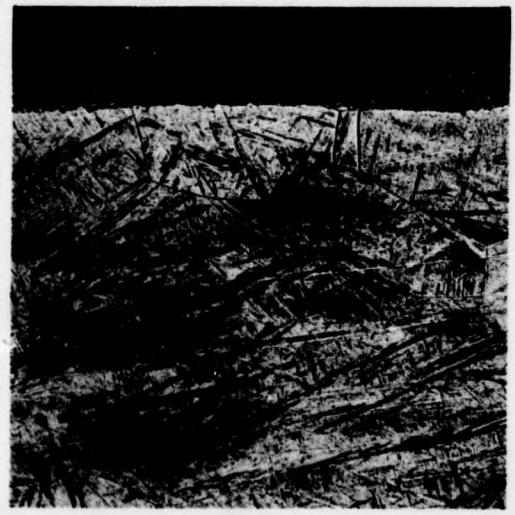


Fig. 6 ½ hr 1000°C Quenched 50X

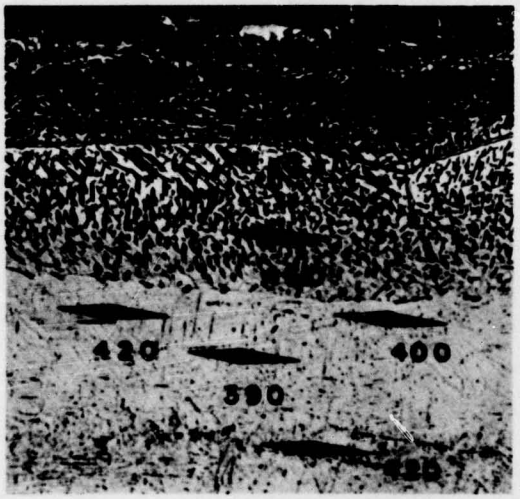



Fig. 7 #6 at 200X Knoop Hardness

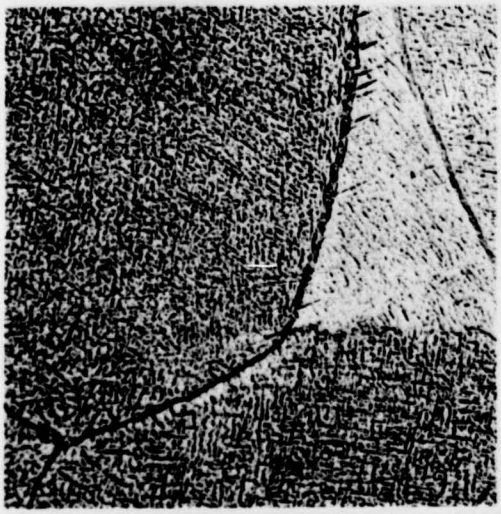


Fig. 8 As Received 500X

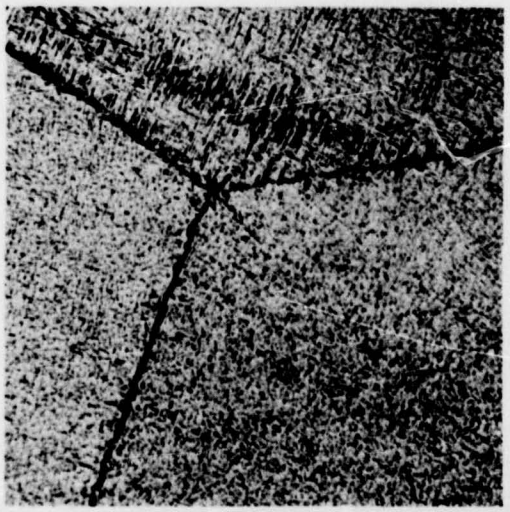


Fig. 9 Heated to 700°C Quenched 500X

Mallory 3% Al - 5% Cr - Ti Alloy Sheet

Fig. 10 Heated to 750° Quenched 500X

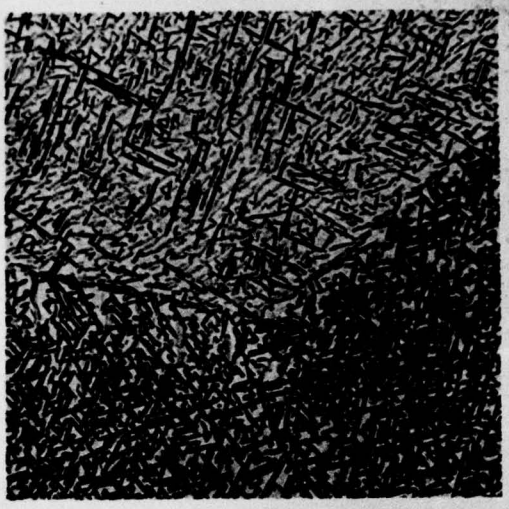


Fig. 11 Heated to 850° Quenched 500X

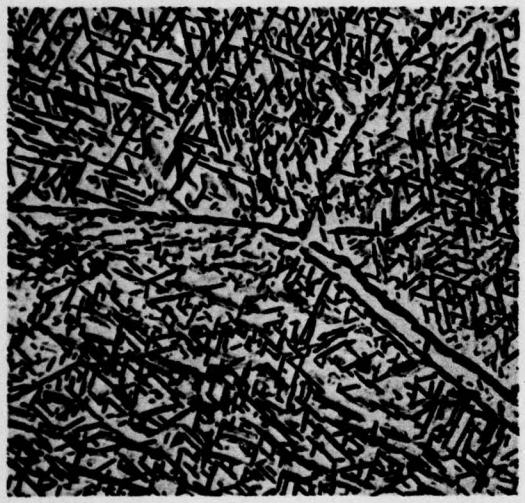


Fig. 12 Heated to 880° Quenched 500X

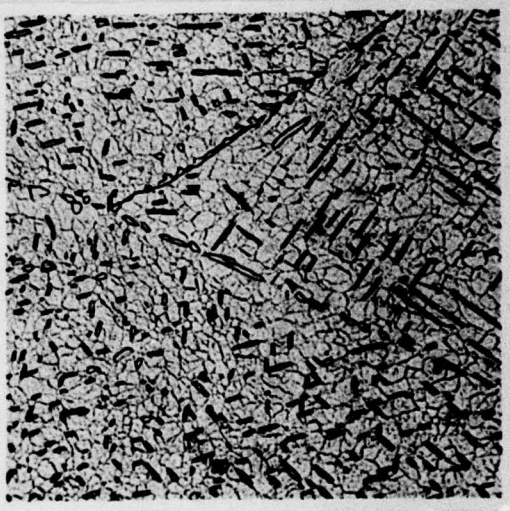


Fig. 13 Heated to 920° Quenched 500X

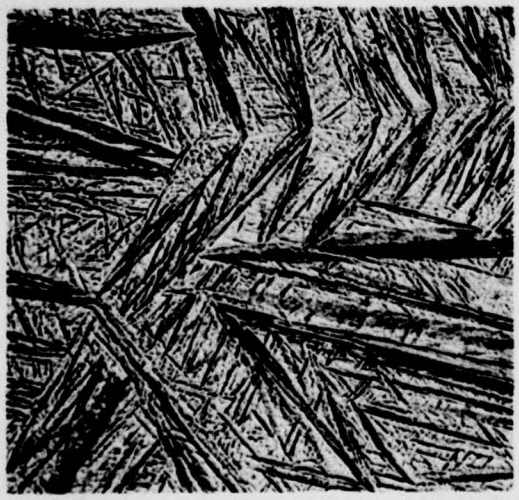


Fig. 14 Heated to 980° Quenched 500X

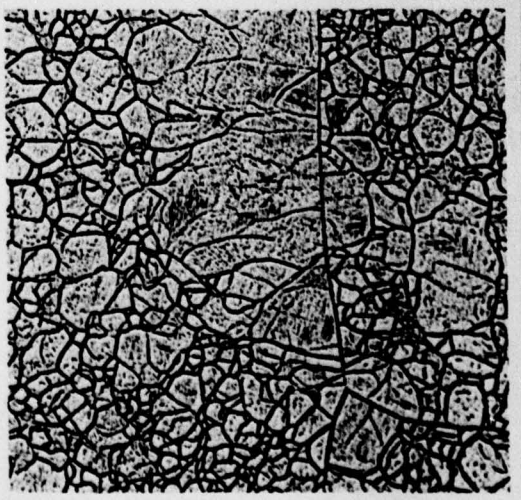


Fig. 15 Quenched Twice from 1000° 100X

Mallory 3% Al -5% Cr - Ti Alloy Sheet Quenched from 1000°C

Fig. 16 Reheated to 700° Quenched 500X

Fig. 17 Reheated to 750°C Quenched 500X

Fig. 18 Reheated to 850° Quenched 500X

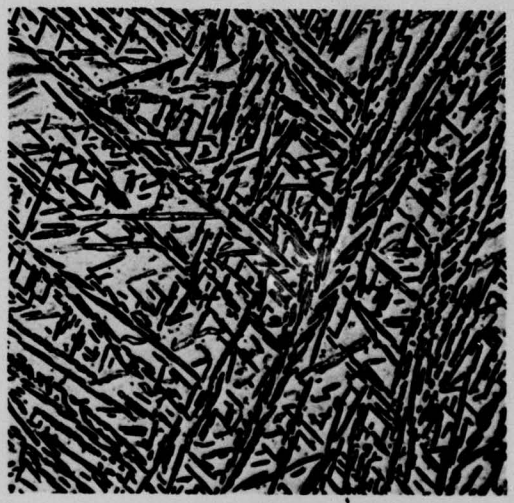


Fig. 19 Reheated to 880° Quenched 500X

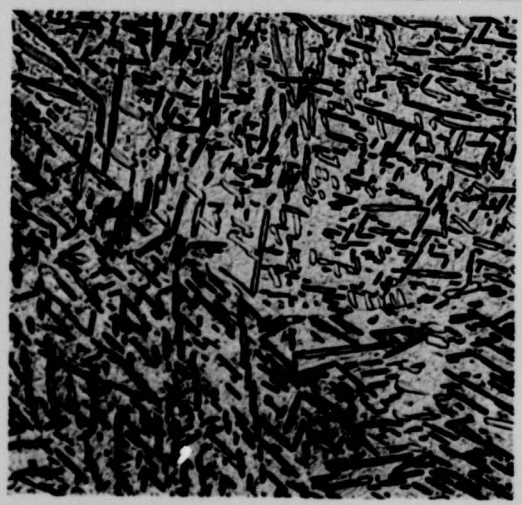


Fig. 20 Reheated to 920° Quenched 500X

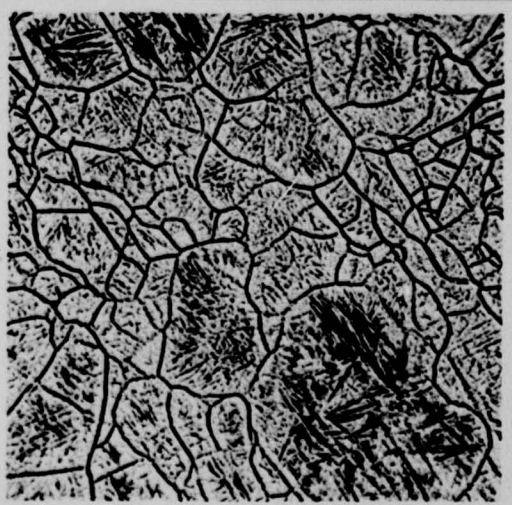


Fig. 21 Reheated to 980° Quenched 500X

Mallory 3% Al-5%Cr-Ti Alloy Sheet Cold Rolled to 20% Reduction

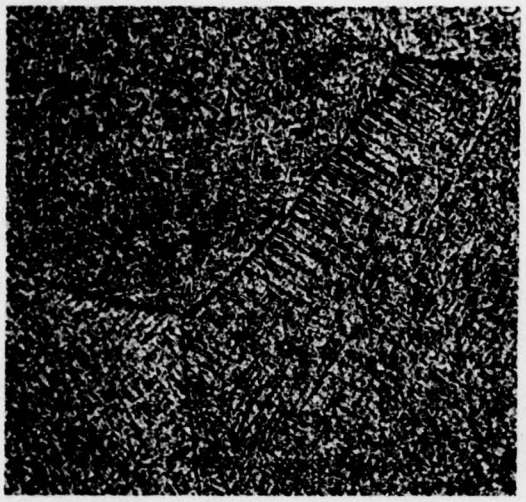


Fig. 22 Heated to 700°C Quenched 500X

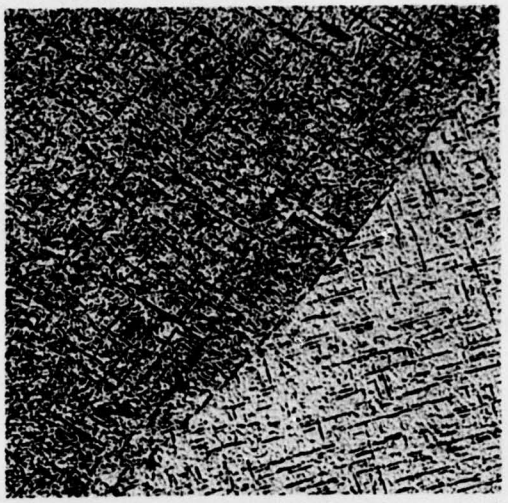


Fig. 23 Heated to 750°C Quenched 500X

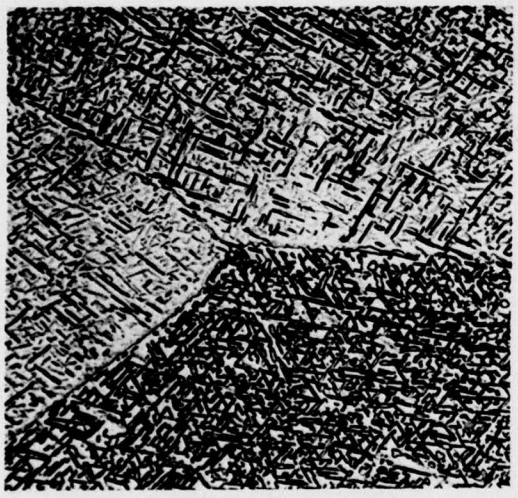


Fig. 24 Heated to 850°C Quenched 500X

Fig. 25 Heated to 880°C Quenched 500X

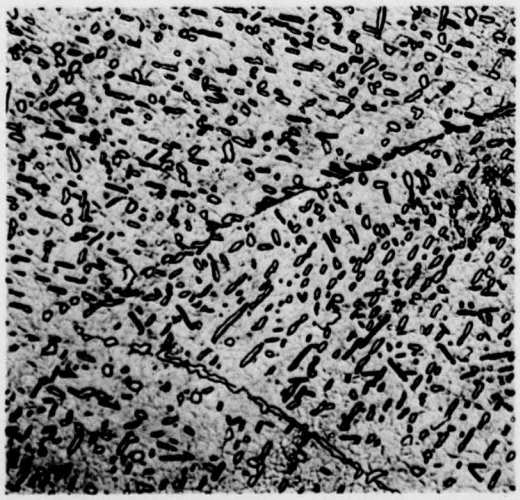


Fig. 26 Heated to 920°C Quenched 500X

Fig. 27 Heated to 980°C Quenched 500X

Fig. 28 3Al-5Cr-Ti Slowly Cooled 500X

Fig. 29 #28 Reheated to 850° 500X

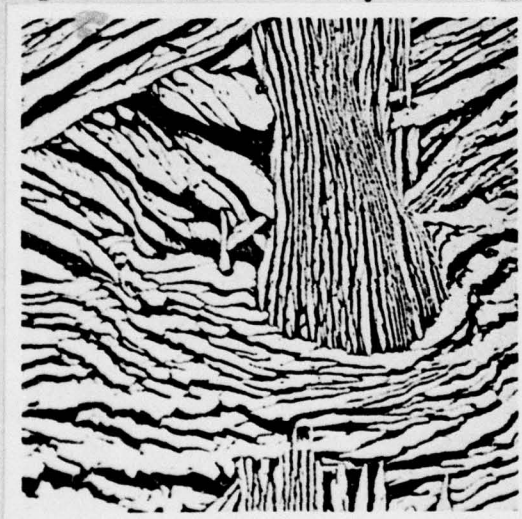


Fig. 30 #28 Reheated to 600°C 500X

Fig. 31 4%Cr-Ti Heated to 850° 500X

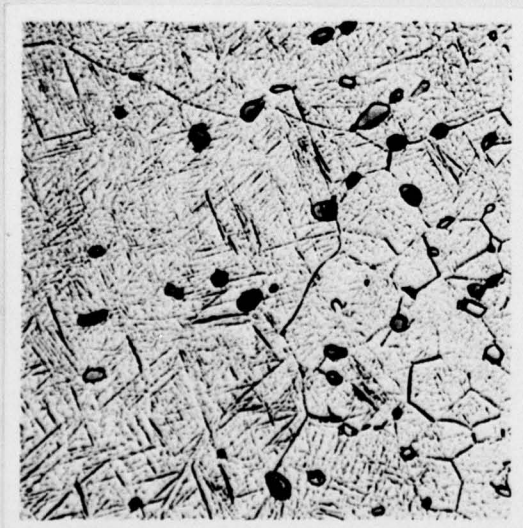


Fig. 32 4%Cr-Ti Heated to 950° 500X

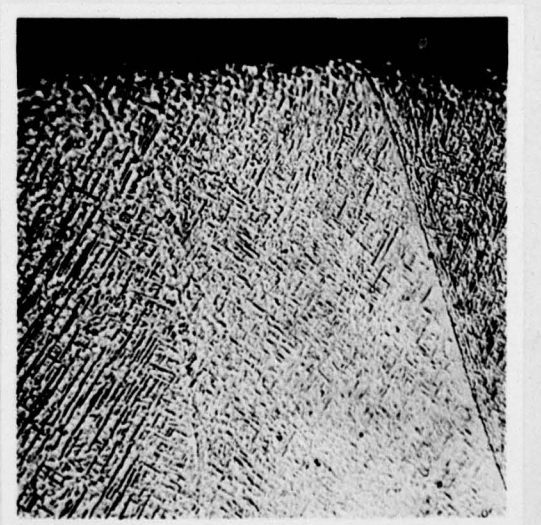
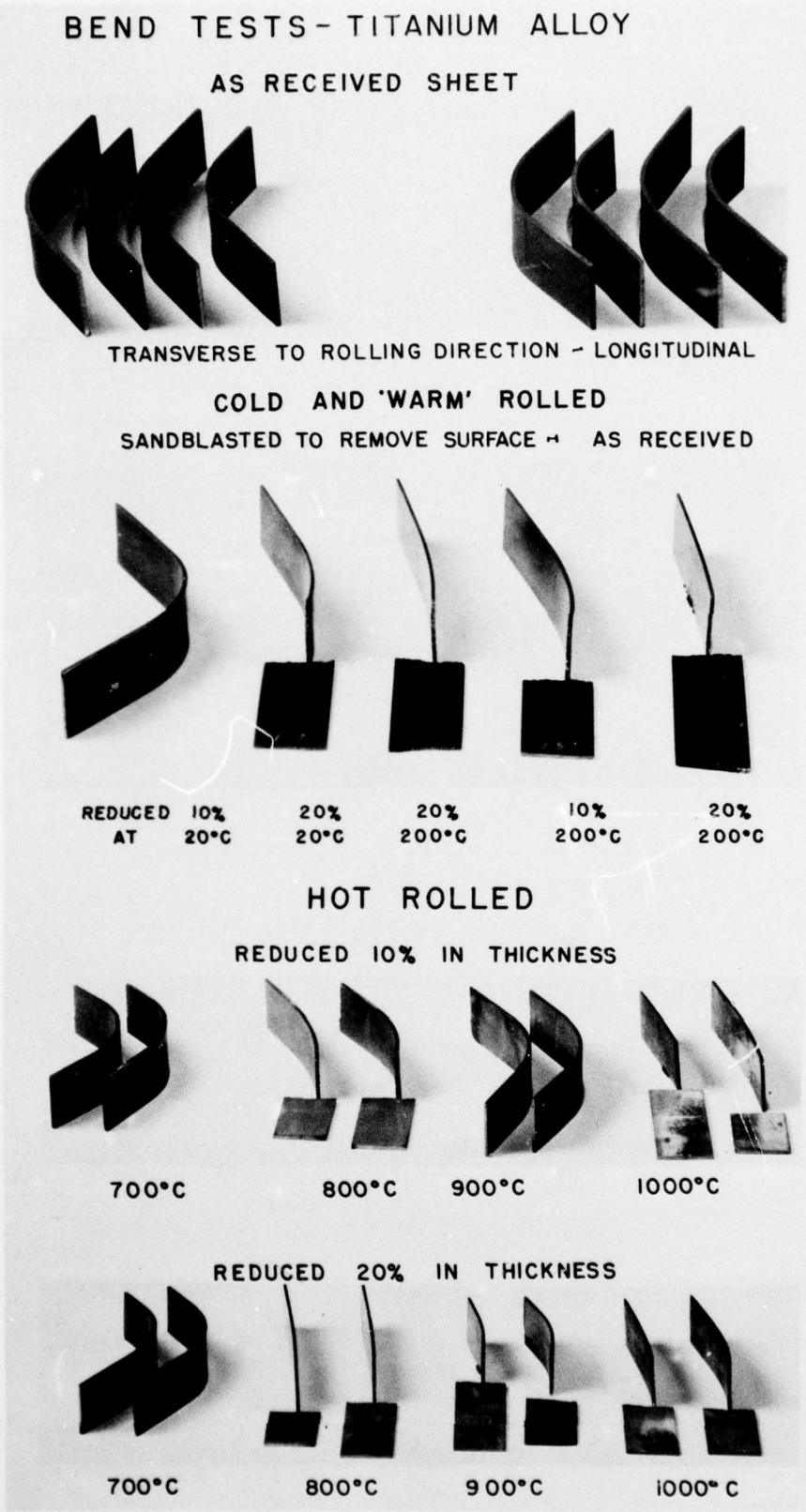


Fig. 33 Sheet heated to 700° 500X




BEND TEST SPECIMENS

FIGURE 34

Figure 35

Microstructure and Rockwell A Hardness of Titanium Alloy Specimens. Specimens were of arc melted bars, cold rolled to 20% reduction, annealed 1 hour at 800°C, cold rolled to 35% reduction, and heated in vacuum for quenching.

| Composition | Structure Before | Quenched from 850°C | Quenched from 900°C | Quenched from 950°C | Quenched from 1000°C |
|---|---|---|---|---|---|
| 3% Al 97% Ti | Worked Alpha grains, single phase 62-64 | Recrystallized Alpha single phase 61-61 | 2% Beta in Alpha grain boundaries 58-59 | 20% Beta at boundaries no growth 59-60 | 75% Acicular Beta, some growth 61-62 |
| 2% Al 98% Ti | Worked Alpha grains, single phase 63-63 | Recrystallized Alpha, single phase 58-61 | Alpha Phase no growth 59.5-60 | 60% Beta at boundaries no growth 59-60 | 95% Beta, very large acicular grains 62-63 |
| 1% Al 99% Ti | Worked Alpha grains, single phase 59-61 | Recrystallized Alpha, single phase 54-54 | 20% Beta in grain bound. no growth 53-55 | 90% Beta, some growth 57-57.5 | 100% Beta, very large acicular grains 57-59 |
| 100% Ti | Worked Alpha grains, single phase 64-64.5 | Recrystallized Alpha, single phase 56.5-57.5 | 25% Beta, in grains and boundaries 58-58 | 80% Beta, some growth 57-59 | 100% Beta, very large acicular grains 57.5-59 |
| 1% Cr 99% Ti | Worked very small Alpha grains 63-64 | 50% Beta, some growth 64.5-66 | 80% Beta, some growth 64-64 | 90% Beta, large grains 63-65 | 100% Beta, very large grains 64.5-66 |
| 2% Cr 98% Ti | Two phases 40% Beta 63-64 | 60% Beta 70-71 | 90% Beta, recrystallizing 68-70 | 100% Beta, large acicular grains 68-69 | 100% Beta, very large acicular grains 67.5-69 |
| 3% Cr 97% Ti | Two phase, large grain network basket weave 66-66.5 | 75% Beta, Alpha is oriented in beta grains 70-71 | 90% Beta, recrystallizing 71-72 | 100% Beta, large acicular grains 68.5-71 | 100% Beta, very large acicular grains 69.5-71 |
| 4% Cr 96% Ti | Two phase, large grain network basket weave 67-68 | 75% Beta, Alpha oriented in lg. Beta grains 72-73 | 85% Beta, recrystallizing 75-76 | 100% Beta, large acicular grains 74-74.5 | 100% Beta, very large acicular grains 74.5-75 |
| 5% Cr 95% Ti | Two phase, large grain network basket weave 64-66 | 75% Beta, large grain network 72-73 | 85% Beta, recrystallizing 75-76 | 100% Beta, large acicular grains 74-74.5 | 100% Beta, very large acicular grains 74.5-75 |
| 3% Al 5% Cr 92% Ti | Two phase, large grains, phases oriented in gr 68-69 | 70% beta, lg. grain network 67.5-67.5 | 85% Beta, small grains 68-68 | 100% Beta, large grains acicular 68-68 | 100% Beta, single phase lg. grains very few needles 67-68 |

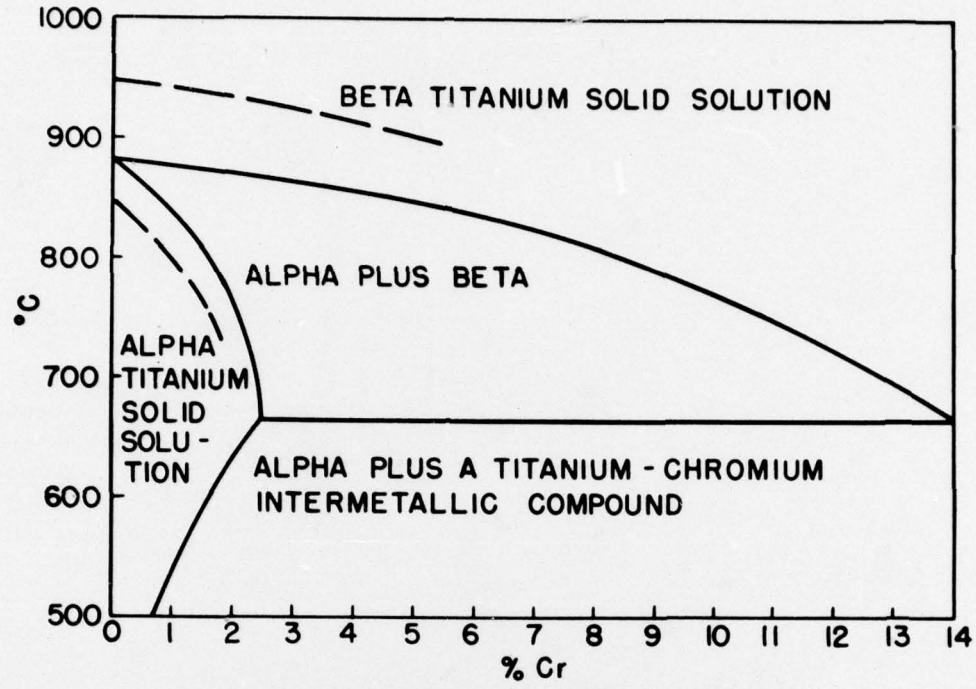



Figure 36. Partial Phase Diagram of Titanium-Chromium System, NYU[2]
Dashed Lines indicate Transformation Range found with commercial sponge Titanium, NRL

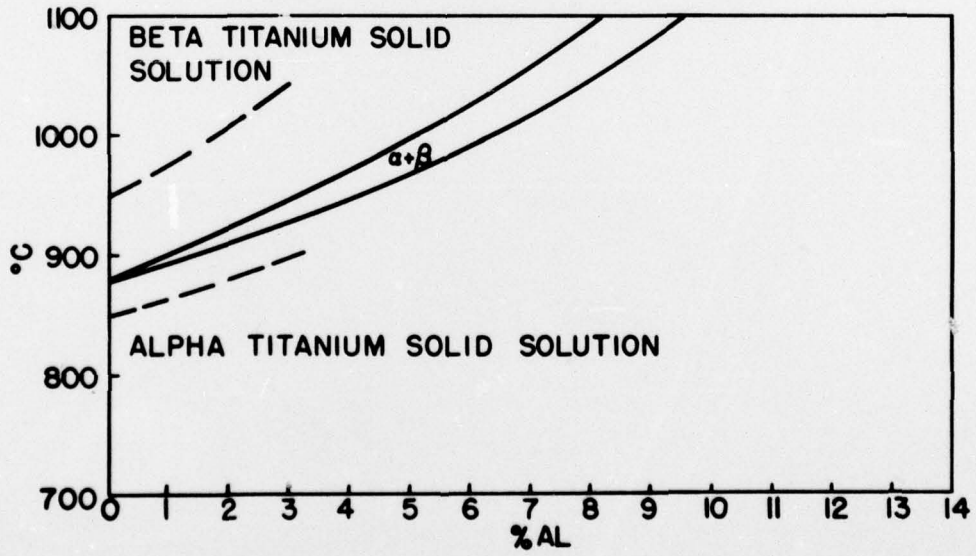



Figure 37. Partial Phase Diagram of Titanium-Aluminum System, NYU[2]
Dashed Lines indicate Transformation Range found with commercial sponge Titanium, NRL

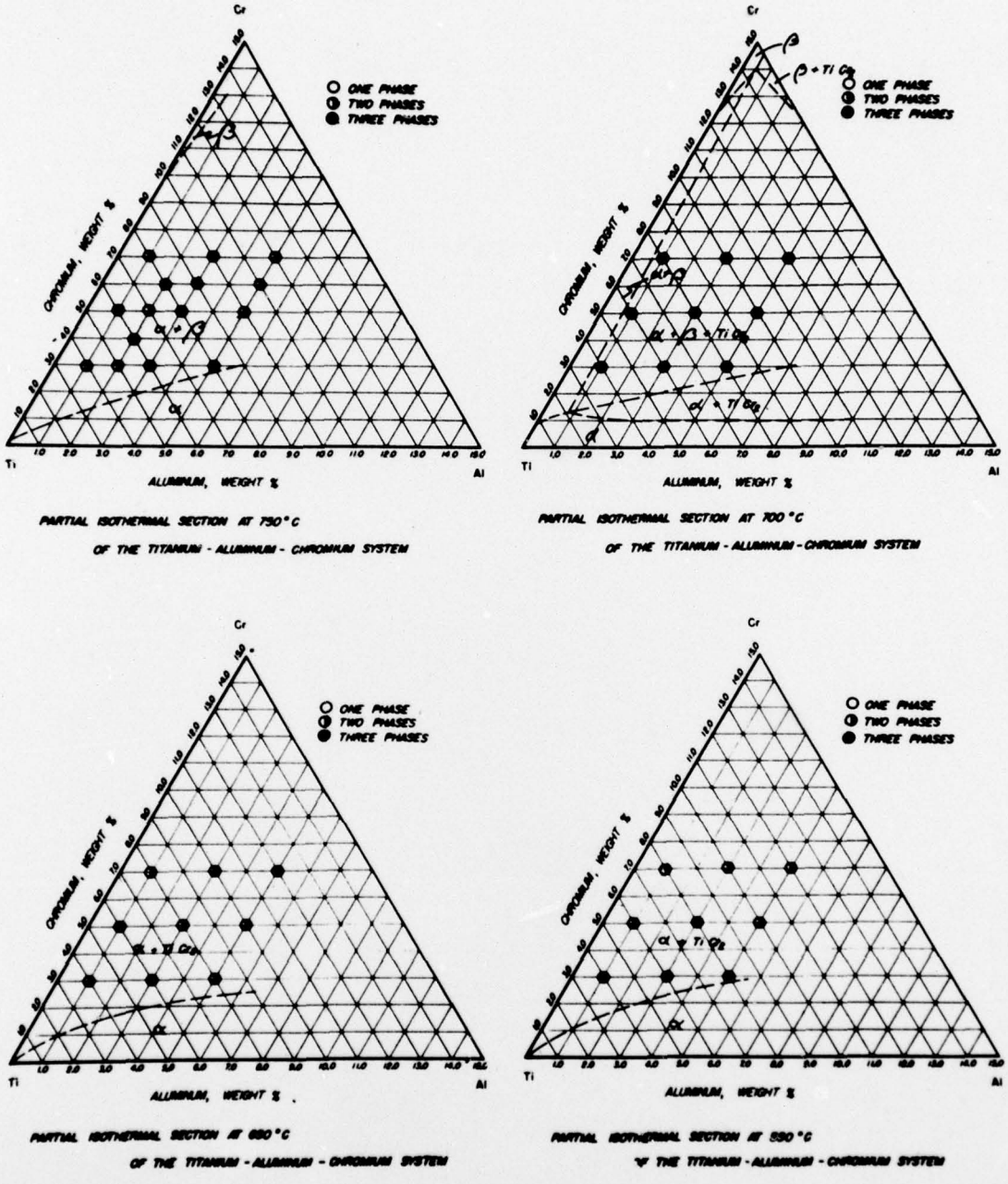



Figure 38. From New York University Report on Navy Contract No. NOa(s)51-331-C

Figure 39. Back Reflection X-ray Pattern - 20% cold Rolled 3 Al-5 Cr- Ti Alloy Sheet

Figure 40. Microstructure of X-rayed area of Figure 39, Electropolished 500X

Figure 41. Back Reflection X-ray Pattern - Sample of Figure 39 after quenching from 920°C.

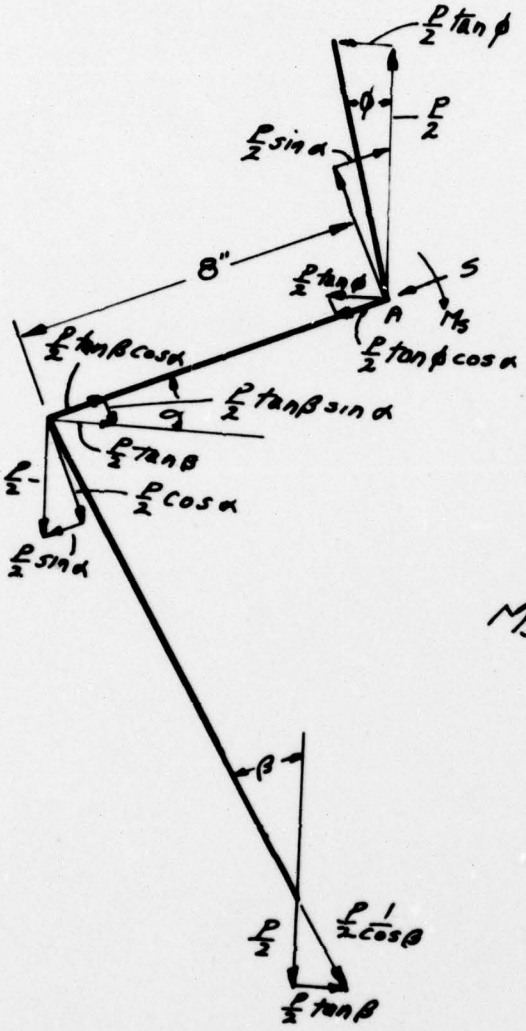
Figure 42. Microstructure of X-rayed area of Figure 41, Electropolished 500X

Figure 43. Tabulation of Tension and Bend Test Results

| Treatment | Ultimate Strength | Measured Elong 2" | Chart Total Strain | Chart Plastic Strain | Fiber Strain in Bend Test |
|---|---|---|---|---|---|
| As rec'd Long. | 163,000 | 3.5% | 2.8% | 2.0% | 2.34%-2.20% |
| Long. | 159,500 | 2.5% | 2.2% | 1.3% | 2.80-2.15% |
| Trans. | 152,560 | 3.0% | 2.55% | 2.0% | 2.44-2.00% |
| Trans. | 152,500 | 3.0% | 2.60% | 1.8% | 2.24-2.30% |
| Rolled 10% 600°C | 159,760 | 4.5% | 3.37% | 2.12% | 2.70%---- |
| Rolled 20% 600°C | 168,960 | 3.5% | 3.13% | 1.90% | 1.71-1.96% |
| Rolled 10% 650°C | 157,000 | x | 3.70% | 2.63% | 3.04--2.61% |
| Rolled 20% 650°C | 162,500 | 3.0% | 2.63% | 1.50% | 2.26%---- |
| Rolled 10% 675°C | 158,400 | 5.0% | 4.44% | 3.32% | 2.99-2.51% |
| Rolled 20% 675°C | 167,700 | 3.5% | 3.00% | 1.82% | 2.70-2.36% |
| Rolled 10% 700°C | 160,000 | 5.0% | 4.25% | 3.25% | 2.56-2.36% |
| Rolled 20% 700°C | 169,280 | 6.0% | 5.0% | 3.66% | 2.36-2.21% |
| Rolled 10% 700°C | 158,740 | 1.5% | x | x | 2.84-3.22% |
| Rolled 20% 700°C | 160,740 | 4.5% | x | x | 2.24-2.44% |
| Rolled 10% 800°C | 190,330 | 1.0% | x | x | 1.0* -1.9%* |
| Rolled 20% 800°C | 190,300 | .75% | x | x | x* - x* |
| Rolled 10% 900°C | 160,350 | 2.0% | x | x | 2.24-2.16% |
| Rolled 20% 900°C | 144,400 | 1.0% | x | x | 1.5%*- x* |
| Rolled 10% 1000°C | x | x | x | x | .5%* -x* |
| Rolled 20% 1000°C | 118,400 | 1.5% | x | x | 1.0%*-.5%* |

(x) Not determined.
(*) Fractured

Appendix I – Side Load and Bending Moment Analysis of 'Constant Moment Bending Jig'.



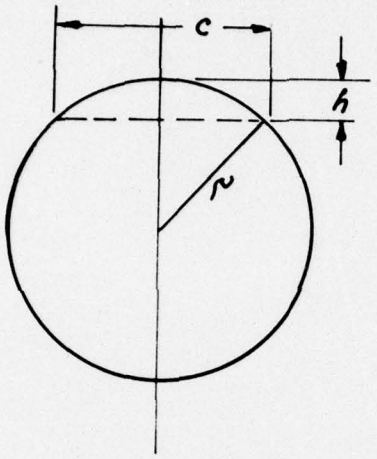

$$\Sigma F_{A\ \text{II to spec.}} = 0 = -S + \frac{P}{2}\tan\beta\cos\alpha - \frac{P}{2}\sin\alpha - \frac{P}{2}\tan\phi\cos\alpha + \frac{P}{2}\sin\alpha$$

$$S = \frac{P}{2}\cos\alpha(\tan\beta - \tan\phi)$$

$$\Sigma M_A = 0 = -M_S + 8\frac{P}{2}\cos\alpha + 8\frac{P}{2}\tan\beta\sin\alpha$$

$$M_S = 4P(\cos\alpha + \tan\beta\sin\alpha)$$

## Appendix II - Mathematical Derivation of Radius of Curvature 'r' from Measurement of Height of Bend of Specimen 'h'.




$$r^2 = (r-h)^2 + \left(\frac{c}{2}\right)^2$$

$$r^2 = r^2 - 2rh + h^2 + \frac{c^2}{4}$$

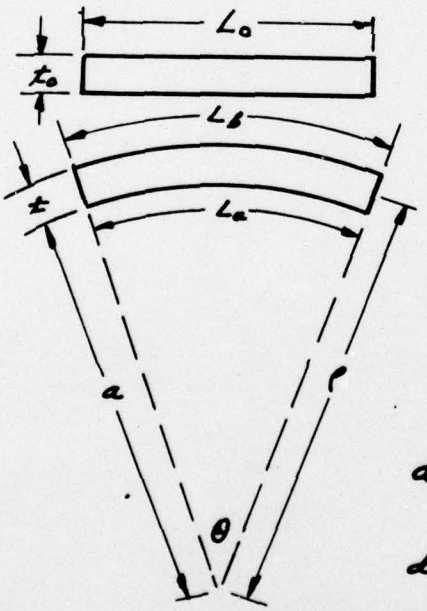
$$2rh = h^2 + \frac{c^2}{4}$$

$$r = \frac{h}{2} + \frac{c^2}{8h}$$

$$c = 1.000''$$

$$r = \frac{h}{2} + \frac{1}{8h}$$

### Derivation of Method of Evaluation of Tensile Strain from Radius of Curvature.




$$A_o = L_o t_o$$

$$A = \frac{L_a + L_b}{2}\, t$$

$$L_a = \theta a \qquad (1)$$

$$L_b = \theta(a+t) \qquad (2)$$

$$\rho = \sqrt{a(a+t)}$$

$$dL_a = -d\theta(\rho - a)$$
$$= -d\theta\left[\sqrt{a(a+t)} - a\right] \qquad (3)$$

$$dL_b = d\theta(a+t-\rho)$$
$$= d\theta\left[a+t-\sqrt{a(a+t)}\right] \qquad (4)$$

Appendix III - Derivation continued

$t$ is assumed constant and differentiating (1) and (2)

$$dL_a = \theta\, da + a\, d\theta$$

$$dL_b = \theta\, da + (a+t)\, d\theta$$

eliminating $\theta$ in (3) and (4)

$$\frac{dL_a}{L_a} = \frac{da}{a} - \frac{da}{\sqrt{a(a+t)}}$$

$$\frac{dL_b}{L_b} = \frac{da}{a+t} - \frac{da}{\sqrt{a(a+t)}}$$

these can be integrated —

$$\frac{L_a}{L_o} = 1 + e_a = \frac{4a}{2a + t + 2\sqrt{a(a+t)}}$$

$$\frac{L_b}{L_o} = 1 + e_b = \frac{4(a+t)}{2a + t + 2\sqrt{a(a+t)}}$$

If a value $z = \frac{a+t}{a} = \frac{r}{r-t}$ is used —

$$e_a = \frac{4}{1 + z + 2\sqrt{z}} - 1$$

$$e_b = \frac{4z}{1 + z + 2\sqrt{z}} - 1$$

$e_b$ is thus the tensile strain in the outer fibers, and the values of $e_b$ corresponding the values of z calculated from the radius of curvature and thickness measurements of the specimens can be read from a graph.