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DEVELOPMENT OF IMPROVED TITANIUM-BASE ALLOYS

HERBERT A. ROBINSON
PAUL D. FROST
WALTER M. PARRIS

BATTELLE MEMORIAL INSTITUTE

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FOREWORD

This report was prepared by the Battelle Memorial Institute under USAF Contract No. AF 33(616)-384. The contract was initiated under Research and Development Order No. 615-11BB, "Titanium Metals & Alloys", and was administered under the direction of the Materials Laboratory, Directorate of Research, Wright Air Development Center, with 1/Lt E. F. Erbin acting as project engineer.

ABSTRACT

The research program during the past year was centered around six alloys which had shown considerable promise in earlier work as potential aircraft structural materials. The tensile properties of the alloys, as affected by variations in hot working procedures and heat treatments, were evaluated.

Outstanding properties were obtained in two alloys:

Ti-3Mn-1Cr-1Fe-1Mo-1V
Ti-5Mn-2.5Cr

Several ingots of the Ti-3Mn-complex alloy were prepared. These ingots have been forged and rolled to bar stock for evaluation by several industrial organizations.

PUBLICATION REVIEW

This report has been reviewed and is approved.

FOR THE COMMANDER:

M. R. WHITMORE
Technical Director
Materials Laboratory
Directorate of Research

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SUMMARY

Exploratory Program

The following six alloys, originally made under Contract No. AF 33(038)-3736, were investigated thoroughly to evaluate the effects of fabricating and heat-treating variables on their mechanical properties:

Ti-3Mn-1Cr-1Fe-1Mo-1V(1)
Ti-4Fe-1Cr-1Mn-1Mo-1V(2)
Ti-5Mn-2.5Cr
Ti-3.5Mn-3.5Cr
Ti-4.5Mo-3.5Fe
Ti-3.5Cr-3.5V

The evaluation program included a study of tensile properties, room-temperature-property stability after exposure to elevated temperatures, fusion weldability, and a few flash-welding tests on certain alloys.

In addition to this more or less routine evaluation, various factors which were known or found to be detrimental to the properties of high-strength alloys were investigated. In this connection, it was found that residual hydrogen content may have a much more deleterious effect on the ductility of beta-stabilized alloys than has been recognized hitherto.

On the basis of the research conducted, the Ti-3Mn-complex alloy was selected as the most promising composition and an industrial evaluation program for this alloy was initiated.

Properties of Selected Alloys as Bar Stock

Excellent strength-ductility relationships were reported in WADC Technical Report 52-334 for 1/2-inch-diameter bar stock of three of the selected alloys solution treated at 1300 or 1400°F and overaged at 800 or 900°F. The same types of heat treatment have been applied to three additional selected alloys and the aging temperature range increased to 1100°F for all six alloys.

For most of the selected alloys, the relationships between tensile strength and ductility are about equal up to a strength level of 180,000 psi. Above this strength level, the Ti-3Mn-complex alloy had somewhat higher ductility for a given strength. This alloy could be heat treated to a range of strengths from 220,000 psi with 8 to 10 per cent elongation down to

(1) Will be referred to as the Ti-3Mn-complex alloy.
(2) Will be referred to as the Ti-4Fe-complex alloy.

135,000 psi with 30 per cent elongation. The Ti-3.5Cr-3.5V alloy had somewhat lower ductilities at all strength levels. The excellent relationships discussed above applied only to alloys rolled in the alpha-beta temperature range and solution treated in this range prior to aging. If solution treated in the beta temperature range, the ductility after aging was very low. The Ti-5Mn-2.5Cr and Ti-3.5Cr-3.5V alloys rolled at a temperature in the beta-phase region (1600°F) and heat treated had very low ductility, even at low strength levels. Under the same conditions, the Ti-3Mn-complex alloy had reasonably good ductility, although its ductility was lower than that obtained after rolling in the alpha-beta region (1450°F). A second heat of the Ti-3.5Cr-3.5V alloy made with high-purity ductile vanadium instead of the usual commercial grade also had good ductility after beta rolling and heat treating.

The section size of these alloys is important, because the cooling rate from the solution temperature must be fast enough to suppress transformation. Heat-treated 7/8-inch-diameter bar stock of the Ti-3Mn-complex and the Ti-3.5Cr-3.5Mn alloys had essentially the same strengths after aging as 1/2-inch-diameter bar stock of these two alloys. The elongations for the 7/8-inch-diameter bar stock were lower, presumably because they were measured over a 2-inch gage length, instead of the 1-inch gage length used for the 1/2-inch-round specimens. Thus, it appears that the selected alloys may be successfully heat treated in section sizes up to at least 7/8-inch diameter.

Initial experiments on isothermal aging of two alloys indicated lower ductility for the isothermally aged specimens than for specimens quenched and aged at the same temperature. The lower ductility obtained by isothermal aging may have occurred because the optimum time of aging for this type of heat treatment was not known. The aging times used were the same as those applied to quenched and aged specimens. The optimum aging time need not be the same for the isothermally aged and quenched and aged conditions.

The Ti-3Mn-complex, Ti-5Mn-2.5Cr, and Ti-3.5Cr-3.5V alloys, heat treated to a strength level of 190,000 to 215,000 psi, were exposed for periods up to 1000 hours to temperatures of 350 and 500°F. The room-temperature properties of all alloys were stable after exposure to 350°F. At 500°F, the Ti-5Mn-2.5Cr and Ti-3.5Cr-3.5V alloys were stable for 1000 hours, but the Ti-3Mn-complex alloy showed a considerable decrease in ductility after exposure periods longer than 500 hours. However, this alloy was heat treated to a higher strength level than the other two, which probably accounts for its lower stability.

A more complete evaluation of stability was made for the Ti-3Mn-complex, the Ti-4Fe-complex, and the Ti-3.5Cr-3.5Mn alloys. These alloys were heat treated to nominal strength levels of 200,000, 170,000, and 140,000 to 160,000 psi and exposed to temperatures of 500, 650, and 800°F for 200 and 1000 hours. The following changes in properties were specified arbitrarily as the limits for acceptable stability:

(1) Maximum variation in strength: ± 20 per cent of as-heat-treated value.

(2) Maximum variation in tensile elongation: ±40 per cent of as-heat-treated value.

(3) Minimum ductility at any strength level: 10 per cent elongation in 1 inch.

Using these limitations, the following conclusions on the stability of the alloys may be made:

(1) The three alloys heat treated to the 140,000- or 170,000-psi level were stable for 1000 hours at temperatures of 500 to 650°F.

(2) The two complex alloys at the same strength levels were stable for 1000 hours at 800°F.

(3) All of the alloys at the 200,000-psi level were unstable at temperatures as low as 500°F.

Properties of Selected Alloys as 14-Gage Sheet

The solution and aging treatments used on bar stock were applied to 14-gage sheet material of four of the selected alloys. As solution treated, the sheet ductilities were a little lower than for bar stock. After aging, the tensile elongation of the sheet material was very low. The balance of the year's work on sheet alloys was spent trying to determine the cause of this low ductility.

The factors investigated which appeared to affect sheet ductility were as follows:

(1) Surface contamination. Removal of about 0.002 inch from the surfaces of heat-treated sheet by etching improved ductility.

(2) Hydrogen content. Reduction of hydrogen content by vacuum annealing to about 220 ppm greatly improved the ductility of heat-treated sheet over that obtained in the same material having hydrogen contents over 400 ppm.

(3) Specimen geometry. Other conditions being equal, rounding the corners of the gage section greatly improved the ductility of heat-treated alloy sheet.

Heat-treated specimens containing the higher hydrogen contents had a third phase present at the interfaces between primary alpha particles and the beta matrix. This phase is believed to be a hydride, since it was not present in vacuum-treated, low-hydrogen material, but reappeared after the material was rehydrogenated.

Beta Embrittlement

Possible causes of the low ductility exhibited by beta-stabilized titanium alloys when fabricated or heat treated in the beta-phase region were investigated. A theory involving partition of the interstitial elements, carbon, oxygen, and nitrogen, between the alpha and the beta phases was tested. Alloys which had been beta-ized previously were re-solution treated in the alpha-beta field for times up to 48 hours prior to aging. No improvement in ductility resulted, indicating that a simple partition of these elements between the two phases did not explain the greater ductility of beta-stabilized alloys heat treated in the alpha-beta-phase region.

Experiments to link hydrogen with beta embrittlement were not successful because the hydrogen content of the test specimens was not reduced sufficiently during the vacuum treatment used.

It is possible that the alpha network existing in beta-embrittled alloys at the prior beta grain boundaries is solely responsible for the low ductility. This investigation will be continued under the extension to the present contract.

Development of a High-Strength Forging Alloy

Beta embrittlement restricts the forgeability of beta-stabilized titanium alloys because they must be finish worked at relatively low temperatures in order to retain ductility. Additions of as little as 0.5 per cent aluminum to raise the beta transus of the Ti-5Mn-2.5Cr alloy gave excellent strength-ductility combinations after rolling at 1600°F. This temperature is in the beta field of the base alloy, and resulted in very low ductilities in the base material. Aluminum additions up to 2.0 per cent to the Ti-3Mn-complex alloy did not improve its strength-ductility relationships over those obtained in the base alloy after rolling at 1600°F. However, the alloy had reasonably good ductility without the aluminum additions.

Weldability

Weldability evaluations were made on tungsten-arc-welded sheet and flash-welded bar stock of the six selected alloys. As welded, none of the specimens had any ductility. The Ti-5Mn-2Mo, Ti-4.5Fe-3.5Mo, Ti-4Fe-complex, and Ti-3.5Cr-3.5V alloys had improved ductility after solution treating at 1300°F and aging 8 hours at 1100°F. The longitudinal bend ductilities increased from greater than 15T to 4 to 7T for the 0.100-inch-thick sheet after this heat treatment. The strength for this condition ranged from 114,000 to 124,000 psi. A solution and aging treatment to produce a strength level of 170,000 psi did not improve the as-welded ductilities appreciably. Two other heat treatments, starting with a beta solution treatment, did not improve the properties over those obtained after the 1100°F aging treatment.

Industrial Evaluation Program

As part of the development program, one alloy, Ti-3Mn-complex, was selected for evaluation by various industrial organizations. In response to an inquiry from WADC, seventeen companies expressed a willingness to participate in this program. Of these, nine have been supplied with small amounts of bar stock from ingots melted at Battelle. The remaining companies will be supplied with bar stock from larger ingots melted at Battelle or from a 225-pound ingot melted by the Bureau of Mines, Albany, Oregon. This ingot was double arc melted by a consumable-electrode method. A section of this ingot was rolled to bar stock at Battelle. The tensile properties were comparable to those obtained on material previously melted as 10- or 20-pound ingots. The remainder of the ingot has been rolled to 7/8-inch bar stock by Mallory-Sharon Titanium, Incorporated, Niles, Ohio.

The results achieved on material which was sent out during the past year are as follows:

(1) Four gear-blank forgings were made at Wyman-Gordon Company from 20-pound ingots of this alloy. The blanks, about 7 inches in diameter and from

1/2 to 1 inch thick, had excellent die fill. After suitable heat treatments, tensile elongations of 12 to 15 per cent in both the tangential and the radial directions were obtained at a strength level of about 160,000 psi.

(2) Superior Tube Company has cold drawn tubing of the Ti-3Mn-complex and the Ti-5Mn-2.5Cr alloys to a total of 50 per cent reduction of area before cracking occurred. This reduction was made in steps at 10 to 15 per cent per pass, with annealing between passes. The annealing consisted of solution treating at 1300°F for 1 hour and aging 8 hours at 1100° F. On the basis of these results, it is believed that a lower alloy material with lower strength should be used for cold-drawn tubing.

(3) The A. O. Smith Corporation evaluated the flash welding of Ti-3Mn-complex alloy bar stock. Very low ductilities were obtained for most of the welded and heat-treated material. However, two specimens, heat treated to about 170,000 psi, fractured in the base metal after attaining a reasonably good tensile elongation.

(4) Wright Air Development Center determined the notch sensitivity of the Ti-3Mn-complex alloy. The Charpy impact values were determined at temperatures from -100 to 375°F. The specimens were heat treated initially to three strength levels. A value of 20 to 25 ft-lb was obtained at a strength of 141,000 psi at temperatures from -100 to 150°F. At a strength level of 191,000 or 212,000 psi, the Charpy impact value was about 9 ft-lb up to 200°F. The energy absorption increased sharply above these temperatures. The room-temperature, notched tensile ratio was found to decrease from 1.61 at the 141,000-psi strength level to 1.33 at the 212,000-psi strength level.

INTRODUCTION

This is the Summary Report on Contract No. AF 33(616)-384, covering the work done during the period from December 8, 1952, to December 7, 1953.

The alloy-development program described in this report is a continuation of the work carried out under Contracts Nos. W 33-038 ac-21229 and AF 33(038)-3736. The results of the work on these two contracts are contained in the following five reports:

Summary Report - Part III, July 30, 1949
AF Technical Report No. 6218-Part II, June 30, 1950
AF Technical Report No. 6623, June 18, 1951
WADC Technical Report No. 52-249, June 18, 1952
WADC Technical Report No. 52-334, December 31, 1952

The present contract provided for the further development of selected titanium alloys first investigated under Contract No. AF 33(038)-3736, which terminated December 31, 1952. The major purpose of this contract was the development of titanium-base alloys with improved strength-ductility relationships, heat treatability, or weldability, or a combination of these properties. In particular, the major effort was to be expended toward the development of alloys classified as F180H(1), S180H(1), and S130W(1) in the Tentative Air Force Classification of Titanium-Base Alloys required for aircraft applications. As part of the developmental work, considerable effort was also expended in melting and fabricating selected high-strength alloys for evaluation by aircraft companies and other industrial organizations. Arrangements for the evaluation work were made through personnel of the Materials Laboratory, Wright-Patterson Air Force Base.

The program set up under the previous contract for evaluation of selected high-strength alloys has been modified to include fewer compositions. The nominal compositions of the alloys tentatively selected for evaluation under the present contract are as follows:

Ti-3Mn-1Cr-1Fe-1Mo-1V
TI-4Fe-1Cr-1Mn-1Mo-1V
Ti-5Mn-2.5Cr
Ti-3.5Cr-3.5Mn
Ti-4.5Mo-3.5Fe
Ti-3.5Cr-3.5V

Particular emphasis was placed on the evaluation of the Ti-3Mn-complex alloy, since it is believed to be the most outstanding of the alloys thus far tested.

(1) In this nomenclature, F indicates a forging alloy and S a sheet alloy. H indicates heat treatable and W weldable. The numerals designate the approximate ultimate tensile strength.

EVALUATION OF SELECTED ALLOYS AS BAR STOCK

At the time that the subject contract was initiated, little was known about the heat treatment of high-strength titanium alloys. Such information is indispensable to the commercial utilization of this type of alloy in order that the desired properties may be attained consistently. The evaluation program described in this section was designed primarily to provide information on the heat treatability of high-strength alloys, and to determine specifically the effects of various heat treatments on the selected compositions.

Heat Treatment and Specimen Preparation

Data presented in WADC-TR-52-334 described the mechanical properties of three selected alloys given a new type of heat treatment. This heat treatment consisted of solution treating at 1300 to 1400°F, quenching, and overaging for relatively long periods of time at 800 or 900°F. It differs from conventional precipitation-hardening heat treatments in that overaging is required to produce good properties in titanium alloys, whereas overaging is usually undesirable in aluminum, magnesium, or other alloys. Outstanding mechanical properties have now been produced in the following alloys by this heat treatment:

Ti-3Mn-1Cr-1Fe-1Mo-1V
Ti-4Fe-1Cr-1Mn-1Mo-1V
Ti-5Mn-2.5Cr
Ti-3.5Cr-3.5Mn
Ti-4.5Mo-3.5Fe
Ti-3.5Cr-3.5V

These alloys were tested in the form of bar stock rolled to various diameters at 1450 or 1600°F. Duplicate specimens cut from the bar stock obtained from different locations in the original ingots were heat treated as follows:

(1) Solution treated for 1 hour at temperatures of 1300, 1400, or 1600°F, and quenched in cold water.

(2) Overaged for varying periods of time at temperatures of 800, 900, 1000, and 1100°F, followed by air cooling.

Heat-treated 1/2-inch-diameter bars were machined to standard 0.250-inch-diameter tensile specimens. Standard 0.500-inch-diameter tensile specimens were made from the 7/8-inch-diameter material.

Mechanical Properties

The tensile specimens were pulled on a Baldwin-Southwark Universal Testing Machine, using a platen speed of 0.02 inch per minute. Where sufficient ductility was indicated by the first specimen of each pair, the 0.2 per cent offset yield strength was determined on the second specimen using a Baldwin-Southwark-Emery stress-strain recorder and strain follower. In all cases, the tensile strengths of the duplicate specimens were reasonably consistent, so that the yield strength given should be fairly representative of both specimens. Average results of the tensile and hardness data are given in Tables 1 to 6 for the six alloys heat treated as 1/2-inch-diameter bar stock.

When rolled at 1450°F and then solution treated at 1300 or 1400°F and overaged, all of the alloys developed excellent strength-ductility combinations over a wide range of tensile strength. In general, the elongations ranged from about 10 per cent in 1 inch at the 190,000 to 200,000-psi strength level up to 30 per cent at the 130,000 to 140,000-psi strength level.

The ratio of yield strength to tensile strength in these tests varied from about 0.88 to 0.99. The Ti-5Mn-2.5Cr alloy and the Ti-4Fe-complex alloy exhibited constant ratios at all strength levels. In the other alloys, the ratio tended to decrease as the tensile strength increased. This trend indicates that aging at higher temperatures decreases the tensile strength more than the yield strength for certain alloys. Heat-treated steels usually show a reverse trend, i.e., the yield-strength ratio increases with increasing strength.

In most cases, the test bars solution treated at 1600°F and overaged were brittle. High strength levels were attained in many instances, but the specimens had zero ductility at hardness levels as low as 350 VHN. The brittleness engendered in titanium-base alloys by heating them in the beta-phase field has been recognized for some time, but no satisfactory explanation for this phenomenon has been advanced.

The data of Tables 1, 2, and 3 also show that, of the heats rolled at 1600°F (WT42A, WT51A, and WT107A), only the Ti-3Mn-complex alloy had any appreciable ductility after solution treating at 1300 or 1400°F and overaging. In the latter alloy, there was little difference in the ductility values obtained from the two rolling temperatures.

The effect of rolling temperature on the properties of the three alloys of Tables 1 to 3 is illustrated graphically in Figure 1, in which tensile strength is plotted against elongation for specimens rolled at the two temperatures, solution treated at 1300 or 1400°F, and overaged. The low ductility of the alloys rolled at 1600°F may be attributed to the same unknown cause

TABLE 1. TENSILE PROPERTIES AND HARDNESSES OF HEAT-TREATED 1/2-INCH-ROUND BARS OF THE Ti-3Mn-1Cr-1Fe-1Mo-1V ALLOY ROLLED AT TWO DIFFERENT TEMPERATURES

Heat No. (Actual Composition, %)	Solution Temp(3), °F	Aging Time, hr	Aging Temp, °F	Elongation(4), % in 1 inch	Reduction of Area(4), %	Yield Strength(5), 0.2% Offset, psi	Ultimate Tensile Strength(4), psi	VHN (10-Kg Load)(6)
WT107A(1)	1300	24	800	10.5	20.0	147,500	177,000	373
(2.95Mn, 1.39Cr, 0.99Fe,	"	48	"	10.5	14.0	141,000	170,000	357
0.92Mo, 0.98V)	"	8	900	22.5	30.5	140,000	152,500	342
	"	24	"	22.5	35.5	129,500	149,500	317
	1400	24	800	4.0	8.5	-	206,500	394
	"	48	"	4.5	7.5	185,500	200,500	397
	"	8	900	11.5	19.0	162,000	176,500	374
	"	24	"	13.5	29.0	150,000	167,000	360
	1600	24	800	0.5	1.0	-	234,000	437
	"	48	"	0.5	1.5	-	226,000	446
	"	8	900	0.5	2.5	-	205,000	409
	"	24	"	0.5	3.5	-	201,000	397
	"	48	"	0.5	3.0	-	188,000	373
WT136A(2)	1300	-	-	22.0(5)	47.0(5)	155,000	155,500(5)	347
(3.10Mn, 1.04Cr, 1.00Fe,	"	24	800	12.5	27.5	167,500	190,500	383
0.90Mo, 0.93V)	"	48	"	13.5	31.5	169,000	186,500	376
	"	8	900	20.0	44.5	149,500	161,500	333
	"	24	"	19.0	37.5	147,000	163,000	336
	"	2	1000	22.0	51.5	148,500	156,500	339
	"	8	"	25.0	56.5	137,000	145,500	325
	"	8	1100	31.5	59.5	130,500	134,000	302
	1400	-	-	10.5(5)	26.0(5)	188,500	189,000(5)	418
	"	24	800	4.0	7.0	-	221,500	435
	"	48	"	8.0	16.0	-	212,000	424
	"	8	900	17.5	42.5	171,500	177,000	370
	"	24	"	16.0	24.5	174,500	180,500	373
	"	2	1000	23.5	52.5	160,500	166,000	345
	"	8	"	24.0	50.5	150,500	154,000	327

TABLE 1. (Continued)

Heat No. (Actual Composition, %)	Solution Temp(3), °F	Aging Time, hr	Aging Temp, °F	Elongation(4), % in 1 inch	Reduction of Area(4), %	Yield Strength(5), 0.2% Offset, psi	Ultimate Tensile Strength(4), psi	VHN (10-Kg Load)(6)
WT136A(2)	1400	8	1100	30.0	58.5	136,000	141,000	312
(3.10Mn, 1.04Cr, 1.00Fe,	1600	-	-	-	-	-	145,000(5)	438
0.90Mo, 0.93 V)	"	24	800	0.5	0.5	-	234,500	446
	"	48	"	0.0	0.0	-	213,500	429
	"	8	900	0.5	3.0	-	202,500	383
	"	24	"	0.5	3.5	-	196,000	394
	"	2	1000	1.0	4.0	-	183,000	390
	"	8	"	5.5	10.0	-	159,000	349
	"	24	"	2.5	6.5	-	175,000	366
	"	8	1100	9.5	9.5	134,500	139,500	313

(1) Rolled at 1600°F.
(2) Rolled at 1450°F.
(3) Held 1 hour at temperature and cold water quenched.
(4) Average of two values.
(5) Single values.
(6) Average of three impressions on one specimen.

TABLE 2. TENSILE PROPERTIES AND HARDNESSES OF HEAT-TREATED 1/2-INCH-ROUND BARS OF THE Ti-5Mn-2.5Cr ALLOY ROLLED AT TWO DIFFERENT TEMPERATURES

Heat No. (Actual Composition, %)	Solution Temp[3], °F	Aging Time, hr	Aging Temp, °F	Elongation[4], % in 1 inch	Reduction of Area[4], %	Yield Strength[5], 0.2% Offset, psi	Ultimate Tensile Strength[4], psi	VHN (10-Kg Load)[6]
WT51A[1]	1300	48	800	0.0	0.0	-	116,500	409
(5.20Mn, 2.63Cr)	"	8	900	1.0	1.5	-	169,500	355
	"	24	"	0.0	0.0	-	111,500	348
	1400	8	"	0.0	0.0	-	189,000	397
	"	24	"	0.0	0.0	-	174,500	384
	1600	24	800	0.0	0.0	-	201,500	455
	"	48	"	0.0	0.0	-	208,000	442
	"	8	900	0.0	0.0	-	197,500	396
	"	24	"	0.0	0.0	-	173,500	394
	"	48	"	0.0	0.0	-	177,000	387
WT132A[2]	1300	-	-	22.0	45.0	153,500	156,000	350
(5.6Mn, 2.8Cr)	"	24	800	5.0	8.5	-	213,000	430
	"	48	"	2.5	7.5	-	206,500	394
	"	8	900	19.0	41.5	164,000	170,500	368
	"	24	"	10.0	14.5	164,500	173,000	369
	"	2	1000	20.0	42.0	161,000	165,000	354
	"	8	"	25.5	46.5	148,000	152,500	345
	"	8	1100	29.5	56.5	131,000	139,500	333
	1400	-	-	10.0	29.5	179,500	181,000	383
	"	24	800	3.0	4.5	-	239,500	458
	"	48	"	3.0	4.0	-	240,500	451
	"	8	900	18.5	45.5	-	183,500	380
	"	24	"	19.0	43.0	168,000	173,500	363
	"	2	1000	18.0	40.5	168,500	172,500	369
	"	8	"	20.0	39.0	155,500	159,000	357
	"	8	1100	30.0	54.0	141,000	147,500	343

TABLE 2. (Continued)

Heat No. (Actual Composition, %)	Solution Temp(3), °F	Aging Time, hr	Aging Temp, °F	Elongation(4), % in 1 inch	Reduction of Area(4), %	Yield Strength(5), 0.2% Offset, psi	Ultimate Tensile Strength(4), psi	VHN (10-Kg Load)(6)
WT132A(2)	1600	-	-	0.0(5)	0.0(5)	-	165,500(5)	401
(5.6Mn, 2.8Cr)	"	24	800	0.0	0.0	-	189,000	468
	"	48	"	0.0	0.0	-	180,000	463
	"	8	900	0.0	0.0	-	194,000	405
	"	24	"	0.0	0.0	-	195,000	397
	"	48	"	0.0	0.0	-	177,500	383
	"	2	1000	0.0	0.0	-	184,000	390
	"	8	"	2.5	6.0	-	158,500	357
	"	24	"	1.5	4.5	-	169,500	357
	"	8	1100	6.0	8.5	148,000	148,500	336

(1) Rolled at 1600°F.
(2) Rolled at 1450°F.
(3) Held 1 hour at temperature and cold water quenched.
(4) Average of two values.
(5) Single values.
(6) Average of three impressions on one specimen.

TABLE 3. TENSILE PROPERTIES AND HARDNESSES OF HEAT-TREATED 1/2-INCH-ROUND BARS OF THE Ti-3.5Cr-3.5 V ALLOY ROLLED AT TWO DIFFERENT TEMPERATURES

Heat No. (Actual Composition, %)	Solution Temp(3), °F	Aging Time, hr	Aging Temp, °F	Elongation(4), % in 1 inch	Reduction of Area(4), %	Yield Strength(5), 0.2% Offset, psi	Ultimate Tensile Strength(4), psi	VHN (10-Kg Load)(6)
WT42A(1)	1400	24	800	1.0	2.0	-	185,500	394
(4.62Cr, 3.34V(7))	"	48	"	0.0	0.0	-	188,000	397
	"	8	900	0.0	0.0	-	153,000	357
	"	24	"	0.0	0.0	-	149,000	339
	1600	24	800	0.0	0.0	-	197,000	409
	"	48	"	0.0	0.0	-	174,000	413
	"	8	900	0.0	0.0	-	150,500	371
	"	24	"	0.0	0.0	-	171,500	380
	"	48	"	0.0	0.0	-	139,000	359
WT125A(2)	1300	-	-	9.5	26.5	172,000	175,500	409
(4.50Cr, 3.22V(7))	"	24	800	10.0	15.5	-	172,500	387
	"	48	"	13.5	27.5	-	168,500	365
	"	8	900	11.5	18.0	-	155,000	379
	"	24	"	16.0	22.5	-	150,000	335
	"	2	1000	18.0	24.0	136,000	142,500	317
	"	8	"	24.0	40.0	127,000	132,500	297
	"	8	1100	29.5	56.0	119,500	126,500	278
	1400	-	-	-	-	-	175,000(5)	505
	"	24	800	4.0	7.5	-	190,500	388
	"	48	"	10.0	16.5	173,500	185,000	383
	"	2	1000	13.0	13.5	143,500	149,000	325
	"	8	"	17.0	27.0	136,500	141,000	307
	"	8	1100	26.5	42.0	126,500	129,500	294
	1600	-	-	-	-	-	94,500(5)	513
	"	24	800	0.0	0.0	-	210,000	420
	"	48	"	0.0	0.0	-	208,000	409
	"	8	900	0.0	0.0	-	174,500	386
	"	24	"	0.0	0.0	-	180,500	381

TABLE 3. (Continued)

Heat No. (Actual Composition, %)	Solution Temp(3), °F	Aging		Elongation(4), % in 1 inch	Reduction of Area(4), %	Yield Strength(5), 0.2% Offset, psi	Ultimate Tensile Strength(4), psi	VHN (10-Kg Load)(6)
		Time, hr	Temp, °F					
WT125A(2)	1600	48	900	0.0	0.0	-	156,500	363
(4.50Cr, 3.22V(7))	"	2	1000	2.0	4.0	-	175,000	366
	"	8	"	6.0	8.5	147,000	153,500	322
	"	24	"	0.0	2.0	-	158,000	342
	"	8	1100	18.0	29.5	126,500	133,500	299
WU76A(2)	1300	-	-	10.0	24.0	164,000	171,000	400
(3.5Cr, 3.5V(8,9))	"	8	900	19.0	40.5	127,500	148,000	333
	"	8	1100	27.0	60.5	109,500	118,000	289
	1400	-	-	-	-	-	160,000(5)	478
	"	24	800	12.0	25.5	178,000	184,000	376
	"	4	1000	20.5	57.0	140,000	150,000	325

(1) Rolled at 1600°F.
(2) Rolled at 1450°F.
(3) Held 1 hour at temperature and cold water quenched.
(4) Average of two values.
(5) Single values.
(6) Average of three impressions.
(7) Commercial vanadium (93 per cent).
(8) Nominal composition.
(9) Ductile vanadium (99.7 per cent).

TABLE 4. TENSILE PROPERTIES AND HARDNESSES OF HEAT-TREATED 1/2-INCH-ROUND BARS OF THE Ti-3.5Cr-3.5Mn ALLOY[1] ROLLED AT 1450°F

Solution Temp, °F	Cooling Medium From Solution Temp	Aging Time, hr	Aging Temp, °F	Elongation[2], % in 1 inch	Reduction of Area[2], %	Yield Strength[3], 0.2% Offset, psi	Ultimate Tensile Strength[2], psi	VHN (10-Kg Load)[4]
1300	Cold water	-	-	18.0	9.0	153,500	156,500	354
"	Air	-	-	20.0	36.5	139,500	153,000	335
"	Cold water	24	800	10.5	13.5	189,000	200,500	417
"	Ditto	48	"	7.5	14.5	173,000	195,500	397
"	"	8	900	17.5	27.5	159,000	170,500	373
"	"	24	"	18.5	30.5	159,500	170,000	370
"	"	48	"	10.0	14.5	158,000	167,500	360
"	"	0.5	1000	17.5	42.5	164,000	171,000	376
"	"	2	"	23.5	50.0	153,000	160,000	353
"	"	8	1100	30.5	53.5	132,500	138,000	309
"	"	24	"	26.5	42.5	127,500	133,500	306
1400	Cold water	-	-	9.5	29.5	174,500	177,000	357
"	Air	-	-	3.0	5.0	-	193,500	413
"	Cold water	24	800	4.5	9.5	208,000	211,500	433
"	Ditto	48	"	4.5	8.0	209,000	213,500	421
"	"	8	900	13.0	26.5	184,500	191,000	383
"	"	24	"	9.5	13.0	173,000	180,000	375
"	"	48	"	9.0	12.5	169,500	176,500	370
"	"	0.5	1000	14.0	32.5	178,500	183,500	392
"	"	2	"	18.0	42.5	160,000	168,500	351
"	"	8	1100	28.5	50.5	138,500	140,000	327
"	"	24	"	26.5	40.0	133,000	135,500	325

(1) Heat No. WT181A; actual composition 3.3Cr-3.3Mn. Heat treatments were made on 1/2-inch-diameter bar stock.
(2) Average of two values.
(3) Single values.
(4) Average of three impressions.

TABLE 5. TENSILE PROPERTIES AND HARDNESSES OF HEAT-TREATED 1/2-INCH-ROUND BARS OF THE Ti-4Fe-1Cr-1Mn-1Mo-1V ALLOY(1) ROLLED AT 1450°F

Solution Temp, °F(2)	Aging Time, hr	Aging Temp, °F	Elongation(3), % in 1 inch	Reduction of Area(3), %	Yield Strength(4), 0.2% Offset, psi	Ultimate Tensile Strength(3), psi	VHN (10-Kg Load)(5)
1300	-	-	22.0	37.0	163,500	165,500	357
"	48	800	1.0	3.0	-	225,000	429
"	8	900	6.0	7.5	204,500	206,000	413
"	"	1000	19.0	32.5	177,500	179,000	383
"	"	1100	25.5	40.0	-	158,000	351
1400	-	-	14.5	29.5	177,500	178,000	360
"	48	800	0.0	1.0	-	227,000	473
"	8	900	1.0	2.5	-	225,000	455
"	"	1000	2.5	4.5	-	188,500	395
"	"	1100	23.5	38.0	161,000	164,500	369
1600	-	-	0.0	0.0	-	181,500	394
"	8	1000	0.0	6.0	-	197,000	418
"							

(1) Heat No. WT170A.
(2) Cold water quenched after 1 hour at solution temperature.
(3) Average of two specimens.
(4) Single values.
(5) Average of three impressions.

TABLE 6. TENSILE PROPERTIES AND HARDNESSES OF HEAT-TREATED 1/2-INCH-ROUND BARS OF THE Ti-4.5Mo-3.5Fe ALLOY(1) ROLLED AT 1450°F

Solution Temp(2), °F	Aging Time, hr	Aging Temp, °F	Elongation(3), % in 1 inch	Reduction of Area(3), %	Yield Strength(4), 0.2% Offset, psi	Ultimate Tensile Strength(3), psi	VHN (10-Kg Load)(5)
1300	-	-	26.0	57.5	148,500	150,500	351
"	24	800	2.0	3.0	-	197,000	414
"	8	900	18.0(4)	46.0(4)	152,500	175,000	366
"	"	1000	25.0	47.0	141,500	150,500	336
"	"	1100	29.5	49.5	138,000	139,500	319
1400	-	-	15.5	17.5	-	170,500	383
"	24	800	-	-	-	(6)	464
"	8	900	-	-	-	(7)	413
"	"	1000	16.0(4)	29.0(4)	-	167,500	370
"	"	1100	25.0	43.5	141,500	147,500	348

(1) Heat No. WU37A.
(2) Cold water quenched after 1 hour at temperature.
(3) Average of two values, except where noted.
(4) Single values.
(5) Average of three impressions.
(6) Specimens broke in radius at values of 187,500 and 181,500 psi.
(7) Specimens broke in threads at values of 197,000 and 201,500 psi.

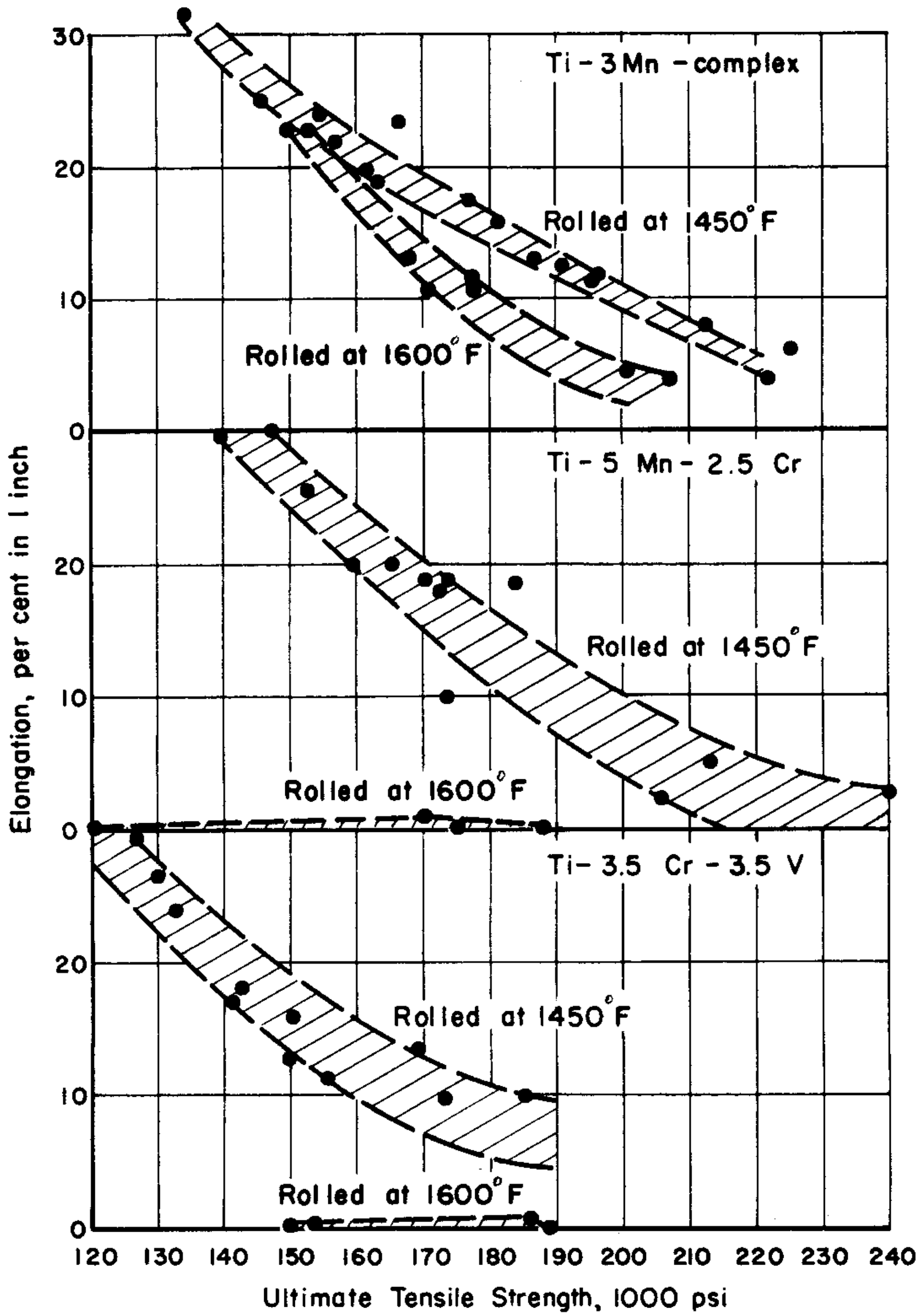

FIGURE 1. EFFECT OF ROLLING TEMPERATURE ON THE DUCTILITY OF THREE TITANIUM ALLOYS

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as may be observed in alloys solution treated at 1600°F and aged. Both may be classed as beta embrittlement.

Plots of tensile strength versus elongation for all six alloys rolled at 1450°F and heat treated are shown in Figures 2 and 3. The median curve of strength versus elongation for the Ti-3Mn-complex alloy has been drawn on the charts of each of the other alloys for comparison purposes. All of the alloys except the Ti-3.5Cr-3.5V alloy have about the same strength-ductility ratios at strengths up to about 190,000 psi. Above this strength, the Ti-3Mn-complex alloy has slightly higher elongations. The Ti-3.5Cr-3.5V alloy had elongation values about 7 or 8 per cent lower than those for the Ti-3Mn-complex for the same tensile strengths.

In Figure 3 (Ti-3.5Cr-3.5Mn alloy), there are six points which fall below the indicated tensile-strength - elongation range. These points represent the properties obtained after longer aging times of 48 hours at 800 or 900°F and 24 hours at 1100°F. Data points representing shorter aging times of 24 or 8 hours, respectively, at these temperatures fall within the indicated range. The low elongations for long aging times indicate a possible optimum aging time with respect to developing the best strength-ductility relationship in a given alloy. This phenomenon was not observed to be consistent for the other alloys and may apply only to the Ti-3.5Mn-3.5Cr alloy. The decrease in ductility for long aging times in this alloy may be due to precipitation of the $TiCr_2$ compound.

The excellent properties of these selected alloys are dependent upon the retention of the high-temperature beta phase by rapid cooling and then a controlled transformation to fine alpha nuclei at some lower temperature. The rate of cooling is dependent upon the section size, so that there is probably some limiting size that will retain the high-temperature phase throughout the section and produce maximum properties. A few experiments were made on the Ti-3Mn-complex and Ti-3.5Cr-3.5Mn alloys to determine the effect of a change in section size from 1/2-inch-diameter to 7/8-inch-diameter bar stock. The tensile properties and hardnesses of these specimens after various heat treatments are given in Table 7. The plots of elongation versus tensile strength for the two specimen sizes and the two alloys are shown in Figure 4. Except for the Ti-3.5Cr-3.5Mn alloy, quenched from 1300°F, the as-quenched strengths and hardnesses were slightly higher for the large-diameter bars. The higher strength values probably indicate slight aging during the quench due to the slower cooling rate. The as-aged tensile strengths and hardnesses were not significantly different for the 1/2-inch- and the 7/8-inch-diameter bars. The elongation and reduction-of-area values were appreciably lower for the 7/8-inch-diameter bars, but these differences may be a result of the specimen size alone. The elongation, of course, would be lower as measured over the 2-inch gage length used for the larger bars. It is apparent from these data that high-strength alloys of the type investigated under this contract may be successfully heat treated in section sizes up to at least 7/8-inch diameter.

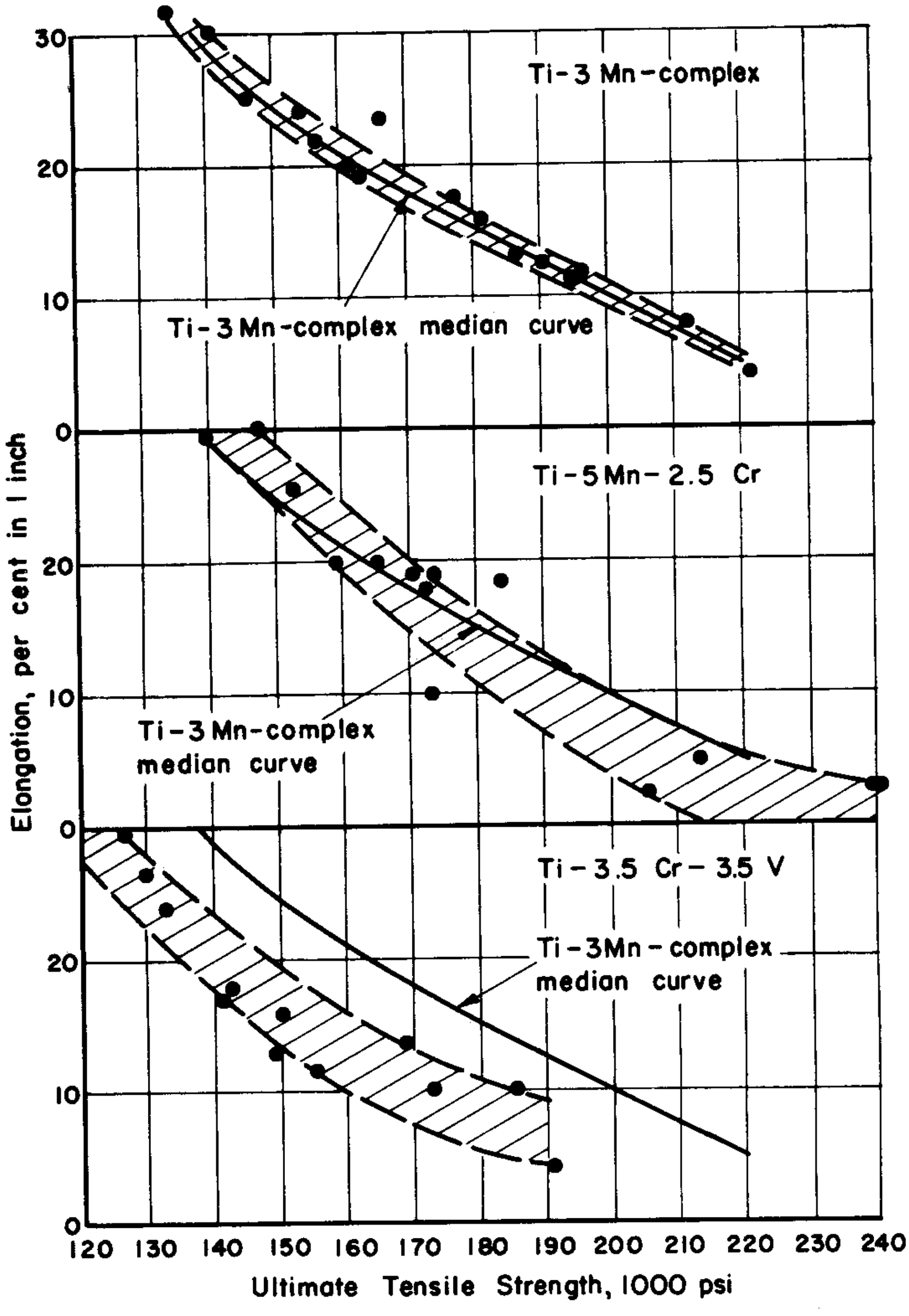

FIGURE 2. TENSILE STRENGTH VERSUS ELONGATION FOR SELECTED ALLOYS ROLLED AT 1450°F TO $\frac{1}{2}$-INCH-ROUND BAR STOCK AND HEAT TREATED

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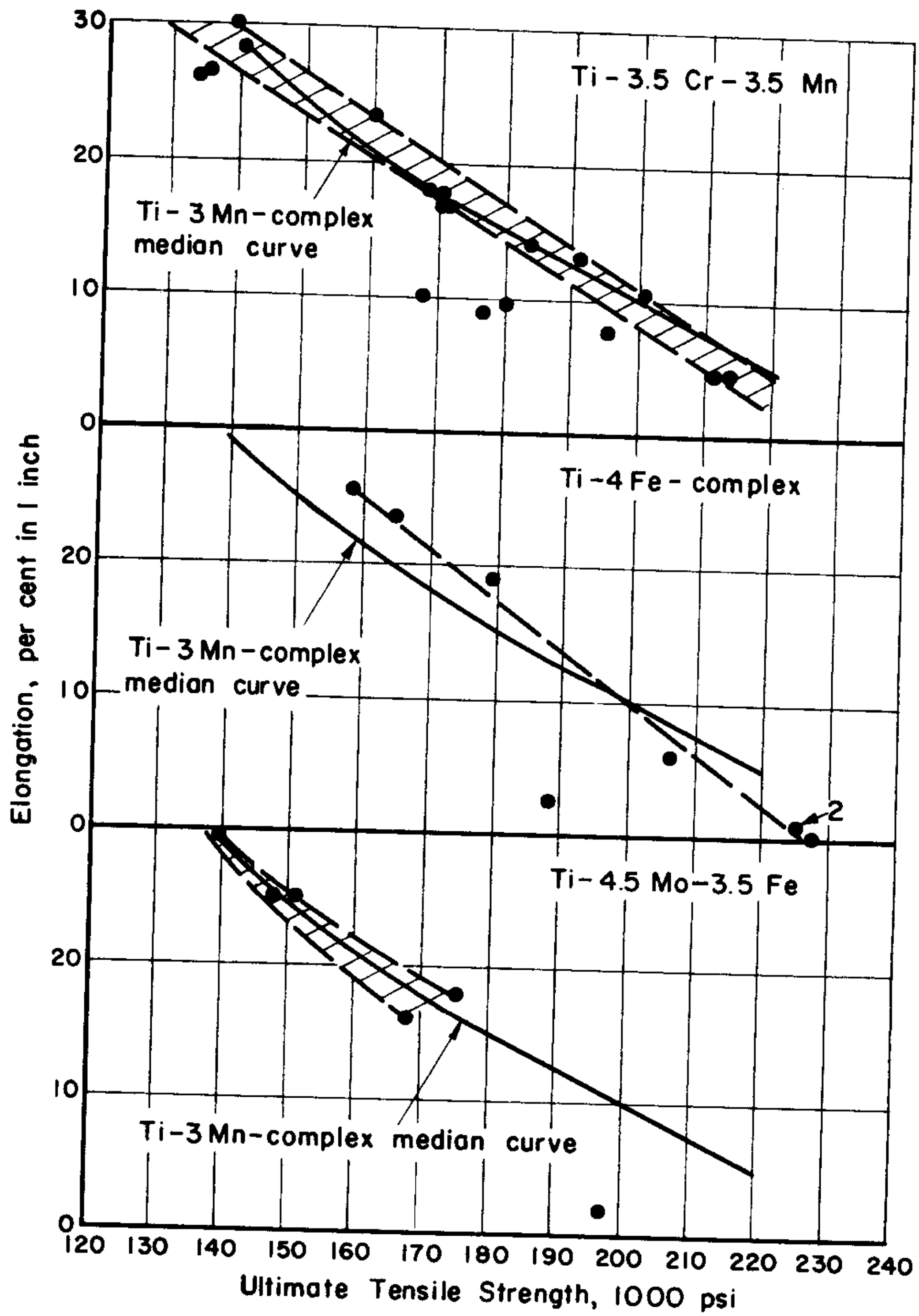

FIGURE 3. TENSILE STRENGTH VERSUS ELONGATION FOR SELECTED ALLOYS ROLLED AT 1450°F TO $\frac{1}{2}$-INCH-ROUND BAR STOCK AND HEAT TREATED

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TABLE 7. EFFECT OF SECTION SIZE ON HEAT-TREATMENT RESPONSE AS SHOWN BY TENSILE PROPERTIES AND HARDNESSES

Nominal Composition, per cent	Solution Temp(1), °F	Aging Time, hr	Aging Temp, °F	0.250-Inch Gage Diameter: Elongation(2), % in 1 inch	Reduction of Area(2), %	Yield Strength(2), 0.2% Offset, psi	Ultimate Tensile Strength(2), psi	VHN (10-Kg Load)(4)	0.505-Inch Gage Diameter: Elongation(2), % in 2 inches	Reduction of Area(3), %	Yield Strength(2), 0.2% Offset, psi	Ultimate Tensile Strength(2), psi	VHN (10-Kg Load)(4)
3Mn, 1Cr, 1Fe, 1Mo, 1V	1300	8	900	20.0	44.5	149,500	161,500	333	12.5	26.0	164,000	174,000	376
	"	"	1100	31.5	59.5	130,500	134,000	302	24.0	44.5	135,000	139,000	325
	1400	-	-	10.5	26.0	188,500	189,000	418	6.0	12.0	-	193,500	416
	"	24	800	4.0	7.0	-	221,500	435	2.0	3.0	-	219,000	425
	"	2	1000	23.5	52.5	160,500	167,500	345	16.0(5)	36.0(5)	-	170,500(5)	370(5)
3.5Cr, 3.5Mn	1300	-	-	18.0	39.0	153,500	156,500	354	20.5	41.5	-	146,500	338
	"	8	900	17.5	27.5	159,000	171,000	373	10.0	16.0	170,000	175,500	373
	"	"	1100	30.5	53.5	132,500	138,000	309	18.5	20.0	131,000	136,000	325
	1400	-	-	9.5	29.5	174,500	175,500	357	2.5	3.5	-	187,000	398
	"	24	800	4.5	9.5	208,000	211,500	433	1.5	2.5	-	216,000	429
	"	2	1000	18.0	42.5	160,000	168,500	351	11.0(5)	16.5(5)	-	165,000(5)	360(5)

(1) Cold water quenched after 1 hour at temperature.
(2) Average of two values.
(3) Single values.
(4) Average of three impressions.
(5) Aged 4 hours, instead of 2 hours.

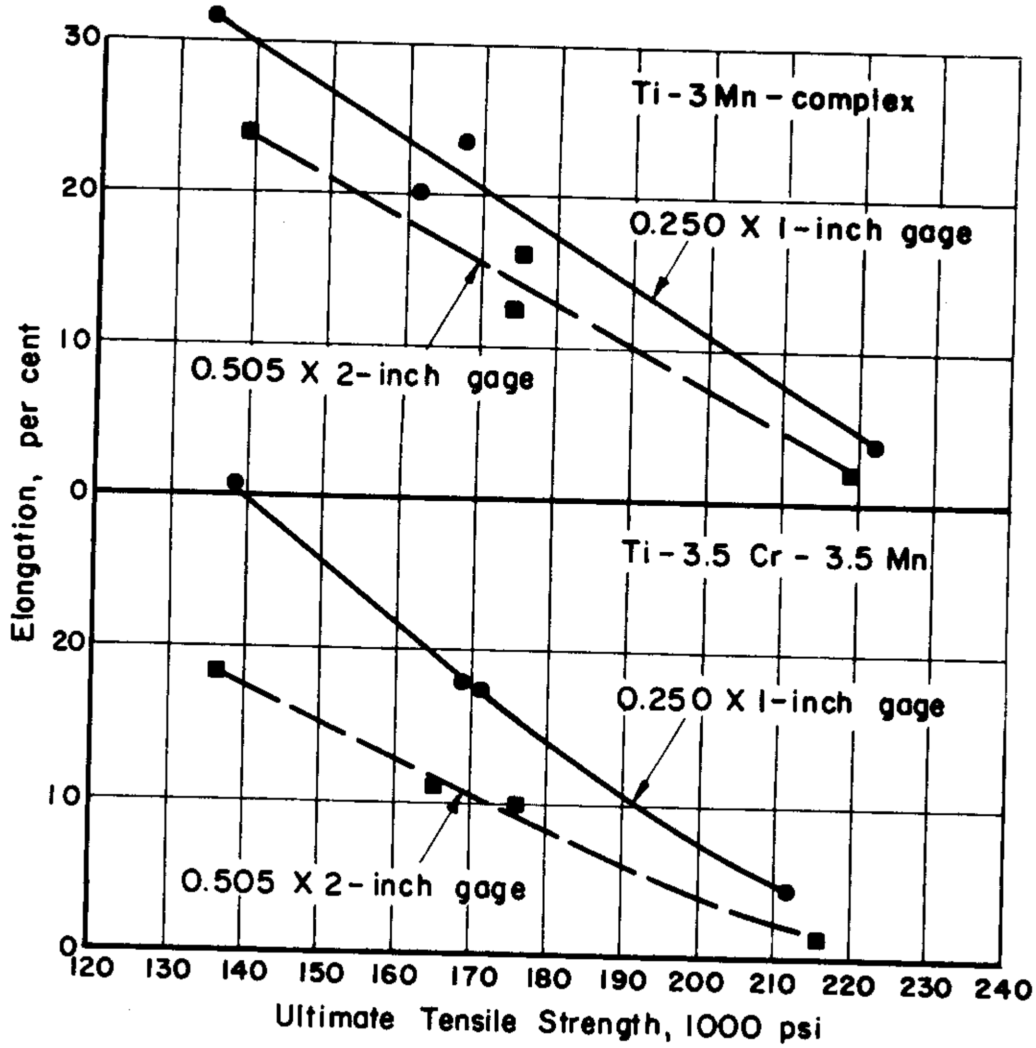

FIGURE 4. EFFECT OF SECTION SIZE ON TENSILE STRENGTH AND ELONGATION OF TWO HIGH-STRENGTH ALLOYS

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Microstructure

The selected alloys all showed good strength-ductility relationships after aging if they were both rolled and solution treated in the alpha-beta-phase region. Solution treating in the beta range decreased ductility severely in the four alloys to which this type of heat treatment was applied. There is no reason to believe that the two alloys not tested in this condition would exhibit better ductility. The Ti-5Mn-2.5Cr alloy and the Ti-3.5Cr-3.5V alloy rolled in the beta field, alpha-beta solution treated, and aged had very poor ductility. However, the Ti-3Mn-complex alloy in this condition had only slightly less ductility than material rolled in the alpha-beta-phase region.

All alloys were examined metallographically to correlate, wherever possible, microstructural changes and mechanical properties. Figures 5 and 6 typify the structure of alloys with the best strength-ductility relationship. These figures are of the Ti-5Mn-2.5Cr alloy rolled at 1450°F, solution treated at 1300°F, and overaged at 800 and 1100°F, respectively. After aging at 800°F, the microstructure consisted of equiaxed, uniformly distributed primary alpha and a fine, unresolved alpha precipitate in a beta matrix. The specimen of Figure 6, aged 8 hours at 1100°F, shows more clearly the precipitate of alpha in the beta matrix. Alloys solution treated at 1400°F had similar structures except for less primary alpha in the beta matrix.

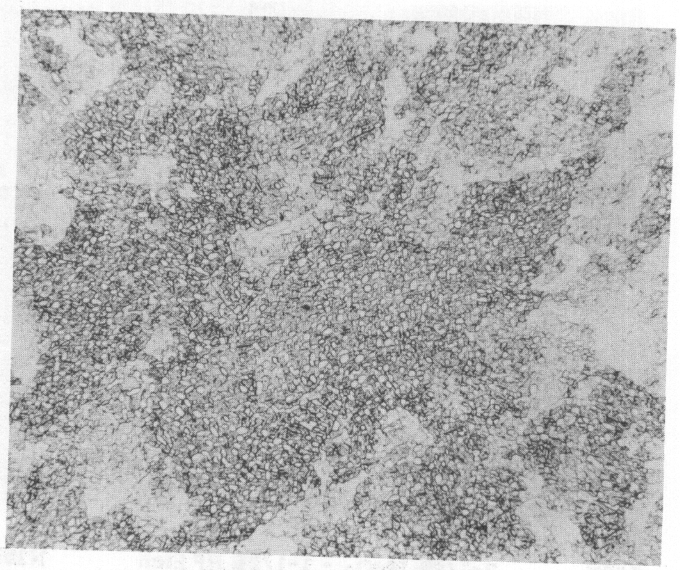
Figure 7 is representative of alloys rolled at 1600°F, solution treated at 1300 or 1400°F, and overaged. Under these conditions, the primary alpha is nearly continuous at the grain boundaries and exists in a Widmanstätten structure within the grains. In the Ti-3Mn-complex alloy, the alpha was not so continuous at the grain boundaries as in the other two alloys, which may account for its somewhat higher ductility. The fracture of the specimen of Figure 7 is shown in Figure 8. It may be noted that the fracture occurs essentially at the grain boundary and seems to be between the alpha and the beta matrixes rather than through the alpha phase. This suggests the possibility of an embrittling precipitate at the interface between the alpha and beta phases. The embrittling effect is intensified in this particular structure because of the continuous nature of the alpha at the grain boundaries.

The proposed precipitate may be the same material as the embrittling needlelike phase in the Ti-3.5Cr-3.5V alloy shown in Figure 9. This alloy was solution treated at 1600°F and aged 24 hours at 800°F. The dark, acicular material may be a hydride. As will be discussed later in this report, a similar phase was removed from sheet specimens by a vacuum-annealing treatment. The dark-etching phase is definitely associated with fracture, as shown in Figure 10, in which the fracture follows the needle-like phase. The phase was not apparent in specimens heated for 8 hours at

500X 3-1/2% HNO_3 - 1-1/2% HF Etch N73

FIGURE 5. Ti-5Mn-2.5Cr ALLOY ROLLED AT 1450°F, SOLUTION TREATED 1 HOUR AT 1300°F, AND AGED 48 HOURS AT 800°F

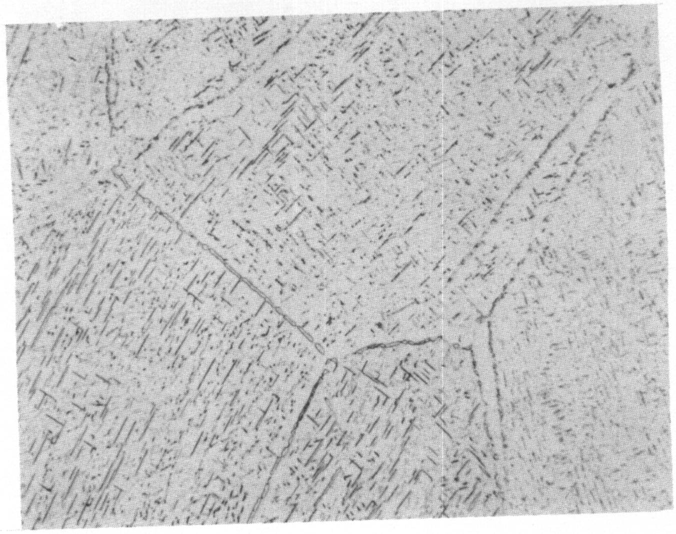
Tensile strength, 206,400 psi, 2.5% elongation in 1 inch
Equiaxed primary alpha and unresolved alpha precipitate in beta matrix



500X 3-1/2% HNO_3 - 1-1/2% HF Etch N74

FIGURE 6. Ti-5Mn-2.5Cr ALLOY ROLLED AT 1450°F, SOLUTION TREATED 1 HOUR AT 1300°F, AND AGED 8 HOURS AT 1100°F

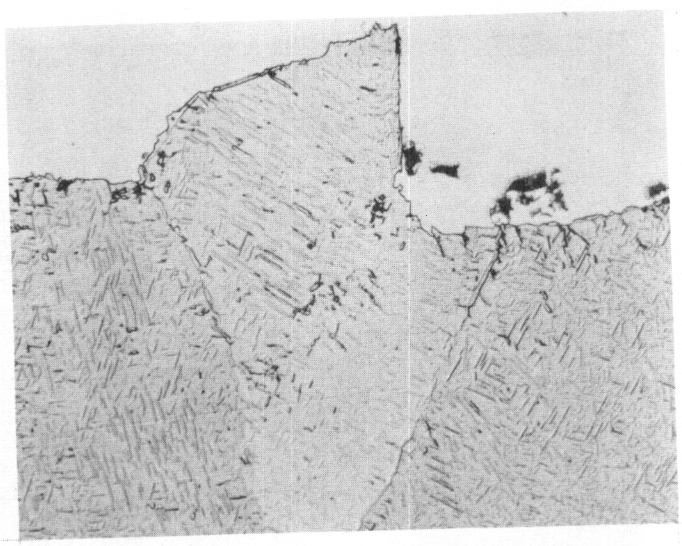
Tensile strength, 139,500 psi, 29.5% elongation in 1 inch
Same structure as in Figure 1 except that alpha precipitate (dark) is better developed



500X   3-1/2% HNO_3 - 1-1/2% HF Etch   N69

FIGURE 7. Ti-5Mn-2.5Cr ALLOY ROLLED AT 1600°F, SOLUTION TREATED AT 1300°F, AND AGED 8 HOURS AT 900°F

Tensile strength, 169,400 psi, 1.0% elongation in 1 inch
Primary alpha at grain boundaries and in Widmanstätten configuration and unresolved alpha precipitate in beta matrix



500X   3-1/2% HNO_3 - 1-1/2% HF Etch   N1591

FIGURE 8. FRACTURE OF STRUCTURE IN FIGURE 3

White is nickel plate

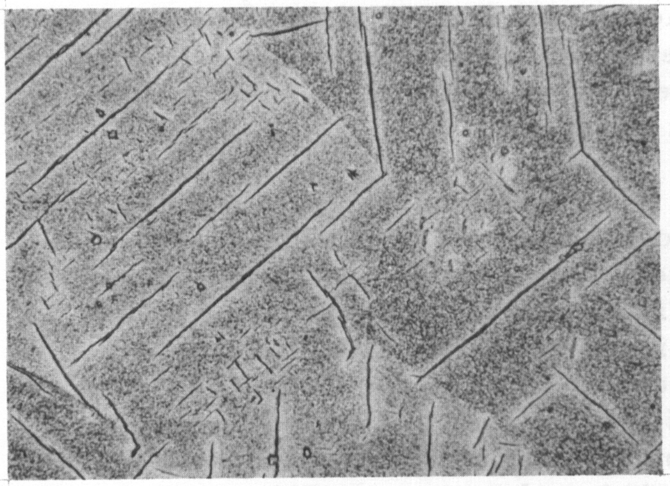


FIGURE 9. Ti-3.5Cr-3.5V ALLOY ROLLED AT 1450°F, SOLUTION TREATED 1 HOUR AT 1600°F, AND AGED 24 HOURS AT 800°F

Tensile strength, 210,000 psi,
0.0% elongation in 1 inch
Needles are unidentified phase which may be TiH; disperse phase is alpha in a beta matrix

500X 3-1/2% HNO_3 - 1-1/2% HF Etch N64

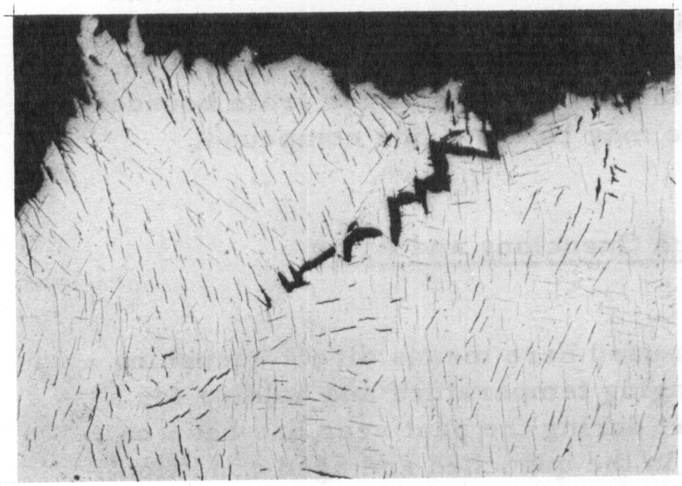


FIGURE 10. FRACTURE OF STRUCTURE IN FIGURE 9

100X 3-1/2% HNO_3 - 1-1/2% HF Etch N444

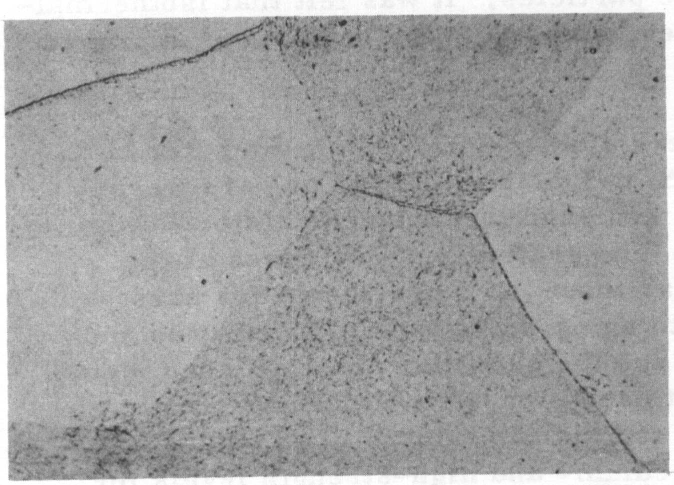


FIGURE 11. Ti-3.5Cr-3.5V ALLOY ROLLED AT 1450°F, SOLUTION TREATED 1 HOUR AT 1600°F, AND AGED 8 HOURS AT 1000°F

Tensile strength, 153,500 psi,
6.0% elongation in 1 inch
Needlelike phase not present
Note nearly continuous grain-boundary phase and alpha finely dispersed in beta matrix

500X 3-1/2% HNO_3 - 1-1/2% HF Etch N67

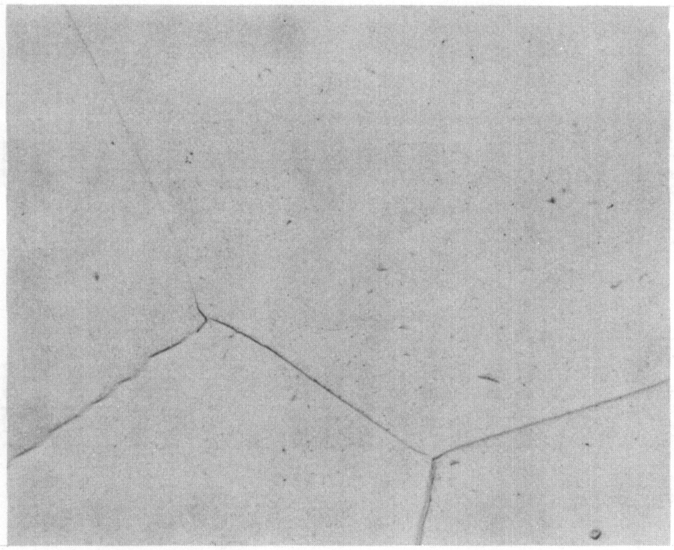
1000°F, as shown in Figure 11. If the specimens aged at 1000°F were then re-aged at 800°F for 24 hours, the needlelike phase reappeared. In spite of the disappearance of the unidentified phase in specimens aged at 1000°F, they showed only a slight increase in elongation.

In Figures 12 and 13, the third type of microstructure which had low ductility is shown. These photomicrographs are of the Ti-5Mn-2.5Cr alloy solution treated at 1600°F and overaged. The specimen of Figure 12 was aged 24 hours at 800°F, and that of Figure 13 was aged 8 hours at 1100°F. The precipitate size may be noted to increase for the 1100°F treatment. Although traces of the dark-etching needles found in the Ti-3.5Cr-3.5V alloy were present in the Ti-5Mn-2.5Cr alloy aged at 800°F, fracture of this specimen seemed to be intergranular. It may be noted from the microstructure that certain areas of the grain boundary seem to have a continuous phase. This continuous phase may be brittle, or, as indicated previously, it may have a precipitate at the interface of the beta causing the embrittlement. In general, these data indicate that the embrittlement of titanium alloys fabricated or solution treated in the beta field and given an overaging treatment may be due to the presence of an unknown phase or phases which appear either in needlelike form within the grains or at the grain boundaries. There is some evidence that the phase may be a hydride compound.

Isothermal Aging Versus Quenching and Aging

The term "isothermal aging" as used here means direct quenching from the solution temperature to the aging temperature and holding for a predetermined time. The major effort during the past year has been on evaluation of the high-strength alloys in the quenched and aged condition. The properties resulting from the aging-type heat treatment are dependent upon the size and distribution of the precipitated particles. A variation of the conditions existing before the aging temperature is reached could alter these by changing the nucleation of the particles. It was felt that isothermal-aging treatments should be investigated, since they might produce improved strength-ductility relationships.

Isothermal-aging treatments were applied to 1/2-inch-diameter bars of the Ti-3Mn-complex and the Ti-3.5Cr-3.5Mn alloys. The bars were held at the solution temperature for 1 hour and then quenched into a lead bath maintained at the selected aging temperature. Aging times at the various temperatures were the same as those used previously for bars quenched to room temperature prior to aging. The tensile properties and hardnesses for the isothermally aged and the quenched and aged specimens are given in Table 8. Tensile-strength and elongation values are plotted in Figure 14. The results were similar for the two alloys. The tensile elongation was appreciably lower at the medium- and high-strength levels for



500X N78

3-1/2% HNO_3 - 1-1/2% HF Etch

FIGURE 12. Ti-5Mn-2.5Cr ALLOY ROLLED AT 1450°F, SOLUTION TREATED 1 HOUR AT 1600°F, AND AGED 24 HOURS AT 800°F

Tensile strength, 189,000 psi,
0.0% elongation in 1 inch

500X N80

3-1/2% HNO_3 - 1-1/2% HF Etch

FIGURE 13. Ti-5Mn-2.5Cr ALLOY ROLLED AT 1450°F, SOLUTION TREATED 1 HOUR AT 1600°F, AND AGED 8 HOURS AT 1100°F

Tensile strength, 148,500 psi,
8.5% elongation in 1 inch

Note nearly continuous grain-boundary phase in Figures 12 and 13
Other structure is alpha finely dispersed in beta matrix

TABLE 8. EFFECT OF ISOTHERMAL AGING ON TENSILE PROPERTIES AND HARDNESSES OF TWO ALLOYS

Heat No. (Nominal Composition, %)	Solution Temp(1), °F	Aging Time, hr	Aging Temp, °F	Elongation(2), % in 1 inch	Reduction of Area(2), %	Yield Strength(3), 0.2% Offset, psi	Ultimate Tensile Strength(2), psi	VHN (10-Kg Load)(4)
				Cold Water Quenched and Aged				
WT136A	1300	48	800	13.5	26.5	169,000	186,500	376
(3Mn, 1Fe, 1Cr,	"	24	900	19.0	37.5	147,000	163,000	336
1Mo, 1V)	"	8	1000	25.0	56.5	137,000	145,500	325
	"	"	1100	31.5	59.5	130,500	134,000	302
	1400	48	800	8.0	16.0	205,000	212,000	424
	"	24	900	14.0	25.0	174,500	180,500	373
	"	8	1000	24.0	50.5	150,500	154,000	327
	"	"	1100	30.0	56.5	136,000	141,500	312
				Isothermal Aging (Quenched Into Lead Bath at Aging Temperature)				
WU24A	1300	48	800	8.5	9.0	160,000	179,000	375
(3Mn, 1Fe, 1Cr,	"	24	900	9.5	11.0	158,500	166,000	354
1Mo, 1V)	"	8	1000	29.0	50.0	139,000	144,000	318
	"	"	1100	31.0	54.0	133,500	138,000	319
	1400	48	800	4.0	4.5	-	199,000	397
	"	24	900	10.5	16.5	161,000	172,000	361
	"	8	1000	21.5	43.0	150,500	157,500	351
	"	"	1100	30.0	53.0	135,000	140,000	342
				Cold Water Quenched and Aged				
WT181A	1300	24	900	18.5	30.5	159,500	170,000	370
(3.5Cr, 3.5Mn)	"	8	1100	30.5	53.5	132,500	138,000	309
	1400	48	800	4.5	8.0	209,000	213,500	421
	"	2	1000	18.0	42.5	160,000	168,500	351

TABLE 8. (Continued)

Heat No. (Nominal Composition, %)	Solution Temp(1), °F	Aging Time, hr	Aging Temp, °F	Elongation(2), % in 1 inch	Reduction of Area(2), %	Yield Strength(3), 0.2% Offset, psi	Ultimate Tensile Strength(4), psi	VHN (10-Kg Load)(4)
				Isothermal Aging (Quenched Into Lead Bath at Aging Temperature)				
WT181A (3.5Cr, 3.5Mn)	1300	24	900	7.0	9.5	157,500	166,500	363
	"	8	1100	32.0	56.5	125,500	134,000	327
	1400	48	800	(5)	(5)	-	207,000	417
	"	8	1000	18.5	31.0	152,500	156,500	351

(1) Quenched after 1 hour at temperature.
(2) Average of two values.
(3) Single values.
(4) Average of three impressions.
(5) Specimen shattered into several pieces.

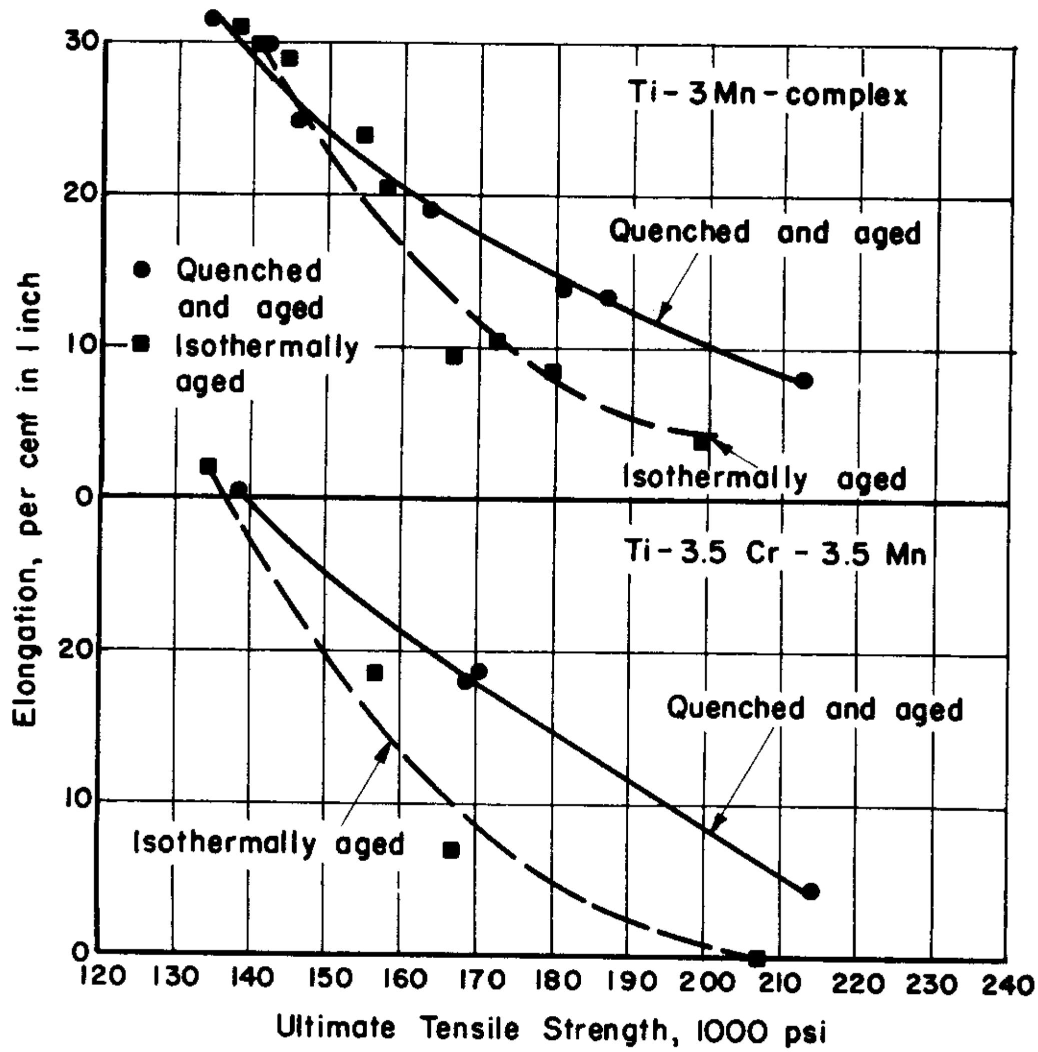


FIGURE 14. EFFECT OF ISOTHERMAL AGING VERSUS QUENCHING AND AGING ON THE STRENGTH-DUCTILITY RELATIONSHIP FOR TWO HIGH-STRENGTH ALLOYS

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the isothermally aged specimens. At a strength of about 140,000 psi, there was little difference in elongation for the two treatments. It should be recognized, however, that the isothermal treatments used may not have given the optimum strength-ductility relationships. It was found that, for the Ti-3.5Cr-3.5Mn alloy, there may be optimum times of aging for maximum properties. This may be true also for the isothermal treatments. Further work on isothermal aging will be done when the present contract is extended.

Thermal Stability of Heat-Treated Alloys

One of the major difficulties in the development of high-strength, beta-stabilized titanium alloys has been their instability at temperatures of 300°F and higher. Upon heating at these relatively low temperatures, an age-hardening reaction(1) takes place in the retained-beta phase which severely embrittles the material. The recently discovered overaging heat treatments produce a partial transformation of this retained beta and might be expected, therefore, to make alloys so treated more stable at temperatures lower than the overaging temperature.

To check this possibility, two series of stability tests were made. The Ti-3.5Cr-3.5V, Ti-5Mn-2.5Cr, and Ti-3Mn-complex alloys heat treated to a strength level of 190,000 to 215,000 psi were exposed at 350 and 500°F for 200, 500, and 1000 hours. Room-temperature tensile tests and hardnesses for these test specimens are given in Table 9, along with the as-heat-treated properties. The Ti-3.5Cr-3.5V alloy had a tensile strength of about 191,000 psi and an elongation of about 10 per cent after all treatments.

The Ti-5Mn-2.5Cr alloy was heat treated to a tensile strength of about 200,000 psi, with an elongation of 9.5 per cent in 1 inch. The specimens that were stability tested showed a slight increase in tensile strength and a decrease in ductility after 1000 hours at 500°F, but neither change was more than 5 per cent.

The Ti-3Mn-complex alloy had an initial as-heat-treated tensile strength of 214,000 psi, with an elongation of 9.5 per cent. No significant change in properties occurred after 1000 hours at 350°F. This alloy was also stable for at least 500 hours at 500°F. However, after 1000 hours at 500°F, a sharp decrease in elongation occurred. The higher strength level

(1) The age hardening of beta-stabilized alloys is believed to be due to the formation of a metastable transition phase, designated as "omega", first discovered in work under Contract No. AF 33(038)-3736. The mechanism of the age-hardening reaction is being investigated under Contract No. AF 33(616)-445.

TABLE 9. ELEVATED-TEMPERATURE STABILITY OF HEAT-TREATED TITANIUM ALLOYS(1)

Heat No. (Actual Composition, %)	Initial Heat Treatment(2)	Stability Treatment Temp, °F	Stability Treatment Time, hr	Elongation(3), % in 1 inch	Reduction of Area(3), %	Yield Strength(4), 0.2% Offset, psi	Ultimate Tensile Strength(3), psi	VHN (10-Kg Load)(5)
WT125A	1 hour at 1400°F,	-	-	9.0	20.5	-	192,000	392
(4.5Cr, 3.22V)	aged 24 hours	350	200	11.0	23.0	-	193,000	389
	at 800°F	"	500	10.0	20.5	184,500	190,000	373
		"	1000	9.0	16.0	184,000	191,500	376
		500	200	9.0	20.5	-	191,600	397
		"	500	11.5	19.5	186,500	193,500	376
		"	1000	12.5	26.5	186,500	192,000	394
WT132A	1 hour at 1400°F,	-	-	9.5	25.0	-	200,000	409
(5.6Mn, 2.8Cr)	aged 8 hours	350	200	11.5	19.0	200,000	204,000	397
	at 900°F	"	500	9.5	17.0	204,500	205,000	413
		"	1000	8.0	12.5	200,000	204,000	397
		500	200	5.5	7.5	-	204,500	405
		"	500	9.0	17.0	-	203,500	335
		"	1000	9.0	15.5	200,000	202,500	350
WT136A	1 hour at 1400°F,	-	-	9.5	21.5	205,000	214,000	421
(1.04CR, 1.00Fe,	aged 48 hours	350	200	8.5	15.0	209,500	215,000	421
3.10Mn, 0.90Mo,	at 800°F	"	500	7.5	10.0	212,000	215,000	377
0.93V)		"	1000	7.0	17.5	218,000	218,500	412
		500	200	10.0	23.0	214,500	215,500	425
		"	500	9.0	18.0	217,000	217,000	399
		"	1000	4.0	9.0	-	217,000	413

(1) Stability tests made on 1/2-inch-round bars rolled from 1450°F. All tensile tests made at room temperature.
(2) Specimens quenched in cold water after 1400°F solution treatment.
(3) Average of two values.
(4) Single values.
(5) Average of three impressions on one specimen.

of this alloy, compared with the other two, is probably a factor in its lower stability.

The promising results of these initial stability tests indicated the need for additional data for higher temperatures and lower strength levels. Therefore, stability tests were made on the Ti-3Mn-complex, Ti-4Fe-complex, and Ti-3.5Cr-3.5Mn alloys, each heat treated to the 200,000, 170,000, and 140,000 to 160,000-psi strength levels. After the initial heat treatment, specimens were exposed for 200 and 1000 hours at 500, 650, and 800°F. Thermal stability was evaluated by changes in room-temperature hardness and tensile properties. Details of the heat treatments used and the results of the stability tests are given in Tables 10, 11, and 12. The ultimate-tensile-strength and elongation values are plotted as a function of exposure time in Figures 15, 16, and 17.

The stability-test data are for single specimens and, therefore, cannot be considered typical. However, the curves indicate trends which undoubtedly would be approximated in repeated tests. The Ti-3Mn-complex alloy, Figure 15, indicates, in general, an increase in strength after 200 hours at the stability-testing temperatures. However, the as-heat-treated and stability-tested bars were from different heats of the alloy and probably had somewhat different strengths as heat treated. The as-heat-treated and stability-tested bars of the Ti-4Fe-complex and the Ti-3.5Cr-3.5Mn alloys were from the same heats.

For convenience in interpreting the results of the stability tests, and because design specifications usually allow some change in properties, the following limiting changes were established arbitrarily as representing a reasonable degree of stability after 1000 hours at the various temperatures:

(1) Maximum variation in strength: ±20 per cent of original value.

(2) Ductility expressed as tensile elongation within ±40 per cent of original value.

(3) Minimum ductility at any strength level: 10 per cent elongation in 1 inch.

Under these limitations, the three alloys may be classified as stable at the temperatures and strength levels given on page 44. Note that none of the alloys were stable at the 200,000-psi strength level, even at an exposure temperature as low as 500°F. This is undoubtedly due to further aging of the retained-beta phase present at this strength level. The instability of the Ti-3.5Cr-3.5Mn alloy at 800°F probably was associated with the precipitation of the $TiCr_2$ compound from the retained-beta phase.

TABLE 10. EFFECT OF ELEVATED-TEMPERATURE EXPOSURE ON ROOM-TEMPERATURE PROPERTIES OF THE Ti-3Mn-1Cr-1Fe-1Mo-1V[1] ALLOY

Heat Treatment			Stability Test[3]						
Solution Temp[2], °F	Aging[3] Time, hr	Aging[3] Temp, °F	Time, hr	Temp, °F	Elongation[4], % in 1 inch	Reduction of Area[4], %	Yield Strength[5], 0.2% Offset, psi	Ultimate Tensile Strength[4], psi	VHN (10-Kg Load)[5]
1300	24	800	-	-	12.5[6]	27.0[6]	167,500	190,500[6]	383
"	"	"	200	500	6.0[7]	-	-	203,000	405
"	"	1000	1000	"	3.0	6.0	180,500	193,500	410
"	"	800	200	650	6.0[7]	-	-	201,500	405
"	"	"	1000	"	7.0	14.0	175,000	197,500	417
"	"	"	200	800	4.0	14.0	161,000	168,000	366
"	"	"	1000	"	15.0	25.0	155,000	159,000	360
1400	2	1000	-	-	23.5[6]	52.0[6]	160,500	166,000[6]	345
"	"	"	200	500	22.0	51.5	166,500	171,500	383
"	"	"	1000	"	15.0	16.0	165,000	172,000	385
"	"	"	200	650	17.0	29.5	161,500	171,500	370
"	"	"	1000	"	17.5	46.5	162,000	172,000	380
"	"	"	200	800	16.0	31.5	146,500	175,000	376
"	"	"	1000	"	18.0	29.5	150,500	159,500	363
1400	8	1100	-	-	30.0[6]	58.5[6]	136,000	141,500[6]	312
"	"	"	200	500	26.0	43.0	140,000	145,500	342
"	"	"	1000	"	28.5	52.0	139,500	145,500	342
"	"	"	200	650	28.0	54.0	140,500	145,500	336
"	"	"	1000	"	28.0	49.5	140,000	147,000	343
"	"	"	200	800	29.0	41.5	133,500	141,000	336
"	"	"	1000	"	21.0	24.0	135,000	141,000	325

(1) Heat-treated specimens, Heat No. WT136 A; stability specimens, Heat No. WU24A.
(2) Cold water quenched after 1 hour at temperature.
(3) Air cooled from treatment temperature.
(4) Single values.
(5) Average of three impressions.
(6) Average of two values.
(7) Elongation for 1/2-inch-gage length on 1/8-inch-diameter specimens remachined from 1/4-inch-diameter specimens that broke in threads.

TABLE 11. EFFECT OF ELEVATED-TEMPERATURE EXPOSURE ON ROOM-TEMPERATURE PROPERTIES OF THE Ti-4Fe-1Cr-1Mn-1Mo-IV[1] ALLOY

Heat Treatment			Stability Test[3]						
Solution	Aging[3]							Ultimate	
Temp[2], °F	Time, hr	Temp, °F	Time, hr	Temp, °F	Elongation[4], % in 1 inch	Reduction of Area[4], %	Yield Strength[4], 0.2% Offset, psi	Tensile Strength[4], psi	VHN (10-Kg Load)[5]
1300	8	900	-	-	6.0[6]	7.5[6]	204,500	206,000	413
"	"	"	200	500	4.0[7]	-	193,000	199,500	409
"	"	"	1000	"	1.5	3.0	189,500	197,500	425
"	"	"	200	650	2.0	5.0	189,500	197,500	413
"	"	"	1000	"	4.5	4.0	-	199,500	429
"	"	"	200	800	0.0[7]	-	-	190,000	390
"	"	"	1000	"	4.0	7.0	169,000	175,000	388
1300	8	1000	-	-	19.0[6]	32.5[6]	177,500	179,000[6]	383
"	"	"	200	500	23.0	34.0	173,000	173,000	387
"	"	"	1000	"	21.0	35.5	167,000	172,000	390
"	"	"	200	650	5.0	-	171,000	171,500	376
"	"	"	1000	"	20.0	32.0	169,500	174,500	397
"	"	"	200	800	12.0	14.5	-	170,500	380
"	"	"	1000	"	14.0	15.5	163,000	167,000	374
1300	8	1100	-	-	25.5[6]	40.0[6]	-	159,000[6]	351
"	"	"	200	500	26.0	39.0	153,500	156,500	360
"	"	"	1000	"	28.0	39.5	151,500	159,000	364
"	"	"	200	650	26.0	35.5	154,500	157,500	366
"	"	"	1000	"	24.5	36.5	152,500	158,000	370
"	"	"	200	800	15.0	15.5	146,000	152,500	357
"	"	"	1000	"	25.0	27.5	147,500	152,000	351

(1) Heat No. WT170A.
(2) Cold water quenched after 1 hour at temperature.
(3) Air cooled from treatment temperature.
(4) Single values.
(5) Average of three impressions.
(6) Average of two values.
(7) Uniform elongation; specimens broke in shoulder.

TABLE 12. EFFECT OF ELEVATED-TEMPERATURE EXPOSURE ON ROOM-TEMPERATURE PROPERTIES OF THE Ti-3.5Cr-3.5Mn(1) ALLOY

Heat Treatment: Solution Temp(2), °F	Heat Treatment: Aging(3) Time, hr	Heat Treatment: Aging(3) Temp, °F	Stability Test(3) Time, hr	Stability Test(3) Temp, °F	Elongation(4), % in 1 inch	Reduction of Area(4), %	Yield Strength(4), 0.2% Offset, psi	Ultimate Tensile Strength(4), psi	VHN (10-Kg Load)(5)
1300	24	800	-	-	10.5(6)	13.5(6)	189,000	200,500(6)	417
"	"	"	200	500	8.0	11.0	185,500	200,000	421
"	"	"	1000	"	5.5	9.5	185,000	201,500	422
"	"	"	200	650	2.0	7.0	187,000	197,000	417
"	"	"	1000	"	5.0	9.5	-	203,000	428
"	"	"	200	800	0.0	0.0	-	183,500	401
"	"	"	1000	"	-	-	-	186,000	397
1300	8	900	-	-	17.0(6)	27.5(6)	159,000	170,500(6)	373
"	"	"	200	500	17.5	-	-	175,000	-
"	"	"	1000	"	16.0	20.5	163,000	175,000	390
"	"	"	200	650	12.0	15.0	163,500	174,500	390
"	"	"	1000	"	15.5	29.5	165,000	175,500	387
"	"	"	200	800	0.0(7)	-	161,000	168,500	376
"	"	"	1000	"	-	-	141,500	165,500	368
1400	8	1100	-	-	28.5(6)	50.5(6)	138,500	140,000(6)	327
"	"	"	200	500	30.0	51.5	136,000	143,000	336
"	"	"	1000	"	26.0	45.5	134,500	142,500	345
"	"	"	200	650	30.0	51.5	139,500	144,000	339
"	"	"	1000	"	28.0	45.5	141,000	147,000	354
"	"	"	200	800	13.0	14.5	131,000	140,500	330
"	"	"	1000	"	1.0	0.0	130,000	137,500	327

(1) Heat No. WT181A.
(2) Cold water quenched after 1 hour at temperature.
(3) Air cooled from treatment temperature.
(4) Single values.
(5) Average of three impressions.
(6) Average of two values.
(7) Uniform elongation; specimen broke in shoulder.

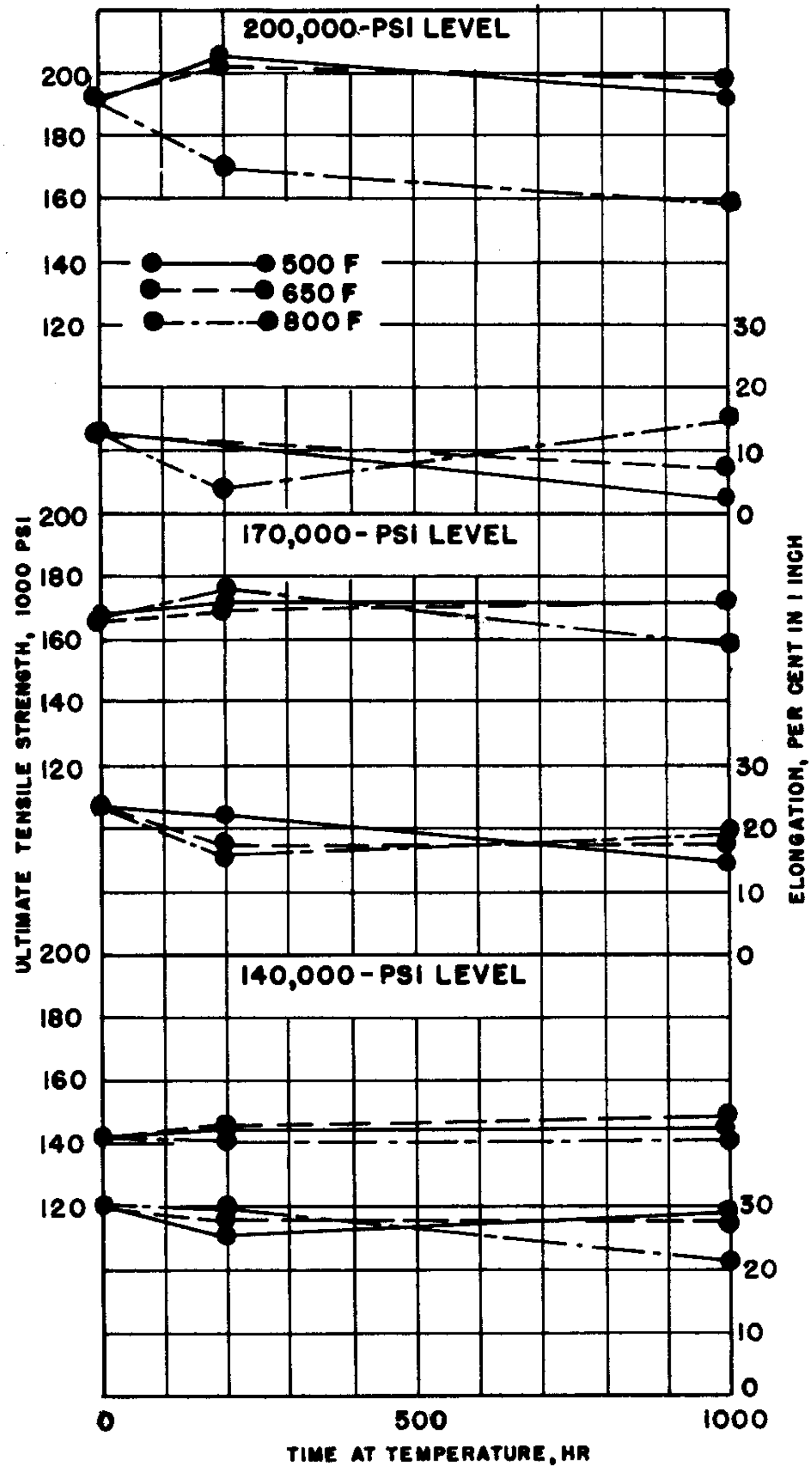


FIGURE 15. EFFECT OF ELEVATED-TEMPERATURE EXPOSURE ON ROOM-TEMPUREATURE TENSILE STRENGTH AND ELONGATION OF THE TI-3 MN-COMPLEX ALLOY A-8544

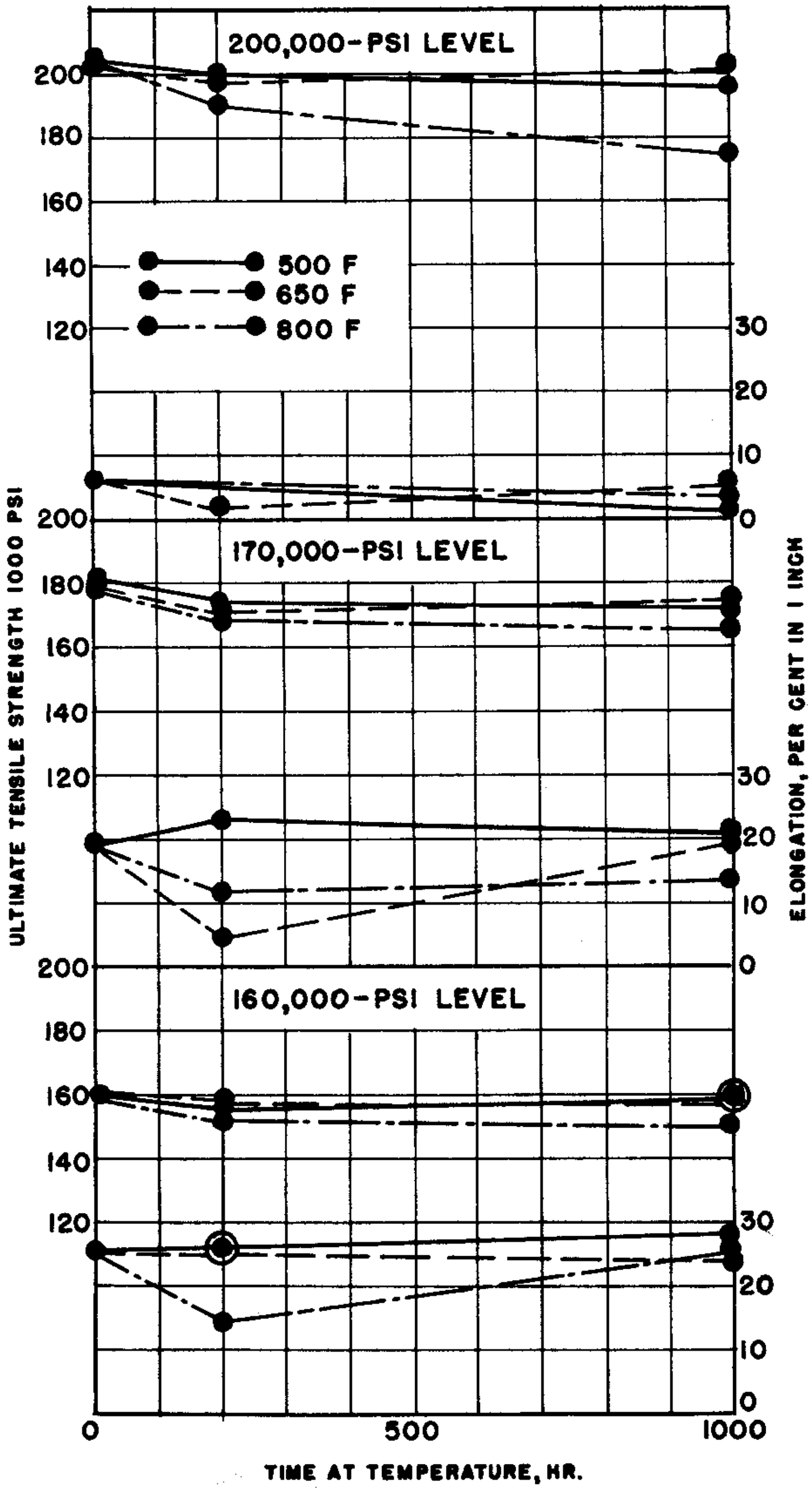


FIGURE 16. EFFECT OF ELEVATED TEMPERATURE EXPOSURE ON ROOM-TEMPERATURE TENSILE STRENGTH AND ELONGATION OF THE TI-4FE-COMPLEX ALLOY A-8545

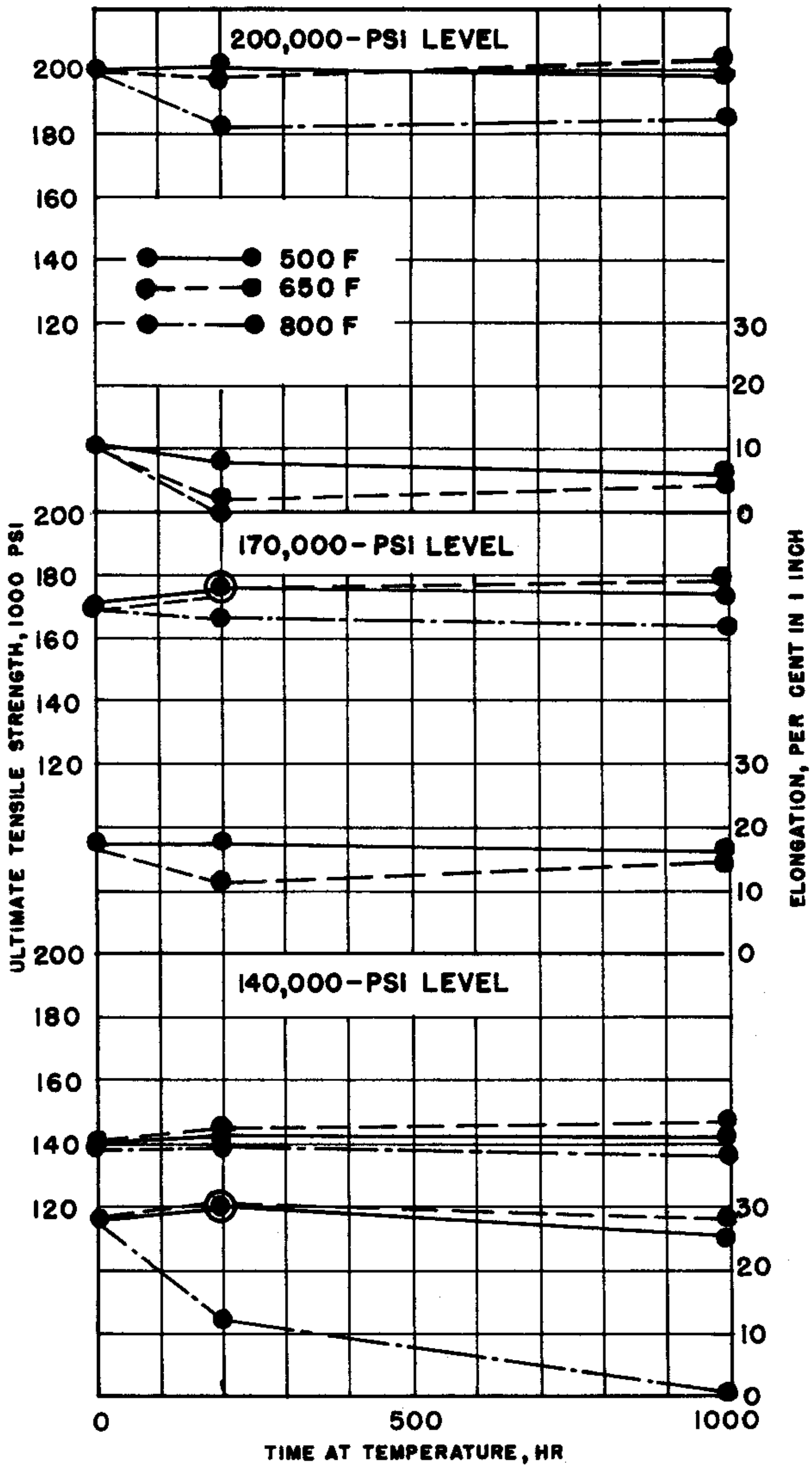


FIGURE 17. EFFECT OF ELEVATED-TEMPERATURE EXPOSURE ON ROOM-TEMPERATURE TENSILE STRENGTH AND ELONGATION OF THE TI-3.5 CR-3.5 MN. ALLOY

A-8546

Alloy	Strength Level, psi	Stable for 1000 Hours at
Ti-3Mn-complex	170,000	500, 650, 800°F
"	140,000	500, 650, 800°F
Ti-4Fe complex	180,000	500, 650, 800°F
"	160,000	500, 650, 800°F
Ti-3.5Cr-3.5Mn	170,000	500, 650°F
"	140,000	500, 650°F

It may be noted that several of the curves of elongation versus time show an initial decrease in elongation after 200 hours' exposure time and then an increase after 1000 hours (for example, the elongation curve for the Ti-4Fe-complex alloy heat treated initially to the 170,000 -psi level and tested at 650°F, Figure 16). However, a few of the curves show an initial increase in elongation and then a decrease, as in the case of the same alloy given the same initial heat treatment, but exposed at 500°F. These results may be attributed to additional aging of the beta phase or to some unknown precipitation reaction. This precipitation reaction may produce initially fine embrittling particles, which agglomerate and restore ductility with further aging. Thus, longer initial aging treatments might produce more nearly equilibrium conditions and, therefore, result in greater stability at elevated temperatures, probably at a sacrifice of some strength. In most cases, however, the reactions had little effect on the strength of the material, as the tensile curves do not show the same deviations that the elongation curves do.

EVALUATION OF SELECTED ALLOYS AS 14-GAGE SHEET

A major objective of the research program was the development of a high-strength, heat-treatable sheet alloy. Bar stock of the selected alloys were shown in the preceding sections to have excellent properties after solution and aging heat treatments. To determine the suitability of these alloys for sheet material, the four compositions listed below were rolled to 14-gage sheet at 1450°F and tested after being heat treated in the same manner as the bar stock:

Ti-3.5Cr-3.5V
Ti-5Mn-2.5Cr
Ti-3Mn-1Cr-1Fe-1Mo-1V
Ti-3.5Cr-3.5Mn

Heat Treatment and Specimen Preparation

All sheet was pickled in an ammonium fluoride - H_2SO_4 - water solution to remove the as-rolled scale, and sheared into 1/2 by 5-inch strips. The Ti-3.5Cr-3.5V, Ti-5Mn-2.5Cr, and Ti-3Mn-complex alloys were then given solution and overaging heat treatments of the type previously described for bar stock. The specimens were heated in a purified argon atmosphere to temperatures of 1300, 1400, and 1600°F and quenched in cold water. Specimens representing each solution treatment were then aged for various times at temperatures of 800, 900, 1000, and 1100°F. After heat treatment, the tensile blanks were machined to substandard sheet tensile specimens having a gage section 0.250 inch wide by about 1-1/4 inches long. The tensile properties and hardnesses of these heat-treated specimens are given in Tables 13, 14, and 15.

The complete series of tests was carried out only on the Ti-3.5Cr-3.5V alloy. Testing of the other two alloys was discontinued because of the very low ductilities which were obtained in the initial tests.

The data in Tables 13, 14, and 15 show that the ductilities obtained in alloy sheet after heat treatment are much lower than those of bar-stock specimens of the same alloy. At strength levels above about 150,000 psi, the ductility of all alloys was unacceptably low. Initially, it was believed that the difference in ductility between sheet and bar stock may have been connected with the difference in cooling rates from the solution temperature. To check this possibility, specimens of the Ti-3.5Cr-3.5Mn alloy were solution treated at 1300 and 1400°F and either quenched or air cooled. The latter treatment should produce a cooling rate comparable with that obtained in a 1/2-inch-diameter bar specimen during a water quench. The sheet specimens were then given various overaging treatments and tested. The results are given in Table 16. There was no observable difference in the tensile properties of specimens cooled at the two rates from the solution temperature. Like the other three alloys, the properties of the Ti-3.5Cr-3.5Mn were very poor compared with those of bar stock of this same composition. Apparently, the cooling rate from the solution temperature within the range studied here was not a factor in the low ductility of alloy sheet.

The balance of the work during the past year on sheet alloys was directed toward determining the causes of the low ductility of alloy sheet as compared with bar stock. The effects of several factors on the ductility of alloy sheet were investigated as described in the following sections of this report.

TABLE 13. PROPERTIES OF HEAT-TREATED 14-GAGE SHEET OF THE Ti-3.5Cr-3.5V ALLOY(1) ROLLED AT 1450°F

Solution Temp(2), °F	Aging Time, hr	Aging Temp, °F	Elongation(3), % in 1 inch	Ultimate Tensile Strength(3), psi	VHN (10-Kg Load)(5)
1300	-	-	9.0	158,000	362
"	24	800	6.5	162,500	364
"	48	"	8.5	160,000	360
"	8	900	8.5	149,000	334
"	24	"	8.5	145,000	352
"	4	1000	10.0	133,000	315
"	8	"	13.0	131,500	311
1400	-	-	1.0(4)	185,000(4)	432
"	24	800	1.5	179,500	380
"	48	"	2.5	178,000	372
"	8	900	5.5	156,000	348
"	24	"	4.0	152,000	337
"	4	1000	6.5	145,500	330
"	8	"	7.0	141,500	321
1600	-	-	0.0(4)	91,000(4)	425
"	24	800	0.0(4)	126,500(4)	427
"	48	"	0.0(4)	159,000	423
"	8	900	0.0(4)	130,000(4)	385
"	24	"	0.0(4)	89,500(4)	377
"	4	1000	1.0(4)	164,500(4)	361
"	8	"	0.5(4)	150,500(4)	356

(1) Heat No. WT141A; actual composition 4.49Cr-3.14V.
(2) Quenched in cold water after 1 hour at the solution temperature.
(3) Average of two values, except where noted.
(4) Single values.
(5) Average of three impressions on each specimen.

TABLE 14. PROPERTIES OF HEAT-TREATED 14-GAGE SHEET OF THE Ti-5Mn-2.5Cr ALLOY(1) ROLLED AT 1450°F

Solution Temp(2), °F	Aging Time, hr	Aging Temp, °F	Elongation(4), % in 1 inch	Ultimate Tensile Strength(4), psi	VHN (10-Kg Load)(5)
1300	-	-	5.5(3)	159,500(3)	345
"	24	800	0.0	146,500	405
"	48	"	0.0	143,000	400
"	8	900	0.0	131,000	376
"	24	"	0.0	120,500	377
"	4	1000	1.0	151,400	354
"	8	"	0.0	139,500	341
1400	-	-	1.0(3)	182,500(3)	373
"	24	800	0.0	46,000	444
"	48	"	0.0	78,000	430
"	8	900	0.0	114,000	401
"	24	"	0.0	148,500	393
"	4	1000	2.0	169,000	364
"	8	"	1.0(3)	143,000(3)	360

(1) Heat No. WT147A: actual composition 4.87Mn-2.52Cr.
(2) Quenched in cold water after 1 hour at the solution temperature.
(3) Average of two values.
(4) Single values, except where noted.
(5) Average of three impressions.

TABLE 15. PROPERTIES OF HEAT-TREATED 14-GAGE SHEET OF THE Ti-3Mn-1Cr-1Fe-1Mo-1V ALLOY(1) ROLLED AT 1450°F

Solution Temp(2), °F	Aging Time, hr	Aging Temp, °F	Elongation(3), % in 1 inch	Ultimate Tensile Strength(3), psi	VHN (10-Kg Load)(4)
1300	-	-	8.0	154,000	350
"	24	800	0.5	178,000	387
"	48	"	1.0	176,500	382
"	8	900	3.0	161,000	362
"	24	"	4.0	157,000	354
"	4	1000	4.0	142,000	338
"	8	"	3.5	140,500	327

(1) Heat No. WT154A: actual composition 3.81Mn-1.39Cr-0.79Fe-1.00Mo-1.01V.
(2) Quenched in cold water after 1 hour at the solution temperature.
(3) Average of two values.
(4) Average of three impressions on each specimen.

TABLE 16. PROPERTIES OF HEAT-TREATED 14-GAGE SHEET OF THE Ti-3.5Cr-3.5Mn ALLOY(1) ROLLED AT 1450°F

Solution Temp, °F	Cooling Medium(2)	Aging Time, hr	Aging Temp, °F	Elongation, % in 1 inch	Ultimate Tensile Strength, psi	VHN (10-Kg Load)
1300	Water	-	-	16.0	146,500	330
"	Air	-	-	20.5	144,000	325
"	Water	12	800	1.0	176,000	413
"	"	24	"	2.5	190,500	401
"	Air	24	"	5.0	181,500	387
"	Water	48	"	2.0	184,000	397
"	"	90	"	1.5	179,000	397
"	"	8	900	7.0	165,000	363
"	Air	8	"	5.5	162,000	363
"	Water	24	"	3.0	162,500	360
"	"	0.5	1000	5.5	154,500	354
"	"	2	"	2.5	148,000	354
"	"	8	1100	10.0	135,000	319
"	"	24	"	16.5	133,500	319
1400	Water	-	-	13.5	168,000	345
"	Air	-	-	3.0	176,000	413
"	Water	12	800	0.5	203,000	468
"	"	24	"	1.5	215,500	437
"	Air	24	"	2.0	203,000	413
"	Water	48	"	0.5	218,500	433
"	"	90	"	0.5	208,000	429
"	"	8	900	3.0	187,500	387

TABLE 16. (Continued)

Solution Temp, °F	Cooling Medium (2)	Aging Time, hr	Aging Temp, °F	Elongation, % in 1 inch	Ultimate Tensile Strength, psi	VHN (10-Kg Load)
1400	Air	8	900	2.5	184,500	401
"	Water	24	"	4.0	182,000	401
"	"	0.5	1000	7.0	177,500	394
"	"	2	"	4.5	170,500	380
"	"	8	1100	22.5	142,500	330
"	"	24	"	22.5	138,500	314

(1) Heat No. WT111A: actual composition 2.9Cr-2.9Mn.

(2) Quenched in cold water or air cooled after 1 hour at the solution temperature.

Effect of As-Cast Hardness on Properties of Alloy Sheet

The as-cast hardnesses of ingots of the Ti-3.5Cr-3.5V, Ti-5Mn-2.5Cr, and Ti-3Mn-complex alloys which were rolled into sheet were 400 BHN or higher. This was considerably higher than the hardnesses of previously made ingots of the same size from which bar stock having excellent properties had been produced. The high hardness of the present ingots was not considered particularly significant until tests described in the preceding section revealed the lack of ductility in the sheet stock made from them. Previously, there had been no definite correlation between high as-cast ingot hardness and low ductility in sheet or bar stock. The ingot hardness may be affected by a great many factors, such as composition, contamination, or the thermal cycles produced during melting.

At this time, a 20-pound ingot of the Ti-3Mn-complex alloy was available which showed considerable variation in as-cast hardness over its length. The top 4 inches of this ingot had hardnesses ranging from 380 to 415 Brinell. The remainder of the ingot, with the exception of about 1-1/2 inches at the bottom, had a hardness of 330 to 345 BHN. The ingot was sectioned at points representing the limits of the two hardness ranges and fabricated into 14-gage sheet. Specimens taken from both sections were given various heat treatments and tensile tested. The hardnesses, chemical compositions, and properties of sheet from the two sections, after various solution and aging treatments, are given in Table 17. It may be seen that specimens taken from the two sections had essentially the same properties after a given heat treatment. Thus, it appears that as-cast ingot hardness variations within the range shown here have little or no effect on the properties of fabricated sheet.

As in the previous tests, this Ti-3Mn-complex sheet had low ductility after aging at 800 or 900°F.

Effect of Hydrogen Content and Surface Contamination

Having shown that the as-cast hardness does not seem to affect properties, chemical analyses were made on the three heats for elements ordinarily determined. The results are presented in Table 18. No reason was indicated by the analyses for the lack of ductility of these heats.

TABLE 17. PROPERTIES OF HEAT-TREATED 14-GAGE SHEET OF THE Ti-3Mn-1Cr-1Fe-1Mo-1V ALLOY TAKEN FROM HIGH- AND LOW-HARDNESS SECTIONS OF A 20-POUND INGOT(1)

Solution	Aging		330 to 345 BHN As-Cast Hardness			380 to 415 BHN As-Cast Hardness		
Temp(2), °F	Time, hr	Temp, °F	Elongation, % in 1 inch	Ultimate Tensile Strength, psi	VHN (10 Kg Load)	Elongation, % in 1 inch	Ultimate Tensile Strength, psi	VHN (10 Kg Load)
1300	-	-	15.0	147,000	312	15.5	146,000	328
"	8	800	2.0	182,500	421	0.5	171,000	413
"	48	"	0.5	179,500	390	1.0	173,500	383
"	4	900	1.0	159,500	361	1.5	159,500	370
"	24	"	4.5	154,500	357	5.5	151,500	360
1400	-	-	7.0	169,000	366	5.0	170,000	366
"	8	800	0.0	187,500	473	0.0	175,000	478
"	48	"	0.5	197,000	429	1.0	197,500	429
"	4	900	2.0	186,500	401	1.5	183,000	417
"	24	"	1.5	172,000	387	2.5	170,000	410

(1) Heat No. WU20A: actual composition 1.41Cr-1.20Fe-2.85Mn-0.75Mo-0.95V; same for both sections of ingot.
(2) Quenched in cold water.

TABLE 18. CHEMICAL ANALYSES OF SELECTED ALLOYS FABRICATED TO 14-GAGE SHEET

Heat	Intended Composition, %	Actual Composition, %							
		Cr	Fe	Mn	Mo	V	W	N	C
WT141A	3.5Cr-3.5V	4.49	-	-	-	3.14	0.56	0.026	0.04
WT147A	5Mn-2.5Cr	2.52	-	4.87	-	-	0.19	0.026	0.04
WT154A	1Cr, 1Fe, 3Mn, 1Mo, 1V	1.39	0.79	3.81	1.00	1.01	0.23	0.021	0.06

Initially, it was suspected that oxygen and nitrogen absorbed during forging after the first melting operation might have resulted in brittleness. However, the nitrogen analyses shown in Table 18 indicate that nitrogen absorption was practically nil. Oxygen determinations by a vacuum-fusion method were made on Heat WT147A, and on a previous heat of the Ti-5Mn-2.5Cr alloy, WT15A. The hydrogen content was determined also during these analyses. The results were as follows:

Heat No.	Oxygen Content, Weight %	Hydrogen Content, Weight %
WT15A	0.12	0.02
WT147A	0.15	0.05

The oxygen analyses again indicate no excessive contamination during forging or melting. However, Heat WT147A contained approximately 2.5 times as much hydrogen as the earlier heat. This high hydrogen content could have lowered the ductility of Heat WT147A.

To determine the role of hydrogen, if any, in the brittleness observed in these heats, vacuum-annealing experiments were conducted on two of the three heats. Duplicate specimens of Heat WT147A (Ti-5Mn-2.5Cr) and WT154A (Ti-3Mn-complex) were vacuum annealed at 1400 or 1600°F to remove hydrogen. The annealing times were 16 hours at 1400°F and 5 hours at 1600°F. At the end of these treatments, the pressure in the furnace was 5×10^{-5} mm of mercury, at which pressure most of the hydrogen was assumed to have been removed. Concurrent with the vacuum-annealing work, other specimens of the same heats were sealed in Vycor tubes under a reduced pressure of argon and given the same thermal cycle.

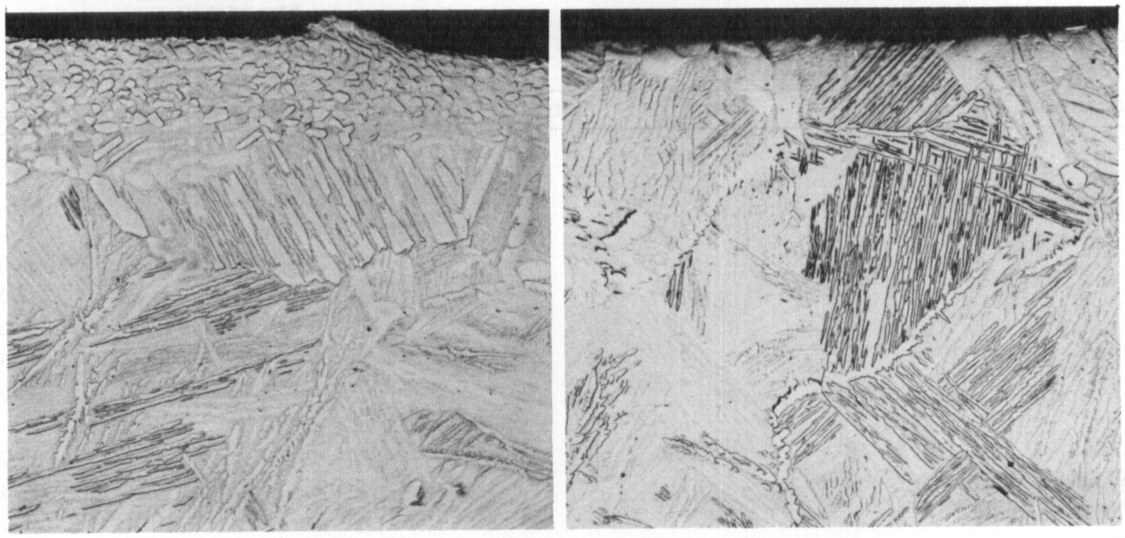
After these initial treatments, all specimens were solution treated at 1300°F and half of the specimens aged 24 hours at 900°F. One specimen of each pair was then tested. In general, the ductility of these specimens was very low. Microscopic examination of the broken specimens revealed that their surfaces were heavily contaminated. Consequently, the remaining specimen of each pair was pickled to remove this contamination, first in the $NH_4F \cdot HF-H_2SO_4$-water solution to remove visible oxide, and then in a 5 per cent HF-15 per cent HNO_3-water solution to produce a smooth, bright surface. Approximately 0.004 inch was removed from each surface by this method. The results of tensile and hardness tests made on all specimens are given in Table 19.

It is evident from these data that both the removal of the contaminated surface and the removal of hydrogen had important effects on the properties of the sheet. With two exceptions, all of the as-solution-treated specimens showed greatly improved ductility after removal of the contaminated surfaces. The depth of contamination of an argon-treated specimen is shown in the left half of Figure 18. The right half of Figure 18 shows the surface after about 0.003 inch was removed by etching. The contaminated layer undoubtedly would have lower ductility, and could initiate fracture of the sheet specimens. The Ti-5Mn-2.5Cr alloy vacuum or argon annealed at 1600°F showed very little or no improvement. In most cases, the over-aged specimens had slightly better ductility after pickling.

Of the specimens which were pickled to remove contamination, those that were vacuum annealed had, in general, better ductility than the argon-annealed specimens. The improvement was particularly striking in the case of the specimens vacuum annealed at 1400°F, solution treated, and overaged. The tensile elongation of the Ti-5Mn-2.5Cr alloy increased from 2 per cent for the argon-annealed specimen to 15 per cent for the vacuum-annealed specimen at a strength level of about 165,000 psi. Corresponding specimens of the Ti-3Mn-complex alloy (Heat WT154A) showed an improvement from 4 per cent to 17 per cent elongation at a strength level of 160,000 psi.

These specimens were examined metallographically. Figures 19 and 20 illustrate the structure of the Ti-3Mn-complex alloy (Heat WT154A) that was argon annealed for 16 hours at 1400°F, solution treated at 1300°F, and overaged 24 hours at 900°F. The structure consists of elongated primary alpha and an unresolved alpha precipitate in a beta matrix. There is also a dark-etching grain-boundary phase which is continuous through the structure. The heavy etch of Figure 19 makes it appear at times like heavy grain boundaries. However, a light etch, as in Figure 20, shows a distinct phase present. This phase etches more easily than the grain boundaries. As seen in the center of the photomicrograph, no alpha grains are visible. The same alloy annealed in vacuum at 1400°F and heat treated contains very little, if any, of this dark-etching phase, as is shown in Figure 21. The disappearance of this phase during vacuum annealing is indicative of a hydride compound.

Specimens treated for 5 hours at 1600°F in argon, solution treated at 1300°F, and overaged 24 hours at 900°F showed the same grain-boundary



250X N2725 250X N2728

1-1/2% HF - 3-1/2% H_2O_2 Etch

FIGURE 18. Ti-3Mn-COMPLEX ALLOY ANNEALED 5 HOURS AT 1600°F IN ARGON, SOLUTION TREATED AT 1300°F, AND OVERAGED AT 900°F FOR 24 HOURS

Upper left half of picture shows alpha-rich surface resulting from contamination during argon anneal
Upper right half shows structure after removal of 0.003 inch from surface by chemical etch

TABLE 19. EFFECT OF VACUUM
PROPERTIES OF TITANIUM-

Heat No. (Actual Composition, %)	Initial Heat Treatment(1)		Aging(2)			Initial
	Time, hr	Temp, °F	Time, hr	Temp, °F	Material Removed From Thickness by Etching(3), inch	Elongation(4), % in 1 inch
WT147A	16	1400	-	-	-	0.0
(4.87Mn, 2.52Cr)	"	"	-	-	0.0099	19.0
	"	"	24	900	-	0.0
	"	"	"	"	0.0117	2.0
	5	1600	-	-	-	10.0
	"	"	-	-	0.0084	4.0
	"	"	24	900	-	0.0
	"	"	"	"	0.0099	0.0
WT154A	16	1400	-	-	-	0.0
(3.81Mn, 1.39Cr,	"	"	-	-	0.0061	18.0
0.79Fe, 1.00Mo,	"	"	24	900	-	0.0
1.01V)	"	"	"	"	0.0062	4.0
	5	1600	-	-	-	17.0
	"	"	-	-	0.0056	19.0
	"	"	24	900	-	3.0
	"	"	"	"	0.0064	1.0

(1) Time was adjusted to give a final vacuum of 0.05 micron.
(2) All alloys were solution treated in argon for 1 hour at 1300°F and cold water quenched prior to aging.
(3) Scale removed by sulfuric acid-ammonium bifluoride etch, followed by nitric acid-hydrofluoric acid etch to remove most of metal.
(4) Single values.
(5) Average of three impressions.
(6) Sealed in Vycor tube with a low-pressure argon atmosphere.

ANNEALING AND ETCHING ON TENSILE
ALLOY SHEET MATERIAL

Atmosphere — Argon(6)		Initial Atmosphere — Vacuum			
Ultimate Tensile Strength(4), psi	VHN (10-Kg Load)(5)	Material Removed From Thickness by Etching(3), inch	Elongation(4), % in 1 inch	Ultimate Tensile Strength(4), psi	VHN (10-Kg Load)(5)
142,000	312	-	1.0	138,500	310
142,500	333	0.0091	21.5	139,000	306
136,500	345	--	0.0	132,000	333
163,000	357	0.0100	15.0	168,000	351
151,000	336	-	2.0	150,500	339
149,500	360	0.0073	4.0	146,500	360
157,500	376	-	0.0	158,000	376
133,500	378	0.0075	1.0	162,500	383
125,500	306	-	0.0	138,000	314
142,000	317	0.0059	24.0	141,500	322
139,000	353	-	0.0	155,500	345
161,000	366	0.0064	17.0	160,000	339
150,500	339	-	6.0	147,500	345
148,500	342	0.0061	19.0	146,500	348
171,000	376	-	5.0	169,500	366
164,000	370	0.0058	8.0	166,000	376

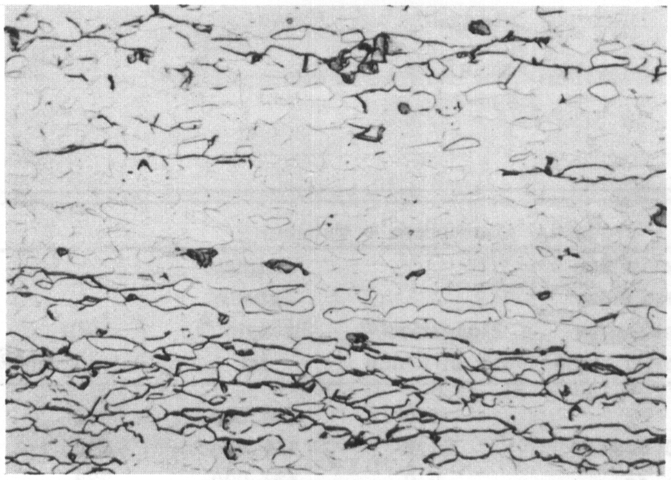


FIGURE 19. Ti-3Mn-COMPLEX ALLOY ANNEALED 16 HOURS AT 1400°F IN ARGON, SOLUTION TREATED AT 1300°F, AND OVERAGED AT 900°F FOR 24 HOURS

Primary alpha in beta matrix containing unresolved alpha precipitate
Dark-etched material probably hydride

500X N2716

1-1/2% HF - 3-1/2% H_2O_2 Etch

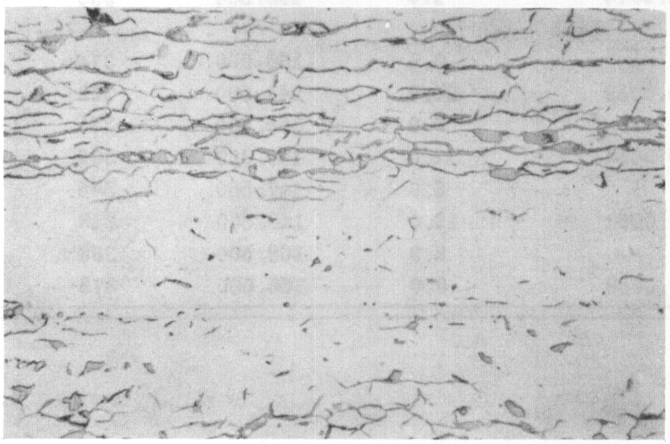


FIGURE 20. Ti-3Mn-COMPLEX ALLOY TREATED IN SAME MANNER AS IN FIGURE 19, BUT ETCHED VERY LIGHTLY

500X N2713

1-1/2% HF - 3-1/2% H_2O_2 Etch

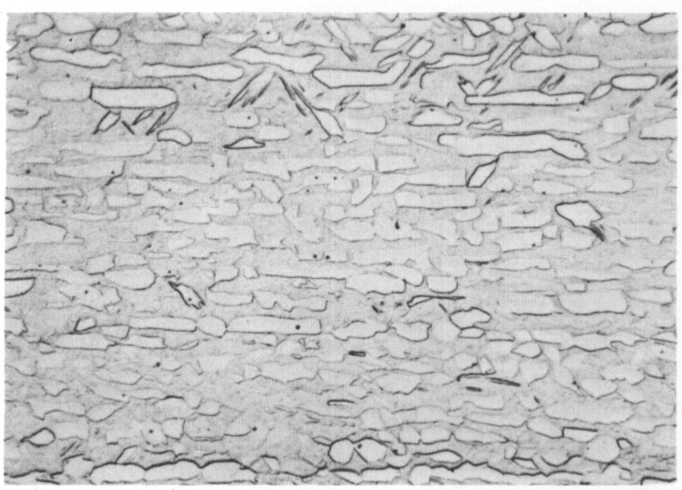


FIGURE 21. Ti-3Mn-COMPLEX ALLOY ANNEALED 16 HOURS AT 1400°F IN VACUUM, SOLUTION TREATED AT 1300°F, AND OVERAGED AT 900°F FOR 24 HOURS

Note very little dark-etching phase at grain boundaries

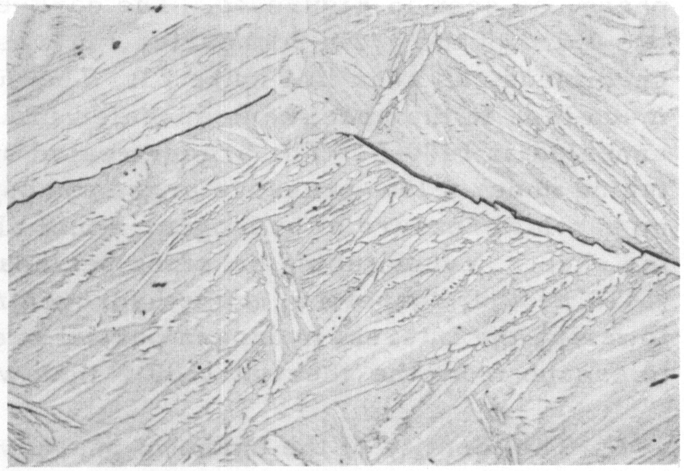
500X N2717

1-1/2% HF - 3-1/2% H_2O_2 Etch

FIGURE 22. TI-3Mn-COMPLEX ALLOY ANNEALED 5 HOURS AT 1600°F IN ARGON, SOLUTION TREATED AT 1300°F, AND OVERAGED FOR 24 HOURS AT 900°F

Unresolved alpha precipitate in beta matrix

The dark-etching phase at the alpha grain boundaries is believed to be a hydride

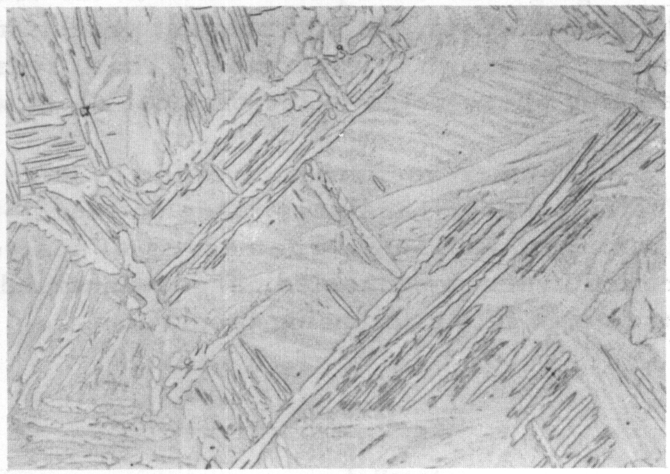


500X N5137

1-1/2% HF - 3-1/2% H_2O_2 Etch

FIGURE 23. TI-3Mn-COMPLEX ALLOY ANNEALED 5 HOURS AT 1600°F IN VACUUM, SOLUTION TREATED AT 1300°F, AND OVERAGED 24 HOURS AT 900°F

Same structure as in Figure 22, except that hydride phase is not present



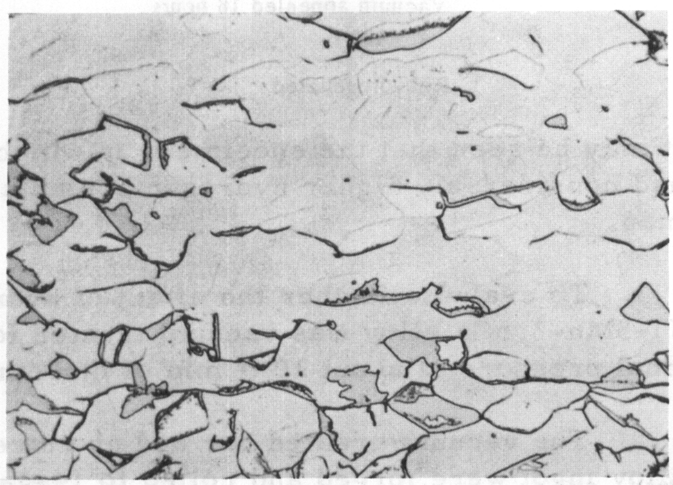
500X N2718

1-1/2% HF - 3-1/2% H_2O_2 Etch

FIGURE 24. TI-3Mn-COMPLEX ALLOY VACUUM TREATED TO 0.05 MICRON OF MERCURY AND REHYDROGENATED WITH 2 ATOMIC PER CENT AT 600°C

Solution treated one hour at 1300°F, cold-water quenched, and aged 8 hours at 900°F

Note darker etching hydride phase



1000X N2719

40% HF - 60% Ethylene Glycol Etch

phase, as seen in Figure 22. The dark-etching phase was visible in about one-third of the grain boundaries. It was not present in the vacuum-treated specimens, as seen in Figure 23. The quantity of hydride would seem to be much less in the specimens annealed in argon at 1600°F than in those annealed at 1400°F. The argon-treated specimens were sealed in Vycor tubing under a low-pressure argon atmosphere. This atmosphere, having no partial pressure of hydrogen, would allow some hydrogen to escape from the sheet. Vycor tubing becomes permeable to hydrogen at about 800°C (1472°F) and, therefore, some hydrogen might pass through the Vycor, lowering the internal hydrogen partial pressure and allowing more hydrogen to escape from the sheet at 1600°F than would be possible at 1400°F.

The identity of the unknown phase was investigated further by re-hydrogenating vacuum-annealed specimens of the Ti-3Mn-complex alloy in a modified Sievert's apparatus. These specimens were then quenched from 1300°F and aged at 900°F, as in the earlier work. After solution treating and aging, the third phase noted in the original sheet was again present in large quantities. It was present predominantly at the alpha grain boundaries, as shown in Figure 24. Thus, it is fairly certain that the dark-etching phase noted in the original sheet was a hydride.

Further confirmation of the hydride theory was obtained from vacuum-fusion hydrogen analyses made on the Ti-3Mn-complex alloy specimens from Heat WT154A. The results of analyses made on specimens in the as-rolled, vacuum-annealed, argon-annealed, and rehydrogenated conditions are given in the following tabulation:

Condition	Hydrogen Content, ppm by weight
As rolled	1000
Argon annealed 16 hours at 1400°F	430
Vacuum annealed 16 hours at 1400°F	220
Rehydrogenated	390

It may be seen that the specimens in which the unknown phase was observed had considerably higher hydrogen contents than the vacuum-annealed specimen.

To evaluate further the effect of hydrogen, a 2 x 2 x 6-inch bar of the Ti-5Mn-2.5Cr alloy was vacuum treated for about 48 hours at 1600°F at a final pressure of about 10^{-4} mm of mercury.

The vacuum-treated bar and an untreated bar from the same 20-pound alloy ingot were forged and rolled to 14-gage sheet. The standard rolling procedure was varied somewhat in an attempt to alter the alpha dispersion of the final sheet. It has been noted that bar stock, which undergoes less reduction of area than sheet during fabrication, has a more random alpha dispersion. In sheet, the alpha is more elongated and seems to form a nearly continuous network through the structure, which may partially account for its lower ductility. The initial 2 x 2 x 6-inch bars were forged at 1750°F to 1-inch-thick slabs and scalped. The slabs were then rolled at 1450°F to 0.180-inch thickness. One half of each sheet was annealed for 10 minutes at 1600°F. The four sections were rolled at 1450°F to 14-gage sheet. The reduction of area from 0.180-inch thickness to 14-gage (0.064-inch) sheet is about the same as the reduction of area in rolling 1/2-inch-diameter bar stock from the forged bar.

Sheet specimens for the four conditions were given various solution and aging treatments. The tensile properties and hardnesses of the heat-treated specimens are given in Table 20. Generally, the vacuum annealing improved the tensile ductility only slightly, except for the specimens solution treated at 1400°F and aged 8 hours at 1100°F. The elongation values increased over 11 per cent for these specimens. The generally poor ductility of the vacuum-treated specimens evidently resulted from insufficient removal of hydrogen from the original 2-inch-square bar. Hydrogen analyses of sheet from the vacuum-treated bar and the untreated bar gave the following results:

Condition	Hydrogen Content, ppm by weight
Untreated bar	577
Vacuum-treated bar	530

The annealing treatment at 1600°F applied to the 0.180-inch sheet was definitely detrimental in all cases. A microexamination showed a continuous alpha phase at the grain boundaries of these specimens. The reduction from 0.180- to 0.065-inch thickness was not enough to break up the grain-boundary alpha which had formed around the beta grains on cooling and reheating after the 1600°F anneal. This resulted in low ductility.

Effect of Specimen Geometry

Specimen geometry is a third possible cause of lower ductility in sheet alloys. The thin, rectangular cross section of the sheet specimens will have a different stress pattern than the circular section of bar-stock specimens. Thus, stress concentrations at the corners of rectangular sections may cause premature failure.

TABLE 20. EFFECT OF VACUUM ANNEALING AND ROLLING PROCEDURE ON PROPERTIES OF 14-GAGE SHEET OF THE Ti-5Mn-2.5Cr ALLOY(1)

Initial Treatment(2)	Solution Temp(3), °F	Aging: Time, hr	Aging: Temp, °F	Rolled From 1 Inch Thick Without Annealing: Elongation(4), % in 1 inch	Ultimate Tensile Strength(4), psi	VHN (10-Kg Load)(6)	Annealed After Rolling to 0.180-Inch-Thick Sheet: Elongation(4), % in 1 inch	Ultimate Tensile Strength(4), psi	VHN (10-Kg Load)(6)
None	1300	-	-	14.5	146,500	317	10.0	146,000	336
Vacuum(7)	"	-	-	15.5	145,000	319	11.0	144,000	342
None	"	8	900	2.5	162,500	360	1.0(5)	157,500(5)	357
Vacuum	"	"	"	1.5	161,000	360	1.0(5)	156,500(5)	370
None	"	4	1100	9.5	127,000	306	4.0	121,500	317
Vacuum	"	"	"	-(8)	125,000	317	9.0(5)	133,500(5)	319
None	1400	-	-	9.0	158,500	357	6.5	165,500	357
Vacuum	"	-	-	10.0	158,000	354	6.5	148,500	339
None	"	8	900	2.0	172,000	373	1.0(5)	174,500(5)	387
Vacuum	"	"	"	4.0	176,000	373	1.5	172,000	383
None	"	4	1100	8.5	135,000	322	3.0	130,000	330
Vacuum	"	"	"	20.0	138,500	325	5.0(5)	134,500(5)	330

(1) Heat No. WT148A.
(2) After forging to 2-inch-square bar and before final forging and rolling to sheet.
(3) Cold water quenched after 1 hour at temperature.
(4) Average of two specimens, except where noted.
(5) Single values.
(6) Average of three impressions.
(7) Heated in vacuum at 1600°F until the pressure was reduced to 10^{-4} mm of mercury.
(8) Specimens broke outside of gage marks.

A few tests have been made of sheet specimens with the sharp corners rounded off. The sheet material from the ingot of the Ti-3Mn-complex alloy melted at the Bureau of Mines was selected for these tests because of its low hydrogen content (0.007 weight per cent). Four specimens were given each of several heat treatments, after which about 0.002 inch was etched off the surfaces before machining. A 0.020-inch radius was filed and sanded on the corners of the gage sections of two specimens from each treatment before testing. The other two specimens were tested as machined. The results of these tests are in Table 21. It may be noted that there was some loss of ductility for the solution-treated specimens with rounded corners. However, there were remarkable improvements in the ductility of the overaged specimens on which the corners were rounded. For example, elongation increased from 0.5 to 4 per cent at the 210,000-psi strength level, and from 4.5 per cent to about 16 per cent at the 160,000-psi strength level.

Conclusions on Factors Affecting Sheet Ductility

From the data presented in this section, it may be concluded that at least three factors contribute to the low ductility obtained in heat-treated, beta-stabilized, high-strength titanium alloy sheet:

(1) Presence of hydrogen.

(2) Surface contamination.

(3) Sharp corners of test section.

The limits of these factors for satisfactory sheet ductility were not established. However, a more thorough investigation will be made under the extension to the present contract.

INVESTIGATION OF BETA EMBRITTLEMENT IN TITANIUM ALLOYS

The term "beta embrittlement" may be defined as the reduction in ductility of beta-stabilized titanium alloys incurred by heating to temperatures in the beta-phase region without subsequent working at lower temperatures. Beta embrittlement poses a serious problem in the use of such alloys because it necessitates fabrication at low hot-working temperatures and very close control of heat-treatment temperatures. As discussed here, beta embrittlement does not necessarily mean a zero elongation value.

TABLE 21. EFFECT OF CORNERS ON TENSILE PROPERTIES OF SHEET SPECIMENS OF THE Ti-3Mn-1Cr-1Fe-1Mo-1V(1) ALLOY

Solution Temp(3), °F	Aging		VHN (10-Kg Load)(4)	Square Corners		Rounded Corners(2)	
	Time, hr	Temp, °F		Elongation(5), % in 1 inch	Ultimate Tensile Strength(5), psi	Elongation(5), % in 1 inch	Ultimate Tensile Strength(5), psi
1300	-	-	341	15.0	155,000	13.5	149,000
"	24	900	365	4.5	160,000	16.0	160,000
"	8	1100	330	12.0	135,000	17.0	138,000
1400	-	-	366	11.0	167,000	9.0	163,500
"	48	800	-	0.5	216,500	4.0	210,500
"	8	1000	-	5.0	164,500	12.0	159,500

(1) From top section of 220-pound ingot melted by The Bureau of Mines.
(2) About 1/64-inch radius filed and sanded on corners of gage section.
(3) Cold water quenched after 1/2 hour at temperature in purified argon.
(4) Average of three impressions.
(5) Average of two values.

However, it was shown in earlier sections in this report, that specimens finish rolled in the beta field or solution treated in this range had ductilities appreciably below the optimum. It was noted that the materials were large grained and fractured intergranularly. It does not seem reasonable that the larger grain size resulting from the beta heat treatment could account for the complete loss in ductility exhibited at the higher strength levels. A more logical hypothesis involves partition of one or more impurity elements between the alpha and the beta phases. In particular, the interstitial elements, oxygen, nitrogen, and carbon, are known to be more soluble in alpha than in beta titanium. Therefore, it is probable that most of these elements remain in solution in the alpha phase during solution treatment in the alpha-beta field. Since alpha is the discontinuous phase, the presence of the interstitials in alpha probably would have little effect. When the alloys are solution treated in the beta-phase field (1600°F), the interstitials would be in solution in the continuous beta phase, and their embrittling characteristics would be fully effective in this structure. It is possible that this embrittling effect is enhanced by segregation of the interstitials at the grain boundaries of the beta phase because of the disordered lattice structure at these locations. A number of the specimens solution treated at 1600°F exhibited intergranular-type fractures, indicating some grain-boundary effect. Of course, trace elements other than oxygen, nitrogen, and carbon might play a part in this embrittlement. However, proof of this statement would be very difficult to obtain.

Experiments have been made on the Ti-3Mn-complex alloy to determine the critical amount of primary alpha necessary to retain ductility during solution and overaging treatments. Half-inch-diameter bar stock was solution treated at 25°F intervals from 1425 to 1550°F. The solution-treated specimens were aged 8 or 24 hours at 900°F. The bars were then machined to standard 0.250-inch tensile bars and tested to obtain ultimate-strength and elongation values. These data are reported in Table 22.

The properties did not change appreciably for solution temperatures from 1425 to 1475°F; the elongation varied from 17 to 14 per cent in 1 inch, and the tensile strength varied from 180,000 to 187,000 psi. The relative quantity of alpha in these structures was estimated by the point-count method to be 15 to 17 per cent. The specimen solution treated at 1500°F showed an increase in strength and a sharp decrease in elongation. This specimen had much less alpha than the more ductile specimens. The two higher solution temperatures produced a further increase in strength to about 207,000-psi, with a corresponding decrease in elongation to a little over 1 per cent. The amount of alpha in these two structures was less than 2 per cent.

Some of the loss in ductility produced by the higher solution temperatures may be attributed to the higher strength and hardness of these specimens. However, it had been shown in earlier work that the Ti-3Mn-complex alloy is capable of developing excellent ductility at the 200,000-psi

TABLE 22. EFFECT OF SOLUTION TEMPERATURE ON TENSILE PROPERTIES AND HARDNESSES OF 1/2-INCH-DIAMETER BAR STOCK OF Ti-3Mn-COMPLEX ALLOY

Solution Temp(1), °F	Alpha in Microstructure(2), per cent	Elongation, % in 1 inch	Ultimate Tensile Strength, psi	VHN (10-Kg Load)
1425(3)	16	17	180,000	383
1450(3)	17	15	183,500	383
1475(3)	15	14	187,000	383
1500(4)	6	5.5	201,000	397
1525(4)	< 2	1.0	208,000	413
1550(4)	< 1	1.5	206,000	397

(1) Held at temperature 1 hour, then cold water quenched.

(2) Estimated by point-count method.

(3) Aged 24 hours at 900°F and air cooled.

(4) Aged 8 hours at 900°F and air cooled.

strength level under the proper conditions of heat treatment. Therefore, it appears that the lower primary-alpha content is the major reason for the low ductility.

The aging times were not the same for all specimens used in these tests. The three alloys solution treated in the 1400°F range were aged for 24 hours, whereas those solution treated in the 1500°F range were aged for 8 hours. However, it has been shown in the section on "Mechanical Properties" that increasing the aging time at 900°F from 8 to 24 hours did not affect properties appreciably.

It has been shown that beta embrittlement may be due to the alpha-stabilizing interstitial elements. If these elements are responsible, two possible methods of minimizing their effects are immediately apparent. One method would be to re-solution treat in the alpha-beta-phase region long enough to allow a partition of the interstitials principally to the alpha. The other would be to add an alloy element that would remove the interstitial elements from the titanium by forming insoluble compounds. The latter possibility will be discussed first.

There is a possibility that oxygen, which is the major interstitial, might be tied up with a strong reducing agent, such as thorium. Thorium additions of 1 or 2 per cent were made to heats of the Ti-5Mn-2.5Cr alloy. Each ingot was split for rolling to 1/2-inch bar stock, with one half being rolled at 1450°F and the balance at 1600°F. (The beta transus of the

Ti-5Mn-2.5Cr alloy is about 1500°F.) Specimens rolled at each temperature were solution treated at 1300, 1400, and 1600°F and overaged.

The results of tensile tests made on the heat-treated bars are given in Table 23. The properties of the base alloy rolled at 1450 and 1600°F are included in the table for comparison.

None of the thorium specimens had as good ductility as the Ti-5Mn-2.5Cr base-alloy specimens rolled at 1450°F. However, specimens of all the thorium-containing alloys solution-treated at 1300°F and overaged had fair ductilities at relatively low strength levels. The 1 per cent thorium alloy had much better tensile elongations after rolling at 1600°F and heat treating than did the base alloy rolled at the same temperature.

There was very little difference in the ductilities of the heat-treated specimens of the thorium-containing alloys rolled at 1450 or 1600°F. Microscopic examination revealed that both rolling temperatures were in the beta-phase region for the alloys, the beta transus being very near 1400°F for the 1 per cent thorium alloy. The coarse-grained structure produced by rolling in the beta field accounts for the relatively low ductility of all of the heat-treated specimens. The 2 per cent thorium alloy was all-beta at 1400°F and about 20 per cent alpha at 1300°F. A large number of angular inclusions were noted in the thorium alloys, and these may have been a compound of thorium with one of the interstitial elements.

The thorium additions failed to overcome beta embrittlement completely. However, they improved the ductilities of specimens rolled in the beta field and subsequently re-solution treated in the alpha-beta field over those obtained by the same type of treatment of the base alloy. Smaller thorium additions would be of interest and will be investigated at a later date.

The hypothesis of re-solution of the interstitials in the alpha phase to overcome beta embrittlement was evaluated in specimens of the Ti-3Mn-complex alloy which were beta solution treated for 1 hour and water quenched. The bars were then held in the alpha-beta field for 48 hours at 1300 or 1475°F, water quenched, and aged 8 hours at 900°F. Tensile and hardness data for these specimens, Heat WU24A, are shown in Table 24. Also tabulated for comparison purposes are the results of aging this alloy directly from the 1600°F quench, as well as the results obtained by heat treating material which was rolled at 1600°F (Heat WT107A). The 1300 or 1475°F solution treatments after the beta quench lowered the strength to values considerably below those obtained by direct quenching from 1600°F and aging, but did not improve the ductility appreciably. However, the heat of this alloy rolled at 1600°F (Heat WT107A) had relatively good ductility after solution treating at 1300 or 1400°F and overaging. Neither heat had any appreciable ductility when aged directly following the 1600°F solution treatment.

TABLE 23. EFFECT ON TENSILE PROPERTIES AND HARDNESSES OF THORIUM ADDITIONS TO THE Ti-5Mn-2.5Cr ALLOY ROLLED AT 1450 OR 1600 °F

Heat No.	Nominal Thorium Content, %	Rolling Temp, °F	Solution Temp(1), °F	Aging Time, hr	Aging Temp, °F	Elongation(2), % in 1 inch	Reduction of Area(2), %	Yield Strength(3), 0.2% Offset, psi	Ultimate Tensile Strength(2), psi	VHN (10-Kg Load)(4)
WT132A	0.0	1450	1300	-	-	22.0	45.0	153,500	156,000	350
		"	"	8	1100	29.5	56.5	131,000	139,500	333
		"	1400	-	-	10.0	29.5	179,500	181,000	383
		"	"	2	1000	18.5	40.5	168,500	172,500	369
		"	1600	-	-	0.0(3)	0.0(3)	-	165,500(3)	401
		"	"	8	900	0.0	0.0	-	194,000	405
WT51A	0.0	1600	1300	8	"	1.0(3)	1.4(3)	-	169,500(3)	355
		"	1400	24	"	0.0(3)	2.0(3)	-	174,500(3)	384
		"	1600	-	-	2.5	2.8	-	178,000	389
		"	"	8	900	0.0(3)	0.5(3)	-	197,500(3)	396
WU74A	1.0	1450	1300	-	-	17.5	31.5	148,500	167,000	331
		"	"	8	1100	26.5	43.5	129,500	136,500	317
		"	1400	-	-	1.5	4.0	-	181,500	407
		"	"	4	1000	5.0	9.0	167,500	178,000	385
		"	1600	-	-	0.0	0.0	-	158,500	401
		"	"	8	900	0.0(3)	0.0(3)	-	212,000(3)	437
		1600	1300	-	-	17.0	30.0	145,000	155,500	342
		"	"	8	1100	24.0	33.0	128,500	135,500	326
		"	"	8	900	5.0	7.0	173,000	187,000	394
		"	1400	-	-	2.0	5.5	-	182,000	397
		"	"	24	800	0.0(3)	0.0(3)	-	219,500(3)	464
		"	"	4	1000	2.0	5.0	-	181,500	372
		"	1600	-	-	1.0(3)	1.5(3)	-	180,500(3)	403
		"	"	8	900	0.0	0.0	-	208,000	429
WU75A	2.0	1450	1300	-	-	13.0	20.5	143,500	154,000	347
		"	"	8	1100	20.0	26.0	124,500	133,500	311
		"	1400	-	-	2.5	7.5	-	168,500	355
		"	"	4	1000	2.0	3.5	-	171,500	363

TABLE 23. (Continued)

Heat No.	Nominal Thorium Content, %	Rolling Temp, °F	Solution Temp(1), °F	Aging Time, hr	Aging Temp, °F	Elongation(2), % in 1 inch	Reduction of Area(2), %	Yield Strength(3), 0.2% Offset, psi	Ultimate Tensile Strength(2), psi	VHN (10-Kg Load)(4)
WU75A	2.0	1450	1600	-	-	0.5	1.5	-	163,000	374
		"	"	8	900	-	-	-	201,000(3)	408
		1600	1300	-	-	11.5	18.5	148,500	156,500	336
		"	"	8	1100	16.5	20.0	114,000	127,000	305
		"	"	-	900	0.0(3)	0.5(3)	-	175,500(3)	389
		"	1400	-	-	0.0	0.0	-	149,500	385
		"	"	24	800	-	-	-	151,500(3)	459
		"	"	4	1000	0.0(3)	0.0(3)	-	174,000(3)	369
		"	1600	-	-	1.0(3)	4.0(3)	-	146,500(3)	379
		"	"	8	900	-	-	-	189,500(3)	420

(1) Held 1 hour at temperature and cold water quenched.
(2) Average of two values, except where noted.
(3) Single values.
(4) Average of three impressions.

TABLE 24. EFFECT OF SUBSEQUENT HEAT TREATMENT ON THE PROPERTIES OF A Ti-3Mn-COMPLEX ALLOY ROLLED OR SOLUTION TREATED IN THE BETA FIELD (1600°F)

Heat No.	Time in Beta Field (1600°F)	Cooling Medium	Solution(1) Time, hr	Solution(1) Temp, °F	Aging Time, hr	Aging Temp, °F	Elongation(2), % in 1 inch	Ultimate Tensile Strength(2), psi	VHN (10-Kg Load)(3)
WT136A(4)	1 hr	Cold-water quench	--	--	8	900	0.5	204,000	-
	Ditto	Ditto	--	--	"	"	1.0	200,500	383
WU24A(4)	"	"	48	1300	"	"	2.0	163,000	-
	"	"	"	"	"	"	1.0	164,500	366
	"	"	"	1475	"	"	1.0	190,500	-
	"	"	"	"	"	"	0.0	180,000	394
WT107A	During rolling	Air	1	1300	"	"	19.0	155,500	-
	Ditto	"	"	"	"	"	26.0	150,000	342
	"	"	"	1400	"	"	11.0	177,500	-
	"	"	"	"	"	"	12.0	175,000	374
	"	"	"	1600	"	"	0.5	210,000	-
	"	"	"	"	"	"	0.5	200,500	409

(1) Cold water quenched from solution temperature.
(2) Single values.
(3) Average of three impressions.
(4) Rolled at 1450°F.

From these results, it would seem that the proposed mechanism based on segregation of oxygen, nitrogen, or carbon at the beta grain boundaries does not completely explain beta embrittlement. If these elements are partitioned between the alpha and the beta phases, then the 48-hour solution treatments in the alpha-beta field would be expected to restore ductility.

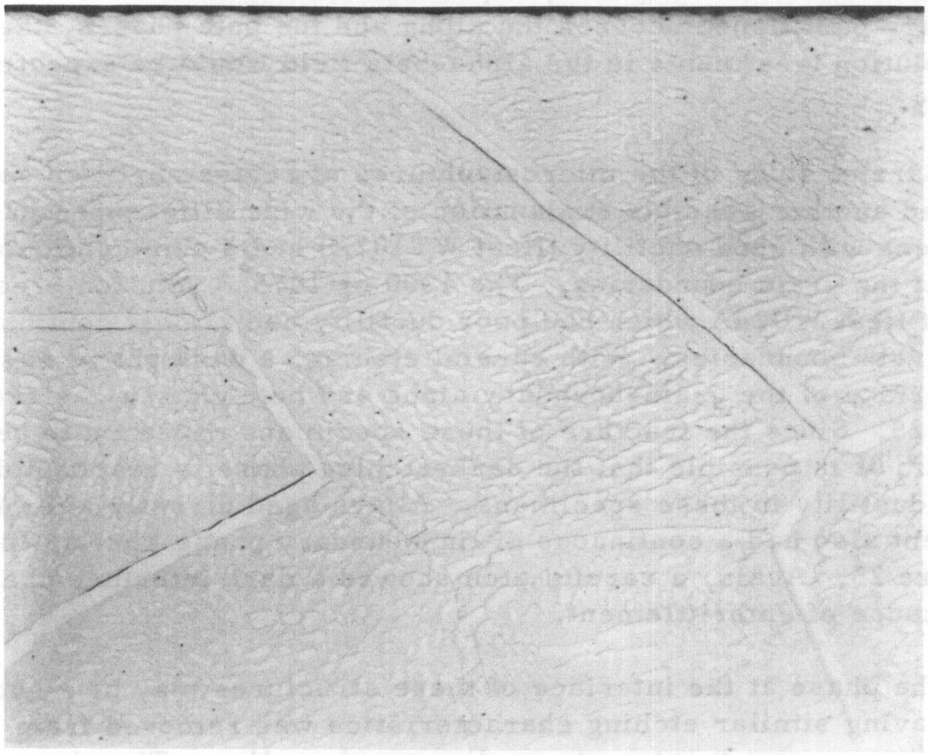
Careful study of the microstructures of beta-embrittled specimens indicated another possible explanation of the wide differences in ductility. Specimens with good ductility (Heat WT107A) had a semicontinuous alpha phase at the grain boundaries. The 1300 or 1475°F solution-treated specimens of Heat WU24A which had poor ductility had almost continuous alpha at the grain boundaries. With careful etching, a dark phase was shown at the interface of the grain-boundary alpha and beta phases, as seen in Figure 25. Since the fracture of these specimens appeared to be intergranular, it is possible that the dark-etching phase is responsible for the lack of ductility in these specimens. Alloys aged directly after the beta treatment also had a continuous grain-boundary phase similar to the alpha in Figure 25. Again, a careful etch showed a dark interface phase that may be the cause of embrittlement.

The phase at the interface of these structures may be a hydride. A phase having similar etching characteristics was removed from sheet alloys by a vacuum treatment, as discussed earlier. The phase precipitates at the alpha-beta interface, and could cause an accentuated embrittlement if the alpha phase were at all continuous in the microstructure. In bar stock rolled at 1450°F, the alpha is completely random and, therefore, the phase precipitation at the alpha-beta interface would not be expected to cause as much embrittlement.

To determine the role of hydrogen in embrittlement caused by solution treating in the beta range, 1/2-inch bar stock specimens of the following alloys were degassed in a high vacuum for 8 hours at 1600°F to remove hydrogen:

Ti-3Mn-1Cr-1Fe-1Mo-1V
Ti-4Fe-1Cr-1Mn-1Mo-1V
Ti-3.5Mn-3.5Cr

All specimens were solution treated at 1600°F for 1 hour and water quenched. Half of the specimens were then re-solution treated for 1 hour at 1400°F and water quenched. These solution treatments were followed by an aging treatment at 900°F for 8 hours for all specimens. The tensile and hardness data for the heat-treated specimens of these alloys are listed in Table 25. The degassed bars had essentially zero ductility after all heat treatments. Microexamination of these specimens showed continuous or semicontinuous alpha at the grain boundaries. The dark-etching phase at the interface of the alpha and beta phases was also observed. This



500X 1-1/2% HF - 3-1/2% H_2O_2 Etch N2720

FIGURE 25. Ti-3Mn-COMPLEX ALLOY SOLUTION TREATED 1 HR AT 1600°F AND WATER QUENCHED, RE-SOLUTION TREATED 48 HR AT 1300°F, WATER QUENCHED, AND AGED 8 HR AT 900°F

Primary Widmanstätten and grain-boundary alpha in a beta matrix containing unresolved alpha precipitate. Note dark-etching phase at α-β interface. This phase is believed to be a hydride.

TABLE 25. PROPERTIES OF SELECTED ALLOYS AFTER VACUUM ANNEALING(1) TO REMOVE HYDROGEN AND THEN HEAT TREATING

Heat No. (Nominal Composition, %)	Heat Treatment: Initial Solution(2) Time, hr	Initial Solution(2) Temp, °F	Final Solution(2) Time, hr	Final Solution(2) Temp, °F	Aging Time, hr	Aging Temp, °F	Elongation(3), % in 1 inch	Ultimate Tensile Strength(3), psi	VHN (10-Kg Load)(4)
WU24A 3 Mn-	1	1600	-	-	8	900	0.0	207,500	-
1Cr-1Fe-	"	"	-	-	"	"	0.0	211,000	433
1Mo-1V	"	"	1	1400	"	"	1.0	139,500	-
	"	"	"	"	"	"	-	164,000	390
WT170A 4Fe-	"	"	-	-	"	"	0.0	176,000	473
1Cr-1Mn-	"	"	-	-	"	"	0.0	171,000	468
1Mo-1V	"	"	1	1400	"	"	0.0	135,500	-
	"	"	"	"	"	"	0.0	144,500	450
WT181A 3.5	"	"	-	-	"	"	0.0	206,500	-
Mn-3.5Cr	"	"	-	-	"	"	0.0	210,000	409
	"	"	1	1400	"	"	1.0	187,500	-
	"	"	"	"	"	"	1.5	186,000	387

(1) Vacuum annealed 8 hours at 1600°F.
(2) Cold water quenched from solution temperature.
(3) Single values.
(4) Average of three impressions.

indicates that either the hydrogen content was not lowered sufficiently by the vacuum treatment used or the dark-etching phase is not a hydride. This part of the investigation will be continued under the extension to the present contract.

DEVELOPMENT OF A HIGH-STRENGTH FORGING ALLOY

The beta embrittlement of the alloys now being evaluated is a serious deterrent to their use as forging alloys. Forging these alloys in the alpha-beta field is difficult because of their stiffness at the low temperatures required. On the other hand, if the alloys are hot worked in the beta range (above about 1500°F) without being finished in the alpha-beta range, they have very limited ductility. After several tests to determine the cause of the beta embrittlement, the cause is still not apparent. In certain commercial alloys, e.g., the Ti-3Al-5Cr produced by Mallory-Sharon Titanium, Incorporated, the alpha-stabilizing element aluminum raises the beta transus temperature and thus permits higher forging temperatures to be used without obtaining an all-beta structure. However, large additions of aluminum also make the alloys more difficult to forge than alloys of the Ti-3Mn-complex type. It was decided that small additions of aluminum might permit somewhat higher forging temperatures without increasing the stiffness of alloys such as the Ti-3Mn-complex and the Ti-5Mn-2.5Cr alloys.

In order to study this possibility, aluminum additions of 0.5, 1.0, and 2.0 per cent were made to the Ti-5Mn-2.5Cr alloy and 0.5 and 1.0 per cent to the Ti-3Mn-complex alloy. The Ti-5Mn-2.5Cr alloys with aluminum were rolled at 1600°F to 1/2-inch-diameter bar stock, solution treated at 1400°F, and aged at 800, 1000, and 1100 F for various times. The Ti-3Mn-complex ingots with aluminum were split and rolled at 1600 and 1700°F to 1/2-inch-diameter bar stock. The specimens were then solution treated at 1300, 1400, or 1600°F and aged at 800, 1000, or 1100°F. Since 1600°F is in the beta-phase region for both base compositions, the properties obtained on the bar stock should be fairly representative of those which would be obtained on small forgings worked in this temperature range.

The results of the tensile tests and the hardnesses for the two alloys are given in Tables 26 and 27, along with those of the base compositions. The Ti-5Mn-2.5Cr-0.5Al alloy showed a substantial increase in ductility over that obtained in the alloy without aluminum, as shown in Table 26. The strength-ductility relationship for the aluminum-bearing alloy rolled at 1600°F was about the same, up to a strength level of 180,000 psi, as that of the base alloy rolled at 1450°F. Aluminum contents greater than 0.5

TABLE 26. EFFECT OF ALUMINUM ON TENSILE PROPERTIES AND HARDNESSES OF 1/2-INCH-DIAMETER BAR STOCK OF THE Ti-5.0Mn-2.5Cr ALLOY ROLLED AT 1600°F

Heat No.	Nominal Al Content, %	Solution Temp(1), °F	Aging Time, hr	Aging Temp, °F	Elongation(2), % in 1 inch	Reduction of Area(2), %	Yield Strength(3), 0.2% Offset, psi	Ultimate Tensile Strength(2), psi	VHN (10-Kg Load)(4)
WT51A	0.0	1350	-	-	9.5(3)	25.7(3)	-	170,000(3)	385
		1300	48	800	0.0(3)	0.0(3)	-	116,500(3)	409
		"	8	900	1.0(3)	1.4(3)	-	169,500(3)	355
		1400	"	"	0.0(3)	0.0(3)	-	189,000(3)	397
WU60A	0.5	1400	-	-	10.5	19.0	161,500	163,500	348
		"	24	800	2.0	3.0	-	235,500	420
		"	4	1000	19.0	38.5	172,500	174,000	370
		"	"	1100	29.0	47.0	140,000	144,000	327
WU61A	1.0	1400	-	-	19.5	33.5	147,500	151,500	336
		"	24	800	0.0	1.5	-	211,500	409
		"	4	1000	20.5	48.5	170,500	175,000	363
		"	"	1100	28.0	57.5	142,500	145,500	332
WU57A	2.0	1400	-	-	19.5	40.0	141,500	145,500	332
		"	24	800	0.0	0.0	-	224,500	389
		"	4	1000	15.0	30.0	183,500	184,500	387
		"	"	1100	22.5	45.0	150,500	152,000	341

(1) Quenched into cold water after 1 hour at temperature.
(2) Average of two values, except where noted.
(3) Single values.
(4) Average of three impressions.

TABLE 27. EFFECT OF ALUMINUM ON TENSILE PROPERTIES AND HARDNESSES OF 1/2-INCH-DIAMETER BAR STOCK OF THE Ti-3Mn-1Cr-1Fe-1Mo-1V ALLOY ROLLED AT TWO DIFFERENT TEMPERATURES

Heat No.	Nominal Al Content, %	Rolling Temp, °F	Solution Temp(1), °F	Aging Time, hr	Aging Temp, °F	Elongation(2), % in 1 inch	Reduction of Area(2), F	Yield Strength(3), 0.2% Offset, psi	Ultimate Tensile Strength(2), psi	VHN (10-Kg Load)(4)
WT107A	0	1600	1300	24	900	22.5	35.5	129,500	149,500	317
		"	1400	"	"	13.5	29.0	150,000	166,500	360
		"	"	48	800	4.5	7.5	185,500	200,500	397
		"	1600	"	900	1.0	1.5	-	226,000	446
WT112A	0.5	1600	1300	8	1100	22.0	36.0	118,000	133,000	319
		"	1400	4	1000	14.5	21.0	133,500	152,000	336
		"	"	48	800	4.0	7.5	154,000	182,000	382
		"	1600	4	1000	5.0	8.5	158,000	170,000	357
		1700	1300	8	1100	18.5	28.0	123,000	133,500	320
		"	1400	4	1000	12.0	17.0	135,500	149,000	342
		"	"	48	800	2.5	5.5	-	183,000	364
		"	1600	4	1000	1.5	4.7	-	175,000	363
WU113A	1.0	1600	1300	8	1100	23.0	34.0	116,500	129,000	307
		"	1400	4	1000	15.5	21.0	129,500	146,000	322
		"	"	48	800	4.5	6.5	145,000	183,000	379
		"	1600	4	1000	2.0	6.0	-	170,000	408
		1700	1300	8	1100	20.5	25.5	120,000	129,000	309
		"	1400	4	1000	11.0	18.0	135,000	148,000	336
		"	"	48	800	1.5	2.5	-	183,000	382
		"	1600	4	1000	2.0	3.0	-	172,500	391

(1) Quenched into cold water after 1 hour at temperature.
(2) Average of two values.
(3) Single values.
(4) Average of three impressions.

per cent seemed to have no additional beneficial effect on ductility. The microstructures of the Ti-5Mn-2.5Cr alloys containing aluminum indicated that they had been hot rolled in the alpha-beta field. This indicates that aluminum additions as small as those used in these experiments raise the beta transus of this alloy significantly.

The Ti-3Mn-complex alloy did not show any improvement in ductility for the aluminum-containing alloys. The microstructures of these alloys indicated that they had been rolled in the beta field and had grain-boundary and Widmanstätten alpha. This was the same structure as that of the base alloy rolled at 1600°F. The base alloy was shown to have somewhat lower tensile elongation when rolled at 1600°F than when rolled at 1450°F, although the ductility of the former was acceptable.

The promising results obtained with the aluminum additions to the Ti-5Mn-2.5Cr alloy will be investigated further.

WELDABILITY EVALUATION

Materials

The nominal compositions of the titanium alloys selected for fusion-welding studies were as follows:

Heat No.	Nominal Composition, per cent
WU43A, WU80A	3.5Cr-3.5V
WU44A, WU79A	5Mn-2.5Cr
WU45A, WU85A	5Mn-2Mo
WU47A, WU83A, WU37A*	4.5Mo-3.5Fe
WU49A, WT170A*	1Cr-4Fe-1Mn-1Mo-1V
WU82A, WU46A, WT181A*	3.5Cr-3.5Mn
WU84A, WU48A, WU24A*	1Cr-1Fe-3Mn-1Mo-1V

Note: Heat numbers marked with an asterisk (*) were fabricated into 1/2-inch bar for flash-welding tests.
All other heats were rolled to 0.100-inch sheet.

The above alloys were upset forged at 1750 to 1800°F to 1-inch slabs and subsequently rolled at 1450°F to 0.100-inch sheet. The alloys for flash welding were forged and rolled to 1/2-inch bar stock. Scale was removed by pickling or by grit blasting.

Welding Procedure

The rolled alloy sheets were cut to provide sufficient material for the welding of longitudinal-bend specimens, tension specimens, and transverse-bend specimens. It was intended to test each alloy in the as-welded condition and in four different heat-treated conditions. However, in some cases, there was not enough sheet material to make all of the test specimens desired.

The orientations of the test specimens in the welded sheet are shown in Figure 26. The specimens were assembled for welding on copper back-up bars as shown in Figure 27. Then they were placed in a controlled-atmosphere welding chamber, which was evacuated to a pressure of about 100 microns and filled with helium gas. The welding chamber is shown in Figure 28. All welds were made manually with a water-cooled, tungsten-arc torch, using straight-polarity direct current. The welding current was 70 amperes, and the arc voltage was 22 to 24 volts. No filler metals were used. Single-layer welded joints were made in the sheet for longitudinal-bend specimens, and penetration from 75 to 100 per cent was obtained. For making tension tests, 100 per cent penetration was desired. Therefore, these welds were made in two passes, one on each side.

For flash-welding studies, the bars were cut to 3-inch lengths and machined to 3/8-inch diameter, with the ends prepared for welding, as shown in Figure 29. Welding was done in a 30-kva, flash-welding machine using the following conditions:

Initial die opening - 11/16 inch
Flash-off - 0.18 inch
Upset - 0.10 inch
Total upset - 0.28 inch
Flashing time - 2.6 seconds
Upset time - minimum
Current on time - 2.71 seconds
Secondary volts - 4.3
Approximate secondary amperes - 18,000
Argon-shielding-gas flow - 28 cfh

Prior to testing, the welds were machined to 1/4-inch-diameter test bars to remove the excess weld metal and to correct any misalignment that might have occurred in welding.

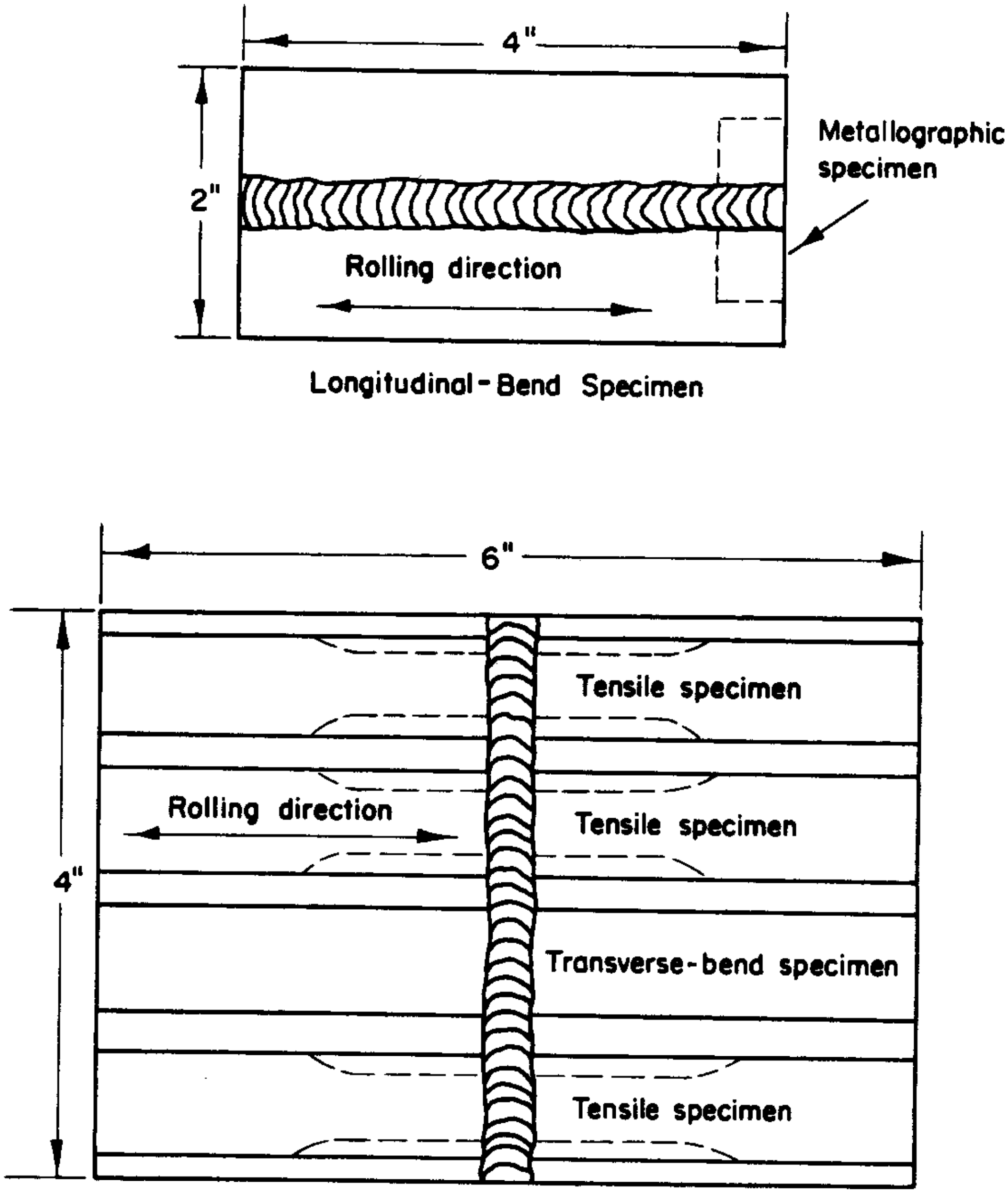


FIGURE 26. ORIENTATION OF TEST SPECIMENS IN $\frac{1}{10}$ -INCH-THICK WELDED SHEET

A-8548

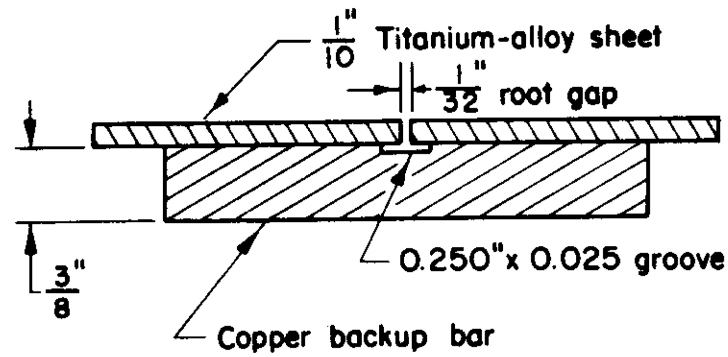


FIGURE 27. SKETCH OF SHEET ASSEMBLED FOR TUNGSTEN-ARC WELDING

A-8549

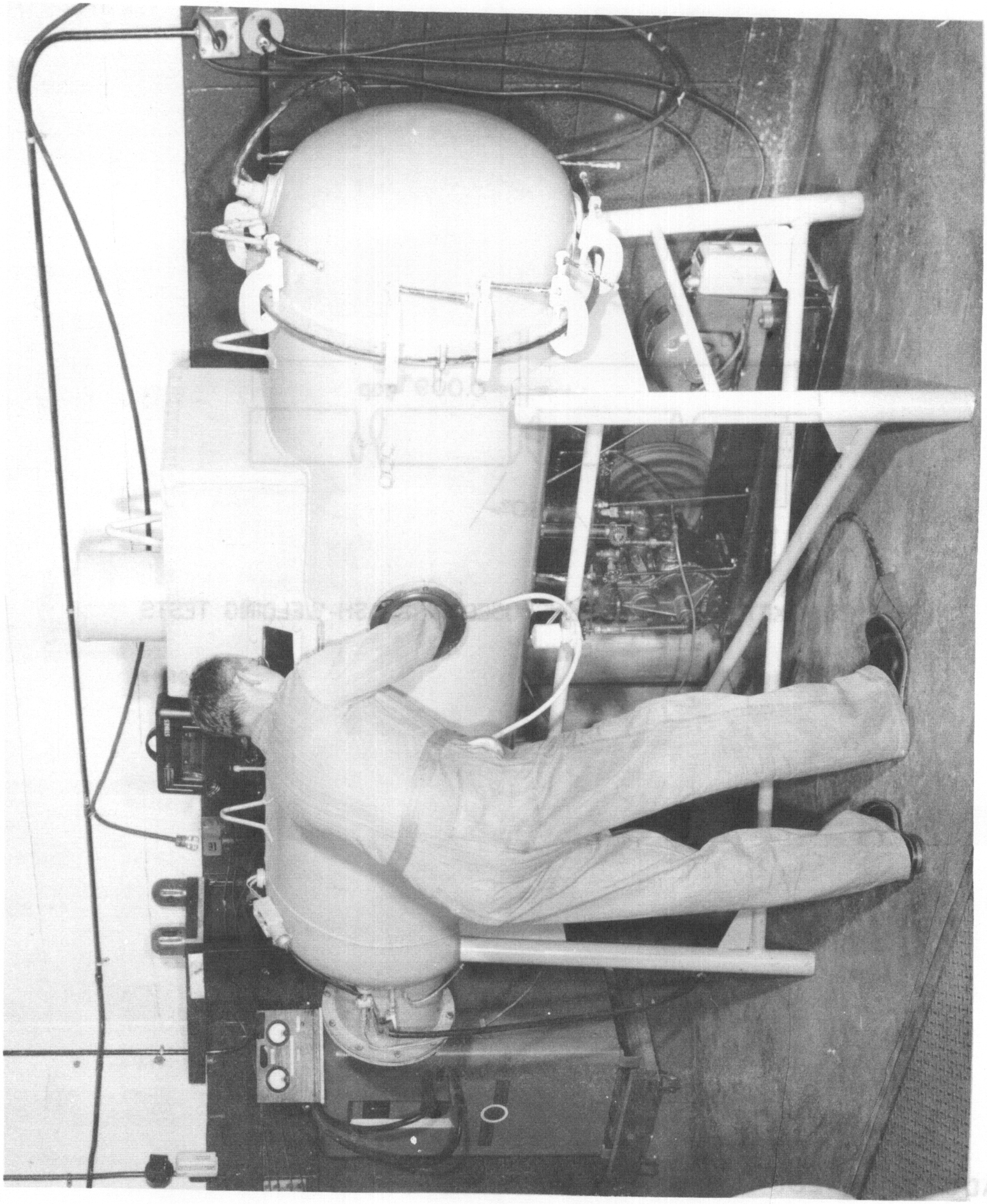


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FIGURE 28. CONTROLLED-ATMOSPHERE WELDING CHAMBER

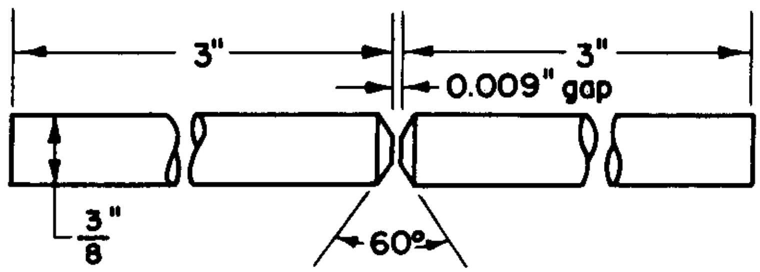

FIGURE 29. SKETCH OF SPECIMENS USED IN FLASH-WELDING TESTS

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Postweld Heat Treatments

The postweld heat treatments applied to the tungsten-arc-welded joints were as follows:

(1) One hour at 1300°F, water quenched, aged 8 hours at 1100°F. This treatment resulted in high ductility and relatively low (130,000 to 150,000 psi) strength in the base material.

(2) One hour at 1400°F, water quenched, aged at temperatures between 900 and 1100°F for appropriate times to give strength levels of about 170,000 psi in the base metal. The aging temperature and time varied somewhat with each alloy.

(3) One hour at 1600°F, furnace cooled.

(4) One hour at 1600°F, furnace cooled to 1350°F, water quenched, and aged at the same temperatures and times as used in Treatment 2.

The initial temperature of 1600°F used in the latter two heat treatments is in the beta-phase region for all of the alloys. The 1300 and 1400°F treatments, on the other hand, resulted in a two-phase, alpha-beta, structure prior to aging. Details of the aging times and temperatures used in Treatments 2 and 4 are given, along with the mechanical properties of the welds, in Table 28.

The flash-welded specimens were given Treatments 1 and 2 only. Details of these treatments are included in Table 29.

Test Procedures

Prior to testing, all specimens were pickled to remove any surface oxides that were formed during welding, heat treating, or machining. The pickling treatment consisted of a quick dip in a 30 per cent sulfuric acid-2 to 3 per cent ammonium bifluoride-water solution to remove scale, followed by immersion in a 15 per cent nitric acid-5 per cent hydrofluoric acid-water solution to remove about 0.002 inch from each surface of the specimen.

TABLE 28. PROPERTIES OF TUNGSTEN-ARC WELDS IN SELECTED TITANIUM ALLOYS

Heat No.	Alloy Composition, %	Heat Treatment	Ultimate Tensile Strength(1), psi	Longitudinal Bend Ductility(2), T	Transverse Bend Ductility(2), T	VHN (10-Kg Load) Weld Metal	VHN (10-Kg Load) Base Metal	Base-Metal Bend Ductility(2), T
WU43A	3.5Cr-3.5V	As welded	155,000	-	-	481	380	3.75(3)
WU43A	3.5Cr-3.5V	1 hr 1300°F; WQ; 8 hr 1100°F	115,000	3.75	-	318	285	1.25
WU43A	3.5Cr-3.5V	1 hr 1400°F; WQ; 2 hr 1000°F	166,000	>15	15	380	360	5
WU80A	3.5Cr-3.5V	1 hr 1600°F; furnace cool	119,000	-	10	302	312	4.5
WU80A	3.5Cr-3.5V	1 hr 1600°F; FC to 1350°F; WQ; 6 hr 900°F	119,000	>15	>15	370	365	11
WU44A	5Mn-2.5Cr	As welded	153,000	>15	-	432	430	3.75(3)
WU44A	5Mn-2.5Cr	1 hr 1300°F; WQ; 8 hr 1100°F	124,000	15	-	330	315	0.5
WU44A	5Mn-2.5Cr	1 hr 1400°F; WQ; 2 hr 1000°F	130,000	>15	>15	383	390	2.5
WU79A	5Mn-2.5Cr	1 hr 1600°F; furnace cool	118,000	7.5	6.25	312	320	1.9
WU79A	5Mn-2.5Cr	1 hr 1600°F; FC to 1350°F; WQ; 2 hr 1000°F	165,000	>15	>15	357	380	3.75
WU45A	5Mn-2Mo	As welded	144,000	>15	-	507	355	2.5(3)
WU45A	5Mn-2Mo	1 hr 1300°F; WQ; 8 hr 1100°F	114,000	3.75	-	306	275	1
WU45A	5Mn-2Mo	1 hr 1400°F; WQ; 2 hr 1000°F	142,000	15	15	362	345	2.2
WU85A	5Mn-2Mo	1 hr 1600°F; furnace cool	116,000	7.5	3	292	305	1.5
WU85A	5Mn-2Mo	1 hr 1600°F; FC to 1350°F; WQ; 2 hr 1000°F	132,000	10	7.5	333	351	3.5

TABLE 28. (Continued)

Heat No.	Alloy Composition, %	Heat Treatment	Ultimate Tensile Strength[1], psi	Longitudinal Bend Ductility[2], T	Transverse Bend Ductility[2], T	VHN (10-Kg Load) Weld Metal	VHN (10-Kg Load) Base Metal	Base-Metal Bend Ductility[2], T
WU47A	4.5Mo-3.5Fe	As welded	146,000	>15	-	497	415	5[3]
WU47A	4.5Mo-3.5Fe	1 hr 1300°F; WQ; 8 hr 1100°F	120,000	7.5	-	309	305	1
WU47A	4.5Mo-3.5Fe	1 hr 1400°F; WQ; 4 hr 1100°F	141,000	15	15	339	325	0.6
WU83A	4.5Mo-3.5Fe	1 hr 1600°F; furnace cool	131,000	10	12.5	298	306	2.5
WU83A	4.5Mo-3.5Fe	1 hr 1600°F; FC to 1350°F; WQ; 4 hr 1100°F	127,000	>15	15	349	345	2.5
WU49A	4Fe-1Cr-1Mn-1Mo-1V	As welded	-	>15	-	412	430	-
WU49A	4Fe-1Cr-1Mn-1Mo-1V	1 hr 1300°F; WQ; 8 hr 1100°F	-	7.5	-	315	305	-
WU49A	4Fe-1Cr-1Mn-1Mo-1V	1 hr 1400°F; WQ; 4 hr 1100°F	-	-	-	326	330	-
WU82A	3.5Cr-3.5Mn	As welded	61,000	>15	-	450	375	1.9[3]
WU82A	3.5Cr-3.5Mn	1 hr 1300°F; WQ; 8 hr 1100°F	119,000	>15	-	329	315	2.5
WU82A	3.5Cr-3.5Mn	1 hr 1400°F; WQ; 2 hr 1000°F	165,000	>15	>15	391	375	7.5
WU46A	3.5Cr-3.5Mn	1 hr 1600°F; furnace cool	129,000	10	15	305	325	7.5
WU46A	3.5Cr-3.5Mn	1 hr 1600°F; FC to 1350°F; WQ; 2 hr 1000°F	139,000	>15	>15	341	342	7.5
WU84A	3Mn-1Cr-1Fe-1Mo-1V	As welded	130,000	>15	-	447	355	2.5[3]
WU84A	3Mn-1Cr-1Fe-1Mo-1V	1 hr 1300°F; WQ; 8 hr 1100°F	53,500	>15	-	320	325	3.75
WU84A	3Mn-1Cr-1Fe-1Mo-IV	1 hr 1400°F; WQ; 2 hr 1000°F	-	>15	>15	387	375	7.5
WU48A	3Mn-1Cr-1Fe-1Mo-1V	1 hr 1600°F; furnace cool	113,000	-	>15	365	360	>15
WU48A	3Mn-1Cr-1Fe-1Mo-1V	1 hr 1600°F; FC to 1350°F; WQ; 2 hr 1000°F	119,000	-	>15	410	418	>15

(1) Testing speed = 0.10 inch per minute. (2) T = sheet thickness. (3) Hot rolled and stress relieved.

TABLE 29. PROPERTIES OF FLASH-WELDED JOINTS

Alloy Composition, %	Heat Treatment	Ultimate Tensile Strength, psi	Elongation[1], % in 1 inch
3Mn-1Cr-1Fe-1Mo-1V	As welded	171,000	<1
Ditto	1 hr 1300°F; WQ; 8 hr 1100°F	138,000	15
"	1 hr 1400°F; WQ; 2 hr 1000°F	140,000	2
3.5Cr-3.5Mn	As welded	182,000	<1
Ditto	1 hr 1300°F; WQ; 8 hr 1100°F	142,000	20
"	1 hr 1400°F; WQ; 2 hr 1000°F	166,000	1
4.5Mo-3.5Fe	As welded	149,000	<1
Ditto	1 hr 1300°F; WQ; 8 hr 1100°F	140,000	19
"	1 hr 1400°F; WQ; 4 hr 1100°F	151,000	8
1Cr-4Fe-1Mn-1Mo-1V	As welded	55,500	<1
Ditto	1 hr 1300°F; WQ; 8 hr 1100°F	159,000	12
"	1 hr 1400°F; WQ; 4 hr 1100°F	52,000	<1

(1) Elongation across weld interface, including weld metal, heat-affected zone, and base metal.

Longitudinal-bend, transverse-bend, tension, and hardness tests were made on the tungsten-arc-welded joints. The bend tests were the guided-bend type and were performed using dies of different radii, starting with those with largest radii and reducing the die radius until failure of the specimens occurred. The criterion for failure was the appearance of an open defect measuring over 1/16 inch in any direction. The longitudinal-bend tests of the welded specimens were made to determine the ability of the metal of the different zones of the welded joints to deform plastically together. Straining was produced in a direction parallel to the weld. The transverse-bend tests were made in such a manner that the direction of straining was normal to the weld. These specimens were placed in the dies with the weld metal directly beneath the plunger, causing maximum straining to occur in the weld metal.

The weld-joint tension specimens were tested at a platen speed of 0.10 inch per minute, with the weld located in the center of the reduced section. The ultimate tensile strength and location of failure were recorded.

Vickers hardness surveys were made along cross sections of the welded joints, using a diamond pyramid indentor and a 10-kilogram load. Hardness readings were taken at 0.03-inch intervals, starting in the approximate center of the weld and proceeding out to the heat-affected zone and base metal. A single traverse was made across each welded joint. Microstructural examinations of the welded joints were made to determine whether a correlation existed between structural variations and weld-joint properties.

The flash-welded specimens were tested in tension in the as-welded condition and in the two alpha-beta, heat-treated conditions.

Discussion and Results

The results of the tension and bend and hardness tests of the tungsten-arc-welded joints are shown in Table 28. In the as-welded condition, all of the alloys were very brittle, as may be seen from the longitudinal-bend-test results. Because of this embrittlement, the tensile strengths shown for the as-welded condition are not representative of the true strength of the welds. The hardnesses of the weld metal in all alloys would indicate much higher strengths than those actually obtained.

In general, the postweld heat treatment, consisting of quenching from 1300°F and aging at 1100°F, improved the ductility of the welds in most of the alloys. The best ductility, as indicated by the longitudinal-bend test, was attained in the Ti-3.5Cr-3.5V and the Ti-5Mn-2Mo alloys, which bent over a 3.75T radius. The strength level after this heat treatment, however, was relatively low, ranging from about 110,000 to 124,000 psi. The ductility of the Ti-3.5Cr-3.5Mn and the Ti-3Mn-complex alloys was not increased appreciably by this heat treatment. It is believed, however, that the heats of the Ti-3Mn-complex alloy were contaminated, because the ductility of the base metal was much lower than obtained for several other heats of this composition. The results given in Table 28 for this alloy should not be considered representative.

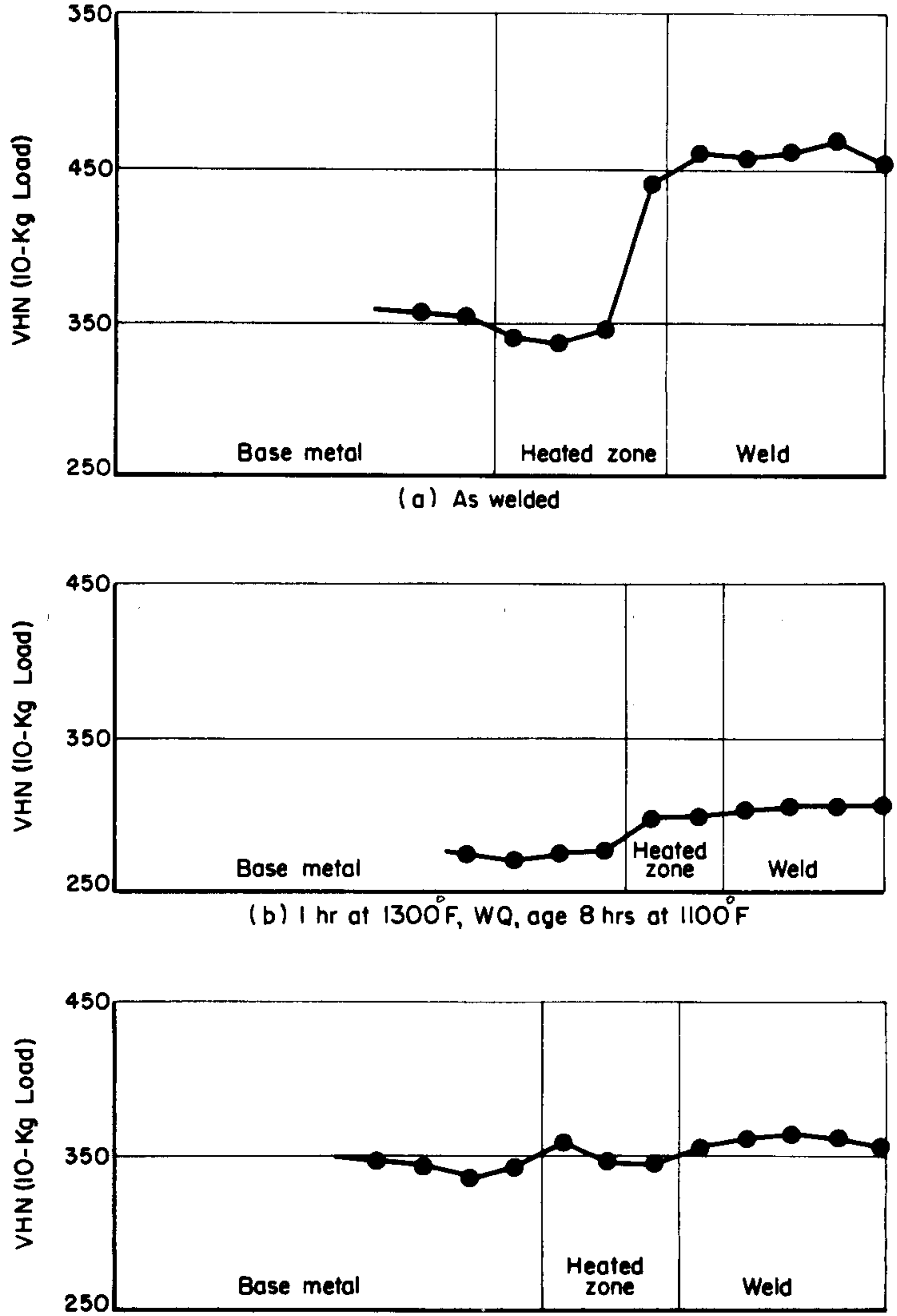
Quenching from 1400°F and overaging to produce a strength level of about 170,000 psi resulted in little improvement in ductility over that obtained in the as-welded condition in any of the alloys. Again, the tensile strengths were lower than would be expected from the hardness data, probably because of extreme embrittlement.

Furnace cooling from 1600°F somewhat improved the bend ductility of four of the seven alloys. The ductilities were not so good as those produced by the first heat treatment, although the resulting strength levels were about the same. The Ti-5Mn-2Mo and the Ti-5Mn-2.5Cr alloys exhibited the best ductility after this heat treatment.

Furnace cooling from 1600 to 1350°F, quenching, and overaging had little effect on the low ductility of the weld joints, with the exception of the weld in the Ti-5Mn-2Mo alloy. Both the longitudinal- and the transverse-bend ductilities of this alloy were improved appreciably.

The data plotted in Figure 30 represent typical hardness surveys made on fusion-welded specimens in the as-welded and heat-treated conditions. Generally, the hardnesses of the weld metal and the heat-affected zone were higher in the as-welded condition than those of the base metal. Heat treatment largely eliminated these differences.

Metallographic examination of the fusion-welded joints showed that, for all alloys, the as-welded structure consisted of finely dispersed alpha particles in a retained-beta matrix. The high hardnesses in the as-welded condition were probably due to age hardening that occurred in the beta phase during cooling. The postweld heat treatments in the alpha-beta range produced large amounts of alpha precipitated in the beta matrix, which resulted in lower hardnesses and greater ductilities than in the as-welded joints. The beta heat treatments also produced large amounts of alpha. However, the alpha was considerably coarser than that obtained in the alpha-beta heat treatments. The series of photomicrographs in Figures 31 through 35 show the microstructures of the Ti-5Mn-2Mo alloy in the as-welded and variously heat-treated conditions. These structures are representative of all of the alloys used in this investigation.



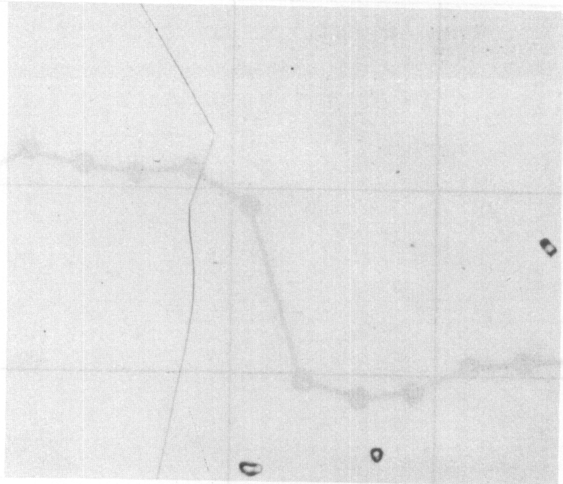
(a) As welded

(b) 1 hr at 1300°F, WQ, age 8 hrs at 1100°F

(c) 1 hr at 1400°F, WQ, age 2 hrs at 1000°F

FIGURE 30. TYPICAL VICKERS HARDNESS SURVEYS ON WELDS MADE IN SHEET OF THE Ti-5 Mn-2 Mo ALLOY

Hardness readings were made after heat treatments indicated



500X 3-1/2% HNO_3 - 1-1/2% HF Etch N5024

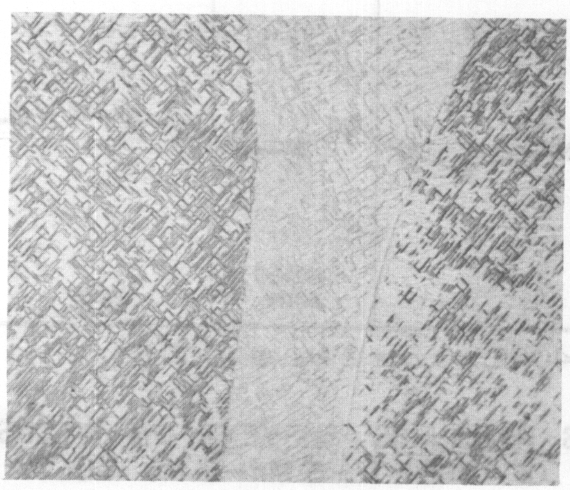
FIGURE 31. Ti-5Mn-2Mo WELD METAL

As welded
Beta prime (hard beta)
VHN 507

500X 3-1/2% HNO_3 - 1-1/2% HF Etch N5025

FIGURE 32. Ti-5Mn-2Mo WELD METAL

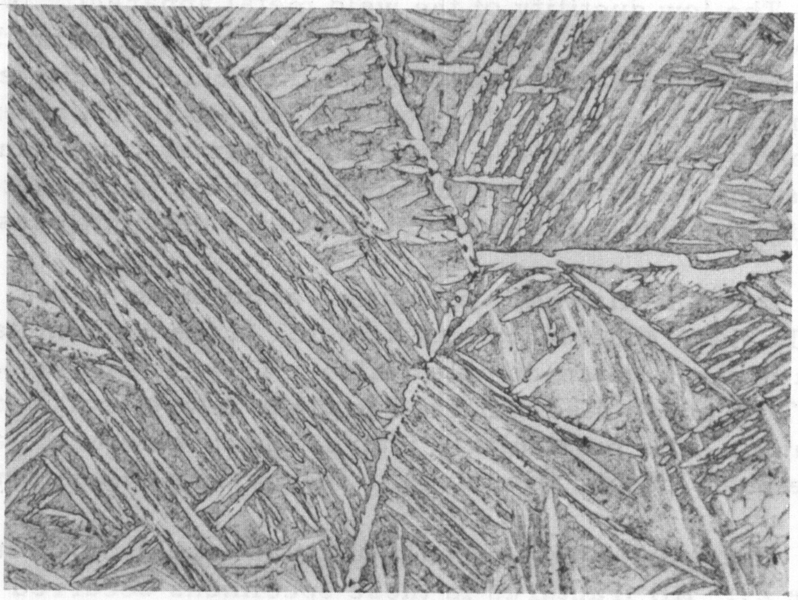
Heat treatment: 1 hr 1300°F, WQ, age 8 hr 1100°F
Widmanstätten alpha in beta matrix
VHN 306



500X 3-1/2% HNO_3 - 1-1/2% HF Etch N5026

FIGURE 33. Ti-5Mn-2Mo WELD METAL

Heat treatment: 1 hr 1400°F, WQ, age 2 hr 1000°F
Widmanstätten alpha in beta matrix
VHN 362



500X 3-1/2% HNO_3 - 1-1/2% HF Etch N8163

FIGURE 34. 5Mn-2Mo ALLOY WELD METAL

Heat treatment: 1 hr 1600°F, furnace cool
Alpha needles in beta matrix
VHN 292

500X 3-1/2% HNO_3 - 1-1/2% HF Etch N8162

FIGURE 35. 5Mn-2Mo ALLOY WELD METAL

Heat treatment: 1 hr 1600°F, furnace cool
to 1350°F, water quench, age 2 hr 1000°F
Alpha needles in beta matrix
VHN 333

The low ductility of the weld metal apparently is associated with the beta-embrittlement phenomenon discussed earlier in this report. The continuous or semicontinuous alpha at the grain boundaries, shown in Figures 32 through 35, apparently produces extreme embrittlement in most conditions of heat treatment. It is not known as yet whether the alpha alone is responsible or whether an unknown phase precipitated at the alpha-beta interface of the grain boundary produces the brittleness.

In connection with the latter, an unknown phase, in the form of small, black spheroids, was detected in the weld-metal-grain-boundary alpha of the Ti-3Mn-complex alloy. This is illustrated in Figures 36 and 37. This phase also occurred in other alloys, but in much smaller amounts. It is particularly interesting in view of the apparent low ductility of the heats of this alloy used in the welding evaluation.

The tensile properties of the flash-welded bar stock are given in Table 29. All fractures in these specimens occurred in the base metal abou. 1/4 inch from the weld. In the as-welded condition, all of the alloys had very low ductility. Quenching from 1300°F and overaging at 1100°F resulted in good ductility at strength levels of 138,000 to 159,000 psi. The second heat treatment shown in Table 29 produced somewhat higher strengths, but much lower ductility.

To summarize briefly, the results of this welding evaluation have shown:

(1) High-strength alloys containing 7 to 8 per cent of beta-stabilizing elements are very brittle as welded.

(2) Reasonable ductility can be restored in some alloys by a heat treatment which reduces the strength drastically.

(3) The brittleness of the weld metal at higher strength levels is associated with the beta-embrittlement phenomenon.

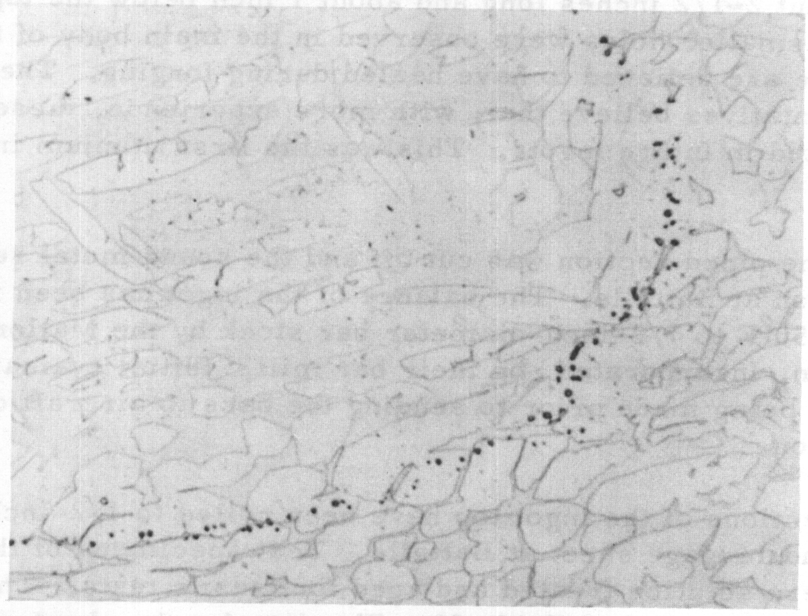
EVALUATION OF SELECTED ALLOYS BY INDUSTRIAL ORGANIZATIONS

One very important step in alloy-development work is the transition from laboratory testing to production. This step has been initiated for the Ti-3Mn-complex alloy. In response to a Wright Air Development

500X 3-1/2% HNO_3 - 1-1/2% HF Etch N8294

FIGURE 36. Ti-3Mn-COMPLEX ALLOY WELD METAL

Heat treatment: 1 hr 1600 F, furnace cool
Alpha needles in beta matrix
VHN 365



1500X 3-1/2% HNO_3 - 1-1/2% HF Etch N8268

FIGURE 37. GRAIN-BOUNDARY AREA IN THE Ti-3Mn-COMPLEX ALLOY WELD METAL

Heat treatment: 1 hr 1600 F, furnace cool
Note globular phase in grain-boundary alpha

Center inquiry, the companies listed in Table 30 have willingness to make evaluation tests if the materials a Half of the material for this evaluation has been melte other half has been melted by the Bureau of Mines as The latter ingot was double melted by a consumable-el About one-quarter of the material has been distributed in the following sections of this report. The balance o nearly ready for distribution.

In addition to this evaluation work, the Mallory-S Incorporated, has melted several 30-pound and one 60 Ti-3Mn-complex alloy. This material will be used for evaluation program.

Evaluation of Bureau of Mines Ingot

A 225-pound ingot of the Ti-3Mn-complex alloy a diameter was double melted by the consumable-electro Bureau of Mines at Albany, Oregon. The as-cast ingo the same as that obtained on Battelle's 20-pound ingots checked ultrasonically and radiographically for soundn pipe about 2-1/2 inches long and about 1 inch below the Several smaller holes were observed in the main body but these are believed to have healed during forging. T representatives believe that, with more experience, th eliminated in future ingots. This was the first titanium melted.

The piped section was cut off and the sound metal evaluation at Battelle. The balance of the ingot has bee from 1550°F to 7/8-inch-diameter bar stock by the Mal Titanium, Incorporated, on their bar mill. Initial eval stock is being made prior to sending the bars to aircra evaluation.

Sections of the ingot top have been rolled to 1/2-i stock and 14-gage sheet at Battelle. Test specimens o rials were solution treated and aged for tensile tests. tests are presented in Table 31. The data for the sheet specimens with rounded corners. The values for squar specimens were given in Table 21. The strength-ductil bar stock are about the same as are usually obtained at ingots. The sheet ductility was lower than that of the b the highest strength level, where it was about the same an effect of the different stress pattern in the rectangul

TABLE 30. INDUSTRIAL ORGANIZATIONS INTERESTED IN EVALUATING EXPERIMENTAL TITANIUM ALLOYS

Company and Location	Type of Material	Evaluation Tests to Be Made
Boeing Airplane Company Seattle, Washington	Forged plate	Tensile, stability, notched tensile, impact, fatigue, creep rupture, compression, shear
Consolidated Vultee Aircraft Corporation San Diego, California	Rolled bars	Tensile, impact, fatigue, prototype part
Curtiss-Wright, Propeller Division Caldwell, New Jersey	Rolled bars	Tensile, fatigue
Curtiss-Wright, Wright Aeronautical Division Wood Ridge, New Jersey	Rolled bars	Tensile, fatigue, impact, creep rupture, compressor blades
Douglas Aircraft Company, Incorporated Santa Monica, California	Rolled bars	Bolt testing
General Electric Company West Lynn, Massachusetts	Ingot for disk forging	Tensile, fatigue, impact, creep
Lockheed Burbank, California	Rolled bars	Tensile, notched tensile, compression, fatigue
Lycoming-Spencer Bridgeport, Connecticut	Rolled bars	Tensile, notched tensile
Mallory-Sharon Titanium, Incorporated Niles, Ohio	Ingots	Forging and rolling, general properties
North American Aviation, Incorporated Columbus Division Columbus, Ohio	Sheet	Prototype part
North American Aviation, Incorporated Downey Plant Downey, California	Rolled bars	General evaluation
North American Aviation, Incorporated Inglewood Plant Inglewood, California	Rolled bars	Tensile, compression, creep rupture
Republic Steel Corporation Massillon, Ohio	Rolled bars	Bolt testing
A. O. Smith Corporation Milwaukee, Wisconsin	Rolled bars	Tensile and bend after flash welding
Superior Tubing Company Norristown, Pennsylvania	Rolled bars	Cold drawing of tubing
Westinghouse Electric Company, Gas Turbine Division Philadelphia, Pennsylvania	Rolled bars	Evaluate for use in jet engines
Wyman-Gordon Company Worcester, Massachusetts	Ingots	Forgeability and properties

TABLE 31. TENSILE PROPERTIES AND HARDNESSES OF BAR STOCK AND SHEET MATERIAL ROLLED FROM 225-POUND INGOT(1) OF THE Ti-3Mn-1Fe-1Cr-1Mo-IV ALLOY MELTED AT THE BUREAU OF MINES

Solution Temp, °F	Aging Time, hr	Aging Temp, °F	Elongation(4), % in 1 inch	Reduction of Area(4), %	Yield Strength(5), 0.2% Offset, psi	Ultimate Tensile Strength(4), psi	VHN (10-Kg Load)(6)
			1/2-Inch-Diameter Bar Stock				
1300(2)	8	900	20.5	42.5	145,000	169,500	369
"	8	1100	30.5	55.0	136,500	146,000	331
1400(2)	24	800	3.5	6.5	-	219,000	424
"	4	1000	21.0	43.5	-	169,000	360
			14-Gage Sheet(7)				
1300(3)	-	-	13.5	-	-	149,000	341
"	24	900	16.0	-	-	160,000	365
"	8	1100	17.0	-	-	138,000	330
1400(3)	-	-	9.0	-	-	163,500	366
"	48	800	4.0	-	-	210,500	439
"	8	1000	12.0	-	-	159,500	363

(1) Designated Heat No. C-1.
(2) Cold water quenched after 1 hour at temperature.
(3) Cold water quenched after 1/2 hour at temperature.
(4) Average of two values.
(5) Single values.
(6) Average of three impressions.
(7) Corners of gage section rounded by hand to about 1/64-inch radius.

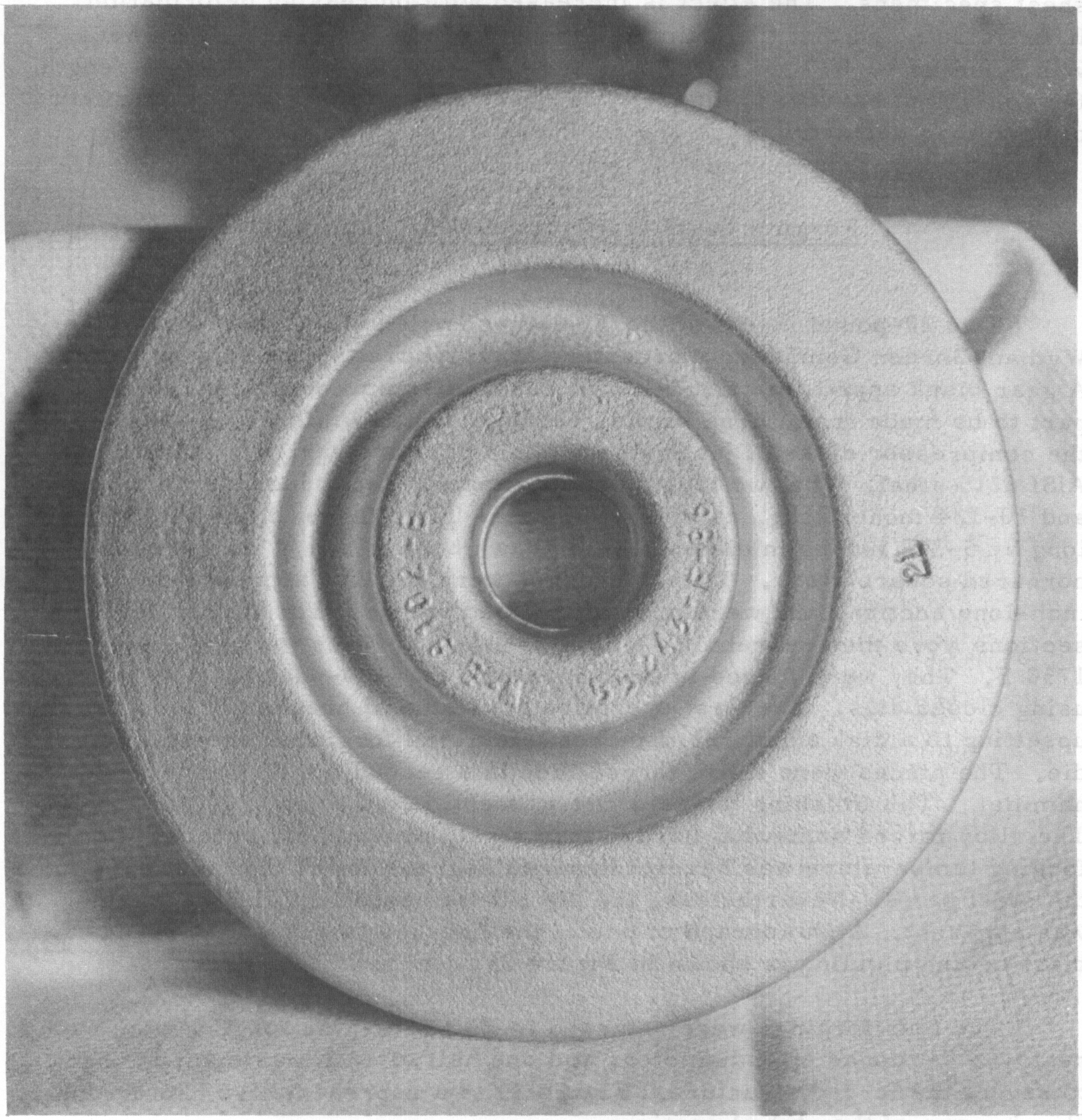
sheet specimens. The effect is increased with increasing deformation. It was also noted that the ultimate strength of the aged sheet specimens was 8,000 to 10,000 psi lower than the bar stock ultimate tensile strength for equivalent hardness values. The lower strength is probably an effect of specimen geometry.

Forging Tests at Wyman-Gordon Company

Two 20-pound ingots of the Ti-3Mn-complex alloy were sent to the Wyman-Gordon Company, Worcester, Massachusetts, for forging tests. A gear blank approximately 7-1/2 inches in diameter was selected as the part to be made from these ingots, because of its similarity in shape to the compressor disks of jet engines. This part ordinarily is made from AISI 6215 steel. The two ingots, approximately 3-3/4 inches in diameter and 10-1/4 inches long, as machined, were upset forged to 5-1/8 inches long by 5-3/8 inches in diameter and drawn back to 3-1/4 inch, round-cornered square stock. Then the billets were cut into approximately 4-inch-long sections and were grit blasted and etched. The four billet sections were then reheated in a rotary-hearth furnace for 1-1/2 hours at 1750°F. They were forged in two operations on 5000-pound steam hammers using closed dies. The first operation consisted of breaking the corners, upsetting to a disk about 1-1/2 inches thick, and forging in the roughing die. The pieces were finish forged then in a second set of dies on another hammer. The finishing temperature was estimated to be 1500 to 1550°F. The alloy forged somewhat harder than steel, presumably because the forging temperature was several hundred degrees lower than that used for the steel parts. Nevertheless, the die fill was excellent and no cracking was apparent. A photograph of one of the finished forgings after sand-blasting and pickling is shown in Figure 38.

The four forgings were returned to Battelle. One forging from each heat was sectioned on a diameter, and one half of each was macroetched to reveal the forged structure. Figure 39 is a representative photograph showing one of the etched sections.

The etched halves were cut again so that four quarter sectors were obtained for heat treating. It was felt that a quarter section would give substantially the same heat-treat response as a full forging. Two sectors were solution treated at 1400°F and one was aged 48 hours at 800°F and the other 8 hours at 875°F to give two strength levels. Two tangential specimens from the outer rim were cut from each sector, and one radial specimen was cut from one sector. The results of tensile tests and hardness surveys of these specimens are given in Table 32, along with those of additional tests made at Battelle, Wyman-Gordon Company, and the

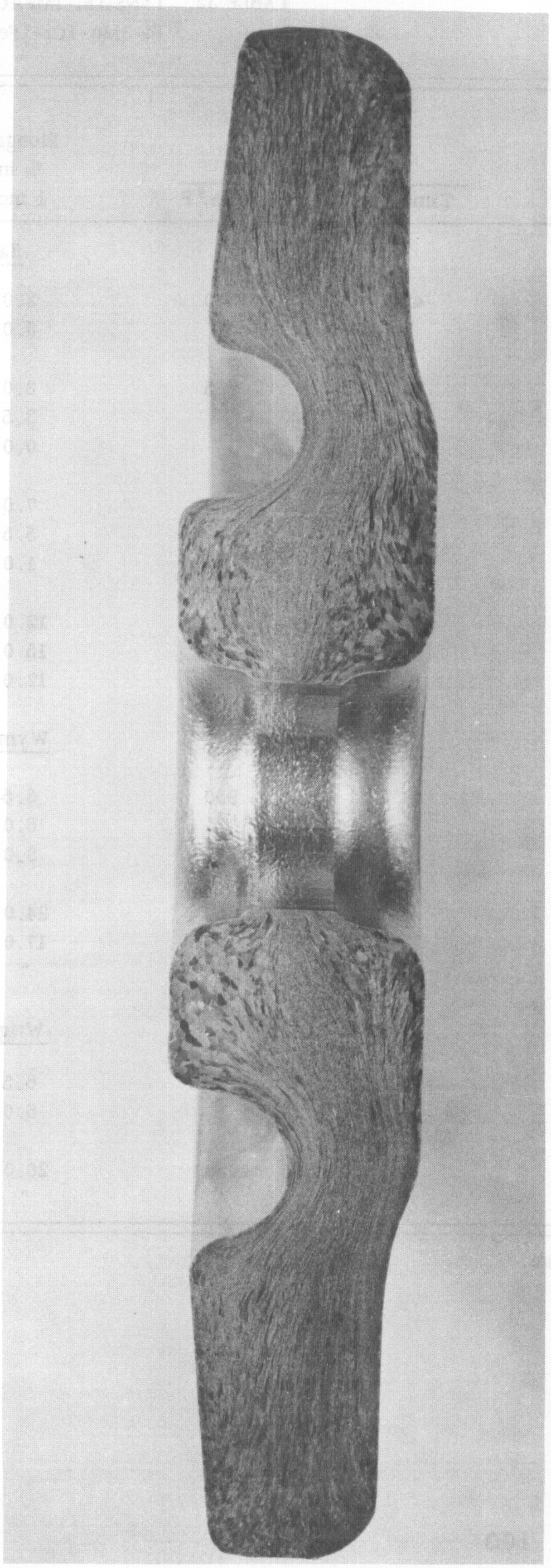


0.75X 98970

FIGURE 38. CLOSED-DIE FORGING OF Ti-3Mn-1Fe-1Cr-1Mo-1V ALLOY FORGED FROM 1750°F

Forging is 7-1/2 inches in diameter

98967



1-1/4X

FIGURE 39. CROSS SECTION OF CLOSED-DIE FORGING

Etch: 2 grams of NH_4FHF in 100 cc of 30% by volume concentrated H_2SO_4

TABLE 32. TENSILE, IMPACT,
Ti-3Mn-1Cr-1Fe-1Mo-

Forging Sector From Heat No.	Orientation of Test Bar in Forging	Solution Temp, °F(1)	Aging Time, hr	Aging Temp, °F	Elongation, % in 1 inch
					Battelle
WU9A	Tangential	1400	48	800	2.0
	"	"	"	"	3.0
WU9A	Tangential	1400	8	875	3.0
	"	"	"	"	3.5
	Radial	"	"	"	0.0
WT160A	Tangential	1400	8	950	7.0
	"	"	"	"	5.5
	Radial	"	"	"	1.0
	Tangential	1400	4	1050	12.0
	"	"	"	"	15.0
	Radial	"	"	"	12.0
					Wyman-
	Tangential	1400	8	950	6.5
	"	"	"	"	6.0
	Radial	"	"	"	3.0
	Tangential	1400	4	1050	24.0
	"	"	"	"	17.0
	Radial	"	"	"	"
					Wright Field
WU9A	Tangential	1350	7	900	6.5
	Radial	"	"	"	6.0
	Tangential	1350	3	1050	20.0
	Radial	"	"	"	"

(1) Cold water quenched after 1 hour at temperature.

(2) Room-temperature test.

AND HARDNESS PROPERTIES OF TWO
1V-ALLOY GEAR-BLANK FORGINGS

Reduction of Area, %	Yield Strength, 0.2% Offset, psi	Ultimate Tensile Strength, psi	Charpy Impact, ft-lb	Hardness VHN (10-Kg Load)	Hardness R_c	Hardness BHN
Data						
4.5	-	205,000	-	417	-	-
5.5	-	203,000	-	429	-	-
7.0	-	205,500	-	405	-	-
"	-	207,500	-	387	-	-
0.0	-	211,500	-	405	-	-
16.5	-	183,000	-	373	-	-
12.0	-	178,000	-	370	-	-
1.0	-	183,500	-	367	-	-
23.5	-	161,500	-	357	-	-
27.0	-	160,000	-	354	-	-
36.0	-	159,000	-	351	-	-
Gordon Data						
11.0	167,000	186,000	-	-	-	363
7.5	174,000	190,000	-	-	-	352
4.0	182,500	197,000	-	-	-	363
42.3	137,500	145,500	-	-	-	302
36.0	147,000	153,000	-	-	-	311
39.2	148,500	155,000	-	-	-	321
Data						
-	164,000	181,500	6.0	-	40	-
-	167,000	181,000	-	-	-	-
-	135,000	144,500	10.0	-	36	-
-	137,000	147,500	-	-	-	-

Materials Laboratory, WADC. The 8-hour age at 875°F was too short to lower the strength, so that both sectors had strengths of about 205,000 psi. The elongation for these tests was low, being about 3 per cent for the tangential specimens and zero for the radial. Two more sectors were heat treated to lower strength levels, as shown in Table 32. At a strength of 180,000 psi, the tangential elongation was about 6.0 per cent and the radial about 1.0 per cent. Lowering the strength by a higher temperature age on another sector to about 160,000 psi gave adequate ductility in both directions of 12 to 15 per cent.

The remaining half gear blanks were sent to Wyman-Gordon Company and to Wright Field for heat treatment and testing. The Wyman-Gordon Company made tangential and radial tensile tests at two strength levels, as in Table 32. At a strength level of 190,000 psi, the tangential elongation was about 6.0 per cent and the radial 3 per cent. At 150,000-psi strength, the elongations were from 17 to 24 per cent. WADC made one tangential and one radial tensile test, as well as one tangential Charpy impact test. At a strength of 181,000 psi, the elongation was about 6 per cent and the Charpy value 6 ft-lb. At 145,000 psi, the elongation was 20 per cent and the Charpy value 10 ft-lb.

The elongations obtained on the gear-blank specimens are somewhat lower than the optimum that has been obtained on 1/2-inch bar stock of the Ti-3Mn-complex alloy. The elongations of the forgings were slightly lower than those of bar stock rolled in the beta field. A microexamination of the gear-blank test bars show them to be finish forged very close to the beta transus. There was semicontinuous alpha at the grain boundaries and Widmanstätten alpha within the grains, which is the structure found in the bar stock rolled in the beta field.

The remaining two forgings are awaiting further evaluation, possibly in a bursting or fatigue test.

Tube-Drawing Tests at Superior Tube Company

An important weight saving could be effected in airplanes if high-strength titanium tubing could be used for the hydraulic system. Superior Tube Company agreed to do some experimental cold drawing of the high-strength alloys developed on this contract. The results of drawing 7/16-inch-OD by 1/8-inch-wall tubing of four alloys have been reported in WADC Technical Report No. 52-334. The Ti-5Mn-2.5Cr alloy was drawn to 0.295-inch-OD by 0.072-inch-wall tubing, a total reduction of area of about 50 per cent, whereas the Ti-3Mn-complex alloy was reduced only about 30 per cent. The reductions per draw varied from 10 to 17 per cent.

These alloys were annealed between each draw by furnace cooling from 1350 to 1150°F, followed by water quenching.

Further drawing has been done on 3/4-inch-OD by 1/4-inch-wall tubing of the Ti-3Mn-complex and the Ti-3.5Mn-3.5Cr alloys. These specimens were annealed by solution treating at 1300°F and aging for 8 hours at 1100°F. This treatment has produced the greatest ductility in tensile testing. The Ti-3Mn-complex alloy was reduced about 50 per cent total reduction of area in steps of 10 to 17 per cent reduction per pass. The tube split during the roll-off operation after the last pass. The Ti-3.5Mn-3.5Cr alloy was also drawn a total of 50 per cent reduction of area without cracking.

Weldability Evaluation by A. O. Smith Corporation

A limited evaluation of the flash-welding characteristics of the Ti-3Mn-complex alloy was made by A. O. Smith Corporation of Milwaukee, Wisconsin. The material for these tests was melted and fabricated into 3/4-inch-diameter bar stock at Battelle. The material was supplied to A. O. Smith Corporation in the as-rolled condition.

Flash-welded specimens 5/8 inch in diameter were machined from the hot-rolled bars. The weld procedure using two upset pressures is the same as that employed to weld Ti-150A. A pressure of 13,300 psi was used in conjunction with a long burn-off time of 12 seconds to make Welds 1, 2, 3, 4, and 5 (Table 33). A pressure of 19,000 psi and a shorter time cycle of 6.5 seconds were used to make Welds 6, 7, 8, 9, and 10. All welds were stress relieved by heating to 1000°F for 1 hour immediately after welding. All welds were free upset; i.e., they were produced without the machine's coming to rest on fixed stops.

All of the welds and two bars of the base metal were solution treated for 1 hour at 1400°F, water quenched, and aged 48 hours at 800°F to produce a high strength level. Welds 3, 4, 5, 6, 7, and 8 and one bar of the base metal, B-1, were machined to standard 0.357-inch-diameter tensile-test bars. The remainder of the test bars were machined to 1/4-inch by 0.470-inch bend bars. Failure of the tensile test bars in the threads was encountered for Specimens 3 and B-1. The threads were ground and probably overheated, causing the failures. By remachining the gage section to 0.225-inch diameter, the thread failure was overcome in Test Bar 7. This specimen had a low ductility and a very high strength. Hardnesses of 474-488 VHN were reported for these specimens, which is appreciably above the maximum recommended for this alloy. Because of

TABLE 33. CONDITIONS OF WELDING AND 1Cr-1Fe-1Mo-1V ALLOY WELDED

Weld No.	Original Length (Both Pieces), inches	Welded Length, inches	Burn-Off Length, inch	Upset Travel, inch	Upset Pressure, psi	Cycle Time, sec	Volts
1	4.752	4.254	0.300	0.198	13,300	13	4.6
2	4.753	4.281	0.300	0.172	13,300	11	4.6
3	3.501	3.043	0.300	0.158	13,300	12.5	4.6
4	3.500	3.064	0.300	0.136	13,300	12	4.6
5	3.500	3.043	0.300	0.157	13,300	10	4.6
6	3.500	3.031	0.300	0.169	19,000	6.5	4.6
7	3.505	3.034	0.300	0.171	19,000	6.5	4.6
8	3.501	3.035	0.300	0.166	19,000	6.5	4.6
9	4.754	4.251	0.300	0.203	19,000	6.5	4.6
10	4.753	4.243	0.300	0.210	19,000	6.5	4.6
B-1	-	-	-	-	-	-	-
B-2	-	-	-	-	-	-	-

(1) Heat Treatment 1 consisted of solution treating 1 hour at 1400°F, water quenching, and aging 48 hours at 800°F.

Heat Treatment 2 consisted of re-aging for 2 hours at 1000°F after Heat Treatment 1.

(2) Broke in threads.

PHYSICAL-TEST RESULTS FOR THE Ti-3Mn-
BY A. O. SMITH CORPORATION

Heat Treatment[1]	Elongation, % in 1 inch	Reduction of Area, %	Tensile Gage Diameter, inch	Yield Strength, psi	Ultimate Tensile Strength, psi	Bend Angle, Over 3T Radius, degrees
1	-	-	-	-	-	1
1	-	-	-	-	-	2
1	-	-	0.357	(2)	(2)	-
1	-	-	0.252	(2)	213,500	-
1 and 2	7.2	28.5	0.252	177,000	178,500	-
1	-	-	-	(2)	(2)	-
1	1.6	-	0.225	-	219,000	-
1 and 2	5.5	17.9	0.252	178,000	178,000	-
1	-	-	-	-	-	5
1	-	-	-	-	-	0.5
1	-	-	0.357	(2)	(2)	-
1	-	-	-	-	-	1

this, Welds 5 and 8 were re-aged for 2 hours at 1000°F to lower the strength and to improve the ductility.

The bend bars were tested in the high-strength condition around a bend die with a radius equal to 3T. The results of these tests and the conditions for welding are also presented in Table 33. Although the data are limited, the ductilities of the double-aged tensile bars were low for the strength level obtained. Test Bars 5 and 8 broke in the base metal about 1/4 inch from the weld line. Part of the low elongation may be attributed to a high-strength weld which would not elongate, thereby reducing the effective gage length by the length of the weld and heat-affected zones.

These tests were too limited for any conclusions to be based upon their results. However, the ductility exhibited by the two tensile specimens heat treated to the 180,000-psi strength level is encouraging.

Notched-Tensile and Impact Testing at Wright Air Development Center

Material for these tests was made and heat treated at Battelle and sent to WADC for testing. Two heats of the Ti-3Mn-complex alloy were used, one of which was rolled to 1/2-inch-square bars for impact specimens and the other to 1/2-inch-diameter bars for tensile testing. The bars for each test were heat treated to three strength levels, as follows:

141,000 psi - 1 hour at 1300°F, cold water quenched,
8 hours at 1100°F, air cooled

191,000 psi - 1 hour at 1400°F, cold water quenched,
8 hours at 900°F, air cooled

212,000 psi - 1 hour at 1400°F, cold water quenched,
48 hours at 800°F, air cooled.

The tensile tests were made on standard 0.250-inch-gage-diameter bars. The notched specimens were machined the same initially, then a 60-degree V-notch with a 0.010-inch root radius was cut at the center of the gage length. This notch was cut deep enough that the area under the notch was equal to one-half the original gage-section area. The tensile tests were made at room temperature in a 20,000-pound Tinius Olsen machine, using a platen speed of 0.005 inch per minute.

The 1/2-inch-square bars were machined to standard Charpy specimens having a 45-degree V-notch cut 0.079 inch deep with a 0.010-inch root radius. The impact tests were made at temperatures from -100 to 375°F. The specimens tested below room temperature were cooled in a dry-ice - alcohol mixture for 15 minutes. The specimens tested at elevated temperatures were heated in silicone oil for 1/2 hour. Transfer from bath to testing machine was accomplished in about 5 seconds. The impact tests were made by using a 264-foot-pound Tinius Olsen machine. Results of these tests are shown in Figure 40. The value of each point on the curves is the average of two or three tests.

The impact value for the 141,000-psi level remained constant at about 23 foot-pounds over a testing-temperature range of -100 to 150°F. Above the latter temperature, the energy absorption increased sharply. At the 191,000- and 212,000-psi strength levels, the alloy had a lower impact value of about 9 foot-pounds over the range -100 to 200°F. Above 200°F, the impact values increased moderately and were inversely proportional to the strength.

Results of room-temperature notched-bar tensile tests for three conditions of heat treatment are reported in Table 34. The notched-to-unnotched tensile ratio decreased from 1.61 for the 141,000-psi strength to 1.33 for the 212,000-psi strength. These notched-tensile ratios approximate those for steel heat treated to about the same strengths.

Evaluation of Bolt Material by Douglas Aircraft Company

A 20-pound ingot of the Ti-3Mn-complex alloy, Heat WU33A, was made for Douglas Aircraft Company. This material was forged and rolled to 9/16-inch-diameter bars at Battelle. Most of the material was solution treated at 1300°F and aged 8 hours at 1100°F. This material is to be made into bolts and tested. About 3 pounds of the stock was heat treated to various higher strength levels, as shown in Table 35, for other tests. It will be noted that the properties are about the same as previously obtained for the Ti-3Mn-complex alloy for similar heat treatments. To date, the results of these tests are not known.

North American Aviation, Incorporated

About 15 pounds of 3/4-inch-diameter bar stock of the Ti-3Mn-complex alloy were sent to North American Aviation, Incorporated, Downey, California. This was sent as hot rolled, and will be heat treated and tested in their plant. The heat treatments will be those developed on this project.

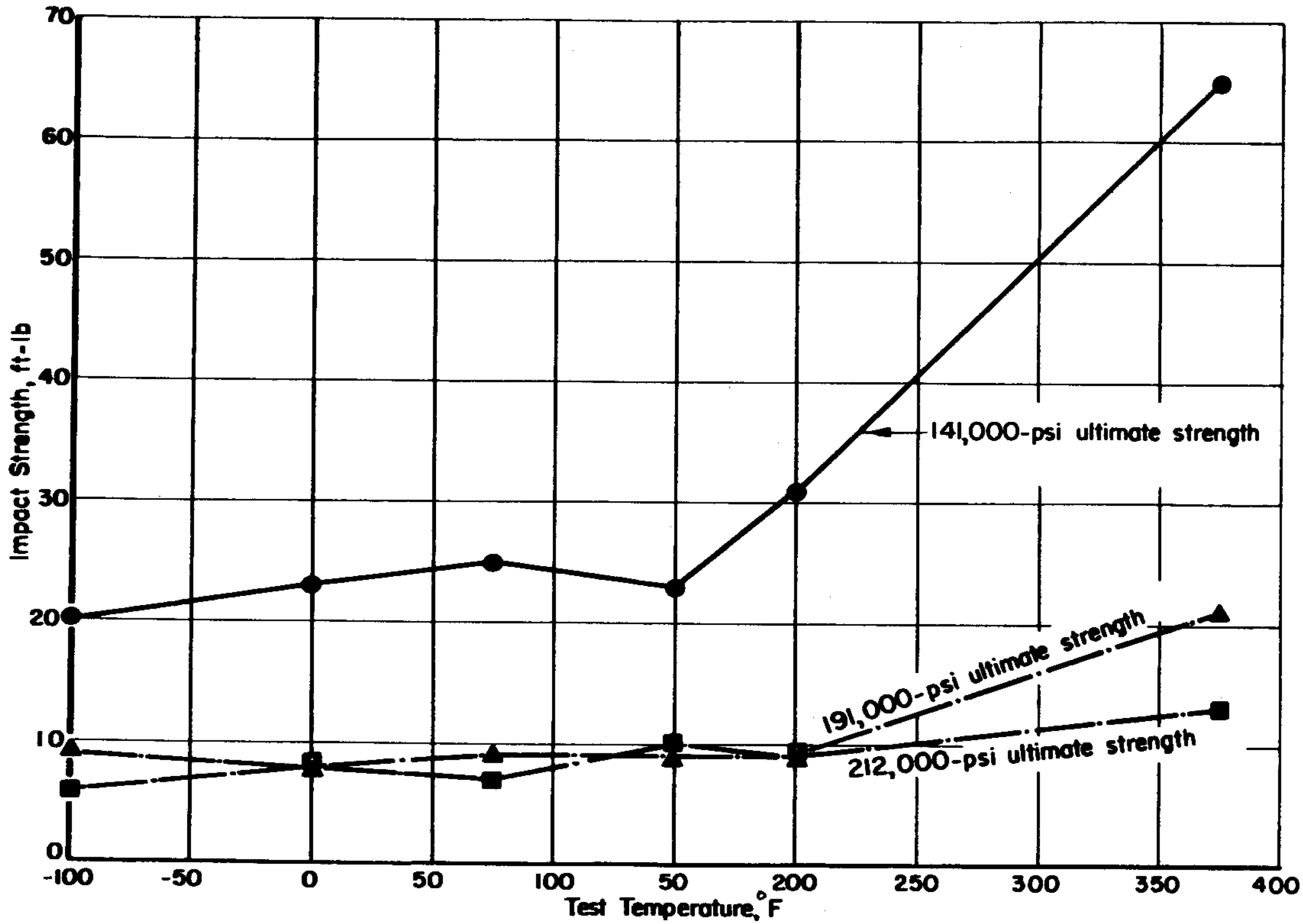


FIGURE 40. EFFECT OF TEMPERATURE ON CHARPY NOTCHED-BAR IMPACT VALUES OF THE Ti-3Mn-COMPLEX ALLOY HEAT TREATED TO THREE DIFFERENT STRENGTH LEVELS

A-8547

TABLE 34. ROOM-TEMPERATURE NOTCHED-BAR TENSILE PROPERTIES AND CHARPY IMPACT PROPERTIES OF THE Ti-3Mn-COMPLEX ALLOY

Heat Treatment	1 hr at 1300°F, CWQ[1], 8 hr at 1100°F, AC	1 hr at 1400°F, CWQ, 8 hr at 900°F, AC	1 hr at 1400°F, CWQ, 48 hr at 800°F, AC
Ultimate Tensile Strength, psi	141,000	190,600	212,500
Yield Strength, 0.2% Offset, psi	134,700	177,800	199,200
Elongation, % in 1 inch	25	12	10
Reduction of Area, %	52	27	32
Notched-Bar Tensile Strength, psi	228,200	275,500	284,800
Notched Tensile / Unnotched Tensile	1.61	1.44	1.33
Measured Tensile-Notch Radius, inch	0.0097	0.010	0.010
Hardness, Rockwell, C	32-33	42-43	46-47
Charpy Impact Values, ft-lb	25	9	7

(1) Symbols: CWQ = Cold water quenched.
AC = Air cooled.

TABLE 35. TENSILE PROPERTIES AND HARDNESSES OF HEAT-TREATED 9/16-INCH-ROUND BARS OF THE Ti-3Mn-COMPLEX ALLOY (WU33A) SENT TO DOUGLAS AIRCRAFT COMPANY

Solution Temp., °F(1)	Aging Time, hr	Aging Temp., °F	Yield Strength(2), 0.2% Offset, psi	Ultimate Tensile Strength(2), psi	Elongation(2), % in 1 inch	Reduction of Area(2), %	VHN (10-Kg Load)(3)
1300	8	1100	-	135,000	29.5	57.5	282
"	"	"	127,000	136,000	30.0	57.5	302
1400	-	-	-	180,500	14.0	28.5	336
"	-	-	160,500	181,000	15.0	36.5	383
"	24	800	-	218,000	6.0	11.0	397
"	"	"	206,000	217,000	2.0	3.0	409
"	8	900	-	189,000	13.5	30.5	349
"	"	"	178,000	187,000	14.0	37.0	345
"	8	1000	-	162,000	21.0	52.0	333
"	"	"	152,000	160,500	20.5	48.5	339

(1) Cold water quenched after 1 hour at solution temperature.
(2) Single values.
(3) Average of three impressions.

Republic Steel Corporation

It was reported in the WADC Technical Report No. 52-334 that about 100 pounds of the Ti-3Mn-complex alloy would be sent to Republic Steel Corporation for rolling to sheet. This sheet was then to be distributed to industrial organizations for evaluation. Because of the relatively poor properties obtained on most sheet alloys to date, this material has been diverted to bar stock. Republic Steel has rolled two 50-pound ingots melted at Battelle to about 7/8-inch-diameter bar stock. They are retaining about 50 pounds for evaluation of bolts. Bolts are to be hot headed and machined from this stock. Then the bolts will be heat treated and tested. The balance of the stock rolled by Republic Steel will be sent to other industrial organizations for evaluation.

Westinghouse Electric Company

The Aviation Gas Turbine Division of Westinghouse Electric Company has agreed to evaluate some of the selected alloys for use in jet engines. Bars of the Ti-3.5Cr-3.5V, Ti-5Mn-2.5Cr, and Ti-3Mn-complex alloys, 1/2-inch-diameter by 22 inches long, have been sent to them for evaluation.

Data on which this report is based are recorded in BMI Laboratory Record Books No. 7585, pages 10 through 100, and No. 8135, pages 3 through 34.

APPENDIX A

MELTING AND FABRICATION OF SELECTED ALLOYS

Melting

The melting procedures were the same as those described in WADC Technical Report 52-249*. No modifications were made in melting technique or alloy additions for alloys listed in Table A-1. All ingots were melted twice in argon using tungsten electrode tips. With one exception, titanium master alloys prepared at Battelle or elements in the pure form (molybdenum, vanadium) were used for alloy additions. One heat of the Ti-3Mn-complex alloy was melted using commercially available master alloys except for the vanadium. The compositions of the alloy additions used for this heat were:

Alloy	Nominal Composition, per cent
Mn-Ti Special	27-31Ti, 8-10Al, 2-4Si, 0.03-0.10C, bal. Mn
Low-carbon ferrochromium	71.1Cr, 0.02O, 0.41Si, 0.17Mn, bal. Fe
Ferromolybdenum	62.8Mo, 0.05C, 0.45Si, bal. Fe
Ductile vanadium	99.7V

Additions of these materials were made to give a calculated composition as follows:

Alloys:	3.0Mn, 1.0Cr, 0.93Fe, 1.0Mo, 1.0V
Residuals:	0.007C, 0.3Si, 0.52Al

In addition to the material just described, the following ingots of the Ti-3Mn-complex alloy were melted for use in the industrial evaluation program:

(1) Four 20-pound ingots and two 50-pound ingots melted by Battelle.

(2) One 225-pound ingot melted by the Bureau of Mines, Albany, Oregon.

*Summary Report on the "Development of Titanium-Base Alloys", dated June 18, 1952, Contract No. AF 33(038)-3736.

TABLE A-1. TITANIUM-ALLOY INGOTS COMPLETED

Number of Ingots	Weight Per Ingot, lb	Nominal Composition, %						Purpose
		Mn	Cr	Fe	Mo	V	Other	
3	20	3	1	1	1	1	-	Alloy evaluation
2	20	1	1	4	1	1	-	Ditto
2	20	-	-	3.5	4.5	-	-	"
1	20	5	2.5	-	-	-	-	"
1	10	-	3.5	-	-	3.5	-	Evaluation of 99.7 per cent vanadium
1	5	3	1	1	1	1	-	Effect of pickling between melts
1	5	3	1	1	1	1	-	Commercial-alloy additions
1	5	5	2.5	-	-	-	1Th	Effect of thorium on beta embrittlement
1	5	5	2.5	-	-	-	2Th	
1	5	5	2.5	-	-	-	1Al	Effect of aluminum on beta embrittlement
1	5	5	2.5	-	-	-	2Al	
2	2	-	3.5	-	-	3.5	-	Weldability
2	2	3.5	3.5	-	-	-	-	"
2	2	5	2.5	-	-	-	-	"
2	2	5	-	-	2	-	-	"
2	2	-	-	3.5	4.5	-	-	"
2	2	3	1	1	1	1	-	"
1	2	1	1	4	1	1	-	"
1	2	5	2.5	-	-	-	0.5Al	Effect of aluminum on beta embrittlement
1	2	5	2.5	-	-	-	1.0Al	
1	2	5	2.5	-	-	-	2.0Al	

TABLE A-1. (Continued)

Number of Ingots	Weight Per Ingot, lb	Nominal Composition, %						Purpose
		Mn	Cr	Fe	Mo	V	Other	
1	2	-	-	-	1	1	-	Alloy evaluation
1	2	-	-	-	1	3.5	-	Ditto
1	2	-	-	-	1	5	-	"
1	2	-	-	-	3.5	1	-	"
1	2	-	-	-	3.5	3.5	-	"
1	2	-	-	-	3.5	5	-	"
1	2	-	-	-	5	1	-	"
1	2	-	-	-	5	3.5	-	"
1	2	-	-	-	5	5	-	"
2	2	-	15	-	-	-	-	Dilatometer studies

The 20- and 50-pound ingots were double melted in inert tungsten-electrode arc furnaces. The 225-pound ingot was double melted in a consumable-electrode furnace normally used for zirconium.

A modification in processing for remelting was made in the material for the three 50-pound Ti-3Mn-complex alloy ingots for industrial evaluation. The rolled single-melted stock was steel grit blasted to remove oxides before remelting, instead of being pickled. This change was made after pickling between melts was shown to be detrimental to the ductility of sheet specimens.

Fabrication

All ingots used for the research and development program were forged at 1750°F and the forged products were rolled at 1450°F in the alpha-beta-phase region as follows:

Forged Bar	Rolled to
1-1/4-inch-square bar	7/8-inch round
3/4-inch-square bar	1/2-inch round
1-inch-thick slab	0.064-inch sheet

A few forgings were rolled at 1600 or 1700°F to determine the effect of fabrication in the beta-phase region on the resulting mechanical properties. In all cases, the bars or slabs were reheated to the rolling temperature between each pass through the mill. The 20-pound ingots used for industrial evaluation were forged and rolled at Battelle to 1/2-inch- and 7/8-inch-diameter rounds. The 50-pound ingots and the 225-pound ingot were forged and rolled to 7/8-inch-diameter bar stock at Republic Steel Corporation and at Mallory-Sharon Titanium, Incorporated, respectively.

APPENDIX B

PROPERTIES OF THE Ti-3Mn-COMPLEX ALLOY MADE WITH COMMERCIAL-ALLOY ADDITIONS

The titanium alloys previously tested were made with additions of master alloys as reported in WADC Technical Report No. 52-249. The master alloys are expensive and time consuming to prepare. For this reason, an alloy made with commercially available alloy additions would be more desirable for production melting.

A heat of the Ti-3Mn-complex alloy was made using materials which are commercially available, except for the vanadium addition. The compositions of the alloy additions used were described in Appendix A.

The alloy was forged at 1750°F and rolled at 1450°F to 1/2-inch-diameter bar stock. The tensile and hardness data for this alloy after various heat treatments are shown in Table B-1. A comparison of the

TABLE B-1. TENSILE PROPERTIES AND HARDNESSES OF THE Ti-3Mn-1Cr-1Fe-1Mo-1V ALLOY MADE WITH COMMERCIAL-ALLOY ADDITIONS(1)

Solution Temp. °F(2)	Aging Time, hr	Aging Temp, °F	Elongation, % in 1 inch(3)	Reduction of Area, %(3)	Yield Strength, 0.2% Offset, psi(3)	Ultimate Tensile Strength, psi(3)	VHN (10-Kg Load)(4)
1300	8	900	13.5	16.0	-	176,000	-
"	"	"	19.0	50.0	174,000	179,000	394
1300	8	1100	28.0	51.5	-	148,500	-
"	"	"	28.0	51.0	146,500	150,500	342
1400	24	800	1.5	3.0	-	221,000	-
"	"	"	5.0	5.5	-	225,000	442
1400	8	1000	22.0	52.0	-	171,000	-
"	"	"	23.0	43.5	169,000	171,000	370

(1) Heat No. WU79A.
(2) Cold water quenched after 1 hour at temperature.
(3) Single values.
(4) Average of three impressions.

strength-ductility relationship for this alloy and that for the same composition made with master-alloy additions is shown in Figure B-1. At low and intermediate strengths, the commercial-alloy additions gave slightly higher elongations. This heat did not soften as much during aging at the higher temperatures as did the heat made from master-alloy material. Both heats had about the same properties at the 220,000-psi strength level. However, the improvement in ductility for the commercial alloy may be lost if commercial vanadium (93 per cent pure) were used, as the balance of this material is principally oxygen.

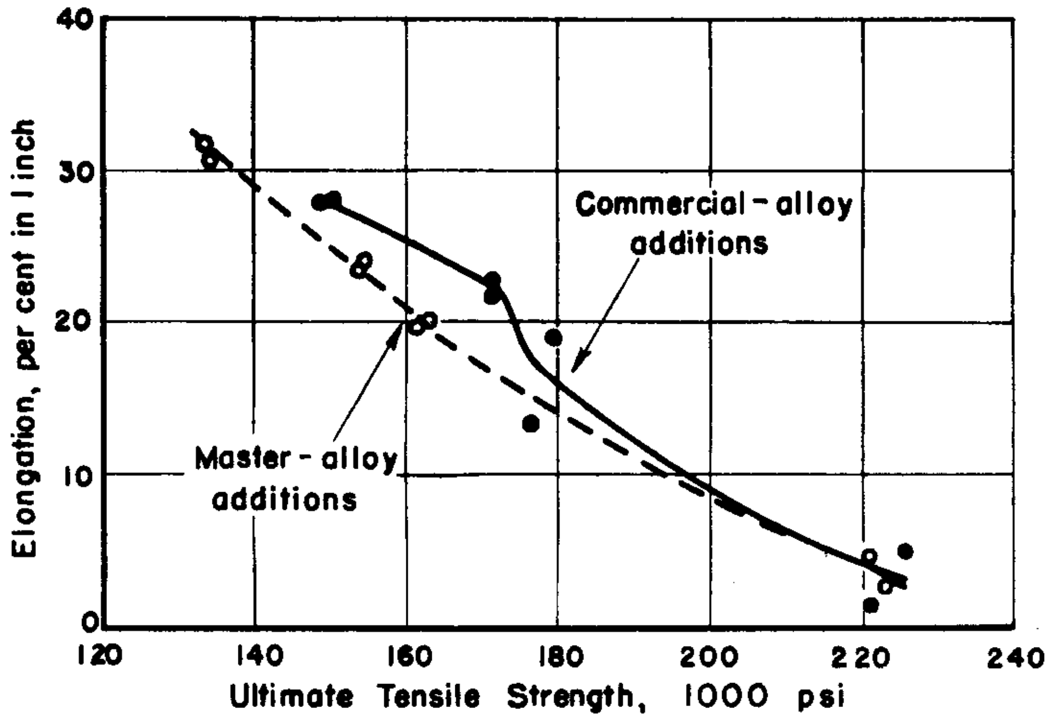


FIGURE B-1. EFFECT ON TENSILE PROPERTIES OF USING COMMERCIAL-ALLOY ADDITIONS IN MELTING THE Ti-3Mn-COMPLEX ALLOY

A-7433

EFFECT OF VANADIUM PURITY ON PROPERTIES OF THE Ti-3.5Cr-3.5V ALLOY

The Ti-3.5Cr-3.5V alloy was shown to have somewhat lower ductility than the other selected alloys. It was thought that the low elongation values of this alloy may have been due to the oxygen in the commercial-grade vanadium used for the alloy addition. The vanadium used was about 93 per cent pure, with almost all of the impurity being oxygen. A second heat with the same nominal composition was made using 99.7 per cent pure vanadium. Half of the ingot was rolled at 1450°F to 1/2-inch-diameter bar stock, and the balance was rolled to 14-gage sheet. Specimens of the 1/2-inch-diameter bar stock and 14-gage sheet were solution treated and aged at various temperatures. The results of tensile tests and the hardnesses of the specimens are given in Table C-1, together with the results of the commercial vanadium alloy for the same heat treatments. The ductile vanadium alloy showed a slightly higher ductility for the bar stock, while the ductility of the sheet material was about the same. A metallographic examination of the ductile vanadium alloy indicated, however, that the rolling temperature of 1450°F was in the beta field for this heat. The structure consisted of semicontinuous grain-boundary and Widmanstätten alpha in a beta matrix. The structure in the commercial vanadium alloy resulting from a beta rolling (1600°F) had essentially zero ductility as solution treated and aged. These results show a definite improvement in ductility for the high-purity vanadium alloy as rolled in the beta field. If rolled in the alpha-beta temperature range, it probably would have better ductility than the alloy made with commercial vanadium. These results also indicate that oxygen may be a factor in the low ductility of beta-embrittled alloys.

Both alloys showed an effect of section size on the quenched properties. It may be noted that the strength and hardness of the 1300°F-quenched sheet material were lower than for the bar stock similarly treated. The lower cooling rate in the bar stock probably allowed some aging during the quench. The aging that occurred in the bars after a 1450°F solution treatment produced very low ductility. Because of this brittle condition, these specimens probably failed prematurely.

TABLE C-1. EFFECT OF PURITY OF VANADIUM OF HEAT-TREATED 1/2-INCH-
Ti-3.5Cr-3.5V ALLOY ROLLED

Heat No.	Solution Temp, °F(1)	Aging Time, hr	Aging Temp, °F	1/2-Inch- Elongation, % in 1 inch(2)	Reduction of Area, %(2)
					93 Per Cent
WT125A	1300	-	-	9.5	26.5
	"	8	900	11.5	18.0
	"	"	1100	29.5	56.0
	1400	-	-	-	-
	"	24	800	9.5	20.5
	"	2	1000	13.0	13.5
					99.7 Per
WU76A	1300	-	-	10.0	24.0
	"	8	900	19.0	40.5
	"	"	1100	27.0	60.5
	1400	-	-	-	-
	"	24	800	12.0	25.5
	"	4	1000	20.5	57.0

(1) Cold water quenched after 1 hour at temperature.
(2) Average of two values, except where noted.
(3) Single values.
(4) Average of three impressions.
(5) Aged 4 hours.

ON THE TENSILE PROPERTIES AND HARDNESSES
ROUND BARS AND 14-GAGE SHEET OF THE
AT 1450°F

Round Bars			14-Gage Sheet			
Yield Strength, 0.2% Offset, psi(3)	Ultimate Tensile Strength, psi(2)	VHN (10-Kg Load)(4)	Heat No.	Elongation, % in 1 inch(2)	Ultimate Tensile Strength, psi(2)	VHN (10-Kg Load)(4)
Vanadium						
172,000	175,500	409	WT141A	9.0	158,000	362
--	155,000	379		8.5	149,000	334
119,500	126,500	278				
--	175,000(3)	505		1.0(3)	185,000(3)	432
--	192,000	392		1.5	179,500	380
143,500	149,000	325		6.5(5)	145,500(5)	330
Cent Vanadium						
164,000	171,000	400	WU76A	6.5	154,000	368
127,500	148,000	333		9.5	142,000	322
109,500	118,000	287		13.5	117,000	274
--	160,000(3)	478		0.5	158,000	446
178,000	184,000	376		0.0	179,500	378
140,000	150,000	325		9.0	140,000	317

APPENDIX D

EFFECT OF PICKLING BETWEEN MELTS ON ALLOY SHEET

The effect of hydrogen on the ductility of sheet alloys was demonstrated in a preceding section. It was also discovered that the hydrogen content of alloys made at Battelle was much higher than that of commercial titanium alloys. The alloys made under this contract have been double melted to get good homogeneity. Between melts, the material has been pickled in sulfuric acid-ammonium bifluoride solution to remove the scale and the contaminated surface layers. This was done to minimize contamination of the alloys by oxygen or nitrogen. However, it will be shown in this section that hydrogen was absorbed on the surface of the material during pickling. This would add hydrogen to the alloy during the second melting operation.

To determine the possible effects of pickling between melts, a 6-pound single-melted ingot of the Ti-3Mn-complex alloy was forged and rolled to sheet. The sheet was steel-grit blasted to remove the oxide. One-half of the sheet was pickled in the usual way. After shearing for remelting, the grit-blasted and pickled materials were melted into separate 3-pound ingots. The ingots were forged and rolled to 14-gage sheet. Duplicate specimens from each ingot were heat treated for tensile testing. One specimen from each heat treatment was etched to remove 0.004 inch of the thickness, whereas the heat-treat scale was left on the other specimen. The tensile and hardness data for these specimens are presented in Table D-1.

The most striking result was the beneficial effect of etching away the contaminated surface of the tensile coupons. The tensile specimens from the grit-blasted ingot which were etched to remove scale had higher elongations than comparable specimens from the pickled ingot for given heat treatments. The strength was lower for the grit-blasted material, presumably because less hydrogen was present, but the loss in strength was proportionately lower than the increase in elongation. The relationship of tensile strength to elongation for the specimens that were etched after heat treatment has been plotted in Figure D-1. It may readily be seen that the grit-blasted material has higher elongations for any strength level than the pickled material. The hydrogen pickup from the pickling between melts definitely seems to lower the tensile elongation of sheet material. Therefore, grit blasting between melts has been adopted as a standard procedure.

TABLE D-1. EFFECT OF PICKLING BETWEEN MELTS ON PROPERTIES OF A DOUBLE-MELTED Ti-3Mn-COMPLEX ALLOY[1]

	Aging			Pickled			Grit Blasted		
Solution Temp, °F[2]	Time, hr	Temp, °F	Surface Condition When Tested	Elongation, % in 1 inch[5]	Ultimate Tensile Strength, psi[5]	VHN (10-Kg Load)[6]	Elongation, % in 1 inch[5]	Ultimate Tensile Strength, psi[5]	VHN (10-Kg Load)[6]
1300	-	-	Scaled[3]	10.5	155,500	322	11.0	147,500	319
"	-	-	Etched[4]	11.5	149,500	318	15.5	139,500	317
"	24	800	Scaled	1.0	170,500	369	1.0	151,500	342
"	"	"	Etched	1.0[7]	180,000	371	12.0	166,500	349
"	8	900	Scaled	2.5	164,000	363	0.0[7]	148,500	333
"	"	"	Etched	8.0	170,000	360	13.0	151,000	345
"	4	1000	Scaled	3.0	144,500	345	1.0	136,500	314
"	"	"	Etched	6.0	151,000	342	15.5	138,000	316
"	"	1100	Scaled	8.0	135,000	314	5.0	128,500	304
"	"	"	Etched	10.5	133,000	304	19.0	119,500	303
1400	-	-	Scaled	9.0	176,500	357	5.0	178,500	365
"	-	-	Etched	11.0	176,500	339	10.0	173,000	369
"	24	800	Scaled	0.0	143,000	363	0.0	115,000	383
"	"	"	Etched	0.5	206,000	373	4.0	184,500	373
"	8	900	Scaled	0.5	149,000	370	0.0	156,000	367
"	"	"	Etched	1.0	188,000	387	5.0	170,000	360
"	4	1000	Scaled	0.5	160,000	360	0.0	142,000	333
"	"	"	Etched	1.0	161,500	372	5.0	152,500	339
"	"	1100	Scaled	3.5	139,000	336	2.0	131,000	317
"	"	"	Etched	3.5	128,500	339	16.0	127,000	312

(1) Heat No. WU100A.
(2) Cold water quenched from solution temperature.
(3) Heat-treat scale.
(4) Tensile coupons etched to remove a minimum of 0.004 inch from the thickness in 30 per cent HCl, 10 per cent HF, and 2 per cent HNO_3 in water solution.
(5) Single values.
(6) Average of three impressions.
(7) Broke outside gage marks; uniform elongation recorded.

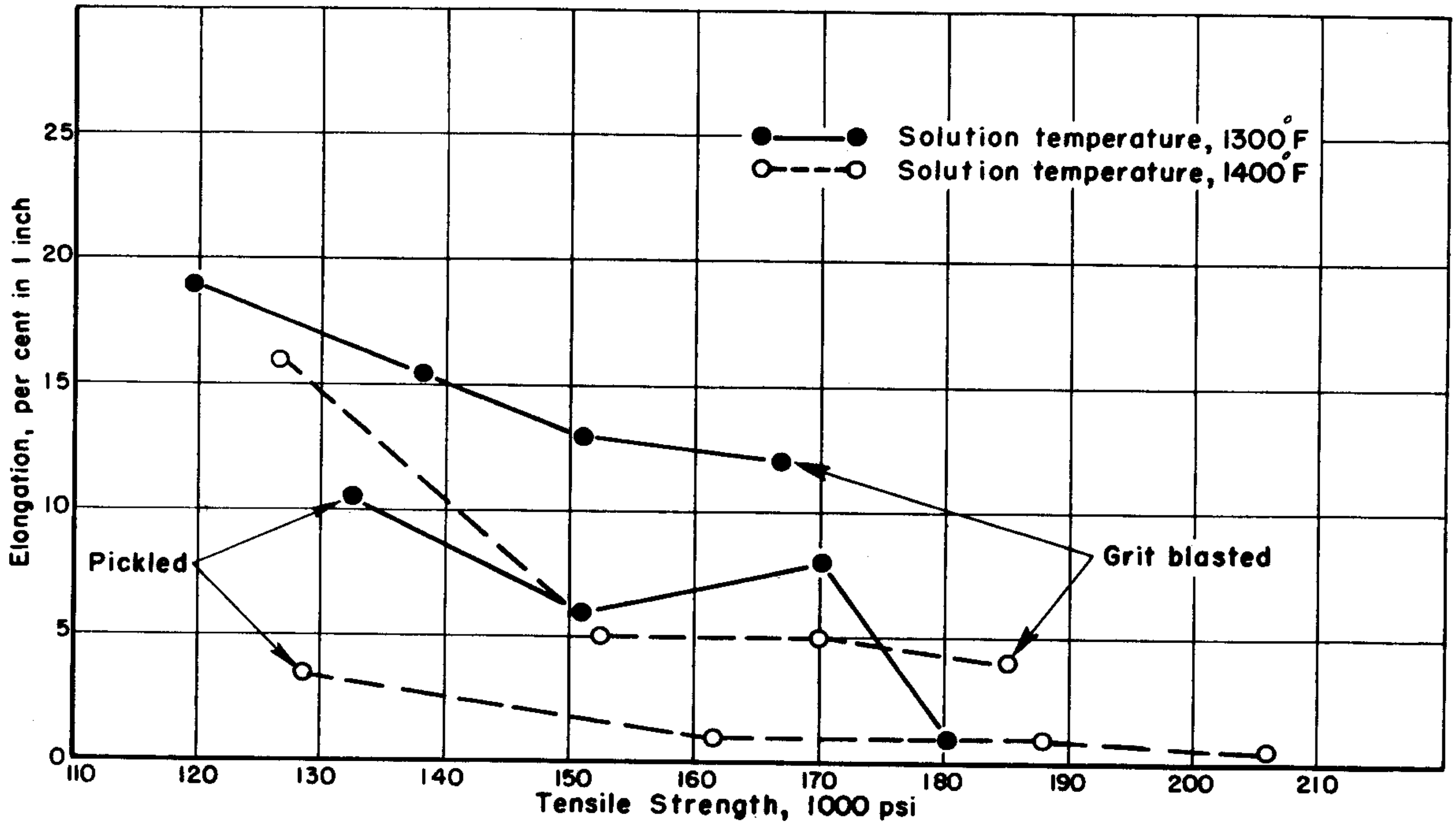


FIGURE D-1. EFFECT OF GRIT BLASTING VERSUS PICKLING BETWEEN MELTS ON THE TENSILE STRENGTH-ELONGATION RELATION FOR A Ti-3 Mn-COMPLEX SHEET ALLOY GIVEN VARIOUS SOLUTION AND OVERAGING HEAT TREATMENTS

A-6516