WADC TECHNICAL REPORT 52-249

DEVELOPMENT OF TITANIUM-BASE ALLOYS

Battelle Memorial Institute
Columbus 1, Ohio

June 18, 1952

Contract No. AF 33(038)-3736
RDO No. 615-11

WRIGHT AIR DEVELOPMENT CENTER
AIR RESEARCH AND DEVELOPMENT COMMAND
UNITED STATES AIR FORCE
WRIGHT-PATTERSON AIR FORCE BASE, OHIO

FOREWORD

This report, covering the period May 19, 1951, to May 18, 1952, was prepared at Battelle Memorial Institute under Wright-Patterson Air Force Base Contract AF 33(038)-3736. The project was administered by the Research Division of the Materials Laboratory, Air Development Center. Dr. H. K. Adenstedt and Lt. J. R. Pequignot of that office were the Project Engineers.

The research was conducted at Battelle by the persons whose names and activities on the project are shown below:

Alloy Development and Heat-Treating Studies
(W. M. Parris, L. L. Hirsch, P. D. Frost, and J. H. Jackson)

Arc-Melting of Alloys
(J. Varga, Jr., G. W. P. Rengstorff, and C. T. Greenidge)

X-Ray Diffraction and Electron Microscopy
(J. R. Doig, G. G. Cocks, D. A. Vaughan, and C. M. Schwartz)

Titanium Extrusion Studies
(G. H. Schippereit, A. M. Sabroff, P. D. Frost, and J. H. Jackson)

Welding Studies
(G. E. Faulkner, G. B. Grable, and C. B. Voldrich)

Project Advisor
C. H. Lorig

ABSTRACT

The experimental work consisted of melting and testing of exploratory alloys. Past work indicated that the addition of the beta stabilizing elements such as chromium, iron, manganese, molybdenum, and vanadium, offered the best possibility in the development of titanium alloys. This observation was verified when ternary and complex alloys containing these elements were tested and found to possess desirable properties.

A number of exploratory alloys were heat treated and tested to determine the general effects of different thermal cycles on tensile properties. The effects of quenching media and low temperature aging were also investigated. As a result of the studies on the binary type alloys such as Ti-Cr and Ti-Mn containing from 5 to 8% of the alloying element, the hypothesis was proposed that the beta phase is susceptible to a hardening phenomenon involving a submicroscopic precipitation of the alpha phase.

Recrystallization, isothermal-transformation studies, and welding experiments were conducted using binary Ti-Cr alloys. Ductile welds in the "as-welded" condition were obtained in these high strength alloys.

PUBLICATION REVIEW

This report has been reviewed and is approved.

FOR THE COMMANDING GENERAL:

M. E. SORTE
Colonel, USAF
Chief, Materials Laboratory
Directorate of Research

TABLE OF CONTENTS

TABLE OF CONTENTS
(Continued)

LIST OF FIGURES

LIST OF FIGURES
(Continued)

LIST OF FIGURES
(Continued)

LIST OF FIGURES
(Continued)

LIST OF FIGURES
(Continued)

LIST OF TABLES

LIST OF TABLES
(Continued)

LIST OF TABLES
(Continued)

WADC-TR-52-249

SUMMARY REPORT

Covering the Period

May 19, 1951, to May 18, 1952

on

DEVELOPMENT OF TITANIUM-BASE ALLOYS

Contract AF 33(038)-3736

to

WRIGHT-PATTERSON AIR FORCE BASE
OHIO

from

BATTELLE MEMORIAL INSTITUTE

June 18, 1952

SUMMARY

Selection and Heat Treatment of Promising Alloys

Experimental work conducted under this contract prior to May, 1951, had shown that the beta-stabilizing elements, chromium, iron, manganese, molybdenum, and vanadium, offered the most promise in the development of titanium alloys. The alpha-stabilizing, interstitial elements, carbon, oxygen, and nitrogen, were potent strengtheners, but decreased ductility disproportionately to their strengthening ability.

Based on these observations, most of the exploratory alloys melted and tested during the past contract year contained one or more of the five metallic elements mentioned above. Several ternary alloys showed much promise as potential high-strength alloys. The Ti-Mn-Cr, Ti-Mn-Mo, and Ti-Fe-Mo systems were of interest in that high-strength properties were obtained over a wide range of compositions. Several complex alloys containing small amounts of all five beta-stabilizing elements also had outstanding properties.

Although there were a number of other compositions which might be of equal interest, twelve alloys were selected for more detailed studies of the effects of heat treatments on tensile properties. Tensile and bend properties of hot-rolled sheet from exploratory heats of these alloys are summarized in Figure I. All of the alloys selected except the Ti-15%Cr alloy had tensile strengths of the order of 200,000 psi. The current heat-treatment program is expected to produce improvements in the ductility of these alloys without excessive loss of strength. The Ti-15%Cr alloy was selected because of its excellent bend ductility at the 150,000 psi strength level.

Several of these alloys have been produced as five- and ten-pound ingots and fabricated into sheet and bar stock for testing. During the next year, ingots weighing 20 or more pounds of the most interesting alloys of this group will be melted, fabricated, and evaluated.

The evaluation of the selected alloys was started shortly before the close of the past contract year. Only a limited amount of data is available at this time. The following seven compositions were melted as duplicate one-pound ingots to test the reproducibility of mechanical properties from heat to heat:

Ti-3.5%Cr-3.5%V
Ti-3.5%Cr-3%Mn
Ti-2.5%Cr-5%Mn
Ti-5%Mn-2%Fe
Ti-5%Mo-4%Fe
Ti-5%Cr-1.5%Fe
Ti-5%Mn-2%Mo

In the as-hot-rolled condition, the mechanical properties of the two heats of each alloy varied considerably. Other specimens from the same heats will be tested in various heat-treated conditions to determine whether such treatments will produce more consistent properties.

Four of the selected alloys, having the nominal compositions listed below, were melted as five-pound ingots, fabricated to sheet, and tested in the as-hot-rolled and annealed conditions.

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

In general, the properties of all four heats were within the high strength range in the as-hot-rolled condition, but considerable variations were found in specimens taken from different sections of the same heat. Chemical analyses of these specimens revealed relatively large differences in composition, in some cases. Annealing at 1300-1400 F followed by air cooling produced much more consistent properties within individual heats. Average properties of the four alloys after annealing at these temperatures are given in Figure II. Specimens annealed at 1300 F had excellent ductility at strength levels of 137,000 to 150,000 psi. Annealing at 1400 F resulted in

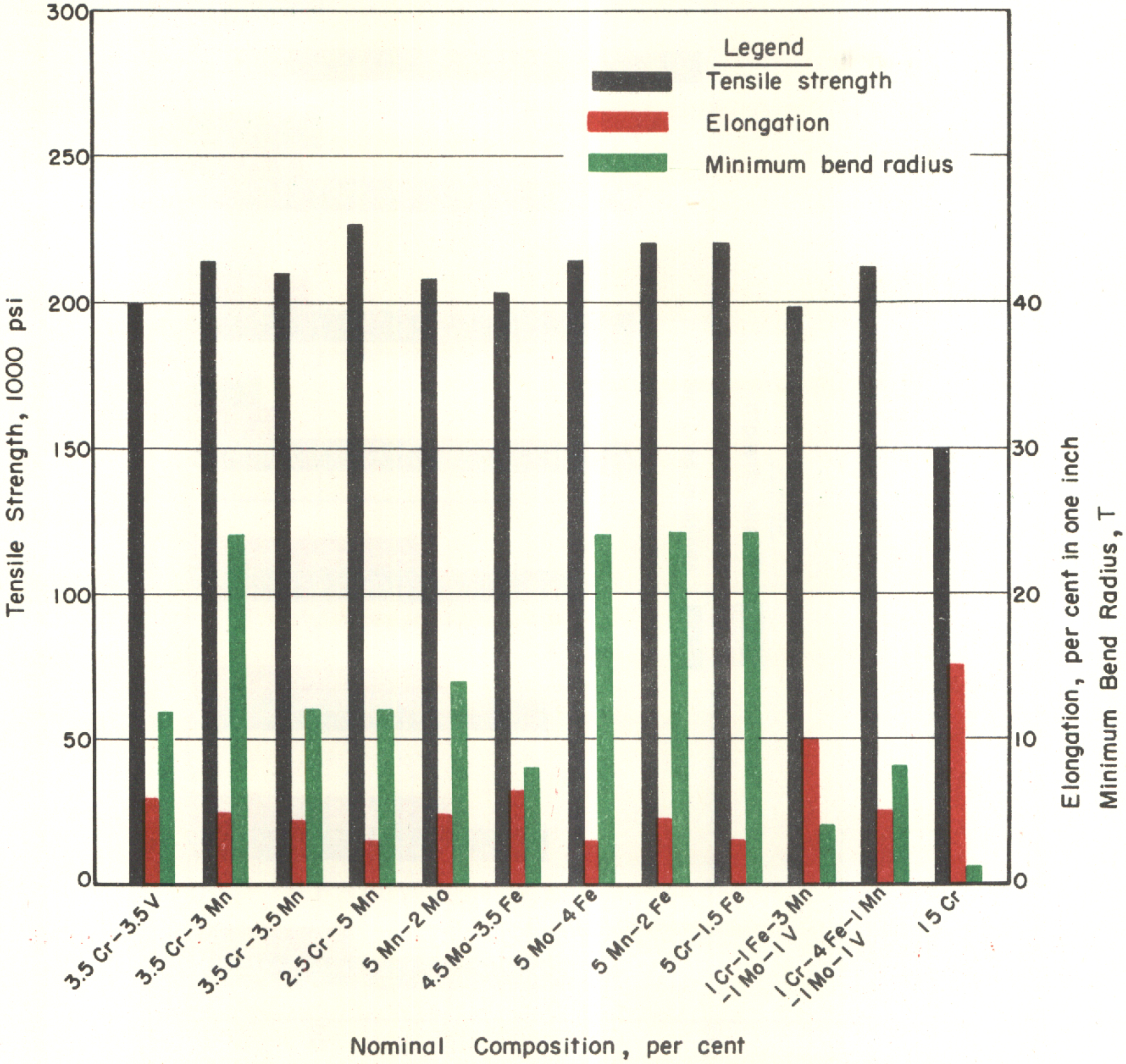


FIGURE I. TYPICAL PROPERTIES OF EXPLORATORY HEATS OF SELECTED ALLOYS IN THE AS-HOT-ROLLED CONDITION, TESTED AS 14-GAGE (0.064") SHEET

A-3245

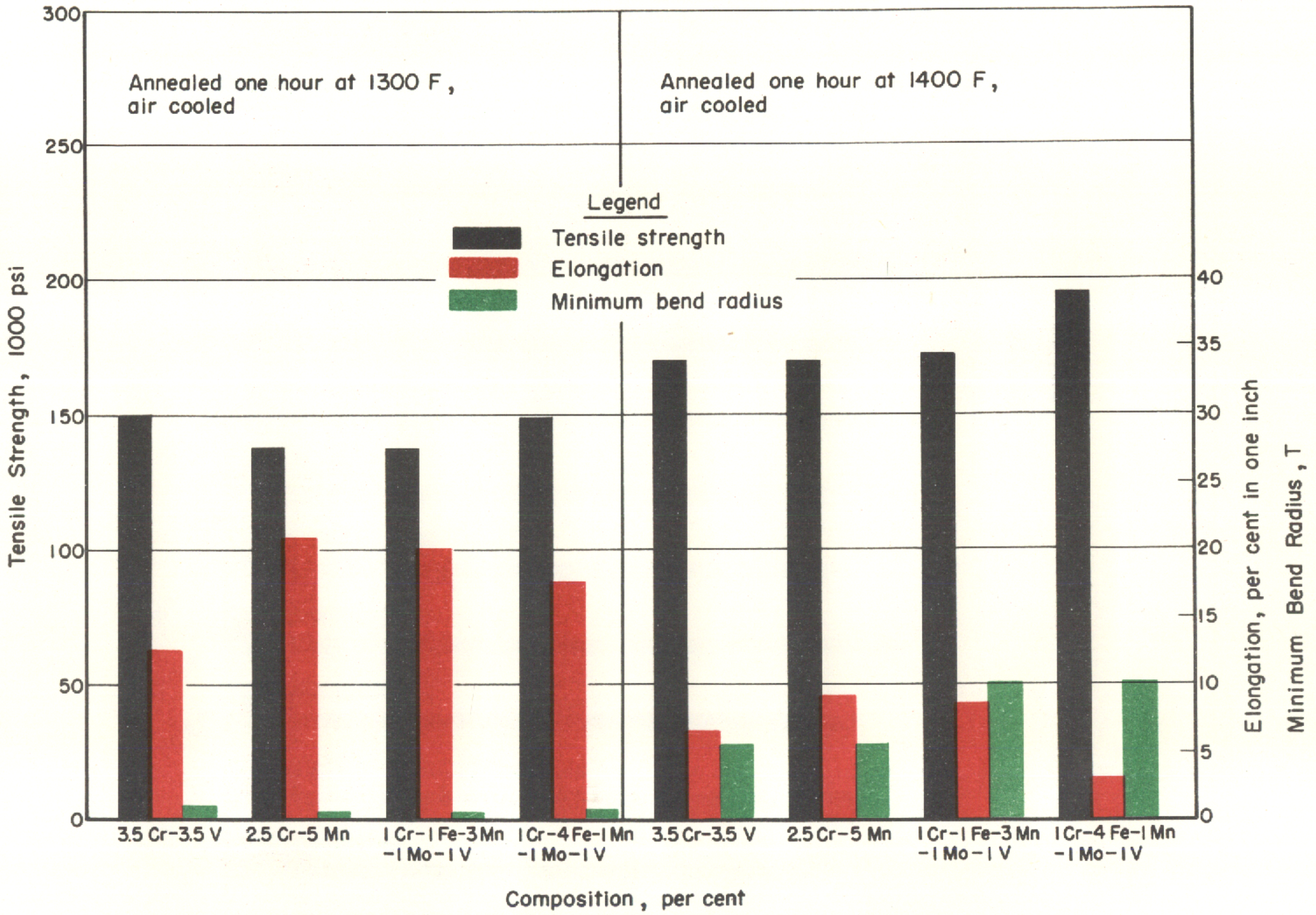


FIGURE II. PROPERTIES OF TITANIUM ALLOYS IN THE FORM OF 0.064-INCH SHEET AFTER ANNEALING AT 1300 AND 1400 F

A-3246

somewhat higher strengths and much lower ductility. A 1500 F anneal produced relatively poor properties in all of the alloys and erratic test results in two individual alloys.

Heat Treatment of Exploratory Alloys

While ingots of the selected alloys were being prepared, several groups of exploratory alloys for which sheet material was already available were heat treated and tested to determine the general effects of different thermal cycles on tensile properties.

Sheet specimens annealed at 1300 F and air cooled generally had excellent ductility but relatively low strength. Increasing the annealing temperature to 1400-1500 F resulted in higher strengths and lower ductility. In some cases, strengths comparable with those obtained in the as-hot-rolled condition were produced by the latter treatments. Solution treatments in the alpha-beta- or beta-phase fields (1300-1600 F) followed by quenching and subsequent aging in the temperature range 600-825 F produced high strengths in many alloys but, in general, ductility was very low. However, some of the alloys, as solution treated, had excellent intermediate strength properties. For example, the Ti-3.5%Cr-3.5%Mn alloy solution treated at 1450 F and water quenched had an ultimate tensile strength of 145,800 psi with an elongation of 24 per cent in one inch.

In conjunction with this exploratory heat-treatment program, the effects of quenching media and low-temperature (200-300 F) aging on the hardnesses of binary Ti-Cr and Ti-Mn alloys were investigated. The hardnesses, as quenched, of heats containing 5 to 8 per cent of alloying element increased sharply with decreasing cooling rate. It was discovered that the highest hardness values were obtained in what appeared to be single-phase beta structures. The hardness was a function of alloy content and quenching rate. As a result of these observations, the hypothesis was proposed that the beta phase, in this type of alloy, is susceptible to a hardening phenomenon involving a submicroscopic precipitation of the alpha phase. This "coherency" hardening may occur during cooling at critical rates from the alpha-beta- or beta-phase fields, as well as during artificial aging treatments.

In connection with the above work, it was found that aging occurred at relatively low temperatures in certain of the binary Ti-Cr and Ti-Mn alloys cooled at various rates from the beta field. Hardness increases up to 100 VHN resulted from aging treatments of 45 minutes at 212 F. Similar hardness increases were produced by the heating involved in mounting metallographic specimens in Bakelite. Alloys in the as-hot-rolled condition did not harden appreciably at these low temperatures.

Recrystallization Experiments

Beta-stabilized titanium alloys develop undesirably large grain sizes when heated in the beta-phase field either during fabrication or subsequent heat treatment. Grain refinement of such alloys by cold working and annealing is desirable from a physical property standpoint. Experiments with a series of binary Ti-Cr alloys indicated that those containing 5 per cent or less of chromium could be recrystallized into a fine-grained structure by suitable thermal-mechanical treatment. However, heats containing 10 to 30 per cent chromium could not be recrystallized by conventional cold rolling and annealing techniques.

Isothermal-Transformation Studies

A study of the isothermal transformation characteristics of three binary Ti-Cr alloys was carried out to provide basic information which has contributed to the proper understanding of the heat-treatment reactions.

Time-temperature-transformation curves were established for three binary Ti-Cr alloys containing 2.35, 4.64, and 7.54 per cent chromium, respectively. The initiation of proeutectoid alpha separation was very rapid in all of the alloys. This phase was detected at transformation times of the order of 10 to 30 seconds at all temperatures investigated. The compound $TiCr_2$ appeared in all of the alloys in about 1 hour at temperatures of 1022-1112 F (550-600 C). At 1292 F, initiation of this reaction was prolonged to very long times (more than 100 hours). M_s temperature ranges were established for the 2.35 and 4.64 per cent chromium alloys at 1004-1040 F (540-560 C) and 932-968 F (500 - 520 C), respectively. The 7.54 per cent Cr alloy did not transform to martensite. Hardnesses of isothermally transformed specimens, in general, decreased with increasing transformation temperature and with increasing time at a given temperature.

Welding Studies

One of the most promising alloys reported early in last year's work was a binary which contained 15 per cent chromium. In the first attempt to weld thin sheets of this alloy, it was found that a single-pass heliarc weld could be bent in the as-welded condition over a radius of 3T. This success led to the welding of several other intermediate and high-strength alloys, all of which could be bent either in the as-welded condition or after a heat treatment. The welding studies are being continued on some of the selected alloys mentioned above.

Extrusion Studies

The contract between Wright Field and Metal Trims, Inc., providing for extrusion studies of titanium was delayed and, therefore, extrusion ingots were not melted at Battelle until late in this contract year. The first shipment of extrusion billets, 4-7/16 inches in diameter by about 6 inches long, has been made.

INTRODUCTION

This report describes the work done under Contract No. AF 33(038)-3736 during the period May 19, 1951, to May 18, 1952.

The general objective of this project has been the development of high-strength structural alloys of titanium. At a recent meeting with the Sponsor, however, this objective was expanded to some degree and the required properties were more closely defined. Briefly, the new program concerns the development of the following types of alloys:

A. Forging Alloys

1. High-Strength Type

180,000 psi ultimate strength
160,000 psi yield strength
10 per cent elongation
Heat treatable from soft (20% elongation) condition

2. Weldable Type

130,000 psi ultimate strength
110,000 psi yield strength
20 per cent elongation

B. Sheet Alloys

1. High-Strength Type

200,000 psi ultimate strength
180,000 psi yield strength
5 per cent elongation
8T bend radius

2. Formable and Heat-Treatable Type

a. Requirements before heat treatment

20 per cent tensile elongation
Minimum bend radius - 2T or less

b. Requirements after heat treatment

180,000 psi ultimate strength
160,000 psi yield strength
10 per cent elongation

3. Weldable Type

130,000 psi ultimate strength
110,000 psi yield strength
20 per cent elongation
Minimum bend radius - 3T

In this report, the emphasis is on the development of the high-strength sheet alloy (Type B-1). Also described are some tests on a formable and heat-treatable sheet alloy (Type B-2). The data obtained on the sheet alloys will be of considerable value in the future development of forging alloys.

During the last several years, a great deal of general information was accumulated on a large number of titanium alloys. With this background, the work during the past year could be devoted to a more intensive study of fewer alloys. More attention was given to the investigation of the effects of annealing treatments, solution and aging treatments, and cooling rates on alloy properties. The isothermal-transformation (TTT) curves of three Ti-Cr binary alloys were established, and a similar study was begun for several Ti-Mn alloys.

The current research is a continuation and expansion of this approach. In addition, large ingots (20-35 pounds) of the most promising structural alloys are being prepared for fabrication and evaluation by commercial companies interested in titanium.

SELECTION AND EVALUATION OF EXPLORATORY ALLOYS

At the beginning of the present contract year, the status of this alloy development program was sufficiently advanced to permit a concentration of effort on a relatively few alloy systems. Nearly every element in the Periodic Table had been investigated in the broad scope of the exploratory screening program conducted during the previous three years(1, 2, 3)*. Although several other elements showed promise, it was concluded that the beta-stabilizing elements chromium, iron, manganese, vanadium, and molybdenum produced alloys having the best combinations of strength and ductility.

The interstitial, alpha-stabilizing elements, carbon, oxygen, and nitrogen, although potent strengtheners, reduce ductility disproportionately to their strengthening ability. Furthermore, these elements, which strengthen the alpha and not the beta phase, are not required for high strength. Thus, it is possible for alloys to derive their strength solely from the presence and condition of the beta phase.

The condition of the beta phase is of great importance. It was found early in the contract year that what appears, microscopically, to be single-phase beta is responsible for the extreme hardness developed in the titanium alloys. This hardness and strength depends not only on the composition of the beta phase, but also on the heat treatment applied to this phase. This will be described in detail in the sections which follow.

In short, then, the past year's work was based on the following background information:

1. The observation that the elements chromium, iron, manganese, vanadium, and molybdenum offer the greatest promise in alloy development.
2. The fact that adequate strength may be obtained in alloys which are predominantly beta.
3. The discovery that the beta phase, stabilized by a suitable heat treatment, can be aged to develop extremely high strength and hardness.

---

* References may be found at the end of the text.

Mechanical Properties of Exploratory Alloys in the As-Hot-Rolled Condition

As in the past, preliminary evaluation of exploratory alloys consisted of the determination of their tensile, hardness, and bend properties in the as-hot-rolled condition. For reasons to be discussed later in the report, the oxide surfaces of all sheet specimens were removed by pickling in NH_4F-H_2SO_4 water solution.

All tensile tests were made using a substandard sheet specimen 0.375 by 0.064 by 5 inches, with a reduced section 0.250 by 0.064 by 1-1/4 inches. Strain rate was 0.04 inch per minute as determined by platen speed. Tensile data presented in this section represent average results of two specimens unless otherwise noted. Bend-test specimens were 0.375 by 0.064 by 4 inches.

Purely for convenience in classifying the alloys as to strength properties, those having tensile strengths in the range 130,000 to 180,000 psi will be considered intermediate-strength alloys, and those with tensile strengths greater than 180,000 psi will be referred to as high-strength alloys.

Binary Alloys

Mechanical properties of binary alloys tested in the past year are given in Table 1. The titanium-chromium and titanium-manganese alloys were intended primarily for use in isothermal-transformation studies. However, as-hot-rolled properties were obtained on these alloys to provide a basing point for any future testing in the heat-treated condition. Heats WR10A, WR25A, and WR12A, which contain 14, 15, and 16 per cent chromium, respectively, are metastable beta alloys.

On the basis of the properties obtained on the recent titanium-chromium alloys, the chart shown in Reference (2), page 37, depicting the tensile properties of earlier titanium-chromium alloys versus composition, has been revised. The revised chart is shown in Figure 1 of the present report. Note that the tensile strength of the Ti-10%Cr alloy shown in the earlier figure has been deleted, because it is now felt that the earlier alloy was too brittle to develop its full tensile strength. The points shown in the chart are, whenever possible, the average of the current results and results obtained previously on similar alloys.

A chart showing the hot-rolled properties of titanium-manganese alloys as a function of alloy content is shown in Figure 2. Experimental points obtained in earlier work are included. Earlier work on this system is summarized in Reference (2), pages 26 and 27.

TABLE 1. AS-HOT-ROLLED PROPERTIES OF BINARY ALLOYS

Heat No.	Intended Composition, %	Ultimate Tensile Strength, psi(1)	Elongation, % in 1 inch	VHN	Minimum Bend Radius, T(2)
WQ47A	Unalloyed	70,700	25.5	186	0.5
WS166A	Unalloyed	73,600	31.0	186	0.5
WR7A	0.5Cr	89,600	26.0	222	1
WR17A	3.0Cr	132,900	14.0	288	1
WR8A	5.0Cr	165,600	9.0	357	1.5
WR152A	8.0Cr	216,300	1.5	421	>24
WS145A	12.0Cr	148,800	11.0	329	1.5
WR10A	14.0Cr	142,000	15.5	294	0.75
WR25A	15.0Cr	138,800	15.0	281	0.5
WR12A	16.0Cr	141,100	17.5	292	0.5
WQ66A	0.5Mn	90,600	20.5	210	1.5
WQ65A	3.0Mn	125,000	19.0	267	2
WR4A	6.0Mn	202,400	5.0	418	24
WS154A	8.0Mn	227,200	1.0	456	24
WQ69A	10.0Mn	171,400	6.0	345	3
WS155A	10.0Mn	177,600	1.0	360	24
WR5A	12.0Mn	154,300	13.0	346	24
WQ17A	0.5 O_2	135,500	1.5	331	4
WS31A	0.1S	85,800	26.0	204	1.5
WS129A	0.25S	82,300	30.0	206	1.5
T994	10.0Bi	91,200	7.0*	229	3
T995	10.0Pb	80,300	12.0*	220	1.5
T1064	10.0In	100,600	16.0	230	3

---

(1) Substandard specimen 0.375 x 0.064 x 5 inches with a reduced section 0.250 x 0.064 x 1-1/4 inches. Sheet rolled from forged 3/4-lb ingots.

(2) Specimen size 0.375 x 0.064 x 4 inches. Bend radius expressed in units of sheet thickness.

* Broke outside gage marks. Uniform elongation recorded.

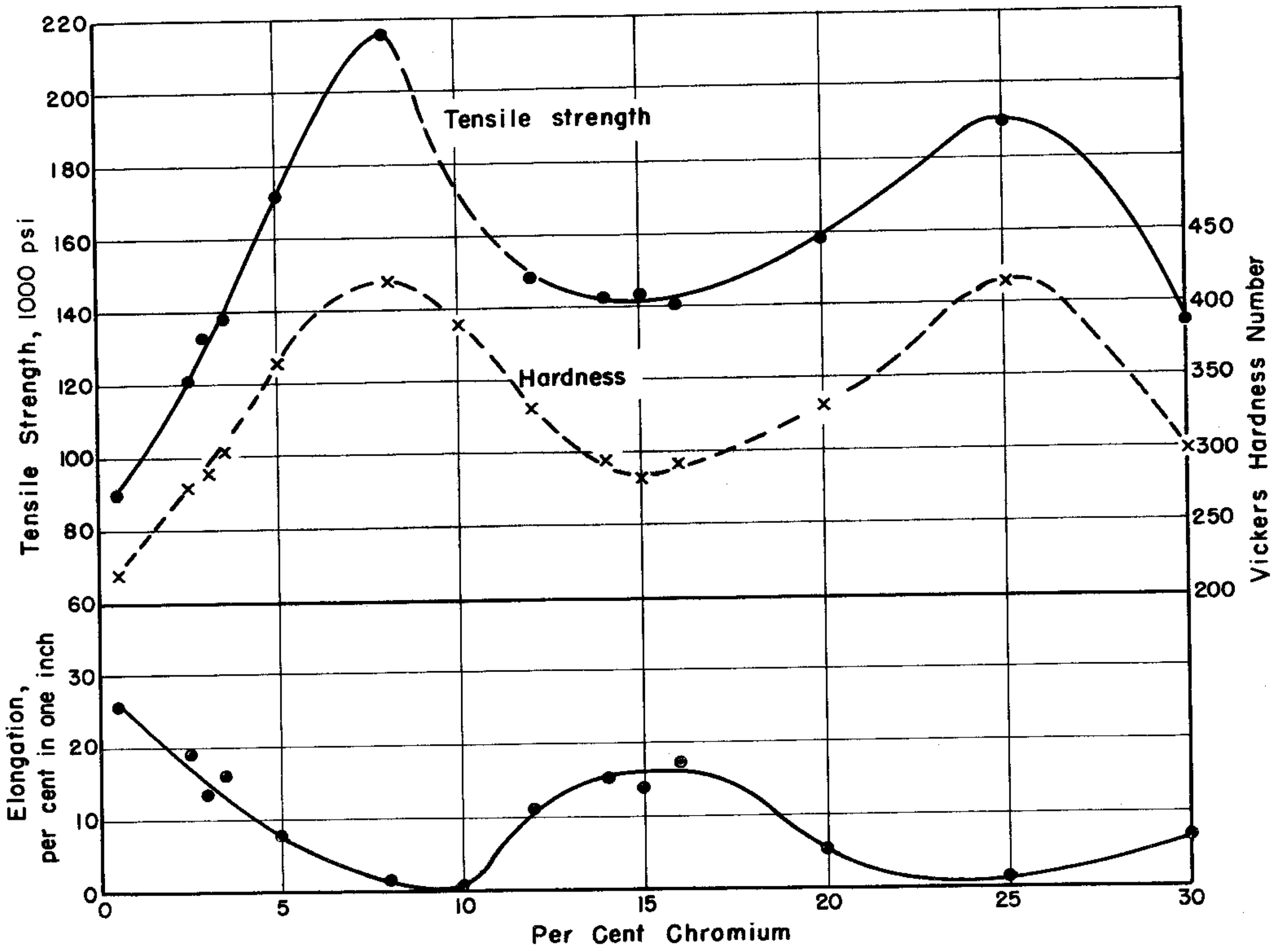


FIGURE I. TENSILE STRENGTH, DUCTILITY, AND HARDNESS OF TITANIUM-CHROMIUM BINARY ALLOYS IN THE AS-HOT-ROLLED CONDITION

A-3247

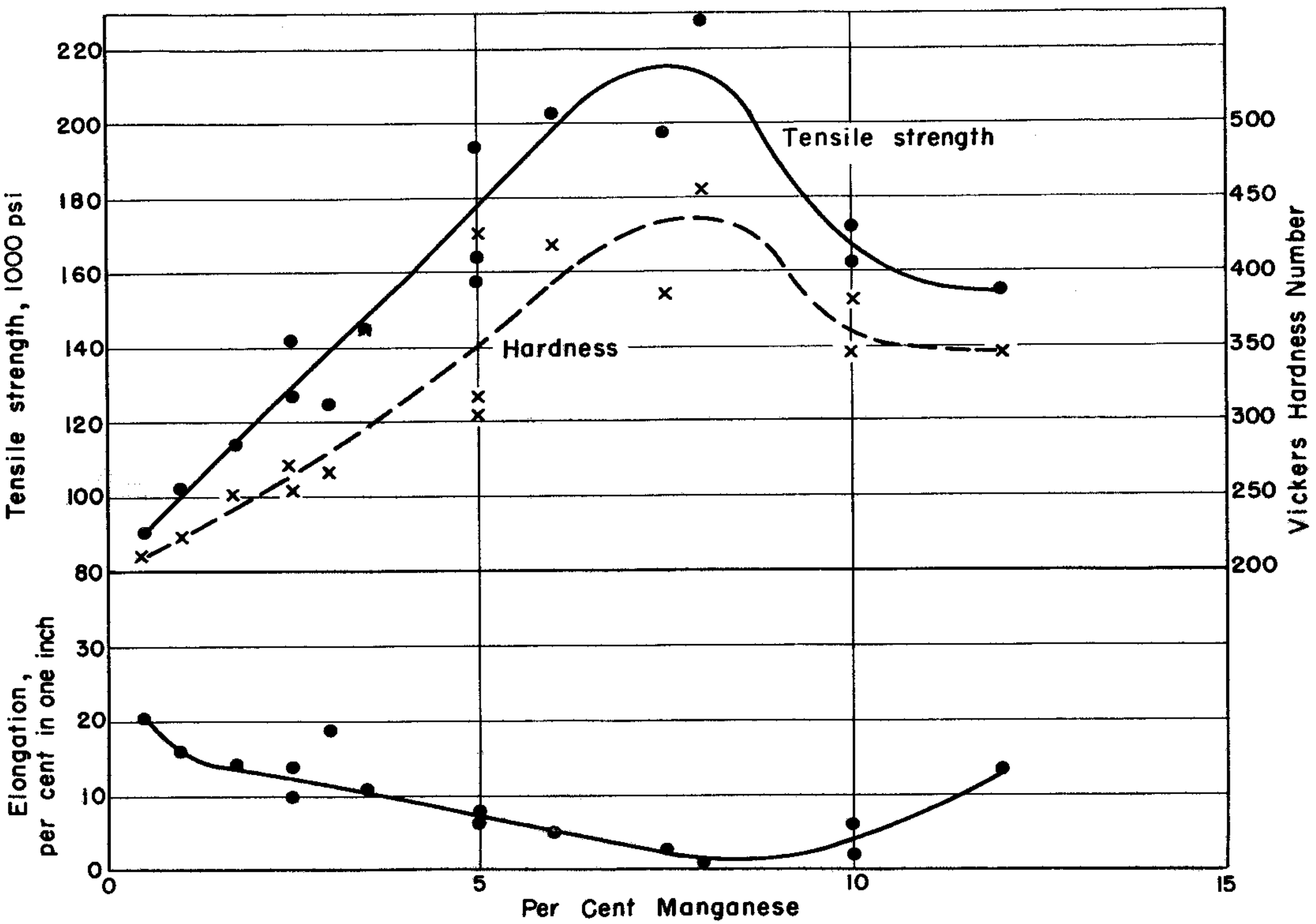


FIGURE 2. TENSILE STRENGTH, DUCTILITY, AND HARDNESS OF TITANIUM-MANGANESE BINARY ALLOYS IN THE AS-HOT-ROLLED CONDITION

A-3248

It is evident from Figures 1 and 2 that a peak in hardness and tensile strength and a consequent minimum ductility are reached in Ti-Cr and Ti-Mn alloys at about 8 per cent alloy content. This maximum in strength and hardness is attributed to coherency hardening produced at relatively low temperatures by a tendency of the alpha phase to separate from the retained beta matrix. For convenience this will be referred to as precipitation or age hardening. At alloy contents above about 8 per cent, the reaction apparently becomes more and more sluggish, resulting in decreasing hardness. A minimum in hardness and tensile strength and a maximum in ductility are reached in as-hot-rolled, titanium-chromium alloys at about 15 per cent chromium. The properties obtained for this alloy content probably represent true solid-solution hardening. The increase in tensile strength and hardness shown by alloys containing more than 15 per cent chromium may result from the precipitation of the $TiCr_2$ compound from the beta phase, but this has not yet been proven.

It must be realized that the maxima shown for these two alloying systems apply only to material in the form of 14-gage sheet fabricated by the particular procedure used for these specimens. A change in rolling temperature or in section size, which would effect a change in cooling rate after rolling, would alter somewhat the shape of these curves.

The titanium-sulfur alloys shown in Table 1 had tensile strengths slightly higher than unalloyed titanium, with about the same tensile and bend ductility. Microscopic examination revealed no second phase. The microstructure appeared essentially the same as that of unalloyed titanium.

The binary bismuth-lead alloys were made to extend the composition range covered in earlier work, Reference (1), pages 127 and 144. The indium alloy was included as one of a series of alloys made, like lead and bismuth alloys, from volatile components. None of the alloys described in the last two paragraphs had particularly good properties and no further work is contemplated on these alloying systems.

Ternary and More Complex Alloys

The alloys discussed in this section were made up by the addition of two or more of the elements chromium, iron, manganese, molybdenum, and vanadium in varying percentages. Previous work on this contract had shown these elements to be the most desirable addition agents from the standpoint of conferring strength, with the least sacrifice of ductility.

Ti-Cr-V Alloys. The mechanical properties of the current series of Ti-Cr-V alloys are given in Table 2. Earlier work on this alloy system had indicated that desirable intermediate and high-strength properties could be obtained within a composition range of 2.5 to 5 per cent chromium and 1 to 3.5 per cent vanadium, Reference (2), pages 66-69. The present series of

TABLE 2. AS-HOT-ROLLED PROPERTIES OF Ti-Cr-V ALLOYS

Heat No.	Intended Composition, %	Ultimate Tensile Strength, psi(1)	Elongation, % in 1 inch	VHN	Minimum Bend Radius, T(2)
WS55A	2.75Cr-2.5V	159,500	10.0	330	3
WS56A	2.75Cr-3.25V	174,400	10.5	357	3
WS58A	2.75Cr-4.0V	168,200	2.0	368	3
WS61A	2.75Cr-4.75V	174,800	7.0	345	3
WS62A	3.5Cr-2.5V	162,000	13.0	348	1.5
WS75A	3.5Cr-3.25V	177,500	5.5	375	8
WS68A	3.5Cr-4.0V	168,200	4.5	375	3
WS69A	3.5Cr-4.75V	172,600	4.5	388	3
WS64A	4.25Cr-2.5V	192,500	6.0	379	4
WS74A	4.25Cr-3.25V	187,500	5.0	374	4
WS71A	4.25Cr-4.0V	163,600	9.0	345	3
WS67A	4.25Cr-4.75V	177,900	6.0	364	3
WS73A	5.0Cr-2.5V	199,100	2.0	390	8
WS66A	5.0Cr-3.25V	196,300	2.0	406	8
WS109A	5.0Cr-4.0V	155,900	7.0	320	8
WS70A	5.0Cr-4.75V	163,000	9.0	335	3

(1) Substandard specimen 0.375 x 0.064 x 5 inches with a reduced section 0.250 x 0.064 x 1-1/4 inches. Sheet rolled from forged 3/4-lb ingots.

(2) Specimen size 0.375 x 0.064 x 4 inches.

alloys was designed to extend the range of vanadium content and to cover intermediate compositions not previously tested. Several alloys of the group had good intermediate-strength properties as hot rolled. In particular, the Ti-3.5%Cr-2.5%V alloy had very good tensile and bend ductility at a strength level of about 160,000 psi. In the high-strength range the alloys which contain 4.25 per cent chromium with 2.5 or 3.25 per cent vanadium had good combinations of properties.

The behavior of the Ti-3.5%Cr-3.25%V alloy was rather disappointing. An earlier heat containing Ti-3.5%Cr-3.5%V had a tensile strength of 200,000 psi, with an elongation of 6 per cent in 1 inch and a bend ductility of 12T. The current heat, WS75A, differing in intended composition by only 0.25 per cent vanadium, had much lower strength. It seems likely that fabrication variables influenced the properties of the two heats more than this small difference in composition. Further investigation of the Ti-3.5%Cr-3.5%V alloy is described later in this report.

Ti-Cr-V-Fe-Base Alloys. Mechanical properties of these alloys are given in Table 3. In general, the Ti-Cr-V-Fe alloys had higher strength and lower ductility than the Ti-Cr-V alloys. The quaternary alloys in the intermediate strength range had somewhat lower ductility than the ternary Ti-Cr-V alloys at about the same strength level. Several alloys exhibited good combinations of high-strength properties.

The effects of the minor additions to the quaternary alloys are uncertain. In general, the addition of 0.1 per cent phosphorus or 0.5 per cent tantalum appeared to increase the ductility without materially affecting strength. The addition of 0.5 per cent columbium apparently had little effect at the 2 or 3 per cent chromium level. It improved ductility slightly in the case of the Ti-4%Cr-1%V-2%Fe-base alloy. It will be shown in a later section that small quantities of a number of different beta-stabilizing elements have an additive effect in improving properties.

Ti-Mn-Cr Alloys. Mechanical properties for these alloys are shown in Table 4. Earlier work on this alloy system may be found in Reference (2), pages 48-49. The current series of alloys covers the composition range which was shown by the earlier work to produce intermediate- and high-strength alloys. Most of the alloys had good properties. The one exception was the Ti-4%Mn-3.5%Cr alloy, which had low ductility. The alloys containing 2.75%Mn-2.5%Cr and 4%Mn-2.5%Cr had particularly good properties. In the high-strength range, the alloys containing 2.75 per cent manganese with 3.5 or 4 per cent chromium, and the Ti-3.5%Mn-3.5%Cr alloy all had good combinations of properties.

Ti-Mn-Cr-Fe-Base Alloys. Mechanical properties for these alloys are presented in Table 6. The addition of 1 or 2 per cent iron to ternary

TABLE 3. AS-HOT-ROLLED PROPERTIES OF Ti-Cr-V-Fe-BASE ALLOYS, SHOWING EFFECTS OF SMALL ADDITIONS OF PHOSPHORUS, COLUMBIUM, OR TANTALUM

Heat No.	Intended Composition, %	Ultimate Tensile Strength, psi(1)	Elongation, % in 1 inch	VHN	Minimum Bend Radius, T(2)
WR43A	2.0Cr-3.0V-1.0Fe	143,200	10.0	292	1.5
WR42A	2.0Cr-2.0V-2.0Fe	158,400	8.5	294	2
WR47A	2.0Cr-1.0V-2.0Fe	170,200	5.0	309	4
WR55A	2.0Cr-1.0V-2.0Fe-0.1P	158,000	7.0	345	3
WR59A	2.0Cr-1.0V-2.0Fe-0.5Cb	158,800	4.5	342	8
WR53A	2.0Cr-1.0V-2.0Fe-0.5Ta	157,200	7.25	287	2
WR56A	3.0Cr-3.0V-1.0Fe	183,100	1.75	315	8
WR54A	3.0Cr-2.0V-2.0Fe	186,600	5.75	357	4
WR58A	3.0Cr-1.0V-2.0Fe	177,800	2.5	366	8
WR57A	3.0Cr-1.0V-2.0Fe-0.1P	175,000	4.5	366	8
T1238A	3.0Cr-1.0V-2.0Fe-0.5Cb	156,000	3.25	354	2
T1189A	3.0Cr-1.0V-2.0Fe-0.5Ta	185,700	4.5	357	24
T1182A(3)	4.0Cr-3.0V-1.0Fe	212,500	1.0	397	>24
T1237A	4.0Cr-2.0V-2.0Fe	193,300	4.0	390	8
T1181A(3)	4.0Cr-2.0V-2.0Fe	85,300(4)	0.0	468	>24
T1183A	4.0Cr-1.0V-2.0Fe	209,600	1.75	397	24
T1185A	4.0Cr-1.0V-2.0Fe-0.1P	200,600	4.75	376	8
T1186A	4.0Cr-1.0V-2.0Fe-0.5Cb	200,900	3.25	376	8
T1187A	4.0Cr-1.0V-2.0Fe-0.5Ta	196,600	3.75	366	12

(1) Substandard specimen 0.375 x 0.064 x 5 inches with a reduced section 0.250 x 0.064 x 1-1/4 inches. Sheet rolled from forged 3/4-lb ingots.

(2) Specimen size 0.375 x 0.064 x 4 inches.

(3) Original ingot abnormally hard. Believed to have been contaminated with oxygen and/or nitrogen during melting.

(4) Results of only one specimen.

TABLE 4. AS-HOT-ROLLED PROPERTIES OF Ti-Mn-Cr ALLOYS

Heat No.	Intended Composition, %	Ultimate Tensile Strength, psi(1)	Elongation, % in 1 inch	VHN	Minimum Bend Radius, T(2)
WS77A	2.75Mn-2.0Cr	158,700	11.0	351	1.5
WS274A	2.75Mn-2.5Cr	162,600	13.0	351	1.5
WS78A	2.75Mn-3.0Cr	168,000	7.5	380	4
WS79A	2.75Mn-3.5Cr	191,400	4.25	414	4
WS80A	2.75Mn-4.0Cr	205,300	5.25	413	8
WS86A	3.5Mn-1.5Cr	159,400	10.5	348	1.5
WS96A	3.5Mn-2.0Cr	171,300	10.5	373	3
WS89A	3.5Mn-2.5Cr	178,900	9.0	376	3
WS88A	3.5Mn-3.0Cr	196,700	3.25	413	8
WS90A	3.5Mn-3.5Cr	209,200	4.25	414	12
WS257A	4.0Mn-1.5Cr	176,600	10.5	380	3
WS99A	4.0Mn-2.0Cr	151,600	13.0	334	1.5
WS101A	4.0Mn-2.5Cr	168,500	11.5	366	1.5
WS102A	4.0Mn-3.0Cr	215,900	1.0	441	24
WS103A	4.0Mn-3.5Cr	164,800	1.0	409	24

(1) Substandard specimens 0.375 x 0.064 x 5 inches with a reduced section 0.250 x 0.064 x 1-1/4 inches. Sheet rolled from forged 3/4- lb ingots.

(2) Specimen size 0.375 x 0.064 x 4 inches.

TABLE 5. AS-HOT-ROLLED PROPERTIES OF Ti-Cr-Mn-Fe ALLOYS SHOWING EFFECTS OF SMALL ADDITIONS OF PHOSPHORUS, TIN, VANADIUM, OR ZIRCONIUM

Heat No.	Intended Composition, %	Ultimate Tensile Strength, psi(1)	Elongation, % in 1 inch	VHN	Minimum Bend Radius, T(2)
WR20A	2.0Cr-1.0Mn-1.0Fe	154,200	6.5	309	1.5
WR21A	2.0Cr-1.0Mn-2.0Fe	169,900	5.25	330	3
WR19A	2.0Cr-3.0Mn-1.0Fe	187,500	1.5	366	8
WR44A(3)	2.0Cr-3.0Mn-1.0Fe-0.1P	179,200	2.25	363	8
WR33A	2.0Cr-3.0Mn-1.0Fe-0.1P	188,700	2.0	370	8
WR60A	2.0Cr-3.0Mn-1.0Fe-0.1Sn	202,900	1.75	394	8
WR65A	2.0Cr-3.0Mn-1.0Fe-0.2Sn	189,400	0.25	336	8
WR51A	2.0Cr-3.0Mn-1.0Fe-0.1V	186,700	2.0	351	8
WR50A	2.0Cr-3.0Mn-1.0Fe-0.25V	168,600	3.25	339	12
WR22A	3.0Cr-1.0Mn-1.0Fe	172,200	3.75	357	4
WR24A	3.0Cr-1.0Mn-2.0Fe	192,100	4.0	348	8
WR23A	3.0Cr-3.0Mn-1.0Fe	191,700	1.0	401	24
WR48A(3)	3.0Cr-3.0Mn-1.0Fe-0.1P	190,700(4)	1.0(4)	429	24
WR67A	3.0Cr-3.0Mn-1.0Fe-0.1P	194,200	0.0	394	>24
WR30A	4.0Cr-1.0Mn-1.0Fe	178,600	6.5	360	3
WS6A	4.0Cr-1.0Mn-2.0Fe	190,300	3.5	397	8
WR62A(3)	4.0Cr-1.0Mn-2.0Fe	203,700	2.0	390	24
WS17A	4.0Cr-3.0Mn-1.0Fe	209,500	1.25	464	>24
WS7A	4.0Cr-3.0Mn-1.0Fe-0.1P	181,500	1.0	425	12
WR49A	4.0Cr-3.0Mn-1.0Fe-0.2Zr	184,200	4.25	376	8
WR46A	4.0Cr-3.0Mn-1.0Fe-0.5Zr	89,700	0.5	488	>24

(1) Substandard specimen 0.375 x 0.064 x 5 inches with a reduced section 0.250 x 0.064 x 1-1/4 inches. Sheet rolled from forged 3/4-lb ingots.

(2) Specimen size 0.375 x 0.064 x 4 inches.

(3) Original ingot abnormally hard. Believed to have been contaminated with oxygen and/or nitrogen during melting.

(4) Results of only one specimen.

Ti-Cr-Mn alloys increased tensile strength and decreased ductility to some extent. The ductilities of all the alloys in this group were lower than those of the Ti-Cr-Mn alloys in Table 4, which had comparable tensile strengths. The Ti-3%Cr-1%Mn-2%Fe alloy had the best high-strength properties.

Small additions of phosphorus, tin, vanadium, and zirconium to the quaternary alloys appeared to have no beneficial effects, except for the addition of 0.2 per cent zirconium to the Ti-4%Cr-3%Mn-1%Fe alloy. The zirconium addition apparently reduced the strength of this alloy somewhat, but increased the ductility considerably.

Ti-Mn-Mo Alloys. The mechanical properties of this group of alloys are given in Table 6. Properties of earlier heats of this alloy system are shown in Reference (2), pages 48 to 51. With the exception of three heats, the earlier alloys were single melted and fabricated by a forging technique which did not provide as much forging reduction as the present process. Only three heats of those in Reference (2), containing 5 per cent manganese, and 2, 3, and 4 per cent molybdenum, respectively, were melted and fabricated using the techniques applied to the current series.

It is evident that all the alloys in this group fall within the intermediate and high-strength range. Both the tensile and bend ductilities of all the alloys are adequate for their respective strength levels. The Ti-5%Mn-2%Mo alloy is one of the most promising high-strength compositions tested to date. An interesting feature of this group of alloys is the relatively wide range of compositions over which high-strength properties are obtained. Variation of the manganese content from 4.25 to 5 per cent and molybdenum content from 2.75 to 3.5 per cent did not greatly affect the mechanical properties. This would be important from a commercial standpoint.

The Ti-8%Mn-8%Mo and Ti-10%Mn-10%Mo alloys were of the metastable beta type. They had exceptional properties for this type of alloy, having medium strength and excellent tensile and bend ductility.

Ti-Fe-Mo Alloys. The mechanical properties for this group of alloys are given in Table 7. Previous work on this alloy system is summarized in Reference (2), pages 76 and 77. As in the case of the Ti-Mn-Mo alloys, however, only a few alloys melted and fabricated under present procedures were tested at that time. The current group of alloys is the most promising yet encountered under this contract. With the exception of the Ti-3.5%Mo-base alloys containing 2 or 2.5 per cent iron, and the Ti-5.25%Mo-3.5%Fe alloy, all of the compositions in this group had tensile strengths of over 190,000 psi. Both tensile and bend ductilities, with a few exceptions, were good for this strength level. Apparently, quite a wide variation in composition can be tolerated in alloys of this type without affecting the

TABLE 6. AS-HOT-ROLLED PROPERTIES OF Ti-Mn-Mo ALLOYS

Heat No.	Intended Composition, %	Ultimate Tensile Strength, psi(1)	Elongation, % in 1 inch	VHN	Minimum Bend Radius, T(2)
WS278A	2.75Mn-2.0Mo	166,300	12.75	339	4
WS279A	2.75Mn-2.75Mo	167,300	11.75	345	3
WS277A	2.75Mn-3.5Mo	180,800	9.5	383	4
WS267A	3.5Mn-2.0Mo	173,400	10.5	345	3
WS269A	3.5Mn-2.75Mo	180,200	10.5	357	3
WS270A	3.5Mn-3.5Mo	182,100	6.5	397	3
WS268A	4.25Mn-2.0Mo	182,500	7.5	394	4
WS265A	4.25Mn-2.75Mo	190,700	6.5	401	4
WS276A	4.25Mn-3.5Mo	191,000	7.25	401	8
WS271A	5.0Mn-2.0Mo	200,000	5.75	437	4
WS272A	5.0Mn-2.75Mo	200,000	5.25	425	8
WS273A	5.0Mn-3.5Mo	200,900	3.5	421	8
WS253A	0.5Mn-0.5Mo	96,300	29.0	302	1.5
WS298A	0.5Mn-9.0Mo	204,500	2.25	390	12
WS283A	8.0Mn-8.0Mo	150,900	16.5	322	1.5
WS254A	9.0Mn-0.5Mo	182,900	6.0	409	4
WS258A	10.0Mn-10.0Mo	151,800	20.5	309	1.5

(1) Substandard specimen 0.375 x 0.064 x 5 inches with a reduced section 0.250 x 0.064 x 1-1/4 inches. Sheet rolled from forged 3/4-lb ingots.

(2) Specimen size 0.375 x 0.064 x 4 inches.

TABLE 7. AS-HOT-ROLLED PROPERTIES OF Ti-Mo-Fe ALLOYS

Heat No.	Intended Composition, %	Ultimate Tensile Strength, psi(1)	Elongation, % in 1 inch	VHN	Minimum Bend Radius, T(2)
WS142A	3.75Mo-2.0Fe	172,200	8.5	342	3
WS139A	3.75Mo-2.5Fe	182,500	9.5	380	3
WS144A	3.75Mo-3.0Fe	197,400	4.5	383	>24
WS150A	3.75Mo-3.5Fe	220,400	3.0	446	12
WS141A	4.5Mo-2.0Fe	194,700	6.0	383	8
WS153A	4.5Mo-2.5Fe	208,000	1.5	433	8
WS163A	4.5Mo-3.0Fe	216,800	4.25	409	12
WS156A	4.5Mo-3.5Fe	204,000	6.5	401	8
WS160A	5.25Mo-2.0Fe	205,200	4.75	409	8
WS158A	5.25Mo-2.5Fe	197,800	4.5	387	12
WS159A	5.25Mo-3.0Fe	213,200	2.5	421	24
WS162A	5.25Mo-3.5Fe	Too brittle to shear after rolling.			
WS264A	2.0Mo-0.5Fe	125,300	20.75	276	1.5
WS248A	9.0Mo-0.5Fe	152,400	14.5	233	3
WS275A	8.0Mo-6.0Fe	155,900	20.0	333	1.5
WS263A	2.0Mo-7.0Fe	188,300	5.75	383	8
WS262A	10.0Mo-8.0Fe	173,300	5.0	347	>24

(1) Substandard specimen 0.375 x 0.064 x 5 inches with a reduced section 0.250 x 0.064 x 1-1/4 inches. Sheet rolled from forged 3/4-lb ingots.

(2) Specimen size 0.375 x 0.064 x 4 inches.

mechanical properties beyond a reasonable range. The data shown in Table 7 were obtained very recently, and an intensive investigation of alloys of this type is planned.

The five alloys shown at the bottom of Table 7 were tested to establish upper and lower composition limits for this alloy system. The Ti-8%Mo-6%Fe composition shows considerable promise as a metastable beta alloy.

Ti-V-Mn and Ti-V-Mn-Cr Alloys. The mechanical properties of this alloy series are shown in Table 8. No previous work had been done on these particular systems. All ternary alloys of this group containing 5 per cent manganese, as well as the Ti-3.5%Mn-5%V alloy, had high strengths. Of these, the Ti-5%Mn-2%V alloy had the best combination of properties. The remaining ternary alloys, with the exception of the Ti-2%V-2%Mn composition, were of the intermediate-strength types. The addition of 1 per cent chromium to the Ti-V-Mn alloys resulted in a considerable increase in strength, except in the case of the Ti-5%V-2%Mn alloy. An addition of 3 per cent chromium produced large increases in strength over those of the ternary alloys, with corresponding decreases in ductility.

Ti-V-Fe Alloys. The mechanical properties of this group of alloys are presented in Table 9. Four alloys of this group had ultimate strengths higher than 180,000 psi. However, several of the Ti-V-Fe alloys had disproportionately low ductility for their respective strength levels. The low ductility appears to be associated with iron contents higher than 1.5 per cent. Thus, the Ti-2%V-1.5%Fe and Ti-3.5%V-1.5%Fe alloys had good ductility and fair strength. The Ti-5%V-1.5%Fe alloy had somewhat lower ductility than might be expected for its strength level, but the combination of properties was satisfactory. However, all alloys containing 3 per cent iron had poor ductility for their particular strengths. This applies also to the 4.5 per cent iron alloys with the exception of the Ti-5%V-4.5%Fe alloy, which had low ductility but unusually high strength. The latter composition and the Ti-3.5%V-1.5%Fe alloy offer the greatest promise for further investigation.

Three of the 5 per cent vanadium alloys were made using high-purity ductile vanadium as an addition agent in place of the lower purity commercial grade. The use of the purer vanadium did not produce consistently better tensile properties. A marked improvement was effected in the bend ductility of the alloys containing 3 and 4.5 per cent iron, but the data are insufficient to conclude that this was caused by the vanadium.

Ti-V-Ni Alloys. Mechanical properties for this group of alloys are also reported in Table 9. The only composition having acceptable high-strength properties was Ti-2%V-3%Ni. This alloy had 7 per cent elongation

TABLE 8. AS-HOT-ROLLED PROPERTIES OF Ti-V-Mn ALLOYS SHOWING EFFECTS OF CHROMIUM ADDITIONS

Heat No.	Intended Composition, %	Ultimate Tensile Strength, psi(1)	Elongation, % in 1 inch	VHN	Minimum Bend Radius, T(2)
WS209A	2.0V-2.0Mn	127,800	16.5	296	1.5
WS210A	2.0V-3.5Mn	157,500	9.75	360	1.5
WS215A	2.0V-5.0Mn	202,100	6.5	421	8
WS247A	2.0V-2.0Mn-1.0Cr	158,600	14.0	351	1.5
WS212A	2.0V-2.0Mn-3.0Cr	179,300	7.5	394	3
WS218A	3.5V-2.0Mn	143,500	12.0	333	1.5
WS216A	3.5V-3.5Mn	174,400	10.25	366	3
WS219A(3)	3.5V-5.0Mn	214,200	3.0	433	8
WS220A	3.5V-2.0Mn-1.0Cr	176,900	10.75	397	3
WS221A	3.5V-2.0Mn-3.0Cr	217,200	4.0	421	8
WS222A	5.0V-2.0Mn	165,700	11.5	387	3
WS223A	5.0V-3.5Mn	202,000	5.0	409	8
WS226A	5.0V-5.0Mn	213,500	3.0	441	24
WS224A	5.0V-2.0Mn-1.0Cr	168,600	4.0	370	4
WS225A	5.0V-2.0Mn-3.0Cr	206,200	3.0	417	12

---

(1) Substandard specimen 0.375 x 0.064 x 5 inches with a reduced section 0.250 x 0.064 x 1-1/4 inches. Sheet rolled from forged 3/4-lb ingots.

(2) Specimen size 0.375 x 0.064 x 4 inches.

(3) Results of only one specimen.

TABLE 9. AS-HOT-ROLLED PROPERTIES OF Ti-V-Fe AND Ti-V-Ni ALLOYS

Heat No.	Intended Composition, %	Ultimate Tensile Strength, psi(1)	Elongation, % in 1 inch	VHN	Minimum Bend Radius, T(2)
WS186A	2.0V-1.5Fe	126,700	18.5	283	1.5
WS194A	2.0V-3.0Fe	172,000	2.0(4)	357	8
WS190A	2.0V-4.5Fe	163,600	0.25(4)	387	12
WS193A	3.5V-1.5Fe	146,600	17.0	336	1.5
WS195A	3.5V-3.0Fe	158,400	0.0(4)	363	4
WS196A	3.5V-4.5Fe	183,300	0.65(4)	417	--
WS199A	5.0V-1.5Fe	156,000	9.25	339	3
WS204A(3)	5.0V-1.5Fe	145,100	9.75	312	3
WS198A	5.0V-3.0Fe	177,200	0.25	425	24
WS206A(3)	5.0V-3.0Fe	186,100	1.75	363	12
WS205A	5.0V-4.5Fe	214,300	2.5	429	24
WS207A(3)	5.0V-4.5Fe	192,600	3.0	414	8
WS239A	2.0V-1.5Ni	116,100	20.0	279	1.5
WS231A	2.0V-3.0Ni	189,600	7.25	394	8
WS201A	2.0V-4.5Ni	140,600	2.75	317	12
WS234A	5.0V-1.5Ni	138,700	14.5	317	1.5
WS232A	5.0V-3.0Ni	125,100	10.5	287	4
WS245A	5.0V-4.5Ni	182,100	1.5	380	--

(1) Substandard specimen 0.375 x 0.064 x 5 inches with a reduced section 0.250 x 0.064 x 1-1/4 inches. Sheet rolled from forged 3/4-lb ingots.

(2) Specimen size 0.375 x 0.064 x 4 inches.

(3) Made with high-purity ductile vanadium instead of the commercially pure grade.

(4) Broke outside gage marks. Uniform elongation recorded.

and could be bent over a radius of 8T at a strength level of 189,600 psi. It may be considered for further evaluation at a later date. The remaining alloys of this group show little promise either as intermediate- or high-strength alloys.

Ti-Mn-V-Fe Alloys. The mechanical properties for these alloys are presented in Table 10. The alloys of this group are outstanding in that nearly every composition offers an excellent combination of properties. Thus, the Ti-2%Mn-2%V-1%Fe alloy had good ductility at an intermediate-strength level, while nearly all the other alloys had considerably higher strength with correspondingly lower ductility. This type of alloy is being investigated in greater detail.

Ti-Mn-Fe Alloys. The mechanical properties for these alloys are given in Table 11. Previous work, Reference (2), pages 50 and 51, indicated that this alloy system held some promise for potential high-strength alloys. Only three of the previously tested alloys had been melted and fabricated in the same manner as the current series. These three alloys, containing 5 per cent manganese and 2, 3, and 4 per cent iron, all had high tensile strengths with fair to good tensile ductility in the as-hot-rolled condition. Several of the current alloys shown in Table 11 had tensile properties in the high-strength range, but their ductilities were relatively low. The Ti-5%Mn-2.75%Fe and Ti-5%Mn-3.5%Fe alloys of the present group had much lower tensile strengths than the earlier alloys having the same manganese content with 3 and 4 per cent iron. These observed differences were attributed to uncontrollable variations in the fabrication procedure rather than to the small differences in chemical compositions.

It is interesting to compare the Ti-Mn-Fe alloys with the corresponding Ti-Mn-Cr alloys described in Table 4. In three out of four cases where alloys had the same manganese content, a given percentage of iron produced higher tensile strength and hardness and lower ductility than an equal percentage of chromium. These differences in properties were large. In the fourth instance, the Ti-Mn-Fe alloy had slightly lower tensile strength but considerably lower ductility. This is further evidence that iron is a more potent hardener than chromium and will require greater compositional control.

Complex Ti-Cr-Fe-Mn-Mo-V Alloys. This series of alloys was made up by holding the composition of four of the five elements constant at 1 per cent and varying the percentage of the fifth element from 0 to 4 per cent in 1 per cent increments. It may be seen from the first six heats listed in Table 12, that the alloys containing 1 per cent of each of any four of the five elements had tensile strengths of about 150,000 psi and fair to good ductility. When 1 per cent of each of all five elements was present, the strength was increased to over 180,000 psi, the tensile ductility was reduced, but the bend ductility was unaffected. By raising the level of any one element from 1 to 4 per cent, the properties of this system could

TABLE 10. AS-HOT-ROLLED PROPERTIES OF Ti-Mn-V-Fe ALLOYS

Heat No.	Intended Composition, %	Ultimate Tensile Strength, psi(1)	Elongation, % in 1 inch	VHN	Minimum Bend Radius, T(2)
WS172A	2.0Mn-2.0V-1.0Fe	156,300	16.0	341	3
WS173A	2.0Mn-2.0V-2.0Fe	179,500	10.5	390	4
WS174A	2.0Mn-4.0V-1.0Fe	183,000	8.25	397	4
WS178A	2.0Mn-4.0V-2.0Fe	196,000	3.0	413	8
WS181A	3.0Mn-2.0V-1.0Fe	184,300	8.25	380	3
WS177A	3.0Mn-2.0V-2.0Fe	202,300	5.0	413	8
WS179A(3)	3.0Mn-4.0V-1.0Fe	219,900	2.5	433	>24
WS182A	3.0Mn-4.0V-2.0Fe	221,800	2.25	433	12
WS187A	4.0Mn-2.0V-1.0Fe	198,400	2.0	401	8
WS184A	4.0Mn-2.0V-2.0Fe	195,400	3.0	405	8
WS183A	4.0Mn-4.0V-1.0Fe	196,700	4.75	413	8
WS185A	4.0Mn-4.0V-2.0Fe	177,000	6.25	366	4

(1) Substandard specimen 0.375 x 0.064 x 5 inches with a reduced section 0.250 x 0.064 x 1-1/4 inches. Sheet rolled from forged 3/4-lb ingots.

(2) Specimen size 0.375 x 0.064 x 4 inches.

(3) Results for only one specimen.

TABLE 11. AS-HOT-ROLLED PROPERTIES OF Ti-Mn-Fe ALLOYS

Heat No.	Intended Composition, %	Ultimate Tensile Strength, psi(1)	Elongation, % in 1 inch	VHN	Minimum Bend Radius, T(2)
WS121A	2.75Mn-2.0Fe	173,200	6.5	397	3
WS122A	2.75Mn-2.75Fe	177,700	6.0	401	4
WS126A	2.75Mn-3.5Fe	226,500	1.0	455	24
WS128A	3.5Mn-2.0Fe	212,100	2.75	421	12
WS127A	3.5Mn-2.75Fe	166,000	1.0	459	24
WS124A	3.5Mn-3.5Fe	200,400	2.25	401	24
WS133A	4.25Mn-2.0Fe	190,500	0.0	446	24
WS169A	4.25Mn-2.75Fe	216,400	1.0	455	24
WS138A	4.25Mn-3.5Fe	201,400	1.75	401	24
WN23A(3)	5.0Mn-2.0Fe	220,700	4.5	444	24
WS140A	5.0Mn-2.0Fe	218,600	1.75	437	24
WS136A	5.0Mn-2.75Fe	175,300	6.0	370	8
WN19A(3)	5.0Mn-3.0Fe	219,700	2.0	439	24
WS134A	5.0Mn-3.5Fe	173,400	2.5	370	>24
WN20A(3)	5.0Mn-4.0Fe	196,800	3.5	400	8

(1) Substandard specimen 0.375 x 0.064 x 5 inches with a reduced section 0.250 x 0.064 x 1-1/4 inches. Sheet rolled from forged 3/4-lb ingots.

(2) Specimen size 0.375 x 0.064 x 4 inches.

(3) Earlier heats, see Reference (3), pp 50-51.

TABLE 12. AS-HOT-ROLLED PROPERTIES OF COMPLEX Ti-Cr-Fe-Mn-Mo-V ALLOYS

Heat No.	Intended Composition, %	Ultimate Tensile Strength, psi(1)	Elongation, % in 1 inch	VHN	Minimum Bend Radius, T(2)
WQ26A	1.0Cr-1.0Fe-1.0Mn-1.0Mo	156,200	10.5	327	3
WQ25A	1.0Cr-1.0Fe-1.0Mn-1.0V	158,800	6.0	330	3
WQ27A	1.0Cr-1.0Fe-1.0Mo-1.0V	151,600	10.5	307	3
WQ23A	1.0Cr-1.0Mn-1.0Mo-1.0V	151,500	14.0	326	3
WS40A	1.0Fe-1.0Mn-1.0Mo-1.0V	141,000	15.5	304	3
WS39A	1.0Cr-1.0Fe-1.0Mn-1.0Mo-1.0V	188,300	2.0(3)	375	3
WS36A	1.0Cr-1.0Fe-1.0Mn-1.0Mo-2.0V	188,500	7.0	361	3
WS60A	1.0Cr-1.0Fe-1.0Mn-1.0Mo-3.0V	181,600	6.0	370	3
WS54A	1.0Cr-1.0Fe-1.0Mn-1.0Mo-4.0V	192,400	4.0	369	4
WS37A	1.0Cr-1.0Fe-1.0Mn-2.0Mo-1.0V	179,600	7.0	359	8
WS43A	1.0Cr-1.0Fe-1.0Mn-3.0Mo-1.0V	198,200	5.0	390	8
WS52A	1.0Cr-1.0Fe-1.0Mn-4.0Mo-1.0V	197,600	4.0	370	8
WS49A	1.0Cr-1.0Fe-2.0Mn-1.0Mo-1.0V	179,900	10.5	363	3
WS53A	1.0Cr-1.0Fe-3.0Mn-1.0Mo-1.0V	196,000	10.5	397	4
WS65A	1.0Cr-1.0Fe-4.0Mn-1.0Mo-1.0V	217,500	1.75(3)	401	12
WS41A	1.0Cr-2.0Fe-1.0Mn-1.0Mo-1.0V	182,500	7.5	363	4
WS42A	1.0Cr-3.0Fe-1.0Mn-1.0Mo-1.0V	220,300	3.0(3)	442	8
WS59A	1.0Cr-4.0Fe-1.0Mn-1.0Mo-1.0V	211,900	5.0	383	8
WS38A	2.0Cr-1.0Fe-1.0Mn-1.0Mo-1.0V	212,000	3.5	397	12
WQ34A	3.0Cr-1.0Fe-1.0Mn-1.0Mo-1.0V	203,200	2.5	408	12
WS63A	4.0Cr-1.0Fe-1.0Mn-1.0Mo-1.0V	200,000	2.0(3)	401	8

(1) Substandard specimen 0.375 x 0.064 x 5 inches with a reduced section 0.250 x 0.064 x 1-1/4 inches. Sheet rolled from forged 3/4-lb ingots.

(2) Specimen size 0.375 x 0.064 x 4 inches.

(3) Broke outside gage marks. Uniform elongation recorded.

be varied over a wide range. All combinations produced high strength and several alloys were outstanding. The alloy containing 3 per cent manganese with 1 per cent of each of the other four elements had the best combination of strength and ductility yet attained in any alloy made under this contract. The alloys containing 3 and 4 per cent iron also had excellent strength-ductility combinations.

This experiment shows that, although each of the five alloying elements has a similar function in titanium, namely, stabilizing the beta phase, a synergistic effect on properties is obtained by combining many elements. Further investigation of larger heats of the 3 per cent manganese and the 4 per cent iron complex alloys is under way.

Ti-O and Ti-N Alloys with Additions of Phosphorus. Previous work on binary titanium-phosphorus alloys, Reference (2), pages 30 to 32, had indicated that phosphorus might have a deoxidizing or denitriding effect on titanium. Additions of phosphorus to unalloyed titanium in amounts up to 0.1 per cent appeared to increase the tensile strength, hardness, and ductility slightly. The present series of alloys was made to determine whether phosphorus would improve the ductility of alloys containing up to 0.5 per cent oxygen or nitrogen. The data in Table 13 indicate that the addition of 0.1 or 0.25 per cent phosphorus to the 0.5 per cent oxygen alloy raised the strength but lowered the ductility.

The intrinsic effect of phosphorus on the properties of a Ti-0.5%N alloy could not be determined because an alloy of this base composition was not available. However, the ductility of both the nitrogen alloys to which phosphorus was added was very low for their respective strength levels. Thus, it may be concluded that phosphorus has no beneficial effect as a deoxidizing or denitriding agent.

Metastable Beta Alloys

The inherently higher ductility of metals having a cubic-type lattice over those having the hexagonal type makes desirable the development of a titanium-base alloy having the body-centered-cubic beta phase stable at ordinary temperatures. Such an alloy should have medium strength and good ductility, hence, good formability. The ability to be age hardened after forming is another desirable characteristic of such an alloy. Previous work had shown that a Ti-15%Cr composition has these characteristics.

In order to investigate additional metastable beta alloys, the heats shown in Table 14 were melted, fabricated, and tested as 14-gage sheet. The compositions shown were selected on the basis of past experience as those most likely to have a retained beta structure in the as-hot-rolled condition. The small additions of silicon, carbon, boron, and beryllium in

TABLE 13. AS-HOT-ROLLED PROPERTIES OF Ti-O AND Ti-N ALLOYS WITH ADDITIONS OF PHOSPHORUS

Heat No.	Intended Composition, %	Ultimate Tensile Strength, psi(1)	Elongation, % in 1 inch	VHN	Minimum Bend Radius, T(2)
RB20	0.5 O_2	117,400	15.5	243	7-1/4
WQ15A	0.5 O_2-0.1P	134,800	12.75	326	8
WQ46A	0.5 O_2-0.25P	132,400	8.25	364	12
WQ18A	0.5N_2-0.1P	151,800	3.5	329	12
WQ44A	0.5N_2-0.25P	150,900	0.0	330	24

(1) Substandard specimen 0.375 x 0.064 x 5 inches with a reduced section 0.250 x 0.064 x 1-1/4 inches. Sheet rolled from forged 3/4-lb ingots.

(2) Specimen size 0.375 x 0.064 x 4 inches.

TABLE 14. AS-HOT-ROLLED PROPERTIES OF EXPLORATORY METASTABLE BETA ALLOYS

Heat No.	Intended Composition, %	Ultimate Tensile Strength, psi(1)	Elongation, % in 1 inch	VHN	Minimum Bend Radius, T(2)
WS15A	4.0Cr-4.0Fe-0.25Si	196,500	1.0	421	12
WS16A	6.0Cr-6.0Fe-0.25Si	156,800	16.0	322	8
WS18A	4.0Cr-4.0Mn-0.25C	218,300	0.0	421	24
WS22A	6.0Cr-6.0Mn-0.25C	147,900	11.0	322	1.5
WS21A	4.0Cr-4.0V-0.25B	193,900	1.0	380	4
WS19A	6.0Cr-6.0V-0.25B	183,800	3.5	409	8
WS24A	4.0Cr-4.0Mo-0.25Be	210,600	2.0	442	12
WS25A	6.0Cr-6.0Mo-0.25Be	165,000	3.5	279	2
WR10A	14.0Cr	142,000	15.5	294	0.75
WR25A	15.0Cr	138,800	15.0	281	0.5
WR12A	16.0Cr	141,100	17.5	292	0.5
WR26A	15.0Cr-0.25Si	140,000	16.0	302	2
WR27A	15.0Cr-0.25C	138,900	14.5	302	1.5

(1) Substandard specimen 0.375 x 0.064 x 5 inches with a reduced section 0.250 x 0.064 x 1-1/4 inches. Sheet rolled from forged 3/4-lb ingots.

(2) Specimen size 0.375 x 0.064 x 4 inches.

some of the alloys were intended to act as grain refiners. The 14 and 16 per cent chromium alloys were included to determine the effect of minor compositional variations on the properties of the Ti-15%Cr-base alloy.

It may be seen from the data in Table 14 that the ternary alloys, with the exception of the Ti-6%Cr-6%Fe-0.25%Si and the Ti-6%Cr-6%Mn-0.25C compositions, did not contain a sufficient quantity of beta-stabilizing elements to develop the desired properties. The binary chromium alloys, including those containing 0.25 per cent silicon or 0.25 per cent carbon, had the best ductility of this group. Varying the compositions of binary chromium alloys from 14 to 16 per cent chromium had no noticeable effect on their mechanical properties. The addition of 0.25 per cent silicon or 0.25 per cent carbon decreased the bend ductility of the 15 per cent chromium alloy slightly, but had no effect on the tensile properties.

Microscopic examination revealed that the addition of 0.25 per cent silicon, carbon, boron, or beryllium had little or no effect on the grain size of these alloys in the as-hot-rolled condition. The alloys containing a total alloy content of 8 per cent had a small amount of the alpha phase present and, as would be expected, their grain sizes were much finer than those of the all-beta alloys.

Further investigation of the 15 per cent chromium alloy is under way at the present time.

Effects of Surface Treatments on the Mechanical Properties of Titanium Alloys

The effect of the contaminated surface layers on the mechanical properties of as-hot-rolled titanium and titanium-alloy sheet has been investigated from time to time under this project. The most recent investigation, Reference (2), pages 110-119, indicated that some improvement in properties was possible in a majority of cases by removing this contaminated layer. Various alloy specimens which had 0.010 inch ground off their as-hot-rolled surfaces showed some improvement in tensile properties in seven instances out of a possible eleven. The effects of etching 0.002 inch from the surfaces of the as-hot-rolled specimens were not conclusive. It was felt that the improvement in properties indicated that a mechanical removal of contaminated surface layers merited further investigation.

Since surface grinding all sheet specimens used on this project would not be feasible, other, more practical methods for removing the contaminated surface and smoothing out surface irregularities were tried. Single tensile blanks from a number of exploratory alloys were given the following surface treatments:

1. Surface pickled in a sulfuric acid-ammonium fluoride-water solution to remove 0.003-0.005 inch from each surface.

2. Specimens pickled as in Item 1, followed by polishing rolled surfaces through 400-grit silicon carbide papers.

3. Specimens pickled as in Item 1, followed by cold rolling to a 4 per cent reduction in thickness.

Complete results of the tests conducted on specimens given in the above treatments are given in Table 15. It may be seen that little or no improvement in tensile properties resulted from the polishing treatment following the removal of surface layers by pickling. The cold-rolling treatment, which was designed primarily to smooth out surface irregularities, resulted in a slight increase in tensile strength and a decrease in ductility in most of the specimens so treated. Since no improvement in properties was evidenced in specimens given the latter two treatments, it was decided to use only the surface pickling treatment for all future work. This treatment has the advantage of removing the oxide layer from the surface of the sheet so that gross irregularities may be detected and avoided in the selection of specimens. All of the as-hot-rolled properties given in the previous sections, with the exception of those for the titanium-oxygen and titanium-nitrogen alloys in Table 13, were obtained on specimens given this surface pickling treatment.

TABLE 15. PROPERTIES OF TITANIUM-BASE ALLOYS IN THE AS-HOT-ROLLED CONDITION WITH VARIOUS SURFACE TREATMENTS

		Surface Treatment					
		Surface Pickled(1)		Pickled and Polished(2)		Pickled and Cold Rolled 4%(3)	
Heat No.	Intended Composition, %	Ultimate Tensile Strength, psi(4)	Elongation, % in 1 inch	Ultimate Tensile Strength, psi(4)	Elongation, % in 1 inch	Ultimate Tensile Strength, psi(4)	Elongation, % in 1 inch
		Chromium-Vanadium Alloys					
WS55A	2.75Cr–2.5V	159,500	10.0	160,600	8.0	168,700	7.0
WS56A	2.75Cr–3.25V	174,400	10.5	169,300	11.0	177,400	8.0
WS58A	2.75Cr–4.0V	168,200	2.0	172,500	2.0	186,300	3.5
WS61A	2.75Cr–4.75V	174,800	7.0	176,400	5.0	183,500	4.0
WS62A	3.5Cr–2.5V	162,000	13.0	165,600	12.0	172,600	8.0
WS75A	3.5Cr–3.25V	177,500	5.5	182,000	6.0	187,900	1.0
WS68A	3.5Cr–4.0V	168,200	4.5	177,300	4.0	169,300	0.0
WS69A	3.5Cr–4.75V	172,600	4.5	172,600	4.0	182,100	1.0
WS64A	4.25Cr–2.5V	192,500	6.0	193,300	6.0	183,200	0.5
WS74A	4.25Cr–3.25V	187,500	5.0	187,500	6.0	191,100	4.0
WS71A	4.25Cr–4.0V	163,600	9.0	165,600	7.0	172,500	6.0
WS67A	4.25Cr–4.75V	177,900	6.0	181,200	6.0	184,300	4.0
WS73A	5.0Cr–2.5V	199,100	2.0	205,500	3.0	205,700	1.0
WS66A	5.0Cr–3.25V	196,300	2.0	194,800	2.0	193,000	2.0
WS109A	5.0Cr–4.0V	155,900	7.0	156,300	7.0	158,900	5.0
WS70A	5.0Cr–4.75V	163,000	9.0	163,000	10.0	174,200	3.0
		Chromium-Iron-Manganese-Molybdenum-Vanadium Alloys					
WQ26A	1Fe–1Cr–1Mn–1Mo	156,200	10.5	–	–	164,700	12.0
WQ25A	1Fe–1Cr–1Mn–1V	158,800	6.0	–	–	169,900	11.0
WQ27A	1Fe–1Cr–1Mo–1V	151,600	10.5	–	–	158,200	10.5
WS40A	1Fe–1Mn–1Mo–1V	141,000	15.5	140,100	16.0	148,300	12.5
WQ23A	1Cr–1Mn–1Mo–1V	151,500	14.0	–	–	159,500	12.5
WS39A	1Cr–1Fe–1Mn–1Mo–1V	188,300	2.0(5)	185,600	9.5	190,700	2.0(5)
WS36A	1Cr–1Fe–1Mn–1Mo–2V	188,500	7.0	189,600	3.5	193,300	1.0(5)
WS60A	1Cr–1Fe–1Mn–1Mo–3V	181,600	6.0	179,100	8.5	194,600	2.0(5)
WS54A	1Cr–1Fe–1Mn–1Mo–4V	192,400	4.0	192,400	4.5	205,200	1.0
WS37A	1Cr–1Fe–1Mn–2Mo–1V	179,600	7.0	180,300	7.5	187,100	5.5(5)
WS43A	1Cr–1Fe–1Mn–3Mo–1V	198,200	5.0	192,800	8.0	200,600	5.0
WS52A	1Cr–1Fe–1Mn–4Mo–1V	197,600	4.0	203,800	4.0	205,000	3.0
WS49A	1Cr–1Fe–2Mn–1Mo–1V	179,900	10.5	177,500	9.0	185,700	7.0
WS53A	1Cr–1Fe–3Mn–1Mo–1V	196,000	10.0	194,600	8.0	205,600	5.5
WS65A	1Cr–1Fe–4Mn–1Mo–1V	217,500	1.75(5)	201,200	0.0	225,300	0.0(5)
WS41A	1Cr–2Fe–1Mn–1Mo–1V	182,500	7.5	181,000	7.5	194,900	4.5
WS42A	1Cr–3Fe–1Mn–1Mo–1V	220,300	3.0(5)	218,100	3.0	216,400	0.0(5)
WS59A	1Cr–4Fe–1Mn–1Mo–1V	211,900	5.0	212,600	5.0	223,600	1.0
WS38A	2Cr–1Fe–1Mn–1Mo–1V	212,000	3.5	213,100	2.0	216,000	2.0
WQ34A	3Cr–1Fe–1Mn–1Mo–1V	203,200	2.5	–	–	219,600	1.0
WS63A	4Cr–1Fe–1Mn–1Mo–1V	200,000	2.0(5)	201,200	3.0	213,400	1.5
WS4A	Unalloyed	74,200	30.5	75,200	18.0	96,000	15.0

TABLE 15. (Continued)

		Surface Treatment					
		Surface Pickled(1)		Pickled and Polished(2)		Pickled and Cold Rolled 4%(3)	
Heat No.	Intended Composition, %	Ultimate Tensile Strength, psi(4)	Elongation, % in 1 inch	Ultimate Tensile Strength, psi(4)	Elongation, % in 1 inch	Ultimate Tensile Strength, psi(4)	Elongation, % in 1 inch
			Metastable Beta Alloys				
WS15A	4.0Cr–4.0Fe–0.25Si	196,500	1.0	–	–	209,200	0.0(5)
WS16A	6.0Cr–6.0Fe–0.25Si	156,800	16.0	–	–	163,200	5.0(5)
WS18A	4.0Cr–4.0Mn–0.25C	218,300	0.0	–	–	226,800	0.0
WS22A	6.0Cr–6.0Mn–0.25C	147,900	11.0	–	–	160,500	8.0
WS21A	4.0Cr–4.0V–0.25B	193,900	1.0	–	–	201,300	0.5
WS19A	6.0Cr–6.0V–0.25B	183,800	3.5	–	–	191,000	3.5
WS24A	4.0Cr–4.0Mo–0.25Be	210,600	2.0	–	–	219,500	0.5
WS25A	6.0Cr–6.0Mo–0.25Be	165,000	3.5	–	–	170,700	4.5
WR10A	14.0Cr	142,000	15.5	–	–	147,400	12.5
WR25A	15.0Cr	138,800	15.0	–	–	155,000	10.0
WR12A	16.0Cr	141,100	17.5	–	–	154,600	13.5
WR26A	15.0Cr–0.25Si	140,000	16.0	–	–	152,800	10.0
WR27A	15.0Cr–0.25C	138,900	14.5	–	–	157,400	7.5
			Special Binary Alloys				
T994	10.0Bi	91,200	7.0(5)	–	–	–	–
T995	10.0Pb	80,300	12.0(5)	–	–	–	–
T1064	10.0In	100,600	16.0	–	–	–	–

(1) 0.003–0.005 inch pickled off the as-hot-rolled surfaces in a NH_4F-H_2SO_4-water solution.

(2) Same pickling treatment as in (1) followed by hand polishing of the rolled surfaces through 600-grit silicon carbide paper.

(3) Same pickling treatment as in (1) followed by cold rolling to a 4 per cent reduction in thickness.

(4) Substandard specimen 0.375 × 0.064 × 5 inches with a reduced section 0.250 × 0.064 × 1-1/4 inches.

(5) Broke outside gage marks. Uniform elongation is given.

HEAT TREATMENT OF EXPLORATORY ALLOYS

An extensive study of the heat-treating characteristics of the more promising titanium alloys has been initiated. This investigation is desirable for a number of reasons. From a commercial standpoint, such information would be very valuable. High-strength alloys which, of necessity, must be high in hardness might be softened for machining or forming operations and, subsequently, heat treated to regain their high strengths. Also, it may be possible to minimize the effects of small compositional variations in a given alloy by suitable heat treatment. Unavoidable variations in fabrication procedures which could affect the properties of the finished product might also be eliminated.

From a research and development viewpoint, a study of the effects of heat treatment will lead to a better knowledge of the fundamental hardening mechanisms occurring in titanium alloys. Consequently, a more intelligent approach to the problem of producing alloys for a given application will be possible. It is now believed that the beta-stabilized titanium alloys are hardened by the precipitation of the hexagonal alpha phase from the body-centered-cubic beta phase. The strengthening mechanism is, apparently, coherency hardening which may occur during cooling at a suitable rate from the alpha-beta- or beta-phase fields, or on reheating of alloys which have been cooled rapidly enough from these regions to retain some or all of the beta phase.

The general types of heat treatments selected for investigation under this phase of the project were as follows:

1. Process Annealing: Heating in the alpha-beta- or beta-phase fields, followed by air cooling.
2. Age Hardening: Heating in the alpha-beta- or beta-phase fields, quenching to room temperature or below, and reheating at a low temperature for a sufficient length of time to obtain the desired properties.
3. Isothermal Transformation: Heating in the alpha-beta- or beta-phase regions, rapidly cooling to a temperature above room temperature, and holding at this temperature for a suitable length of time.

Only limited data could be obtained on the exploratory alloys, because of the small amounts of available stock of a given composition. However, it was felt that sufficient information could be obtained to determine the most promising types of heat treatment.

Process Annealing

Process-annealing treatments were carried out on a number of miscellaneous alloys for which sheet material was available after screening tensile tests had been made. A wide variety of compositions was available, but only a limited quantity of material was on hand from each. The experiments must, therefore, be considered of an exploratory nature, intended to point the way for later, detailed study of the most interesting alloys and treatments.

Specimens were annealed in two different prior conditions, namely, (1) as hot rolled, and (2) as quenched from the all-beta region. Half of the specimens in the latter group were not given a true process anneal, because they were quenched rather than air cooled from the annealing temperature.

Prior Condition: As Hot Rolled

The quaternary Ti-Cr-V-Fe and Ti-Cr-Mn-Fe-base alloys, described earlier in this report, were used for these experiments. Specimens of each alloy were heated in a dried-argon atmosphere at temperatures in the range 1200-1500 F for times of 1/2 and 1 hour and air cooled. Wherever possible, specimens having axes in the direction of the last rolling pass were used. However, in many cases, it was necessary to use specimens cut 90 degrees to the last rolling direction. Complete results of these tests are presented in Tables 16 and 17, along with the as-hot-rolled properties of each alloy.

The 1/2-hour annealing treatment at 1200 F invariably produced lower strength than that obtained in the as-hot-rolled condition. In most cases, the loss in strength was accompanied by very little improvement in ductility. In fact, in many alloys, the ductility actually decreased. Increasing the annealing time to 1 hour at 1200 F reduced the tensile strength still further, but increased the ductility of most of the alloys. The increase in ductility was particularly striking for the Ti-Cr-V-Fe alloys containing 2 per cent chromium, shown in Table 16. Annealing for 1/2 hour at 1300 F resulted in somewhat different effects for the two series of alloys. The Ti-Cr-Mn-Fe alloys generally showed increased strength and rather large increases in ductility over the results obtained in the 1-hour 1200 F treatment. The alloys had lower strength and much better ductility in this condition than as hot rolled. The Ti-Cr-V-Fe alloys, on the other hand, had slightly higher strength with about the same or lower ductility than that obtained in the 1-hour 1200 F treatment. The ductility was generally better than that of the as-rolled specimens, but the strength was disproportionately lower. Annealing for 1/2 hour at 1400 or 1500 F produced higher strength than the other annealing treatments. In some cases, annealing at these higher temperatures resulted in properties essentially the same as or better than those obtained in the as-hot-rolled condition.

TABLE 16. TENSILE PROPERTIES OF Ti-Cr-V-Fe ALLOYS AS HOT ROLLED AND PROCESS ANNEALED AT VARIOUS TEMPERATURES

Heat No.	Intended Composition, %	Orientation(2)	As Hot Rolled		Heat Treatment(1)									
					Annealed 1/2 Hour 1200 F		Annealed 1 Hour 1200 F		Annealed 1/2 Hour 1300 F		Annealed 1/2 Hour 1400 F		Annealed 1/2 Hour 1500 F	
			Ultimate Tensile Strength, psi(3)	Elongation, % in 1 inch	Ultimate Tensile Strength, psi(3)	Elongation, % in 1 inch	Ultimate Tensile Strength, psi(3)	Elongation, % in 1 inch	Ultimate Tensile Strength, psi(3)	Elongation, % in 1 inch	Ultimate Tensile Strength, psi(3)	Elongation, % in 1 inch	Ultimate Tensile Strength, psi(3)	Elongation, % in 1 inch
WR47A	2.0Cr–1.0V–2.0Fe	L	170,200	5.0	146,700	8.0	132,500	8.0	136,400	7.0(4)	–	–	–	–
WR42A	2.0Cr–2.0V–2.0Fe	L	158,400	8.5	141,700	12.0	128,700	19.5	129,500	14.0	168,700	6.0	–	–
WR43A	2.0Cr–3.0V–1.0Fe	L	143,200	10.0	126,700	14.0	115,000	21.0	120,300	16.0	–	–	–	–
WR55A	2.0Cr–1.0V–2.0Fe–0.1P	L	158,000	7.0	137,600	15.0	129,900	16.5	131,400	7.0	150,600	4.0(4)	–	–
		T	–	–	–	–	–	–	146,800	11.0	–	–	141,800	8.0
WR59A	2.0Cr–1.0V–2.0Fe–0.5Cb	L	158,000	4.5	129,800	5.5	121,700	13.0	132,700	11.0	–	–	–	–
		T	–	–	–	–	–	–	154,100	6.0	155,000	5.0	161,000	6.0
WR53A	2.0Cr–1.0V–2.0Fe–0.5Ta	L	157,200	7.25	141,400	6.5	126,500	14.0	130,400	16.0	–	–	–	–
		T	–	–	–	–	–	–	143,800	14.0	160,600	6.0	136,700	8.0
WR58A	3.0Cr–1.0V–2.0Fe	L	177,800	2.5	115,700	0.0(4)	127,500	3.5	100,600	1.0	–	–	–	–
		T	–	–	–	–	–	–	154,000	3.0(4)	163,300	4.5	182,900	1.0
WR54A	3.0Cr–2.0V–2.0Fe	L	186,600	5.75	155,500	2.5	128,000	4.0(4)	144,400	16.5	–	–	–	–
		T	–	–	–	–	–	–	160,800	7.0(4)	171,400	10.0	–	–
WR56A	3.0Cr–3.0V–1.0Fe	L	183,100	1.75	153,900	3.0	141,300	13.5	164,100	4.0(4)	179,500	1.0(4)	–	–
		T	–	–	–	–	–	–	170,000	1.0(4)	188,700	2.5	167,700	5.5
WR57A	3.0Cr–1.0V–2.0Fe–0.1P	L	175,000	4.5	153,900	2.5	134,000	3.0(4)	140,500	11.0	169,700	2.0	–	–
T1238A	3.0Cr–1.0V–2.0Fe–0.5Cb	L	156,000	3.25	130,300	12.0	125,000	7.0(4)	130,400	3.0	–	–	–	–
		T	–	–	–	–	–	–	148,100	15.5	153,400	6.0	–	–
T1188A	3.0Cr–1.0V–2.0Fe–0.5Cb	L	176,500	1.5	–	–	–	–	146,500	10.0(4)	–	–	–	–
		T	–	–	–	–	–	–	174,400	–	184,600	5.5(4)	–	–
T1189A	3.0Cr–1.0V–2.0Fe–0.5Ta	L	185,700	4.5	–	–	138,500	7.5	–	–	–	–	–	–
T1183A	4.0Cr–1.0V–2.0Fe	L	209,600	1.75	167,100	2.0	140,200	6.0	146,200	10.0	–	–	–	–
T1237A	4.0Cr–2.0V–2.0Fe	L	193,300	4.0	170,200	6.5	137,400	14.0	148,800	11.0	–	–	–	–
T1182A	4.0Cr–3.0V–1.0Fe	L	212,500	1.0	158,200	2.0(4)	137,600	14.0	138,600	1.0	–	–	–	–
T1185A	4.0Cr–1.0V–2.0Fe–0.1P	L	200,600	4.75	151,200	2.0(4)	133,500	5.5	156,800	4.0(4)	–	–	–	–
T1186A	4.0Cr–1.0V–2.0Fe–0.5Cb	L	200,900	3.25	160,500	2.0(4)	136,800	10.5	146,900	11.0	194,700	4.0	–	–
T1187A	4.0Cr–1.0V–2.0Fe–0.5Ta	L	196,600	3.75	186,100	3.0	133,700	10.0	151,200	11.0	–	–	–	–
		T	–	–	–	–	–	–	–	–	181,300	5.0	–	–

---

(1) All heat treatments were carried out in an argon atmosphere and followed by cooling in still air.

(2) L = Specimen axis in direction of the last rolling pass.
T = Specimen axis 90° to the direction of the last rolling pass.

(3) Substandard specimen 0.375 × 0.064 × 5 inches with a reduced section 0.250 × 0.064 × 1-1/4 inches.

(4) Broke outside gage marks. Uniform elongation recorded.

TABLE 17. TENSILE PROPERTIES OF Ti-Cr-Mn-Fe BASE ALLOYS AS HOT ROLLED AND PROCESS ANNEALED AT VARIOUS TEMPERATURES

			As Hot Rolled		Heat Treatment(1): Annealed 1/2 Hour 1200 F		Annealed 1 Hour 1200 F		Annealed 1/2 Hour 1300 F		Annealed 1/2 Hour 1400 F		Annealed 1/2 Hour 1500 F	
Heat No.	Intended Composition, %	Orientation(2)	Ultimate Tensile Strength, psi(3)	Elongation, % in 1 inch	Ultimate Tensile Strength, psi(3)	Elongation, % in 1 inch	Ultimate Tensile Strength, psi(3)	Elongation, % in 1 inch	Ultimate Tensile Strength, psi(3)	Elongation, % in 1 inch	Ultimate Tensile Strength, psi(3)	Elongation, % in 1 inch	Ultimate Tensile Strength, psi(3)	Elongation, % in 1 inch
WR20A	2.0Cr-1.0Mn-1.0Fe	L	154,200	6.5	129,300	10.0	113,500	16.0	111,500	3.0	–	–	–	–
		T	–	–	–	–	–	–	–	–	131,800	10.0	–	–
WR21A	2.0Cr-1.0Mn-2.0Fe	L	169,600	5.25	144,900	4.0(4)	124,200	3.5	130,800	15.5	–	–	–	–
WR19A	2.0Cr-3.0Mn-1.0Fe	L	187,500	1.5	160,300	0.5(4)	136,100	2.0	150,600	9.0	175,500	5.0	–	–
WR33A	2.0Cr-3.0Mn-1.0Fe-0.1P	L	188,700	2.0	159,000	0.0	137,300	2.0	156,200	9.5	173,200	5.0	187,100	3.0
WR60A	2.0Cr-3.0Mn-1.0Fe-0.1Sn	L	202,900	1.75	181,700	0.5(4)	131,000	8.0	142,200	11.0	–	–	–	–
WR65A	2.0Cr-3.0Mn-1.0Fe-0.2Sn	L	189,400	0.25	149,700	3.0	134,100	5.5	138,700	17.0	158,400	11.0	–	–
		T	–	–	–	–	–	–	150,300	12.0	–	–	147,200	0.5
WR51A	2.0Cr-3.0Mn-1.0Fe-0.1V	L	186,700	2.0	164,000	0.0(4)	134,300	5.0	139,200	10.0	153,800	10.5	166,900	6.0
WR50A	2.0Cr-3.0Mn-1.0Fe-0.25V	L	168,600	3.25	150,600	4.0	133,200	9.0	132,800	10.0	–	–	–	–
		T	–	–	–	–	–	–	–	–	158,800	10.5	–	–
WR22A	3.0Cr-1.0Mn-1.0Fe	L	172,200	3.75	148,100	2.0(4)	127,400	6.5	131,800	14.5	–	–	–	–
		T	–	–	–	–	–	–	129,400	11.0	–	–	–	–
WR24A	3.0Cr-1.0Mn-2.0Fe	L	192,100	4.0	150,000	1.5(4)	127,900	3.0	136,600	14.0	183,500	6.0	–	–
WR23A	3.0Cr-3.0Mn-1.0Fe	L	191,700	1.0	143,700	0.5	138,600	3.5	150,900	7.0	–	–	–	–
		T	–	–	–	–	–	–	171,100	7.0	187,500	6.0	–	–
WR48A	3.0Cr-3.0Mn-1.0Fe-0.1P	L	190,700	1.0	154,400	0.0(4)	136,000	2.0	145,300	6.0	–	–	–	–
WR30A	4.0Cr-1.0Mn-1.0Fe	L	178,600	6.5	158,900	5.0	135,800	9.5	136,500	15.0	154,100	11.5	–	–
		T	–	–	–	–	–	–	156,800	9.0	–	–	–	–
WR62A	4.0Cr-1.0Mn-2.0Fe	L	203,700	2.0	179,400	0.0(4)	132,900	5.5	143,500	9.0	182,100	4.0	–	–
WS17A	4.0Cr-3.0Mn-1.0Fe	L	209,500	1.25	169,000	2.0	133,700	3.0	146,400	16.0	–	–	–	–
WR31A	4.0Cr-3.0Mn-1.0Fe	L	174,500	3.0	–	–	–	–	142,400	11.0	–	–	–	–
WR63A	4.0Cr-3.0Mn-1.0Fe	L	113,600	0.5	–	–	–	–	163,200	7.5	187,500	5.0	–	–
		T	–	–	–	–	–	–	188,300	5.0	–	–	–	–
WS7A	4.0Cr-3.0Mn-1.0Fe-0.1P	L	181,500	1.0	169,600	1.0(4)	137,300	4.0	154,000	6.0	–	–	–	–
		T	–	–	–	–	–	–	–	–	187,900	5.5	–	–
WR49A	4.0Cr-3.0Mn-1.0Fe-0.2Zr	L	184,200	4.25	161,900	4.0	132,500	3.0	141,600	12.0	–	–	–	–

(1) All heat treatments were carried out in an argon atmosphere and followed by cooling in still air.

(2) L = Specimen axis in the direction of the last rolling pass.
T = Specimen axis 90° to the direction of the last rolling pass.

(3) Substandard specimen 0.375 × 0.064 × 5 inches with a reduced section 0.250 × 0.064 × 1-1/4 inches.

(4) Broke outside gage marks. Uniform elongation recorded.

Other specimens of the same two groups of alloys were cold rolled prior to being annealed. To obtain the requisite ductility for cold rolling, specimens of all heats were heated for 4 hours at 1250 F in an argon atmosphere and furnace cooled. They were then cold rolled to 15, 20, and 25 per cent reduction in thickness and annealed at various temperatures in the range 500 to 1500 F. Two cooling rates were used from annealing temperatures of 1100 F and higher. Some specimens were individually air cooled in a rack. Others were placed in a stainless steel boat approximately 1-1/2 inches in diameter, packed in titanium chips, and annealed and cooled in this container. The tensile properties and hardnesses obtained in these tests, as well as those obtained on specimens in the as-hot-rolled condition, are given in Table 18. It is again pointed out that these data are exploratory, and are intended only to suggest avenues for further detailed investigation.

Tensile properties superior to those obtained in the as-hot-rolled condition were produced in Heat WR19A, cold rolled 20 and 25 per cent, and subsequently annealed at 1400 F; in Heat WR56A, cold rolled 25 per cent and annealed at 1400 F; and in Heat TI188A, cold rolled 20 per cent and annealed at 1500 F. All of these specimens were individually air cooled from the annealing temperature.

Air cooling in the stainless steel boat generally produced higher ductility and lower strength in all the alloys at all annealing temperatures.

Prior Condition: As Quenched From the Beta Field

Ternary Ti-Cr-V-base alloys and complex Ti-Cr-Fe-Mn-Mo-V alloys were used for these experiments. All specimens were quenched in ice water from 1700 F (927 C) prior to being annealed at a lower temperature. The conditions under which the 1700 F treatment was carried out on the two groups of alloys varied somewhat. The complex alloys were heated in a dried-argon atmosphere for a time of 1/2 hour prior to quenching. The ternary Ti-Cr-V alloys were beta-ized by heating in a lead bath for three minutes prior to water quenching. This short-time treatment was used to maintain as small a grain size as possible.

Preliminary experiments indicated that all of the ternary Ti-Cr-V alloys were 100 per cent beta after three minutes in a molten lead bath maintained at 1700 F. Subsequent microscopic examination of the quenched specimens indicated that the short-time treatment produced only a slightly smaller grain size than holding at temperature for 1/2 hour.

After the above treatments, specimens of each heat were reheated for one hour in a dried-argon atmosphere at temperatures of 1350, 1450, and 1550 F, and either air cooled or quenched in ice water. Complete results of tensile and hardness tests made on these specimens, along with the as-hot-rolled properties of each alloy, are given in Tables 19 and 20. The properties of specimens water quenched from 1700 F are also included.

TABLE 18. PROPERTIES OF Ti-Cr-Mn-Fe- AND Ti-Cr-V-Fe-BASE ALLOYS PROCESS ANNEALED(1), COLD ROLLED, AND REANNEALED(2)

Heat No.	Intended Composition, %	As Hot Rolled at 1450 F			15% Cold Reduction			20% Cold Reduction			25% Cold Reduction		
		Ultimate Tensile Strength, psi(3)	Elongation, % in 1 inch	VHN	Ultimate Tensile Strength, psi(3)	Elongation, % in 1 inch	VHN	Ultimate Tensile Strength, psi(3)	Elongation, % in 1 inch	VHN	Ultimate Tensile Strength, psi(3)	Elongation, % in 1 inch	VHN
Cold Rolled, Annealed 1 Hour at 500 F (260 C), Individually Air Cooled(4)													
WR33A	2Cr–3Mn–1Fe–0.1P	188,700	2.0	370	158,600	3.5	328	170,200	5.5	316	167,000	7.0	347
WR48A	3Cr–3Mn–1Fe–0.1P	190,700	1.0	429	136,400	2.0	284	–	–	–	–	–	–
WR63A	4Cr–3Mn–1Fe	113,600	0.5	462	152,000	2.0	317	160,300	6.0	257	165,300	7.0	354
WS7A	4Cr–3Mn–1Fe–0.1P	181,500	1.0	425	135,500	2.0	311	105,300	1.0	353	159,300	4.0	358
WR47A	2Cr–1V–2Fe	170,200	5.0	309	141,500	11.0	309	148,200	10.0	336	150,400	10.0	330
WR58A	3Cr–1V–2Fe	177,700	2.5	366	140,400	8.0	301	167,500	5.0	282	153,000	7.0	319
T1183A	4Cr–1V–2Fe	209,550	1.75	397	138,500	5.0	313	165,300	2.0	350	157,600	4.0	354
Cold Rolled, Annealed 1 Hour at 900 F (482 C), Individually Air Cooled(4)													
WR60A	2Cr–3Mn–1Fe–0.1Sn	202,900	2.0	394	131,000	11.0	316	154,700	6.0	323	142,700	7.0	308
WR51A	2Cr–3Mn–1Fe–0.1V	186,700	2.0(6)	351	129,600	12.0	284	138,000	11.5	320	140,200	8.0	316
WS17A	4Cr–3Mn–1Fe	209,500	1.25	464	134,200	8.0	308	130,500	6.0	–	147,800	11.0	326
WR49A	4Cr–3Mn–1Fe–0.2Zr	184,200	4.25	376	129,600	13.5	291	136,900	12.0	298	141,600	10.0	308
WR55A	2Cr–1V–2Fe–0.1P	158,000	7.0	345	–	–	–	135,000	17.0	297	133,600	20.0	304
WR57A	3Cr–1V–2Fe–0.1P	175,000	4.5	366	138,500	12.0	316	137,000	11.0	319	143,900	7.0(6)	323
T1189A	3Cr–1V–2Fe–0.5Ta	185,700	4.5	357	136,400	13.0	310	151,100	4.0	317	143,100	7.0	315
T1185A	4Cr–1V–2Fe–0.1P	200,500	4.75	376	137,600	13.0	302	151,000	7.0	382	145,200	14.0	318
Cold Rolled, Annealed 1 Hour at 1100 F (593 C), Individually Air Cooled(4)													
WR65A	2Cr–3Mn–1Fe–0.1Sn	189,400	0.25	336	128,900	7.0	268	135,000	11.0	295	133,400	10.0	298
WR50A	2Cr–3Mn–1Fe–0.2Zr	168,600	3.25	339	131,500	8.0	286	–	–	–	128,600	8.0	313
WS6A	4Cr–1Mn–2Fe	190,300	3.5	397	128,500	10.0	284	143,200	11.0	309	143,900	11.0	340
Cold Rolled, Annealed 1 Hour at 1100 F, Air Cooled in Boat(5)													
WR59A	2Cr–1V–2Fe–0.5Cb	158,800	4.5	342	122,500	16.0	259	122,700	22.0	273	127,900	17.0	291
T1186A	4Cr–1V–2Fe–0.5Cb	200,900	3.25	376	131,500	15.5	288	138,000	3.0	260	136,000	13.0	310
T1181A	4Cr–2V–2Fe	85,300	0.0	468	136,700	13.0	302	–	–	–	–	–	–
Cold Rolled, Annealed 1 Hour at 1300 F (704 C), Individually Air Cooled(4)													
WR20A	2Cr–1Mn–1Fe	154,200	6.5	309	121,500	17.0	263	124,800	17.0	256	126,300	15.0	266
WR42A	2Cr–2V–2Fe	158,400	8.5	294	139,700	8.0	280	140,800	21.0	287	143,100	18.0	285
WR54A	3Cr–2V–2Fe	186,600	5.75	357	145,700	18.0	300	137,300	2.0	318	138,800	19.0	309

TABLE 18. (Continued)

Heat No.	Intended Composition, %	As-Hot-Rolled at 1450 F			15% Cold Reduction			20% Cold Reduction			25% Cold Reduction		
		Ultimate Tensile Strength, psi(3)	Elongation, % in 1 inch	VHN	Ultimate Tensile Strength, psi(3)	Elongation, % in 1 inch	VHN	Ultimate Tensile Strength, psi(3)	Elongation, % in 1 inch	VHN	Ultimate Tensile Strength, psi(3)	Elongation, % in 1 inch	VHN
					Cold Rolled, Annealed 1 Hour at 1300 F, Air Cooled in Boat(5)								
WR22A	3Cr–1Mn–1Fe	172,200	3.75	357	125,200	18.0	274	124,300	21.0	258	126,500	24.0	280
WR30A	4Cr–1Mn–1Fe	178,600	6.5	360	126,900	23.5	260	124,000	19.0	268	127,700	22.0	301
T1237A	4Cr–2V–2Fe	193,300	4.0	390	142,600	14.0	321	141,000	15.5	309	143,000	15.0	308
					Cold Rolled, Annealed 1 Hour at 1400 F (760 C), Individually Air Cooled(4)								
WR19A	2Cr–3Mn–1Fe	187,500	1.5	366	161,900	(6)	410	200,300	3.5	401	195,800	3.0	408
WR44A	2Cr–3Mn–1Fe–0.1P	179,200	3.0	363	170,800	1.0	347	–	–	–	–	–	–
WR56A	3Cr–3V–1Fe	183,100	1.5	315	160,600	0.0	412	153,200	2.0	394	210,200	1.0(6)	467
					Cold Rolled, Annealed 1 Hour at 1400 F, Air Cooled in Boat(5)								
WR31A	4Cr–3Mn–1Fe	174,500	3.0	366	141,700	15.0	313	142,800	14.0	343	141,600	13.0	320
WR43A	2Cr–3V–1Fe	143,200	10.0	292	116,500	19.0	258	115,800	22.0	345	116,300	21.0	260
T1182A	4Cr–3V–1Fe	212,500	0.75	397	151,000	11.0	340	156,000	7.0	349	156,000	10.0	353
					Cold Rolled, Annealed 1 Hour at 1500 F (816 C), Individually Air Cooled(4)								
WR21A	2Cr–1Mn–2Fe	169,900	5.0	330	156,900	4.5	311	168,000	4.0	314	150,000	6.0	293
WR53A	2Cr–1V–2Fe–0.5Ta	157,200	7.25	287	148,600	2.0	284	152,000	8.0	355	148,300	1.0(6)	287
T1188A	3Cr–1V–2Fe–0.5Cb	176,500	1.5	390	170,800	1.5	357	185,500	5.0	342	172,900	5.0	336
					Cold Rolled, Annealed 1 Hour at 1500 F, Air Cooled in Boat(5)								
WR24A	3Cr–1Mn–1Fe	171,100	4.0	348	131,200	12.0	299	131,800	18.5	297	133,300	18.0	310
WR62A	4Cr–1Mn–2Fe	203,700	2.0	390	143,600	12.0	312	–	–	–	–	–	–
T1187A	4Cr–1V–2Fe–0.5Ta	196,600	3.75	366	144,000	6.0	330	147,400	11.0	336	147,500	13.0	346

(1) Annealed 4 hours at 1250 F (675 C) and furnace cooled.

(2) All heat treatments were done in a dried-argon atmosphere.

(3) Substandard specimen 0.375 × 0.064 × 5 inches with a reduced section 0.250 × 0.064 × 1-1/4 inches.

(4) Specimens were cooled in still air in a stainless steel rack.

(5) Specimens were cooled in a stainless steel boat 1-1/2 inches in diameter, packed with titanium chips.

(6) Broke outside gage marks. Uniform elongation recorded.

TABLE 19. TENSILE PROPERTIES AND HARDNESSES OF COMPLEX Ti-Cr-Fe-Mn-Mo-V ALLOYS, AS HOT ROLLED AND HEAT TREATED

								Water Quenched (0 C) From 1700 F, Reheated 1 Hour at(1)									
		As Hot Rolled at 1450 F			Water Quenched (0 C) From 1700 F (925 C)(1)				1350 F (732 C)(2)			1450 F (787 C)(2)			1550 F (842 C)(2)		
Heat No.	Intended Composition, %	Ultimate Tensile Strength, psi(3)	Elongation, % in 1 inch	VHN	Ultimate Tensile Strength, psi(3)	Elongation, % in 1 inch	VHN	Cooling Medium	Ultimate Tensile Strength, psi(3)	Elongation, % in 1 inch	VHN	Ultimate Tensile Strength, psi(3)	Elongation, % in 1 inch	VHN	Ultimate Tensile Strength, psi(3)	Elongation, % in 1 inch	VHN
WS39A	1Cr–1Fe–1Mn–1Mo–1V	188,300	2.0(5)	375	127,800(6)	1.0	421	Ice water	143,500	6.0	325	160,200	5.0	366	150,300	4.5	322
WS36A	1Cr–1Fe–1Mn–1Mo–2V	188,500	7.0	361	126,100(6)	0.0	380	Ice water	164,800	6.0	358	153,800	6.0	422	101,900(6)	0.0	409
WS60A	1Cr–1Fe–1Mn–1Mo–3V	181,600	6.0	370	(4)	(4)	426	Still air	152,500	5.0	357	171,700	4.5	369	159,700	4.0	351
WS54A	1Cr–1Fe–1Mn–1Mo–4V	192,400	4.0	369	119,500(6)	1.0	427	Ice water	162,400	2.0	432	163,200	2.0	442	117,200(6)	0.0	446
WS37A	1Cr–1Fe–1Mn–2Mo–1V	179,600	7.0	359	150,000	3.0	385	Still air	131,300	11.0	309	147,500	6.0	314	144,400	5.0	324
WS43A	1Cr–1Fe–1Mn–3Mo–1V	198,200	5.0	390	106,000(6)	1.0	420	Ice water	166,100	2.5	371	155,200	2.5	410	140,500	2.5	432
WS52A	1Cr–1Fe–1Mn–4Mo–1V	197,600	4.0	370	145,700	2.0	386	Still air	150,600	7.0	351	173,800	2.0	412	160,400(6)	1.0	401
WS49A	1Cr–1Fe–2Mn–1Mo–1V	179,900	10.5	363	113,000(6)	0.5	461	Ice water	158,100	11.0	370	157,900	3.0	402	87,900(6)	1.0	459
WS53A	1Cr–1Fe–3Mn–1Mo–1V	196,000	10.0	397	136,400(6)	1.0	433	Still air	148,100	11.0	333	167,500	2.0	373	163,000	3.0	357
WS65A	1Cr–1Fe–4Mn–1Mo–1V	217,500	1.75(5)	401	191,500	5.5	409	Ice water	170,000	2.5	378	182,100	1.5	396	191,600	3.0	410
WS41A	1Cr–2Fe–1Mn–1Mo–1V	182,500	7.5	363	49,100(6)	0.5	478	Ice water	155,900	8.0	351	168,400	4.0	401	120,700(6)	1.0(5)	459
WS42A	1Cr–3Fe–1Mn–1Mo–1V	220,300	3.0(5)	442	202,600	3.0	409	Still air	161,500	2.0	390	175,600	2.0	425	166,000(6)	2.0	452
WS59A	1Cr–4Fe–1Mn–1Mo–1V	211,900	5.0	383	205,800	4.0	437	Ice water	173,600	4.0	361	188,500	2.0	418	202,800	3.0	428
WS38A	2Cr–1Fe–1Mn–1Mo–1V	212,000	3.5	397	155,400(6)	1.0	436	Still air	153,500	5.0	360	174,400	2.5	374	178,200	2.0	401
WS34A	3Cr–1Fe–1Mn–1Mo–1V	203,200	2.5	408	130,500	3.0	426	Ice water	148,500	2.5	401	141,300	2.5	420	133,900(6)	2.0	421
WS63A	4Cr–1Fe–1Mn–1Mo–1V	200,000	2.0(5)	401	146,200	6.0	400	Still air	162,100	3.5	345	179,300	2.0	383	187,400	3.0	392
WQ26A	1Fe–1Cr–1Mn–1Mo	156,200	10.5	327	161,000(6)	1.0	437	Ice water	137,000	10.0	309	139,500	12.0	349	161,500	2.0	424
WQ25A	1Fe–1Cr–1Mn–1V	158,800	6.0	330	166,300(6)	1.0	412	Still air	127,000	6.0	299	138,700	5.5	322	123,000	8.0	288
WQ27A	1Fe–1Cr–1Mo–1V	151,600	10.5	307	164,000	2.0	390	Still air	124,800	13.5	279	129,300	9.0	285	119,800	11.0	274
WS40A	1Fe–1Mn–1Mo–1V	141,000	15.5	304	172,700	3.0	387	Ice water	126,500	14.0	318	126,500	11.0	326	164,400	2.0	374
WQ23A	1Cr–1Mn–1Mo–1V	151,500	14.0	326	145,300	0.5	346	Ice water	130,900	8.0	314	123,800	7.0	354	161,100	2.0	358

(1) Specimens heated in dried argon at 1700 F for 1/2 hour and quenched in ice water (0 C).

(2) Heated in a dried-argon atmosphere.

(3) Substandard specimen 0.375 × 0.064 × 5 inches with a reduced section 0.250 × 0.064 × 1-1/4 inches.

(4) Specimen too brittle to test.

(5) Broke outside gage marks. Uniform elongation recorded.

(6) Specimens believed to be too brittle to develop a true tensile strength.

TABLE 20. TENSILE PROPERTIES AND HARDNESSES OF TERNARY, Ti-Cr-V ALLOYS, AS HOT ROLLED AND HEAT TREATED

		As Hot Rolled at 1450 F			Water Quenched (0 C) From 1700 F (925 C)(1)			Water Quenched (0 C) From 1700 F, Reheated 1 Hour at(1)									
									1350 F (732 C)(2)			1450 F (787 C)(2)			1550 F (842 C)(2)		
Heat No.	Intended Composition, %	Ultimate Tensile Strength, psi(3)	Elongation, % in 1 inch	VHN	Ultimate Tensile Strength, psi(3)	Elongation, % in 1 inch	VHN	Cooling Medium	Ultimate Tensile Strength, psi(3)	Elongation, % in 1 inch	VHN	Ultimate Tensile Strength, psi(3)	Elongation, % in 1 inch	VHN	Ultimate Tensile Strength, psi(3)	Elongation, % in 1 inch	VHN
WS55A	2.75Cr–2.5V	159,500	10.0	330	161,300	4.0	345	Still air	136,700	12.0	317	143,100	8.0	325	143,000	5.0	286
WS56A	2.75Cr–3.25V	174,400	10.5	357	151,500	3.0	357	Ice water	146,300	10.0	339	158,500	1.5	339	133,100	2.0	219
WS58A	2.75Cr–4.0V	168,200	2.0	368	150,900	4.0	333	Still air	162,500	8.0	342	160,000	8.0	433	157,100	5.0	333
WS61A	2.75Cr–4.75V	174,800	7.0	345	150,300	5.0	330	Ice water	143,400	7.0	390	124,800	2.0	446	133,300	2.0	312
WS62A	3.5Cr–2.5V	162,000	13.0	348	156,800	4.5	348	Still air	133,700	14.0	303	148,500	7.0	320	141,500	7.0	304
WS75A	3.5Cr–3.25V	177,500	5.5	375	76,700(4)	1.0	344	Ice water	145,700	4.5	397	–	–	–	137,000(4)	2.0	401
WS68A	3.5Cr–4.0V	168,200	4.5	375	116,300(4)	1.5	417	Still air	158,100	6.0	297	153,500	4.5	306	160,800	2.0	314
WS69A	3.5Cr–4.75V	172,600	4.5	388	112,800(4)	1.0	433	Ice water	143,900	6.0	397	158,800(4)	1.0	413	88,400(4)	0.0	314
WS64A	4.25Cr–2.5V	192,500	6.0	379	116,200(4)	1.0	430	Still air	154,200	6.0	315	163,900	7.0	342	161,700	6.0	309
WS74A	4.25Cr–3.25V	187,500	5.0	374	62,000(4)	1.0	374	Ice water	142,300	8.0	369	148,200	2.0	327	82,800(4)	1.0	366
WS71A	4.25Cr–4.0V	163,600	9.0	345	155,200	5.5	342	Still air	138,800	10.0	309	147,900	6.0	447	145,500	7.0	303
WS67A	4.25Cr–4.75V	177,900	6.0	364	146,300	4.0	336	Ice water	142,900	9.5	405	150,900	1.0	376	119,600(4)	1.0	317
WS73A	5.0Cr–2.5V	199,100	2.0	390	146,700(4)	1.0	425	Still air	164,800	9.0	349	161,200	4.5	433	168,500	6.0	357
WS66A	5.0Cr–3.25V	196,300	2.0	406	102,400(4)	1.0	422	Ice water	144,800	8.0	360	147,600	1.5	327	109,500(4)	1.0	373
WS109A	5.0Cr–4.0V	155,900	7.0	320	121,600(4)	1.5	342	Still air	124,900(4)	1.0	339	129,300(4)	1.0	443	140,500(4)	1.0	330
WS70A	5.0Cr–4.75V	163,000	9.0	335	116,300(4)	0.0	425	Ice water	144,400	7.0	383	–	–	–	112,300(4)	1.0	401

(1) Specimens heated at 1700 F for 3 minutes in molten lead and quenched in ice water (0 C).

(2) Heated in a dried-argon atmosphere for 1 hour.

(3) Substandard specimen 0.375 × 0.064 × 5 inches with a reduced gage section 0.250 × 0.064 × 1-1/4 inches.

(4) Specimens believed to be too brittle to develop a true tensile strength.

The properties of the specimens of both groups of alloys quenched from the beta region were poor. In general, both the tensile strengths and the ductilities were much lower than those obtained in the as-hot-rolled condition. In many cases, very low tensile strengths were obtained at relatively high hardness levels, indicating that these specimens were too brittle to attain a true tensile strength.

Heats WS42A, WS65A, and WS59A of the complex alloy group offered notable exceptions to this general trend in that they had high strength and fair ductility in the as-quenched condition. The properties of these three alloys closely approached those obtained in the as-hot-rolled condition.

For most alloys, annealing at 1350 to 1550 F generally resulted in an improvement in properties over those obtained in the as-quenched condition. The tensile properties of these specimens were still decidedly inferior to those obtained in the as-hot-rolled condition, however. In particular, the ductilities of the annealed specimens were, in general, low for their particular strength levels. This may be due, in part, to the large grain size produced in the specimens by the 1700 F beta-izing treatment. The complex alloy containing 4 per cent iron (Heat WS59A) was again an exception, in that, after quenching and annealing at 1550 F, it had tensile properties approaching those obtained in the as-hot-rolled condition. Heats WS65A and WS63A also had fair ductility at high strength levels after the 1550 F anneal. The effect of cooling rate subsequent to the annealing treatment is uncertain, since both high and low strengths and ductilities were produced by each method of cooling.

It is evident that quenching from the beta field prior to a process-annealing treatment results in generally poorer tensile properties than can be obtained by annealing after hot rolling. Consequently, no further work on this type of heat treatment is contemplated. There is considerable evidence that annealing at certain temperatures in the alpha-beta field, followed by air cooling or water quenching, produces desirable properties in a wide variety of alloys, and this treatment will be investigated further. There is a strong indication that certain alloys, e.g., the complex Ti-Cr-Fe-Mn-Mo-V type, may be amenable to a solution treatment in the beta field followed by a rapid quench. This treatment will be studied in detail.

Age Hardening

The age-hardening mechanism in beta-stabilized titanium alloys is unique in two respects. First, it is believed that the hardening mechanism involves the precipitation of one allotropic form of the metal, the hexagonal alpha phase, from another allotropic form, the body-centered-cubic beta phase. Second, the amount and the composition of the beta phase in a given alloy can be varied over fairly wide ranges by suitable heat treatments. This coherency hardening is complicated by the more obvious transformation mechanisms by which the alpha phase assumes a martensitic or a Widmanstätten structure. A further complication is the precipitation of third phases, e.g., $TiCr_2$ from solid solution. It is apparent that any study of the age hardening of titanium alloys involves a very large number of variables.

Some work was done during the early part of this contract on solution treating and aging titanium alloys, using hardness as a criterion.(1) Tensile properties were obtained, however, only on specimens aged from the as-hot-rolled condition. Further work on the age-hardening characteristics of beta-stabilized alloys was considered mandatory.

For preliminary work, a series of ternary Ti-Cr-Mn alloys having manganese contents of 1 to 5 per cent and chromium levels of 3.5 and 5 per cent were used. The as-hot-rolled properties of these alloys were given in the previous summary report, Reference (2), pages 48 and 49. Small specimens of each alloy were solution treated in a dried-argon atmosphere for one hour at 1300 F (704 C), 1450 F (788 C), and 1600 F (871 C) and quenched in ice water. The solution-treated specimens were then aged at various temperatures in the range 375 to 825 F (190 to 440 C) for periods of time ranging from 1/2 to 48 hours. As-hot-rolled specimens were also included in these aging treatments. The hardness data obtained on the aged specimens are presented in the form of aging curves for the various solution treatments. The majority of these curves will be found in Appendix II. The curves for the Ti-5%Mn-2.5%Cr alloys, Figures 3 to 6, are included in this section as examples of the general behavior of these alloys.

It is evident from the data that all of the alloys used in this investigation age hardened appreciably after solution treating at 1300 F and above. Maximum hardening was attained in specimens solution treated at 1600 F, which is in the beta field for all of these alloys. Solution treating at 1450 F resulted in hardnesses only slightly lower than those attained after the 1600 F treatment. The microstructure of all the alloys after the 1450 F solution treatment consisted of about 50 to 70 per cent beta phase.

The 1300 F solution treatment resulted in much lower maximum hardnesses in the aged condition for all alloys. The rate of hardening was also appreciably lower than for the 1450 and 1600 F treatments. The microstructures of the specimens solution treated at 1300 F consisted of an alpha-beta mixture containing about 20 to 30 per cent of the beta phase. These specimens might be expected to have lower hardnesses as aged, because of the small amount of beta present. The slower rate of hardening at a given aging temperature is probably due to the higher alloy content of the beta in equilibrium with the alpha phase at the 1300 F solution temperature. This higher alloy content should result in greater stability of the beta, with consequent slower aging rates. Of the specimens aged from the as-hot-rolled condition, only those containing about 7 per cent or more of total alloy content age hardened appreciably. The materials having less than this total alloy content evidently did not retain enough of the beta phase to age harden to any great degree.

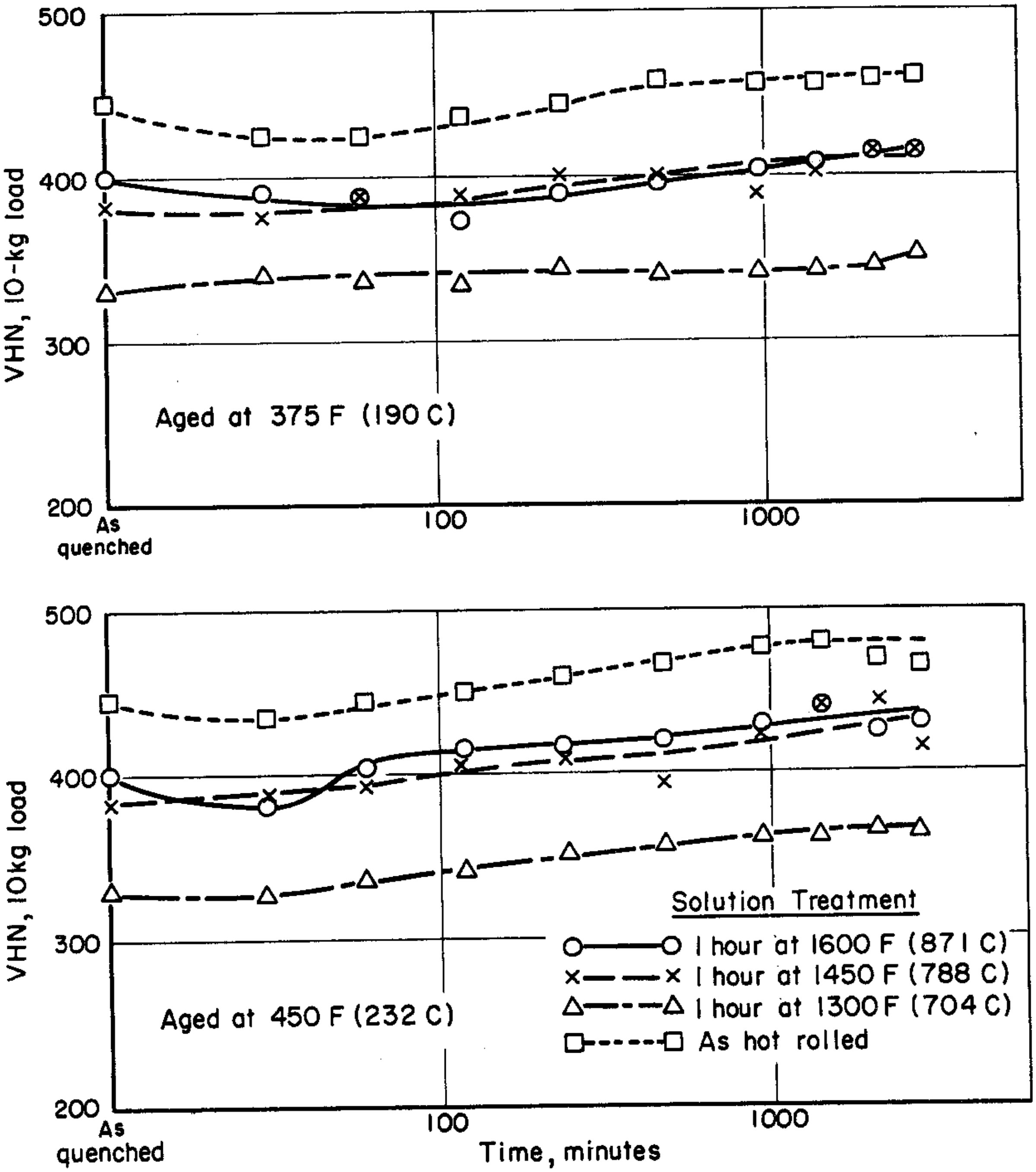


FIGURE 3. HARDNESS VERSUS AGING TIME FOR A Ti-5% Mn-2.5% Cr ALLOY SOLUTION TREATED AT VARIOUS TEMPERATURES

A-3249

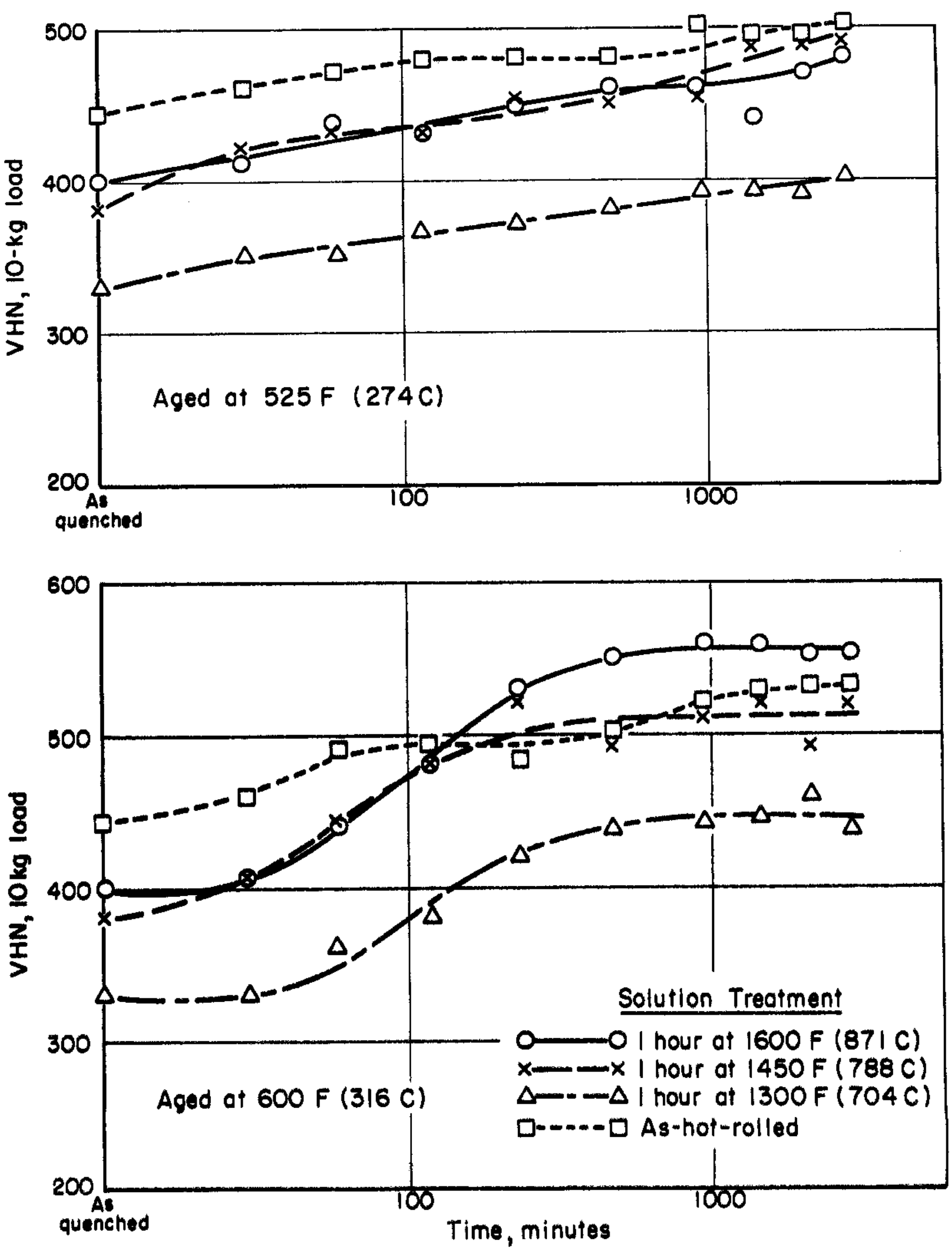


FIGURE 4. HARDNESS VERSUS AGING TIME FOR A Ti-5% Mn-2.5% Cr ALLOY SOLUTION TREATED AT VARIOUS TEMPERATURES

A-3250

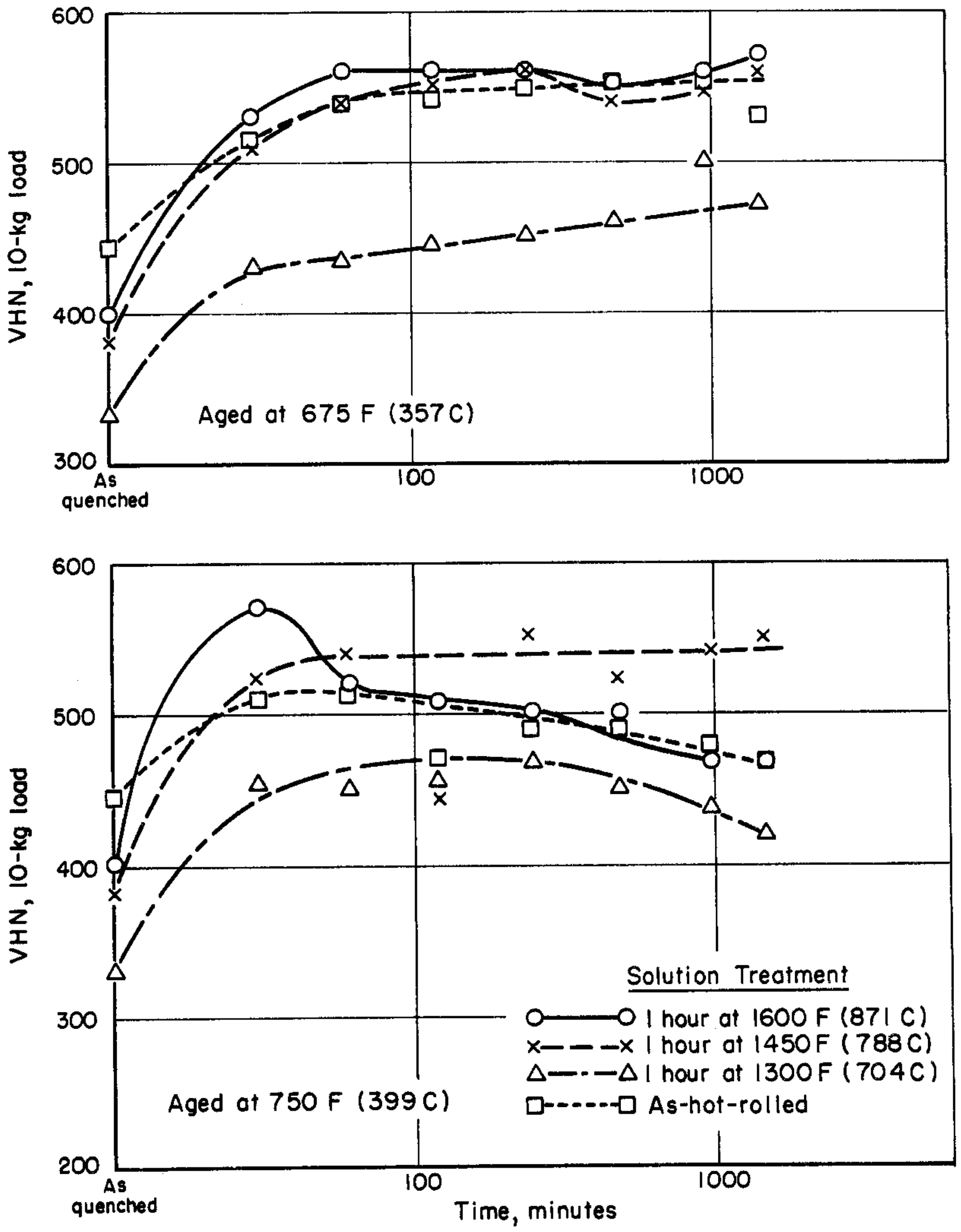


FIGURE 5. HARDNESS VERSUS AGING TIME FOR A Ti-5% Mn-2.5% Cr ALLOY SOLUTION TREATED AT VARIOUS TEMPERATURES

A-3251

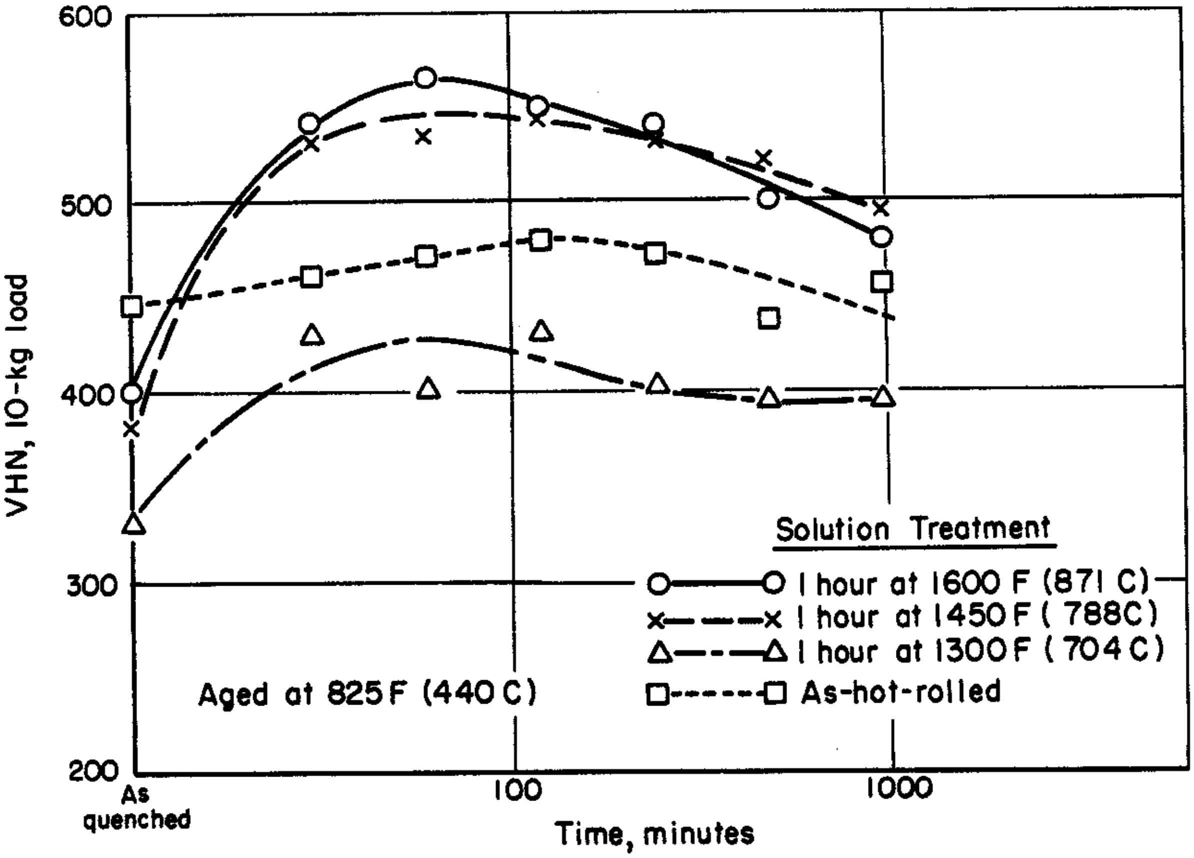


FIGURE 6. HARDNESS VERSUS AGING TIME FOR A Ti-5% Mn-2.5% Cr ALLOY SOLUTION TREATED AT VARIOUS TEMPERATURES

A-3249

Maximum hardness was attained in all alloys at an aging temperature of 675 or 750 F under all conditions of prior treatment. The rate of hardening at these temperatures is relatively fast, maximum hardness being reached in times of one hour or less. It is interesting to note also that the maximum hardness attained at these aging temperatures and the rate of hardening were essentially the same for all alloys solution treated at 1450 or 1600 F. Thus, it is indicated that alloys of relatively low total alloy content, e.g., the Ti-1%Mn-3.5%Cr alloy, may be heat treated to high strength levels. The maximum hardness for specimens in the as-hot-rolled condition or solution treated at 1300 F varied somewhat with alloy content. In general, higher alloy contents produced higher maximum hardnesses. Some hardening occurred in all alloys at aging temperatures as low as 450 F when they had been solution treated at 1450 or 1600 F. A few alloys exhibited appreciable hardening at an aging temperature of 375 F after the same solution treatment. All alloys overaged at a temperature of 825 F.

Some alloys solution treated at 1450 or 1600 F and aged at 375 or 450 F showed very erratic hardness results. Microscopic examination of these specimens revealed gross differences in microstructure, apparently from alloy segregation.

On the basis of the foregoing data, specimens of a second series of Ti-Cr-Mn alloys were solution treated at 1300 and 1450 F and aged at 600, 675, and 825 F to give, respectively, an underaged, fully aged, and overaged condition. An aging time of 4 hours was used at all temperatures. The as-hot-rolled properties for this group of alloys may be found in Table 4 of this report. Solution and aging treatments were carried out in a dried-argon atmosphere to prevent surface contamination. Complete results of tensile and hardness tests made on these specimens are given in Table 21.

Specimens of all alloys as solution treated at 1300 F had medium strength and very good ductility. Aging such specimens at 600 F resulted in rather large increases in tensile strength, with consequent sharp reductions in ductility. Aging at 675 F increased the strength of most of the alloys still further and, in most cases, reduced the ductility to a very low value. In many instances, the strength of the specimens aged at 675 F were higher than those obtained in the as-hot-rolled condition, but, of course, their ductilities were much lower. Specimens treated at 825 F were, in general, overaged, as indicated by their lower tensile strengths and hardnesses. The ductility of the latter specimens was somewhat greater than that of the specimens aged at 675 F, but this increase in ductility was not commensurate with the rather large drop in tensile strength.

The specimens quenched from 1450 F had rather unusual properties. With the exception of the 4 per cent manganese alloys containing 3 or 3.5 per cent chromium, the tensile strengths and hardnesses of all specimens fell within a relatively narrow range. The ductility, however, varied widely from 1 per cent elongation for the Ti-2.75%Mn-3%Cr alloy to 24 per cent elongation for the Ti-3.5%Mn-3.5%Cr alloy. The variation in ductility was

TABLE 21. TENSILE PROPERTIES AND HARDNESSES OF Ti-Cr-Mn ALLOYS SOLUTION TREATED AND AGED

Heat No.	Intended Composition, %	As Quenched			Aged 4 Hours at 600 F (316 C)(2)			Aged 4 Hours at 675 F (358 C)(2)			Aged 4 Hours at 825 F (441 C)(2)		
		Elongation, % in 1 inch	Tensile Strength, psi(3)	VHN	Elongation, % in 1 inch	Tensile Strength, psi	VHN	Elongation, % in 1 inch	Tensile Strength, psi	VHN	Elongation, % in 1 inch	Tensile Strength, psi	VHN
					1 Hour at 1300 F (704 C), Ice-Water Quenched(1)								
WS77A	2.75Mn–2.0Cr	20.0	125,600	283	5.0	154,000	338	2.0	164,100	372	5.0	141,200	319
WS78A	2.75Mn–3.0Cr	12.0	134,100	292	2.0	175,900	406	1.5	172,100	409	5.0	152,500	342
WS79A	2.75Mn–3.5Cr	16.0	140,500	309	2.0	178,400	397	1.5	203,600	455	4.5	159,300	351
WS80A	2.75Mn–4.0Cr	11.5	148,200	309	0.5	172,900	425	1.0	178,900	464	2.0	161,200	373
WS86A	3.5Mn–1.5Cr	18.0	127,200	281	6.0	155,700	345	1.5	143,600	370	8.0	146,000	317
WS96A	3.5Mn–2.0Cr	20.0	139,100	289	6.0	167,800	360	2.0	182,400	409	5.0	156,000	339
WS89A	3.5Mn–2.5Cr	20.0	137,700	297	3.0	173,700	376	2.0	190,500	418	3.0	157,500	342
WS88A	3.5Mn–3.0Cr	16.0	142,200	302	2.0	184,400	383	1.0	117,600	450	4.0	169,500	357
WS90A	3.5Mn–3.5Cr	15.0	146,700	314	1.0	187,900	421	1.5	195,300	456	3.0	174,700	354
WS99A	4.0Mn–2.0Cr	21.0	125,800	275	4.5	155,100	348	2.0	162,700	361	7.0	138,700	302
WS101A	4.0Mn–2.5Cr	22.0	134,400	287	8.0	167,700	360	3.0	182,500	388	12.0	154,400	333
WS102A	4.0Mn–3.0Cr	14.0	157,100	336	0.0	179,700	438	2.0	154,300	491	1.0	190,600	417
WS103A	4.0Mn–3.5Cr	12.5	158,200	331	–	–	–	(4)	(4)	512	1.0	171,200	442
					1 Hour at 1450 F (788 C), Ice-Water Quenched(1)								
WS77A	2.75Mn–2.0Cr	17.0	144,900	345	1.0	194,900	468	2.0	177,300	344	5.0	145,800	297
WS78A	2.75Mn–3.0Cr	1.0	132,900	373	–	(6)	534	1.5	180,500	381	3.0	158,900	322
WS79A	2.75Mn–3.5Cr	4.0	149,400	361	–	(6)	530	–	–	409	5.0	166,700	333
WS80A	2.75Mn–4.0Cr	5.0	131,600	357	–	(6)	530	0.0	176,000	413	2.0	171,200	332
WS86A	3.5Mn–1.5Cr	7.5	140,700	368	1.0	164,400	485	2.0	173,700	360	7.0	147,600	302
WS96A	3.5Mn–2.0Cr	17.0	147,600	351	0.0	176,000	498	2.0	191,700	370	7.5	160,000	345
WS89A	3.5Mn–2.5Cr	14.0	145,100	354	–	(6)	514	1.0	162,300	390	5.0	164,200	359
WS88A	3.5Mn–3.0Cr	12.0	140,100	361	–	(6)	514	1.0	172,600	401	4.5	174,400	376
WS90A	3.5Mn–3.5Cr	24.0	145,800	357	–	(6)	507	0.5	188,200	430	5.0	182,400	390
WS99A	4.0Mn–2.0Cr	11.5	146,200	358	1.0	158,200	458	0.5	109,200	327	7.0	141,300	309
WS101A	4.0Mn–2.5Cr	19.0	144,700	363	–	(6)	508	1.0	170,200	352	7.0	158,700	330
WS102A	4.0Mn–3.0Cr	2.0	174,300	373	–	(6)	523	1.0	131,800	459	3.0	198,900	421
WS103A	4.0Mn–3.5Cr	(5)	(5)	351	–	(6)	514	–	–	464	1.5	155,100	405

(1) Heated in a dried-argon atmosphere.

(2) Heated in a dried-argon atmosphere and water quenched.

(3) Substandard specimen, 0.375 × 0.064 × 5 inches with a reduced section 0.250 × 0.064 × 1-1/4 inches.

(4) Defective specimen.

(5) Specimen broke in grips.

(6) Specimens too hard to grip.

first believed to be a function of the relative amounts and distribution of the alpha and beta phases. Microscopic examination of these specimens, however, revealed no consistent correlation between visible microstructure and tensile ductility. Thus, it was indicated that some submicroscopic reaction was responsible, and this is undoubtedly the coherency hardening of the beta phase.

It is evident from the data in Table 21 that the desired conditions of aging were not attained in the specimens solution treated at 1450 F. Those aged at 600 F had very high hardnesses and were, for the most part, extremely brittle. Aging at the higher temperatures resulted in a progressive drop in hardness and tensile strength. The ductilities of the specimens aged at 825 F were low for their respective strength levels. With the exception of the Ti-4%Mn-3.5%Cr alloy, the specimens solution treated at 1450 F and aged at 825 F had somewhat higher strengths with equal or higher ductility than those solution treated at 1300 F and given the same aging treatment.

The aging reaction apparently proceeds at an appreciably slower rate after the 1300 F solution treatment than after quenching from 1450 F. Thus, the hardnesses of specimens treated at 1300 F increased when the aging temperature was increased from 600 to 675 F, while the hardness of the specimens quenched from 1450 F decreased under the same condition, indicating overaging. The high strength attained by a number of the alloys solution treated at 1300 F is rather surprising, in view of the fact that their microstructures were estimated to contain only about 20 to 30 per cent beta phase.

Lack of sufficient stock of the Ti-Cr-Mn alloys prevented investigation of the aging characteristics of these compositions when solution treated in the beta field. However, a very limited investigation was carried out on the group of complex alloys described earlier in this report (Table 12). The available stock had been annealed for 4 hours at 1250 F, furnace cooled, and cold rolled to 25 per cent reduction. Specimens of the cold-rolled stock were heated for 1/2 hour at 1700 F and quenched in ice water. Quenched specimens of each alloy were then aged for one hour at 932 F (500 C) and water quenched. Results of tensile and hardness tests made on as-quenched and quenched and aged specimens are given in Table 22.

The tensile properties of the as-quenched specimens were very poor, with the exception of the heats containing 4 per cent manganese (WS65A) or 3 and 4 per cent iron (WS42A and WS59A). This is in agreement with the data shown in Table 19 for the same alloys given the same solution treatment from the as-hot-rolled condition. The cold reduction given the current specimens did not improve the as-quenched properties of these alloys. Aging at 932 F produced very high tensile strengths in a majority of the alloys; in many instances, they were actually higher than those obtained in the as-hot-rolled condition. The ductility of all aged specimens was disproportionately low.

TABLE 22. TENSILE PROPERTIES AND HARDNESSES OF COMPLEX Ti-Cr-Fe-Mn-Mo-V ALLOYS SOLUTION TREATED AND AGED AFTER 25% COLD REDUCTION(1)

		Heated 1/2 Hour at 1700 F (927 C) and Quenched in Water (0 C)(2)					
		As Quenched			Aged 1 Hour at 932 F (500 C)(2)		
Heat No.	Intended Composition, %	Ultimate Tensile Strength, psi(3)	Elongation, % in 1 inch	VHN	Ultimate Tensile Strength, psi(3)	Elongation, % in 1 inch	VHN
WQ26A	1Cr–1Fe–1Mn–1Mo	153,700	2.0	429	161,000	1.0	437
WQ25A	1Cr–1Fe–1Mn–1V	150,000	1.0	483	166,300	1.0	412
WQ27A	1Cr–1Fe–1Mo–1V	157,000	1.5	405	164,000	2.0	390
WQ23A	1Cr–1Mn–1Mo–1V	142,900	2.0	370	145,300	0.5	346
WS40A	1Fe–1Mn–1Mo–1V	160,300	1.0	390	172,700	3.0	387
WS39A	1Cr–1Fe–1Mn–1Mo–1V	141,000	1.0	401	–	–	–
WS36A	1Cr–1Fe–1Mn–1Mo–2V	121,000	0.5	435	186,500	1.5	370
WS60A	1Cr–1Fe–1Mn–1Mo–3V	126,000	2.0	450	202,000	1.5	409
WS54A	1Cr–1Fe–1Mn–1Mo–4V	114,000	1.0	425	–	–	–
WS37A	1Cr–1Fe–1Mn–2Mo–1V	142,600	2.0	387	186,500	1.0	376
WS43A	1Cr–1Fe–1Mn–3Mo–1V	128,900	1.5	424	195,100	0.5	405
WS52A	1Cr–1Fe–1Mn–4Mo–1V	145,500	2.5	401	208,000	2.0	413
WS49A	1Cr–1Fe–2Mn–1Mo–1V	99,200	0.0	430	197,300	3.0	383
WS53A	1Cr–1Fe–3Mn–1Mo–1V	137,900	1.0	429	206,900	1.0	405
WS65A	1Cr–1Fe–4Mn–1Mo–1V	193,500	4.0	401	214,900	1.0	446
WS41A	1Cr–2Fe–1Mn–1Mo–1V	125,600	1.0	451	193,400	1.0	409
WS42A	1Cr–3Fe–1Mn–1Mo–1V	198,300	2.5	409	161,000	0.0	421
WS59A	1Cr–4Fe–1Mn–1Mo–1V	201,500	4.0	413	81,400	0.0	442
WS38A	2Cr–1Fe–1Mn–1Mo–1V	158,000	3.5	397	150,700	1.0	429
WQ34A	3Cr–1Fe–1Mn–1Mo–1V	124,600	2.5	410	207,700	2.0	403
WS63A	4Cr–1Fe–1Mn–1Mo–1V	148,400	4.5	414	218,500	2.0	401

(1) Prior to cold rolling, specimens were heated for 4 hours at 1250 F (677 C) in an argon atmosphere and furnace cooled.
(2) Heated in a dried-argon atmosphere.
(3) Substandard specimen 0.375 × 0.064 × 5 inches with a reduced section 0.250 × 0.064 × 1-1/4 inches.

It is evident from the foregoing results that solution-treating and age-hardening treatments hold considerable promise for possible commercial application. In general, the most promising treatment evolved thus far consists of solution treating at 1300 F. This may be followed by aging at 675 F for 4 hours to produce high strength. The solution treatment produced high ductility in all alloys to which it was applied. In fact, this good ductility was often accompanied by fairly high strength. The forming characteristics in this condition would probably be good. Following forming, several compositions could be age hardened to a tensile strength of the order of 200,000 psi. Ductility would be drastically reduced by such a treatment, but, in many applications, high ductility is not required. In a few cases, the 1400 F solution treatment produced excellent combinations of ductility and strength. For example, the Ti-3.5%Mn-3.5%Cr alloy had 29 per cent elongation at a strength level of 145,800 psi. A more thorough investigation of age hardening is planned on larger heats of selected high-strength alloys.

Isothermal Transformation

A detailed study of the effects of isothermal transformation on the microstructure and hardness of binary titanium-chromium alloys will be described later in this report. Only a very limited investigation of the mechanical properties of isothermally transformed exploratory alloys has been made. This consisted of applying arbitrarily selected isothermal treatments to the complex alloys whose as-hot-rolled properties were described in Table 12.

Specimens of the complex alloys were heated for 1/2 hour at 1700 F in a dried-argon atmosphere and quenched into a molten lead bath held at 750 F or 932 F. After one hour in the lead bath, the specimens were quenched in ice water at 0 C. Tensile properties and hardnesses of the heat-treated specimens are shown in Table 23. As-hot-rolled properties of these alloys are repeated in the table for comparison purposes.

The specimens isothermally transformed at 750 F were extremely brittle and could not be subjected to tensile testing. This is rather surprising in view of the relatively low hardnesses of the alloys containing four alloying elements. Transformation at 932 F resulted in hardnesses and tensile strengths comparable, in many instances, to those obtained in the as-hot-rolled condition. The ductilities of all specimens so treated were very low, however. The alloys containing 3 per cent iron or 2 per cent chromium obviously did not attain a true tensile strength, since their strengths are much lower than would be expected from their respective hardnesses.

Comparison of the properties of the specimens transformed at 932 F with those given in Table 22 for the same alloys quenched to 0 C from 1700 F and reheated to 932 F for one hour shows that, in general, the latter treatment produced higher tensile strengths with about the same or slightly lower

TABLE 23. TENSILE PROPERTIES AND HARDNESSES OF COMPLEX Ti-Cr-Fe-Mn-Mo-V ALLOYS AS HOT ROLLED AND ISOTHERMALLY TRANSFORMED AT 750 AND 932 F

					Heated 1/2 Hour at 1700 F (927 C) in Argon and:					
		As Hot Rolled at 1450 F			Transformed 1 Hour at 750 F (400 C)(1)			Transformed 1 Hour at 932 F (500 C)(1)		
Heat No.	Intended Composition, %	Ultimate Tensile Strength, psi(2)	Elongation, % in 1 inch	VHN	Ultimate Tensile Strength, psi	Elongation, % in 1 inch	VHN	Ultimate Tensile Strength, psi(2)	Elongation, % in 1 inch	VHN
WQ26A	1Cr–1Fe–1Mn–1Mo	156,200	10.5	327	All specimens too brittle to test		416	160,300	3.5	354
WQ25A	1Cr–1Fe–1Mn–1V	158,800	6.0	330			–	147,100	2.0	357
WQ27A	1Cr–1Fe–1Mo–1V	151,600	10.5	307			376	142,100	0.5	378
WQ23A	1Cr–1Mn–1Mo–1V	151,500	14.0	326			383	146,600	2.0	349
WS40A	1Fe–1Mn–1Mo–1V	141,000	15.5	304			342	138,200	2.0	339
WS39A	1Cr–1Fe–1Mn–1Mo–1V	188,300	2.0*	375			–	179,500	4.0	397
WS36A	1Cr–1Fe–1Mn–1Mo–2V	188,500	7.0	361			478	–	–	–
WS60A	1Cr–1Fe–1Mn–1Mo–3V	181,600	6.0	370			–	–	–	–
WS54A	1Cr–1Fe–1Mn–1Mo–4V	192,400	4.0	369			–	–	–	–
WS37A	1Cr–1Fe–1Mn–2Mo–1V	179,600	7.0	359			451	179,200	3.5	373
WS43A	1Cr–1Fe–1Mn–3Mo–1V	198,200	5.0	390			536	186,500	2.5	408
WS52A	1Cr–1Fe–1Mn–4Mo–1V	197,600	4.0	370			542	198,300	1.0	418
WS49A	1Cr–1Fe–2Mn–1Mo–1V	179,900	10.5	363			537	186,600	3.5	393
WS53A	1Cr–1Fe–3Mn–1Mo–1V	196,000	10.5	397			–	198,300	2.0	393
WS65A	1Cr–1Fe–4Mn–1Mo–1V	217,500	1.75*	401			530	183,500	1.5	413
WS41A	1Cr–2Fe–1Mn–1Mo–1V	182,500	7.5	363			518	175,200	1.0	378
WS42A	1Cr–3Fe–1Mn–1Mo–1V	220,300	3.0*	442			–	133,300	1.5	404
WS59A	1Cr–4Fe–1Mn–1Mo–1V	211,900	5.0	383			554	169,800	1.5	422
WS38A	2Cr–1Fe–1Mn–1Mo–1V	212,000	3.5	397			542	148,000	0.5	416
WQ34A	3Cr–1Fe–1Mn–1Mo–1V	203,200	2.5	408			542	187,000	2.0	380
WS63A	4Cr–1Fe–1Mn–1Mo–1V	200,000	2.0*	401			536	198,500	3.0	400

(1) Transformation carried out in a well deoxidized lead bath.

(2) Substandard specimen 0.375 × 0.064 × 5 inches with a reduced section 0.250 × 0.064 × 1-1/4 inches.

* Broke outside gage marks. Uniform elongation recorded.

ductility. Thus, it was indicated that the solution-treating and aging heat treatment was superior to direct isothermal transformation in producing high strengths in these alloys. Further investigation of isothermal heat treatments is planned, however, and this will include the study of lower transformation temperatures.

Effect of Quenching Media on Hardness of Titanium Alloys

Experiments were carried out on binary Ti-Cr and Ti-Mn alloys to establish the effects of changing the cooling rate on their hardness. Alloys having compositions ranging from 3 to 15 per cent of manganese or chromium were used. Small specimens of each alloy were heated at 1742 F (950 C) in an argon atmosphere for 1/2 hour and quenched in various media to obtain a range of cooling rates. The quenching media, in order of decreasing cooling capacity were: (1) iced 10% brine, (2) ice water (0 C), (3) medium viscosity oil, (4) liquid nitrogen, and (5) still air. The quenched specimens were mounted in a room-temperature-setting resin and subjected to Vickers hardness tests. Complete results of these tests are presented in Figures 7 and 8.

Since the specimens employed in these experiments were only 1/16 inch thick, the cooling rates obtained were rapid in all cases. However, it is evident from Figures 7 and 8 that large differences in hardness were produced in some of the alloys by the various quenching treatments. Since the iced 10% brine quench resulted in the lowest hardnesses in most specimens, the hardness results for this treatment were used as a basis for comparison. Hardness variations of 25 VHN on a single specimen were common, so only hardness differences greater than this amount were considered significant.

The curves indicate that a wide variation in hardness and, presumably, in other physical properties, may be obtained by adjusting the cooling rates of suitable alloy compositions. The coherency hardening mechanism suggested earlier in this report to explain the variation of as-hot-rolled properties in binary Ti-Cr and Ti-Mn alloys also applies to these tests. In the faster cooling media (brine and water), the cooling rate is sufficiently rapid to suppress the formation of the alpha phase except by a shear-type transformation. At the slower cooling rates, however, alpha nuclei form which are coherent with the beta matrix and produce hardening. The number of the latter nuclei apparently increases with increasing alloy content up to about 8 per cent chromium or 6 per cent manganese. At high alloy contents, the number again decreases, because of the increased stability of the beta phase, until at about 12 to 15 per cent chromium or manganese little or no hardening of this type occurs. Thus, at 15 per cent chromium, the hardnesses of all specimens were the same regardless of cooling rate. The 15 per cent manganese alloy had essentially the same hardness for all quenching media except still air, which produced a lower hardness. This lower

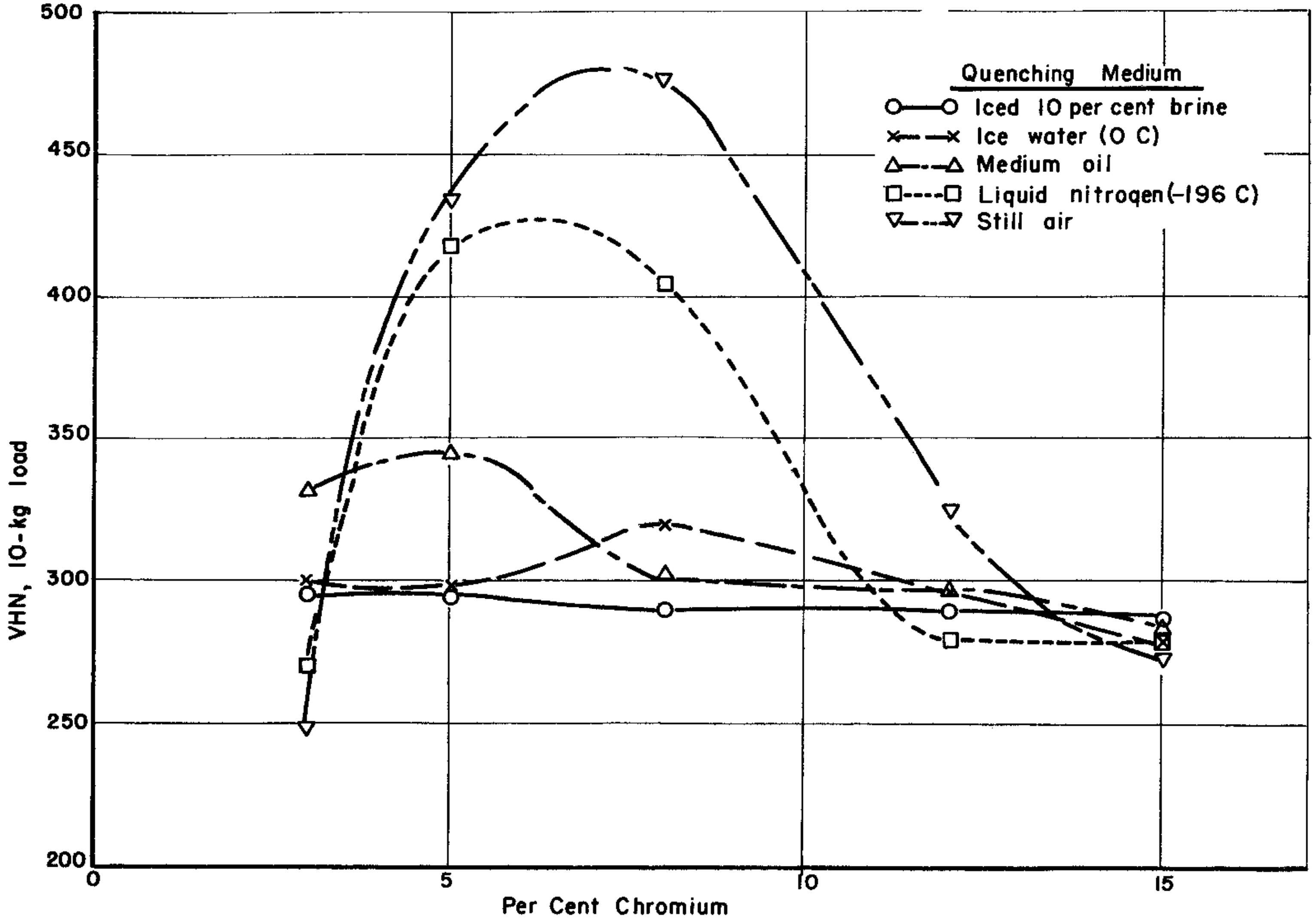


FIGURE 7. EFFECT OF QUENCHING MEDIA (COOLING RATE) ON HARDNESS OF BINARY Ti-Cr ALLOYS QUENCHED FROM 1700F (927C)
Specimen Size : $\frac{1}{2}$ x $\frac{3}{8}$ x $\frac{1}{16}$ inch

A-3253

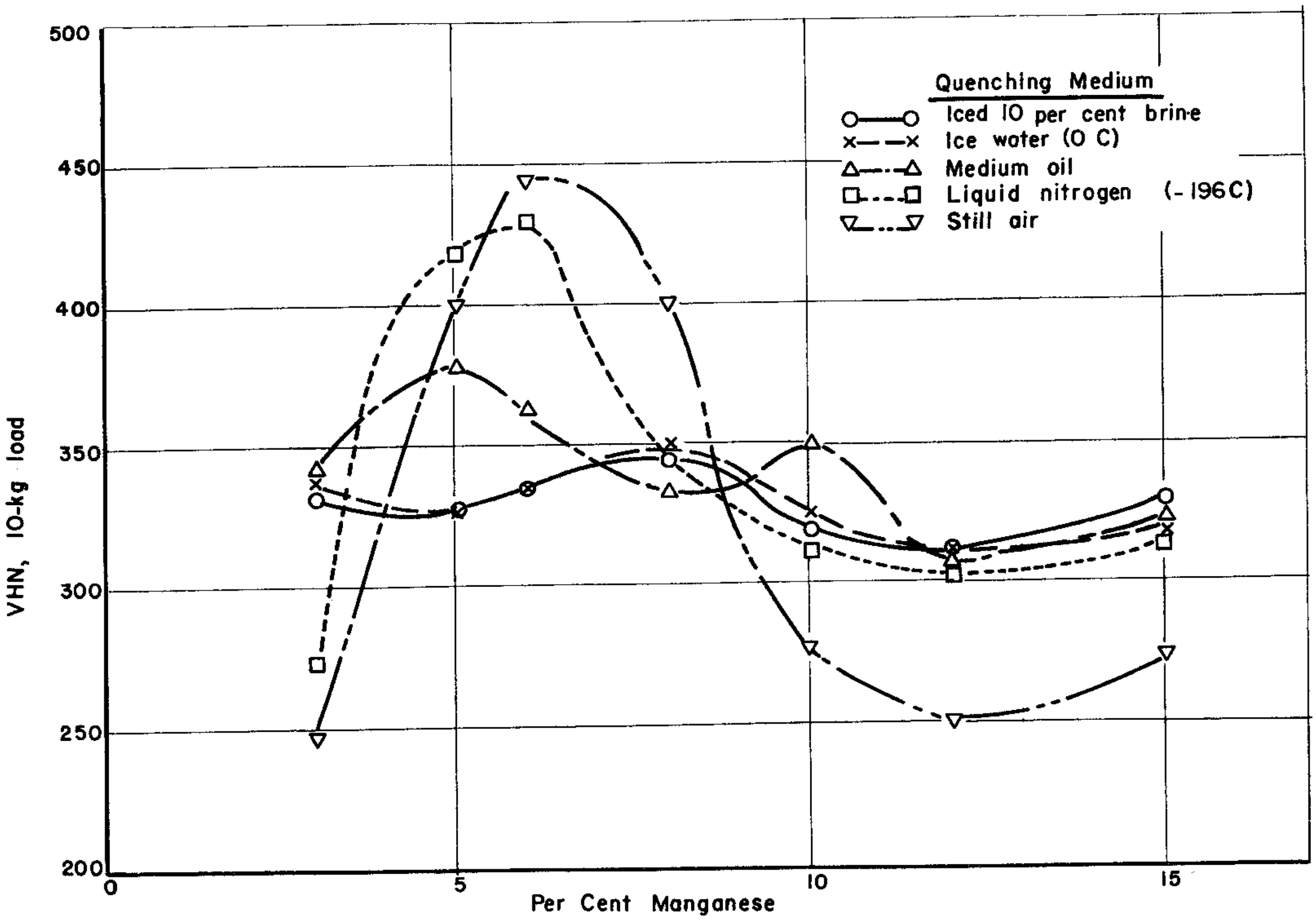


FIGURE 8. EFFECT OF QUENCHING MEDIA (COOLING RATE) ON HARDNESS OF BINARY Ti-Mn ALLOYS QUENCHED FROM 1700 F (927C)
Specimen Size - $\frac{1}{2} \times \frac{3}{8} \times \frac{1}{16}$ inches

A-3254

hardness of the air-cooled 10 to 15 per cent manganese alloys is rather puzzling. The absence of quenching stresses, such as would be found in the liquid-quenched specimens, might result in slightly lower hardnesses in these specimens, but would not be expected to account for the relatively large differences shown in Figure 8. It is conceivable that the alloys containing 10 and 15 per cent manganese are sufficiently sluggish to prevent hardening entirely, whereas in all the other alloys a slight amount of hardening occurs regardless of the cooling rate.

Low-Temperature Age Hardening of Titanium Alloys

Low-temperature age hardening of complex alloys was mentioned earlier in this report in the section entitled "Process Annealing Heat Treatments". It was stated that mounting specimens in Bakelite prior to determining their hardnesses resulted in hardness increases in some instances. This section of the report contains the data upon which this conclusion was based. Since further investigation of this phenomenon seemed warranted, the specimens used in the investigation of the effect of quenching media on hardness, described in the previous section, were also subjected to low-temperature aging tests.

Complex Alloys

Small specimens cut from tensile specimens of complex alloys quenched from 1700 F (Table 19) were mounted in a room-temperature-setting resin and given Vickers hardness tests. The mounted specimens were then heated for one hour at 212 F (100 C) by immersing them in boiling water, and again given hardness tests. Average results of these tests, along with the hardnesses taken from Table 19 for specimens cut from the same tensile bars and mounted in Bakelite, are presented in Table 24.

It is evident that some hardening occurred in all of the specimens heated at 212 F. Hardness increases of 20-25 VHN were common, and in the case of Heat WS54A, which contains 4 per cent vanadium, an increase of about 60 VHN was obtained.

The procedure for mounting specimens in Bakelite involved heating to a temperature of about 320 F (160 C) over a period of 15 minutes and rapidly cooling to room temperature. Comparison of the hardnesses of specimens so mounted with those of similar specimens mounted at room temperature shows a somewhat erratic trend. The hardnesses of the Bakelite-mounted specimens were somewhat higher in the majority of the alloys, however. Thus, it is indicated that some age hardening may take place during the brief heating period employed in this procedure.

TABLE 24. EFFECT OF LOW-TEMPERATURE AGING ON HARDNESS OF COMPLEX Ti-Cr-Mn-Mo-V ALLOYS QUENCHED FROM 1700 F (927 C)

		VHN		
Heat No.	Intended Composition, %	As Mounted in Room-Temperature Mount	Heated 1 Hour at 212 F	Bakelite Mount
WQ26A	1Cr-1Fe-1Mn-1Mo	430	450	437
WQ25A	1Cr-1Fe-1Mn-1V	388	427	412
WQ27A	1Cr-1Fe-1Mo-1V	385	413	390
WQ23A	1Cr-1Mn-1Mo-1V	374	405	346
WS40A	1Fe-1Mn-1Mo-1V	373	399	387
WS39A	1Cr-1Fe-1Mn-1Mo-1V	392	413	421
WS36A	1Cr-1Fe-1Mn-1Mo-2V	414	433	380
WS60A	1Cr-1Fe-1Mn-1Mo-3V	436	450	426
WS54A	1Cr-1Fe-1Mn-1Mo-4V	401	462	427
WS37A	1Cr-1Fe-1Mn-2Mo-1V	348	360	385
WS43A	1Cr-1Fe-1Mn-3Mo-1V	421	444	420
WS52A	1Cr-1Fe-1Mn-4Mo-1V	362	387	386
WS49A	1Cr-1Fe-2Mn-1Mo-1V	442	478	461
WS53A	1Cr-1Fe-3Mn-1Mo-1V	433	450	433
WS65A	1Cr-1Fe-4Mn-1Mo-1V	400	413	409
WS41A	1Cr-2Fe-1Mn-1Mo-1V	449	493	478
WS42A	1Cr-3Fe-1Mn-1Mo-1V	410	455	409
WS59A	1Cr-4Fe-1Mn-1Mo-1V	421	462	437
WS38A	2Cr-1Fe-1Mn-1Mo-1V	407	432	436
WQ34A	3Cr-1Fe-1Mn-1Mo-1V	370	409	426
WS63A	4Cr-1Fe-1Mn-1Mo-1V	368	394	400

Binary Ti-Cr and Ti-Mn Alloys

The specimens used in this investigation were those described in the preceding section on the effect of quenching media on hardness. They had been quenched in the various media from 1742 F and given Vickers tests after mounting in a room-temperature-setting plastic. The aging treatment given these specimens consisted of immersion in boiling water for 45 minutes. Specimens of the Ti-Cr series which had been quenched in ice water or liquid nitrogen from 1742 F were also mounted in Bakelite. Averaged results of Vickers hardness tests made on all specimens are presented in Figures 9 through 13.

Some hardening occurred in all of the alloys during the aging treatment at 212 F. Hardness increases in the specimens quenched in brine, water, or oil were moderate, ranging from 10 to 50 VHN with an average of about 25 VHN (Figures 9, 10, and 11). Specimens of some of the alloys quenched in liquid nitrogen or air cooled showed, however, hardness increases of more than 100 VHN (Figures 12 and 13). The 5 and 8 per cent chromium and the 5, 6, and 8 per cent manganese alloys increased in hardness much more than alloys having higher or lower alloy contents. With the exception of the water-quenched Ti-8%Cr alloy, the hardnesses of the Ti-Cr specimens mounted in Bakelite agreed very well with those of the specimens aged at 212 F. Thus, it is indicated that large errors in hardness may be introduced by the treatment employed in the Bakelite-mounting procedure.

In determining the significance of these results, it must be remembered that, in most instances, the alloys which age hardened appreciably were in their most unstable condition in that they had been cooled, rather rapidly, from the beta-phase region. Thus, any retained beta present in their microstructures would have the lowest alloy content attainable in a given alloy; whereas heating at temperatures in the alpha-beta field would result in higher alloy content and greater stability of the beta phase.

As-Hot-Rolled Exploratory Alloys

The hardnesses of most of the alloys tested thus far under this contract have been determined on metallographic specimens mounted in Bakelite. In view of the foregoing results, the possibility of appreciable errors in hardness values reported for alloys in the as-hot-rolled condition could not be ignored. To determine the effect of low-temperature aging on the hardness of alloys in the as-hot-rolled condition, the alloys whose properties are given in Tables 5, 7, 8, and 9 were given tests similar to those previously described; i.e., hardnesses were determined on as-hot-rolled specimens before and after aging for 45 minutes at 212 F. Results are given in Table 25.

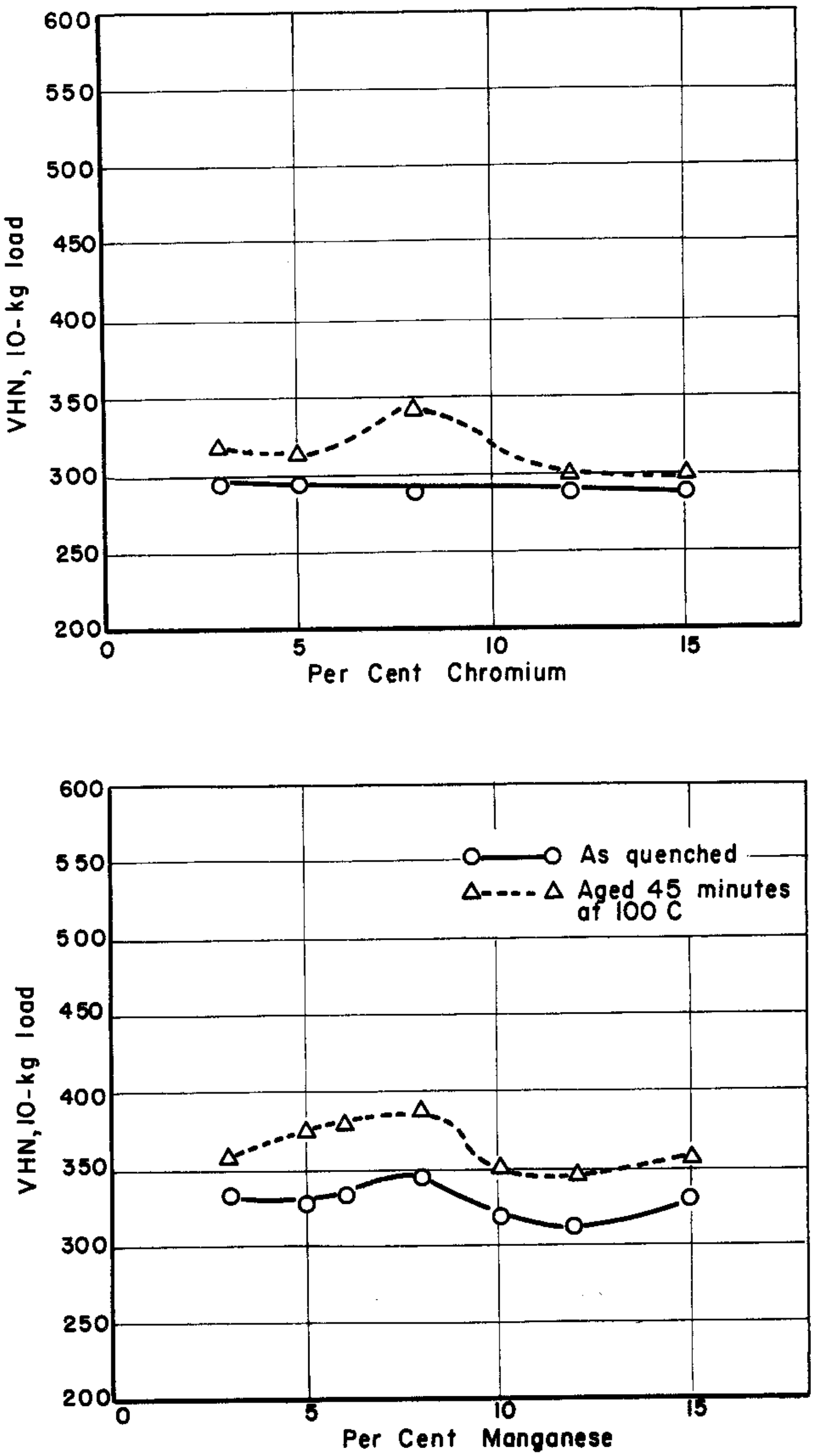
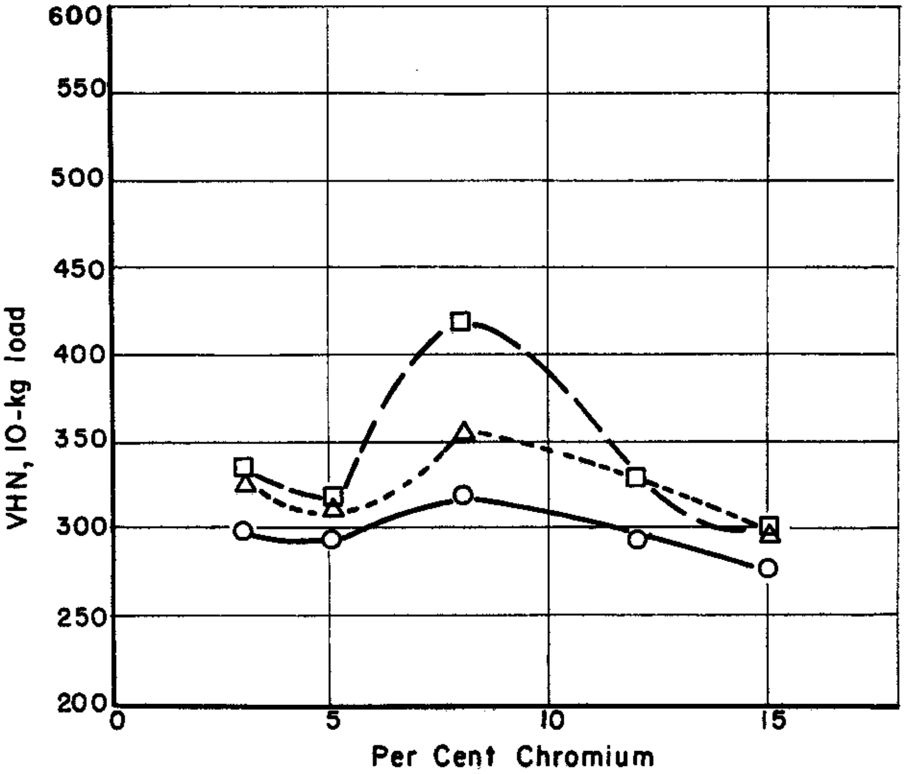


FIGURE 9. EFFECT OF LOW-TEMPERATURE AGING ON HARDNESS OF BINARY Ti-Cr AND Ti-Mn ALLOYS QUENCHED IN ICED 10 PER CENT BRINE FROM 1742 F (950 C)

A-3255



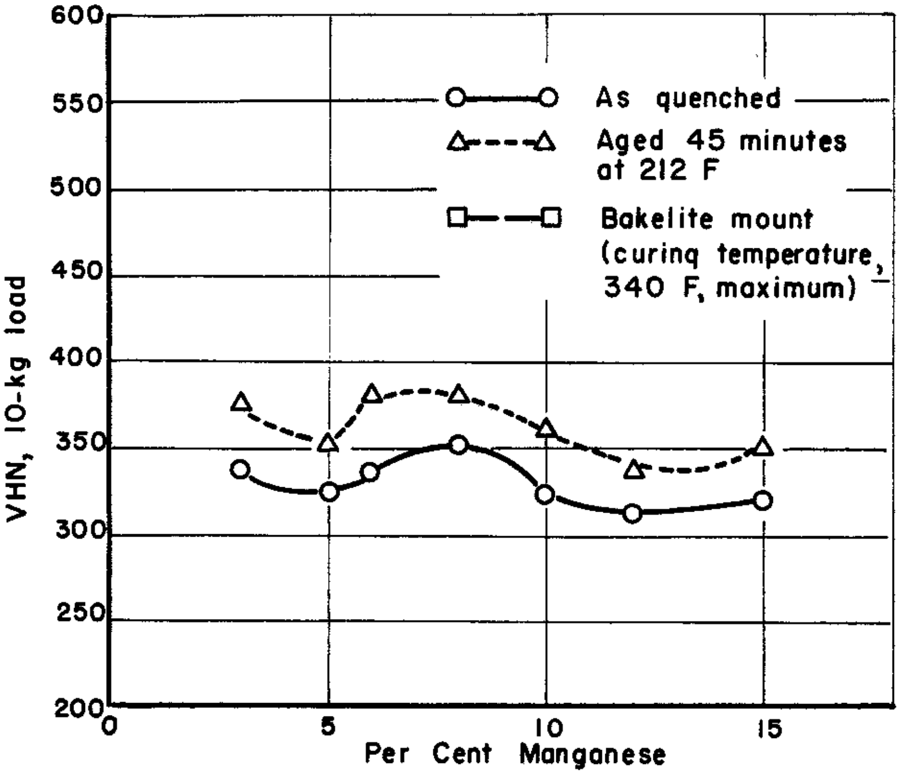
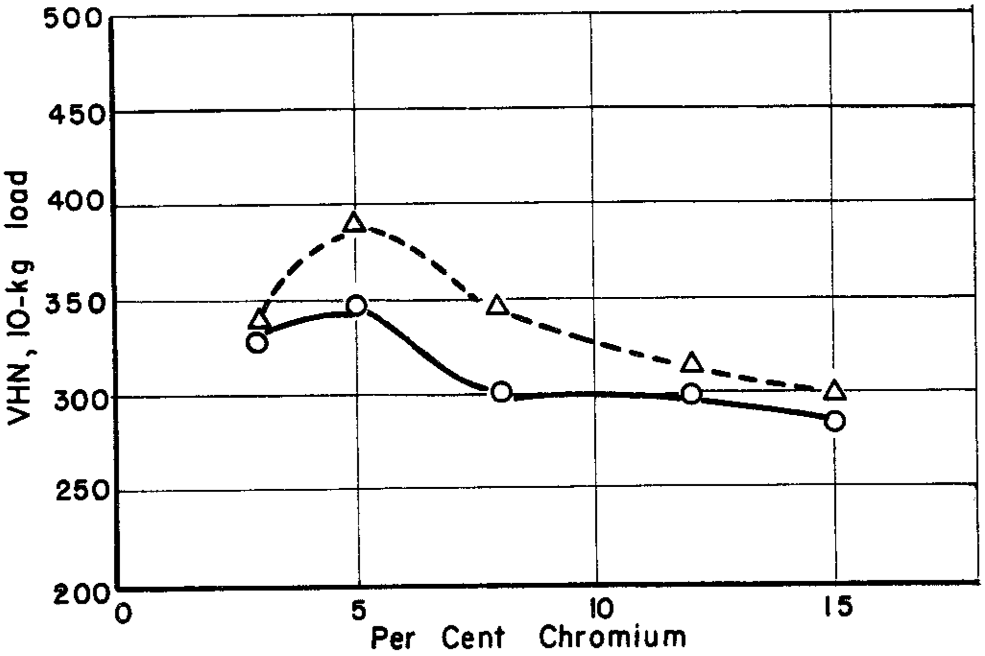


FIGURE 10. EFFECT OF LOW-TEMPERATURE AGING ON HARDNESS OF BINARY Ti-Cr AND Ti-Mn ALLOYS QUENCHED IN ICE WATER (0 C) FROM 1742 F (950 C)

A-3256



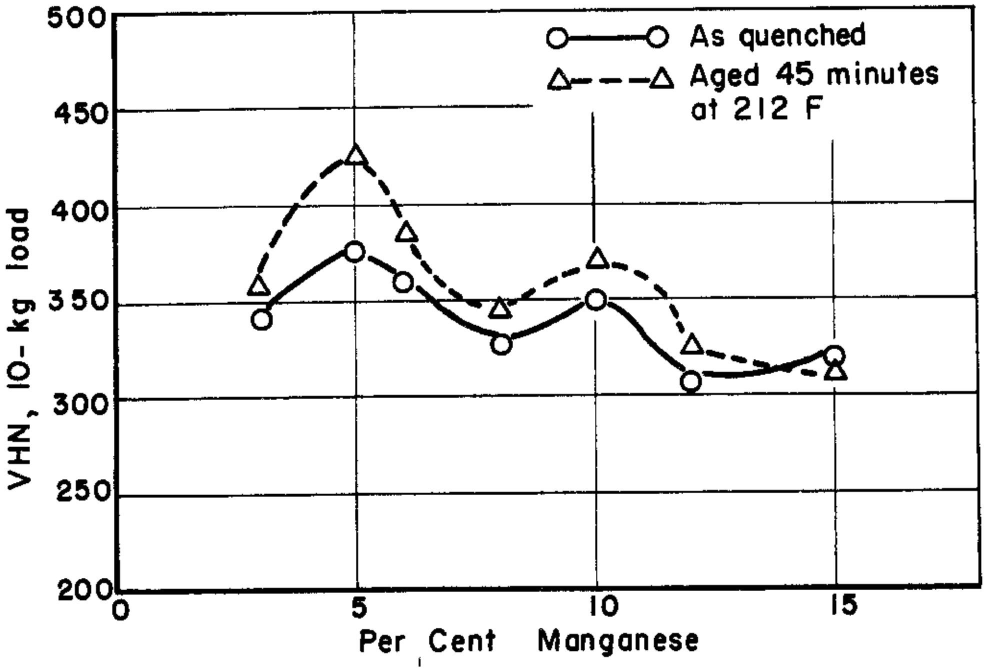


FIGURE 11. EFFECT OF LOW-TEMPERATURE AGING ON HARDNESS OF BINARY Ti-Cr AND Ti-Mn ALLOYS QUENCHED IN MEDIUM OIL FROM 1742 F (950C)

A-3257

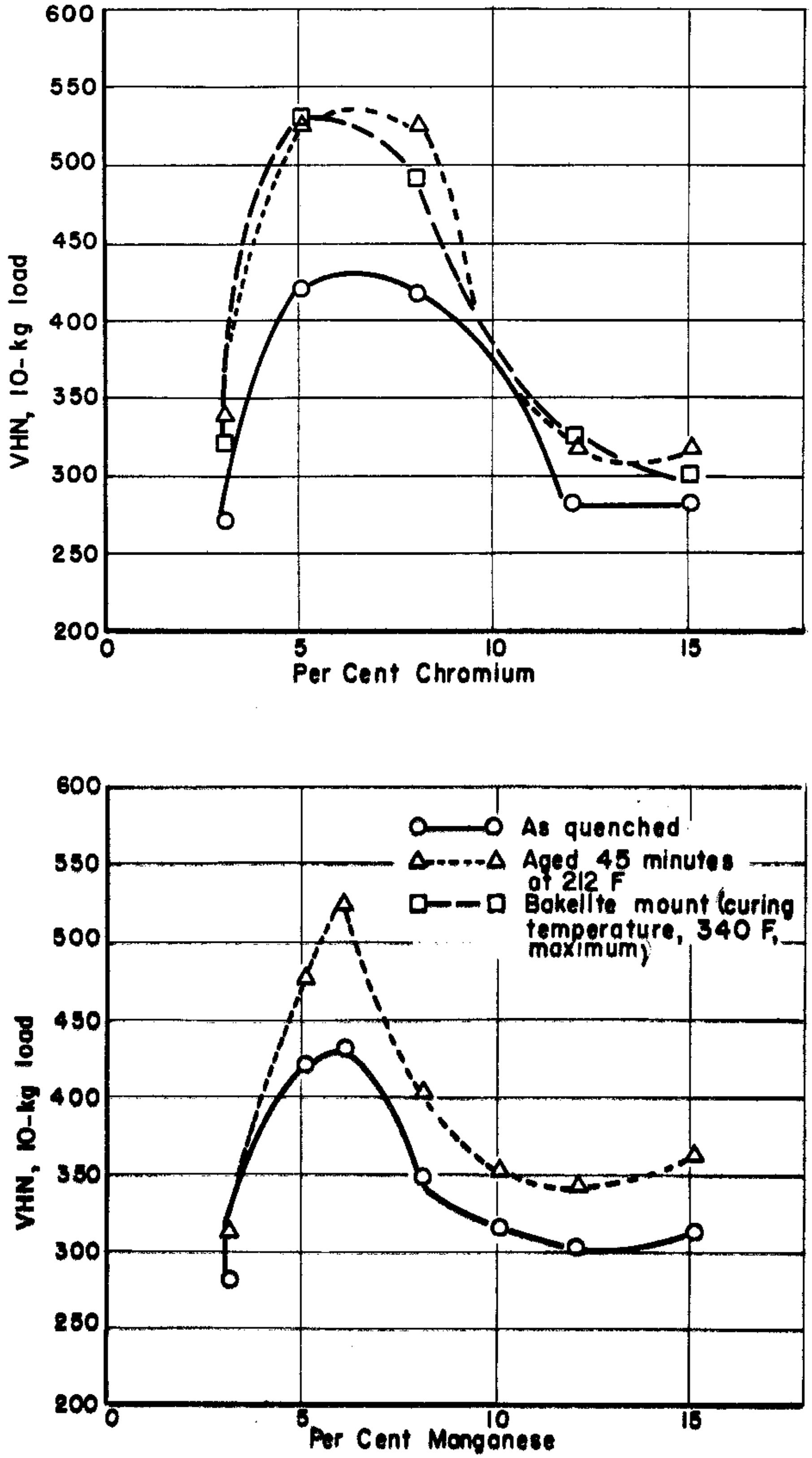


FIGURE 12. EFFECT OF LOW-TEMPERATURE AGING ON HARDNESS OF BINARY Ti-Cr AND Ti-Mn ALLOYS QUENCHED IN LIQUID NITROGEN FROM 1742 F (950C)

A-3258

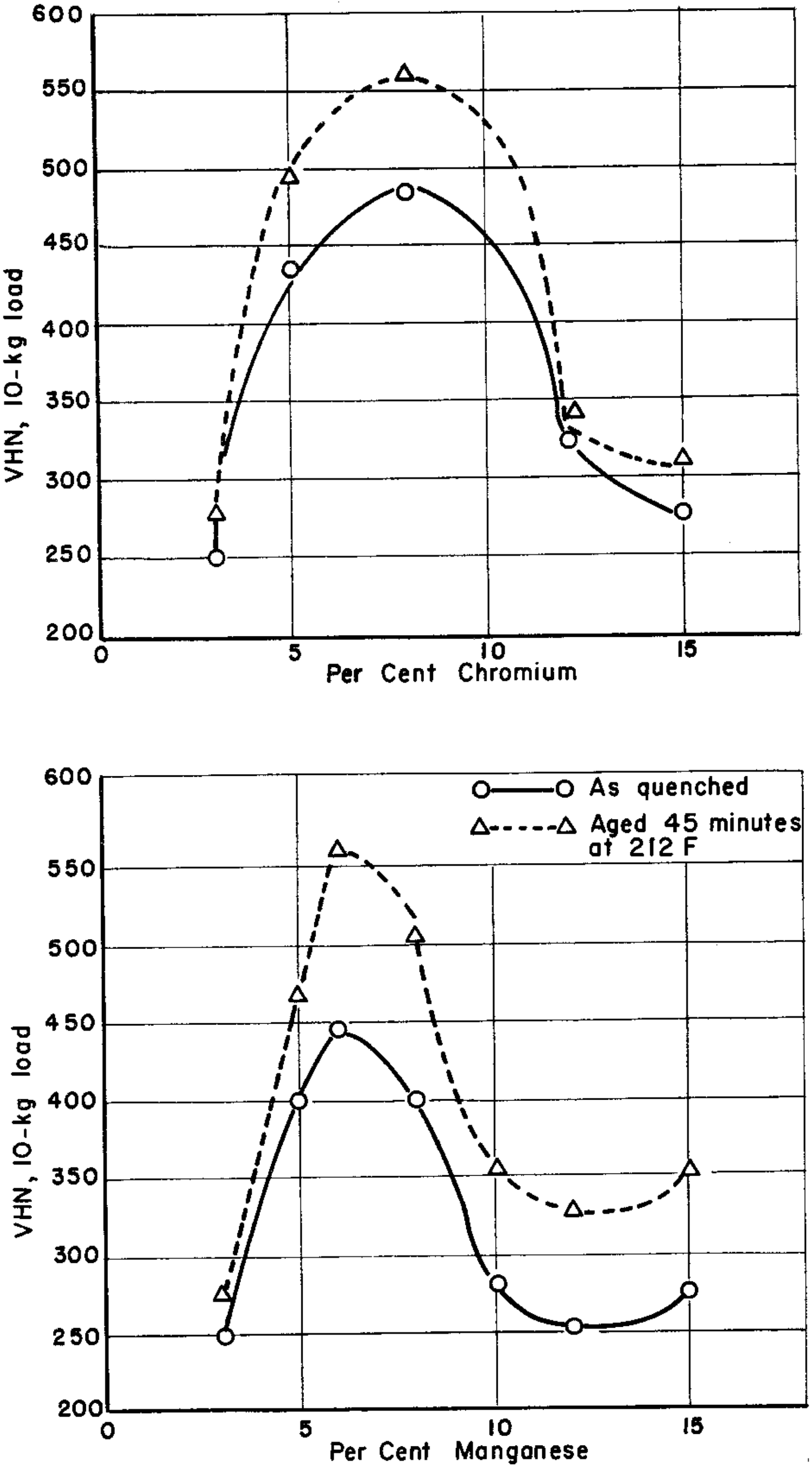


FIGURE 13. EFFECT OF LOW-TEMPERATURE AGING ON HARDNESS OF BINARY Ti-Cr AND Ti-Mn ALLOYS COOLED IN STILL AIR FROM 1742 F (950C)

A-3260

TABLE 25. EFFECT OF LOW-TEMPERATURE AGING ON HARDNESS OF TITANIUM ALLOYS IN THE AS-HOT-ROLLED CONDITION

Heat No.	Intended Composition, %	VHN As Hot Rolled	VHN Aged 45 Min at 100 C	Heat No.	Intended Composition, %	VHN As Hot Rolled	VHN Aged 45 Min at 100 C
WS278A	2.75Mn-2.0Mo	339	352	WS199A	5.0V-1.5Fe	339	342
WS279A	2.75Mn-2.75Mo	345	344	WS204A	5.0V-1.5Fe	312	323
WS277A	2.75Mn-3.5Mo	383	388	WS198A	5.0V-3.0Fe	425	435
				WS206A	5.0V-3.0Fe	363	392
WS267A	3.5Mn-2.0Mo	345	352	WS205A	5.0V-4.5Fe	429	442
WS269A	3.5Mn-2.75Mo	357	370	WS207A	5.0V-4.5Fe	414	437
WS270A	3.5Mn-3.5Mo	397	396				
WS268A	4.25Mn-2.0Mo	394	405	WS239A	2.0V-1.5Ni	279	257
WS265A	4.25Mn-2.75Mo	401	421	WS231A	2.0V-3.0Ni	394	405
WS276A	4.25Mn-3.5Mo	401	414	WS201A	2.0V-4.5Ni	317	318
WS271A	5.0Mn-2.0Mo	437	446	WS234A	5.0V-1.5Ni	317	333
WS272A	5.0Mn-2.75Mo	425	437	WS232A	5.0V-3.0Ni	287	291
WS273A	5.0Mn-3.5Mo	421	433	WS245A	5.0V-4.5Ni	380	391
WS186A	2.0V-1.5Fe	283	287	WS209A	2.0V-2.0Mn	296	299
WS194A	2.0V-3.0Fe	357	363	WS210A	2.0V-3.5Mn	360	369
WS190A	2.0V-4.5Fe	387	405	WS215A	2.0V-5.0Mn	421	433
				WS247A	2.0V-2.0Mn-1.0Cr	351	363
WS193A	3.5V-1.5Fe	336	351	WS212A	2.0V-2.0Mn-3.0Cr	394	397
WS195A	3.5V-3.0Fe	363	381				
WS196A	3.5V-4.5Fe	417	433				

TABLE 25. (Continued)

		VHN				VHN	
Heat No.	Intended Composition, %	As Hot Rolled	Aged 45 Min at 100 C	Heat No.	Intended Composition, %	As Hot Rolled	Aged 45 Min at 100C
WS218A	3.5V-2.0Mn	333	347	WS172A	2.0Mn-2.0V-1.0Fe	341	311
WS216A	3.5V-3.5Mn	336	383	WS173A	2.0Mn-2.0V-2.0Fe	390	396
WS219A	3.5V-5.0Mn	433	437	WS174A	2.0Mn-4.0V-1.0Fe	397	405
WS220A	3.5V-2.0Mn-1.0Cr	397	392	WS178A	2.0Mn-4.0V-2.0Fe	413	430
WS221A	3.5V-2.0Mn-3.0Cr	421	449				
				WS181A	3.0Mn-2.0V-1.0Fe	380	390
WS222A	5.0V-2.0Mn	387	389	WS177A	3.0Mn-2.0V-2.0Fe	413	433
WS223A	5.0V-3.5Mn	409	421	WS179A	3.0Mn-4.0V-1.0Fe	433	459
WS226A	5.0V-5.0Mn	441	446	WS182A	3.0Mn-4.0V-2.0Fe	433	433
WS224A	5.0V-2.0Mn-1.0Cr	370	384				
WS225A	5.0V-2.0Mn-3.0Cr	417	449	WS187A	4.0Mn-2.0V-1.0Fe	401	410
				WS184A	4.0Mn-2.0V-2.0Fe	405	409
				WS183A	4.0Mn-4.0V-1.0Fe	413	437
				WS185A	4.0Mn-4.0V-2.0Fe	366	366

With very few exceptions, hardness before and after aging differed by less than 20 VHN, which is considered within experimental error. In no case was the hardness increase after aging more than 30 VHN. Therefore, the effect of mounting in Bakelite on the hardness of alloys in the as-hot-rolled condition is believed to be of little significance.

Recrystallization Study of Binary Titanium-Chromium Alloys

These experiments were designed to provide basic information on the response of the titanium-chromium alloys to thermal treatment following plastic straining. Of particular interest was the possibility of refining the beta grain size in alloys exhibiting a retained beta structure at room temperature. The compositions chosen included chromium levels from 2.5 to 30 per cent.

Strips of each alloy, 1/2 by 4 by 0.064 inch, were annealed four hours at 1250 F in argon, furnace cooled, and cold rolled until edge cracking occurred. The Ti-10%Cr alloy broke up during rolling and only a limited amount could be used. Cold rolling of the unalloyed titanium specimens, included for comparison purposes, and the Ti-2.5%Cr alloy was discontinued at 71 and 59.5 per cent reductions, respectively, even though no edge cracking was observed. Rolling data and hardness test results for the rolled strips are given in Table 26.

TABLE 26. ROLLING DATA ON RECRYSTALLIZATION EXPERIMENTS FOR TITANIUM-CHROMIUM ALLOYS

Heat No.	Intended Composition, %	Reduction in Thickness by Cold Rolling, %	Degree of Edge Cracking From Rolling	Vickers Hardness(1) As Hot Rolled	Vickers Hardness(1) As Cold Rolled
WL87	Unalloyed	71.0	None	185**	258*
WL60A	2.5Cr	59.5	None	279**	301**
WL64A	3.5Cr	59.0	Slight	306**	339
WL122A	5.0Cr	24.5	Moderate	357**	385
WL21A	10.0Cr	31.0	Failure	390**	371
WL22A	15.0Cr	28.5	Slight	329**	370
WL20A	20.0Cr	30.0	Severe	330**	387
WL130A	25.0Cr	28.0	Severe	417**	381
WL132A	30.0Cr	39.5	Moderate	299**	336

(1) Vickers Hardness: Hardness taken at center of cross section of the sheet 90 degrees to the rolling surface. Loads (*) 2.5 kg, (**) 10 kg, all others 5 kg. Average of 5 Vickers readings reported.

Small specimens sheared from the cold-rolled strips were heated in a well-deoxidized lead bath at 932 F (500 C), 1112 F (600 C), and 1292 F (700 C) for periods of 1/4, 1/2, 2, 6, and 60 hours. They were then quenched in ice water to retain the phase relationships existing at the elevated temperature. Results of hardness tests made on the quenched specimens are plotted in Figures 14 through 18.

Hardness changes could not be used as an indication of recrystallization. The hardness of the cold-rolled unalloyed titanium (Figure 14) decreased from about 260 to 190 VHN after 2 hours at 932 F. However, no recrystallization was detected microscopically until an annealing time of six hours had transpired. After 60 hours at 932 F, recrystallization was complete, as shown by a fine, equiaxed grain structure throughout the specimen. At 1112 and 1292 F, recrystallization of the unalloyed specimens was complete after annealing for 1/4 hour.

Like the unalloyed material, the hardnesses of the 2.5, 3.5, 5, 20, and 25 per cent chromium alloys decreased with increasing annealing time at 932 and 1112 F. On the other hand, the 10 and 15 per cent chromium alloys exhibited initial hardness increases at 932 F, and the 30 per cent chromium alloy increased in hardness at both 932 and 1112 F.

All of the alloys which were annealed at 1292 F decreased in hardness initially; at annealing times of 6 to 60 hours, the hardnesses again increased. The 10 and 15 per cent chromium alloys were not annealed at 1292 F because of a shortage of stock. The increase in hardness at 1292 F of the binary alloys containing up to 5 per cent chromium undoubtedly resulted from the transformation of some of the alpha phase into beta of a relatively high alloy content. This was then hardened by precipitation of the alpha phase during the quench.

The reason for the increase in hardness of the 20 to 30 per cent chromium alloys annealed at 1292 F is believed to be associated with the precipitation of $TiCr_2$. This will be more fully discussed later in this section.

The microstructures of the 2.5, 3.5, and 5 per cent chromium alloys were very similar after the various annealing treatments. The structures of the alloys, as cold rolled, could not be resolved microscopically, but probably consisted largely of alpha with a small amount of beta. Little change was observed in the cold-rolled structure at annealing times up to 6 hours at 932 F. After 60 hours at this temperature, a small amount of recrystallized alpha was detected along with an unresolvable precipitate of alpha or $TiCr_2$. At 1112 F, some recrystallization was observed after six hours, as shown in Figure 19, which is a photomicrograph of the Ti-3.5%Cr alloy. Sixty hours at 1112 F produced a structure which was apparently about 70 per cent recrystallized, as illustrated in Figure 20. Complete recrystallization, as denoted by the formation of equiaxed grains throughout the specimen, was attained only after 60 hours at 1292 F. At this temperature, the

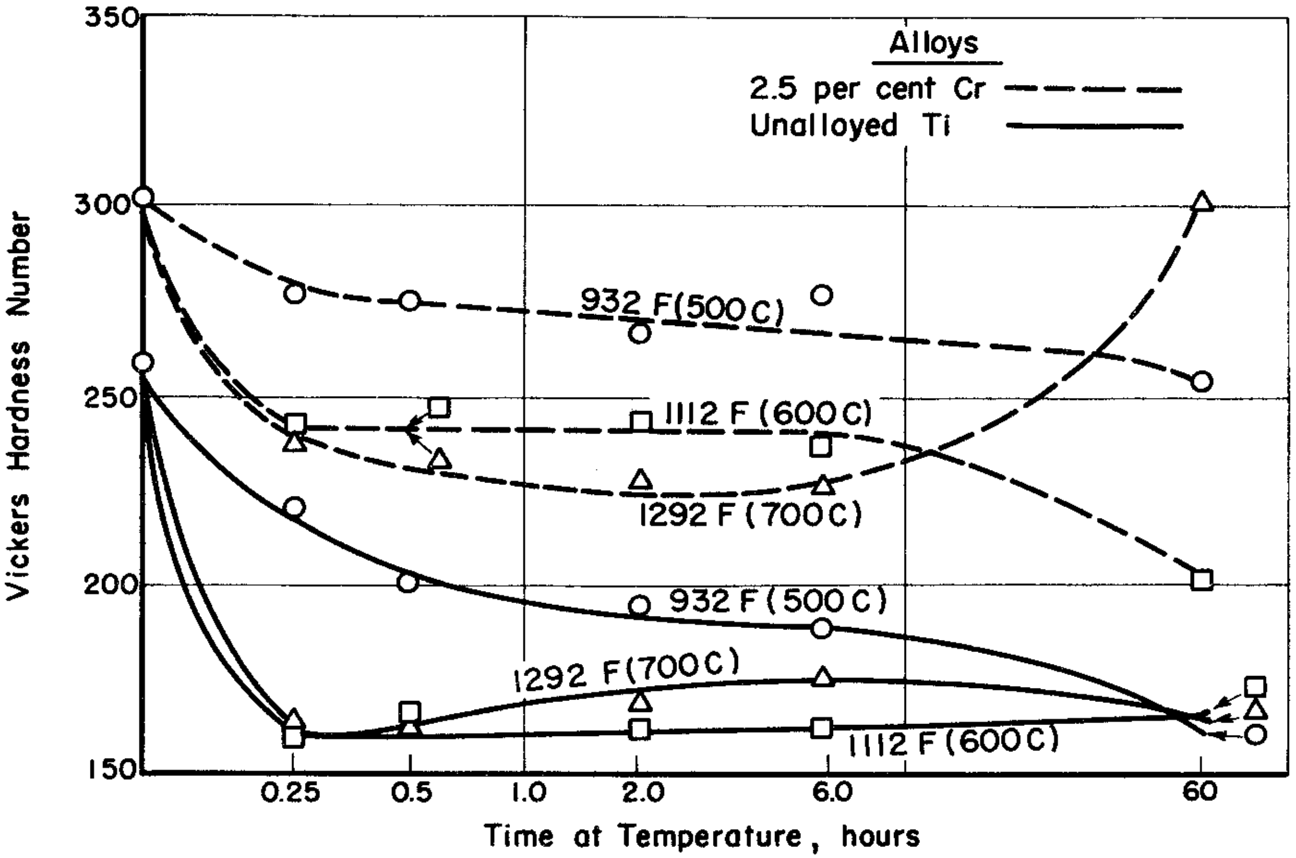


FIGURE 14. EFFECT OF THERMAL TREATMENTS ON THE HARDNESS OF COLD-ROLLED TITANIUM-CHROMIUM ALLOYS

A-3260

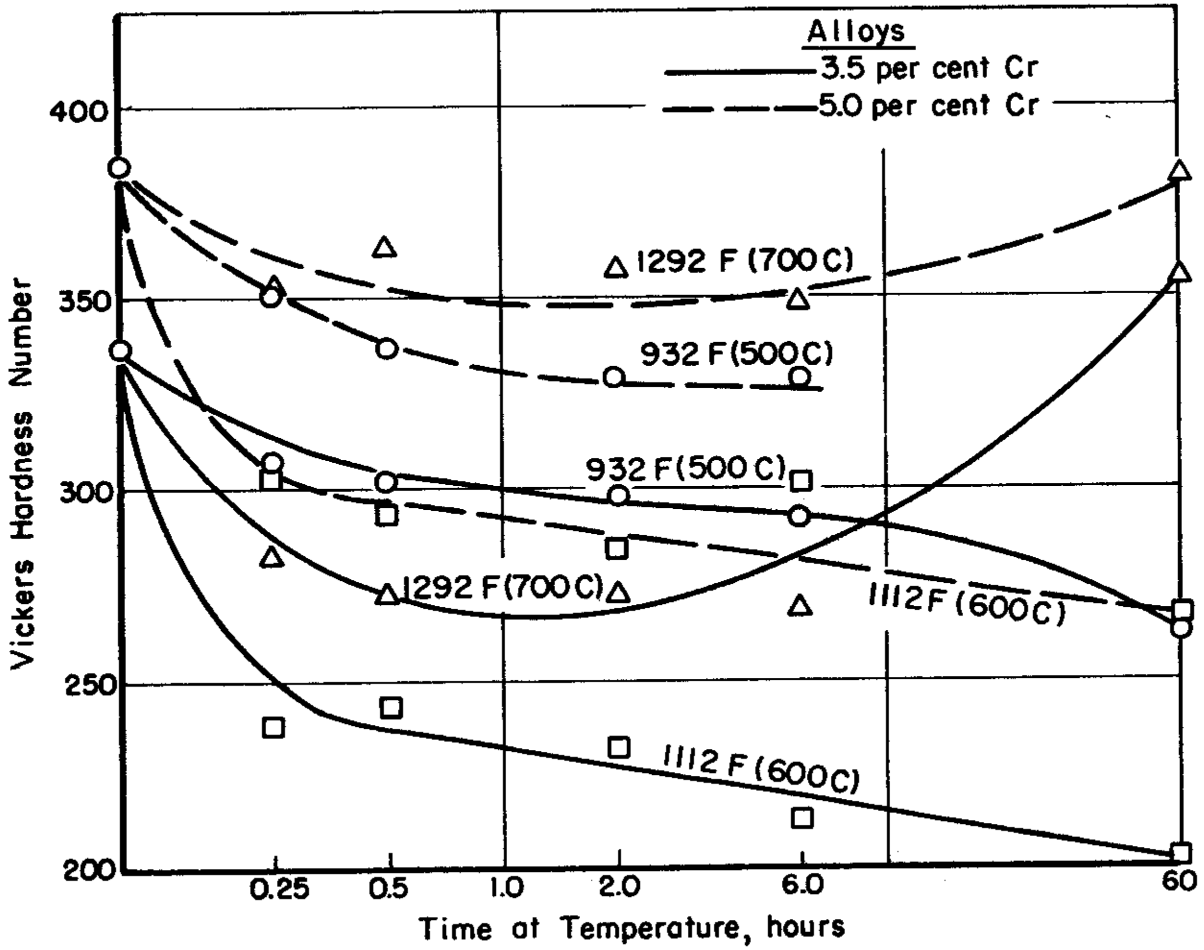


FIGURE 15. EFFECT OF THERMAL TREATMENTS ON THE HARDNESS OF COLD-ROLLED TITANIUM-CHROMIUM ALLOYS

A-3261

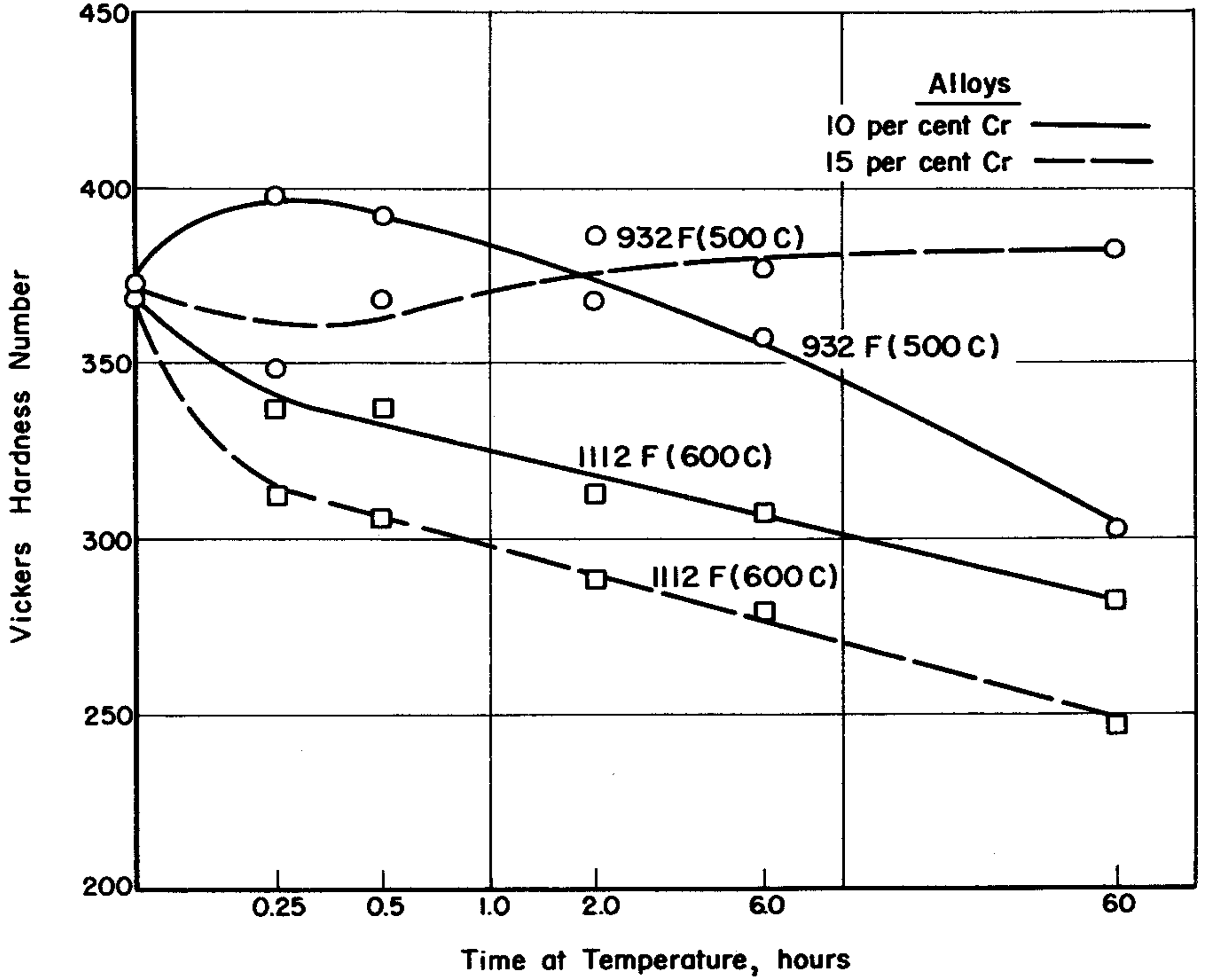


FIGURE 16. EFFECT OF THERMAL TREATMENTS ON THE HARDNESS OF COLD-ROLLED TITANIUM-CHROMIUM ALLOYS

A-3262

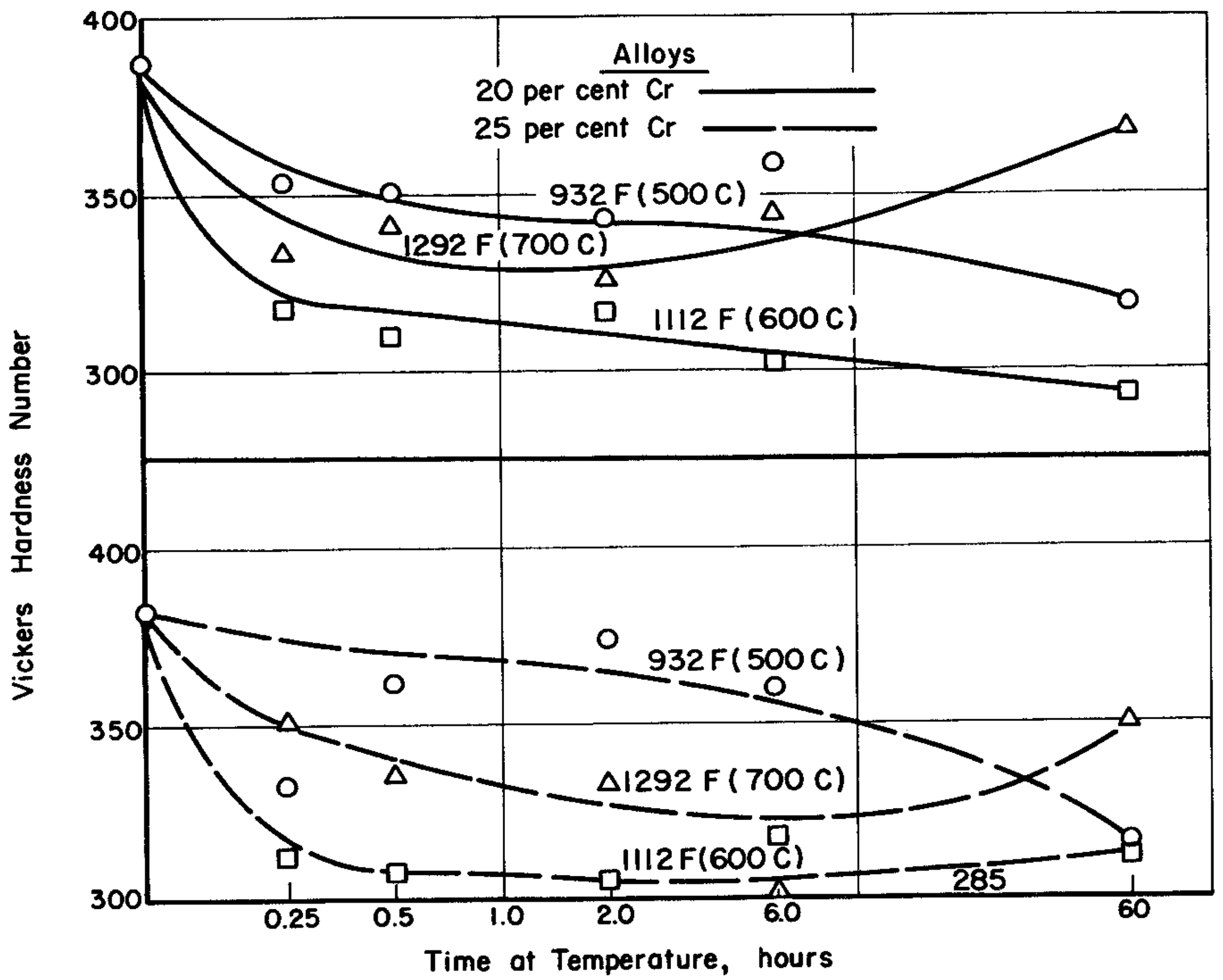


FIGURE 17. EFFECT OF THERMAL TREATMENTS ON THE HARDNESS OF COLD-ROLLED TITANIUM-CHROMIUM ALLOYS

A-3263

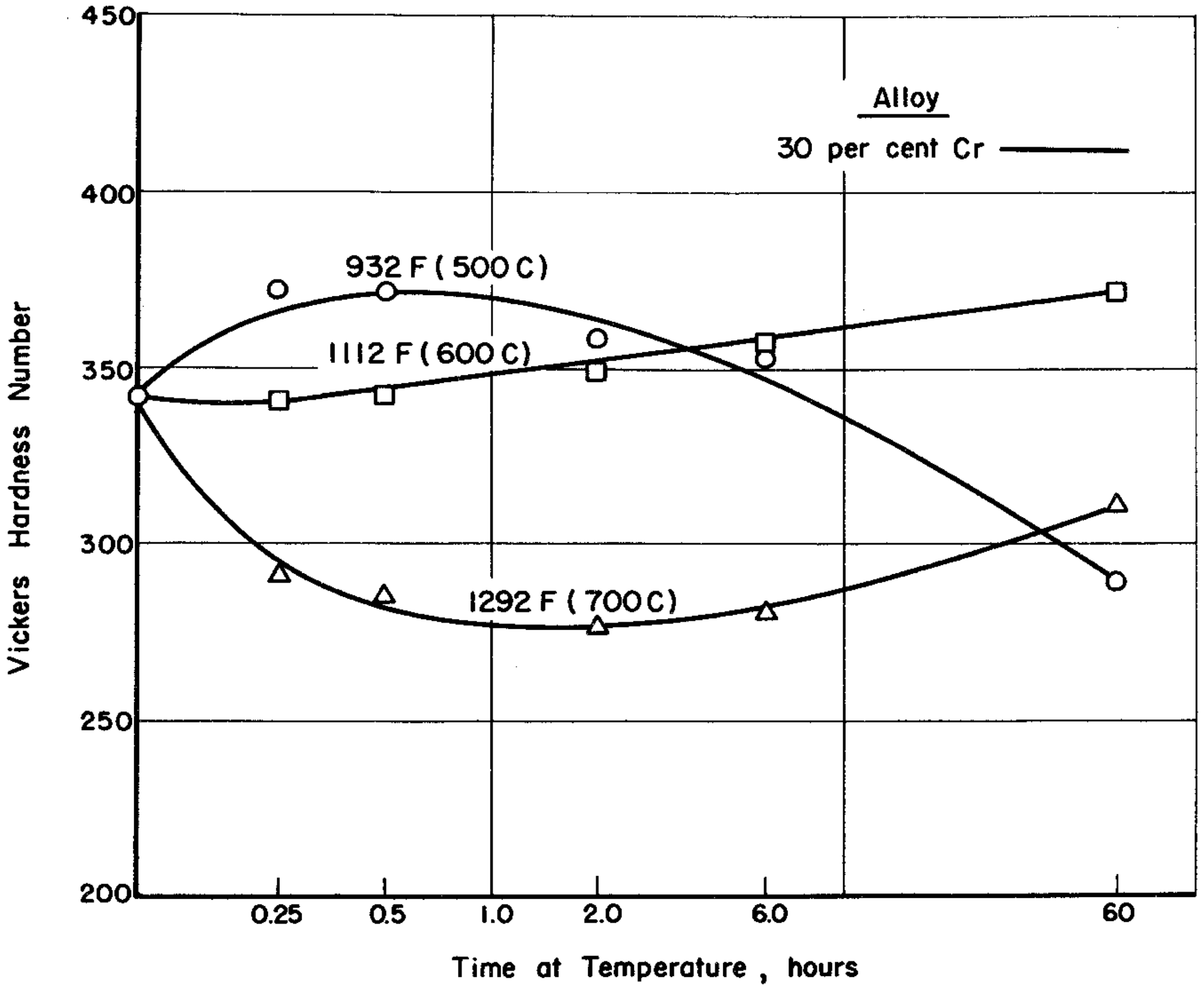


FIGURE 18. EFFECT OF THERMAL TREATMENTS ON THE HARDNESS OF COLD-ROLLED TITANIUM-CHROMIUM ALLOYS

A-3264

recrystallization was accompanied by the transformation of part of the alpha to the beta phase which was retained after the quench. The structure of the completely recrystallized Ti-3.5%Cr alloy is shown in Figure 21.

The structure of the Ti-10%Cr alloy, as cold rolled, was largely retained beta with perhaps 10 per cent alpha present as discrete particles. That of the Ti-15%Cr alloy was all beta. Annealing at 932 F produced an unresolved precipitate, presumably alpha, in the beta matrix. The amount of the precipitate increased with increasing time. Figure 22 illustrates the structure of the Ti-10%Cr alloy after annealing six hours at 932 F. The corresponding Ti-15%Cr specimen had a smaller amount of precipitate as annealed six hours at 932 F. Annealing the Ti-10%Cr alloy at 1112 F resulted in a structure consisting of alpha and eutectoid in a beta matrix, as shown in Figures 23 and 24. The Ti-15%Cr alloy was almost completely transformed to the eutectoid after 60 hours at 1112 F. No recrystallization was observed in either of these two alloys.

The alloys containing 20 to 30 per cent chromium were of the hypereutectoid type. The thermal history of these alloys was such that the cold-rolled structure consisted of discrete particles of $TiCr_2$ in a beta matrix. The amount of the compound increased with increasing alloy content. Annealing at 932 F for times up to 60 hours produced an unresolved precipitate of $TiCr_2$ in the beta matrix. As in the 10 and 15 per cent chromium alloys, the eutectoid appeared in specimens annealed at 1112 F. Time for initiation of the eutectoid reaction increased with increasing alloy content. Thus, the first evidence of eutectoid appeared in 1/2, 2, and 6 hours, respectively, in the 20, 25, and 30 per cent chromium alloys. The microstructure of the Ti-20%Cr alloy annealed for six hours at 1112 F is shown in Figure 25. Nucleation of the eutectoid occurred generally within the grains as well as at the grain boundaries. In the Ti-10%Cr alloy, however, nucleation was predominantly at the grain boundaries (Figure 23).

The only visible change in the structure of the 20 to 30 per cent chromium alloys during the anneal at 1292 F was an increase in the number and size of the $TiCr_2$ particles. A typical structure produced at 1292 F is shown in Figure 26. The increase in the amount of $TiCr_2$ probably accounts for the increase in hardness of these alloys after 6 to 60 hours at 1292 F. Since $TiCr_2$ is harder than the beta matrix, an increase in the amount of this phase would tend to increase the average hardness of the alloy. The initial drop in hardness exhibited by these alloys at 1292 F was evidently a stress-relief effect.

In view of the above discussion, the following general conclusions may be drawn regarding the recrystallization of cold-worked and annealed Ti-Cr alloys:

1. Alloys containing 5 per cent chromium or under may be completely recrystallized in reasonably short times (60 hours or less) only by heating above the eutectoid temperature.

2. Alloys containing more than 10 per cent chromium cannot easily be recrystallized by the type of thermal-mechanical treatment commonly employed.

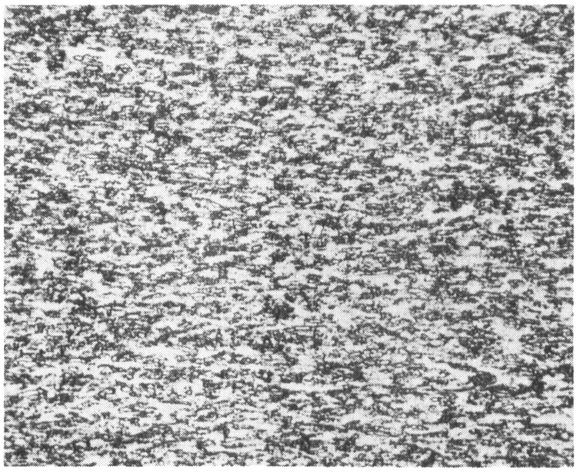


FIGURE 19. 3.5 PER CENT Cr ALLOY COLD ROLLED 59 PER CENT AND ANNEALED 6 HOURS AT 1112 F (600 C)
Structure: Some Recrystallized Alpha Plus Unresolved Precipitate of Alpha and/or $TiCr_2$

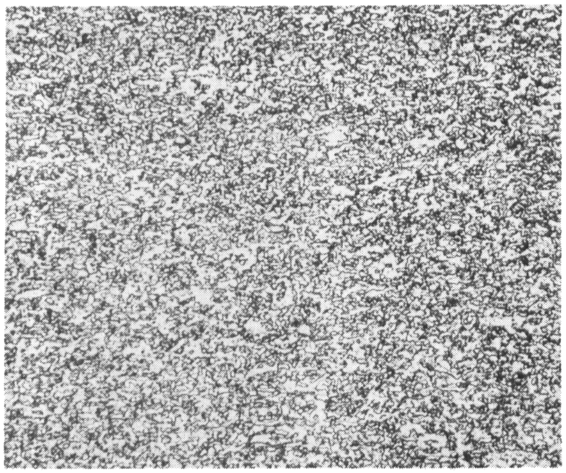


FIGURE 20. 3.5 PER CENT Cr ALLOY COLD ROLLED 59 PER CENT AND ANNEALED 60 HOURS AT 1112 F (600 C)
Structure: Same as Figure 24 Except More of the Alpha is Recrystallized

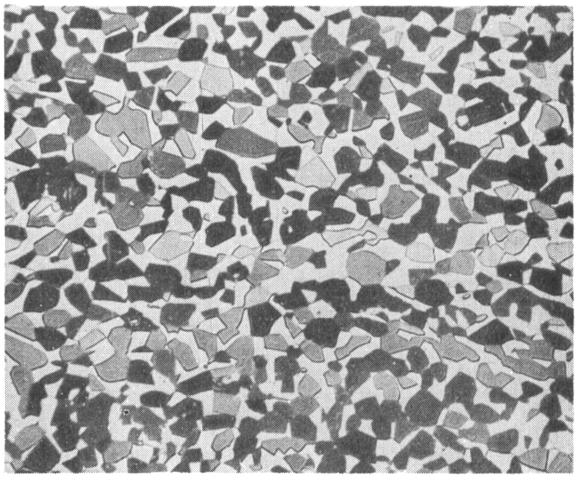


FIGURE 21. 3.5 PER CENT Cr ALLOY COLD ROLLED 59 PER CENT AND ANNEALED 60 HOURS AT 1292 F (700 C)
Structure: Equiaxed Alpha in Beta Matrix, Completely Recrystallized

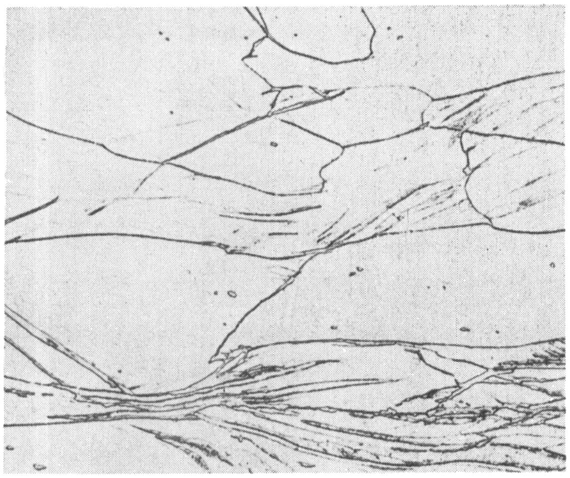
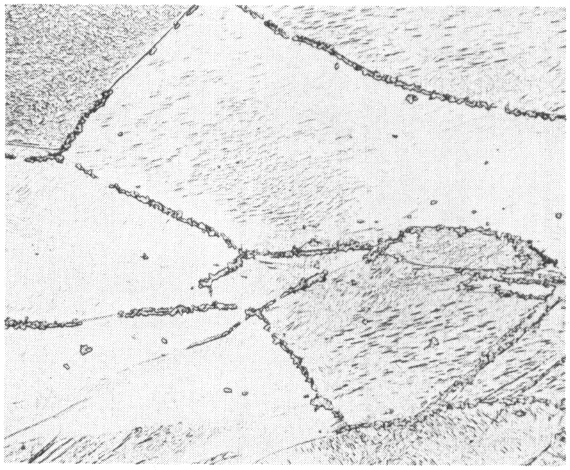
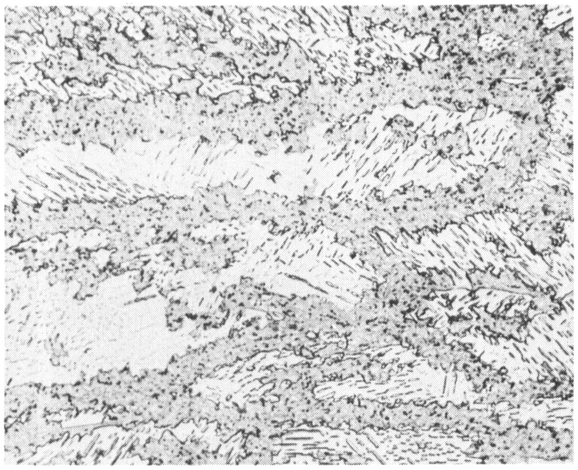


FIGURE 22. 10 PER CENT Cr ALLOY COLD ROLLED 31 PER CENT AND ANNEALED 6 HOURS AT 932 F (500 C)
Structure: Unresolved Precipitate of Alpha in Unrecrystallized Beta Matrix



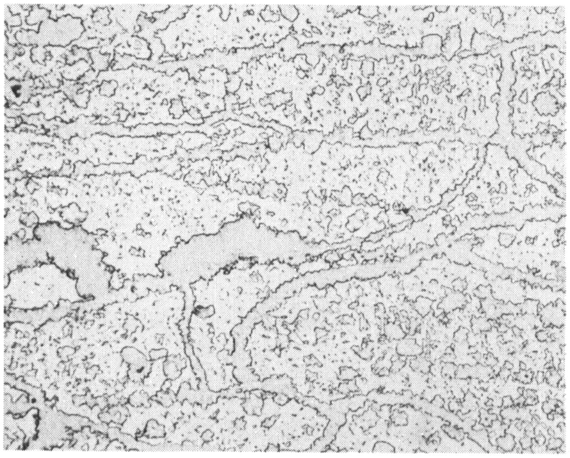
500X 83121

FIGURE 23. 10 PER CENT Cr ALLOY COLD ROLLED 31 PER CENT AND ANNEALED 6 HOURS AT 1112 F (600 C)
Structure: Alpha Plus Eutectoid in a Beta Matrix, No Recrystallization



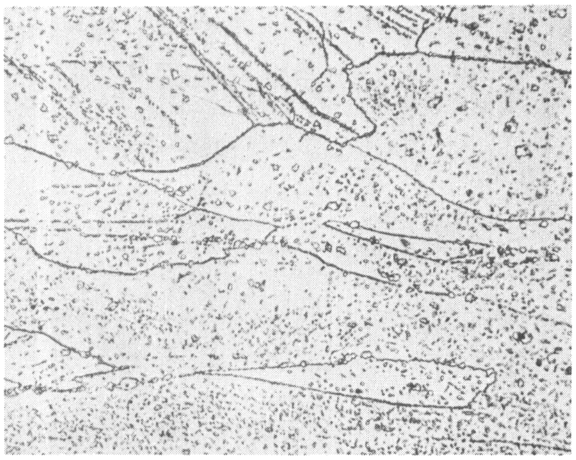
500X 83120

FIGURE 24. 10 PER CENT Cr ALLOY COLD ROLLED 31 PER CENT AND ANNEALED 60 HOURS AT 1112 F (600 C)
Structure: Same as Figure 23 Except for More Eutectoid



500X 83124

FIGURE 25. 20 PER CENT Cr ALLOY COLD ROLLED 30 PER CENT AND ANNEALED 6 HOURS AT 1112 F (600 C)
Structure: $TiCr_2$ Plus Eutectiod in Beta Matrix, No Recrystallization



500X 83107

FIGURE 26. 20 PER CENT Cr ALLOY COLD ROLLED 30 PER CENT AND ANNEALED 1/4 HOUR AT 1292 F (700 C)
Structure: $TiCr_2$ in Beta Matrix, No Recrystallization

FURTHER EVALUATION OF SELECTED ALLOYS

On the basis of their as-hot-rolled properties, twelve alloys were selected for further evaluation. The intended compositions and the ingot sizes in which additional heats of these alloys have been prepared to date are given in Table 27. All of the compositions listed are high-strength alloys with the exception of the Ti-15% Cr composition, which is a medium-strength, ductile, metastable beta alloy.

Evaluation of the alloys in several section sizes is contemplated. Initially, tensile properties and bend ductility will serve as a basis for evaluation of the sheet alloys, while only the tensile properties will be considered for bar-stock compositions. Later, the program will be broadened to include additional tests such as elevated-temperature creep, notch sensitivity, fatique, and others. Several well-known companies have indicated that they will assist in the evaluation of some of the most interesting alloys. Republic Steel has agreed to roll ingots into sheet and bar stock which will then be given additional tests. The Wyman-Gordon Company will evaluate the forgeability of certain alloys. The Aviation Gas Turbine Division of Westinghouse will evaluate some of the properties of sheet and bar stock.

The primary objective of the early part of this program is to determine, by easily applied mechanical tests, how the alloys respond to fabrication and heat treatments.

For convenient presentation, this section of the report has been divided according to the ingot size of the various alloys. During the period covered by this report, tests were made only on the one-pound and the five-pound ingots.

One-Pound Ingots

The primary purpose of these heats was to determine the reproducibility of mechanical properties from heat to heat in the as-hot-rolled condition and after various process-annealing treatments. Chemical analyses have been determined on the first ingot of each composition. The analyses of the other ingot are not yet available.

Fabrication

The fabrication of the one-pound ingots was essentially the same as that described earlier in the report for the 3/4-pound ingots of the exploratory alloys.

TABLE 27. INTENDED COMPOSITIONS AND INGOT SIZES OF ALLOYS SELECTED FOR FURTHER EVALUATION

1-Pound Ingots

5.0Mo-2.0Fe
5.0Mn-2.5Cr
5.0Mn-2.0Fe
5.0Mn-2.0Mo
5.0Cr-1.5Fe
3.5Cr-3.5V
3.5Cr-3.0Mn

5-Pound Ingots

3.5Cr-3.5V
5.0Mn-2.5Cr
1Cr-1Fe-3Mn-1Mo-1V
1Cr-4Fe-1Mn-1Mo-1V

10-Pound Ingots

3.5Cr-3.5V
5.0Mn-2.5Cr
1Cr-1Fe-3Mn-1Mo-1V
1Cr-4Fe-1Mn-1Mo-1V
4.5Mo-3.5Fe
5.0Mn-2.0Mo
3.5Mn-3.5Cr

20-Pound Ingots

4.5Mo-3.5Fe
5.0Mn-2.0Mo
3.5Mn-3.5Cr

35-Pound Ingots

15.0Cr

Mechanical Properties - As-Hot-Rolled Condition

The tensile properties, bend ductility, and hardness of each heat of these alloys are presented in Table 28. Actual chemical compositions, where available, are also included. For comparative purposes, the mechanical properties of earlier heats of the same intended compositions are given in the table. Of the earlier heats, only those fabricated by the same procedure used on the current series are considered here.

All of the alloys showed considerable variation in mechanical properties from heat to heat. The best reproducibility was obtained in the Ti-5% Mo-4%Fe alloy. Since only one ingot of each composition has been analyzed, no evaluation of the effects of chemical composition on properties is possible at the present time. It is known, however, that uncontrollable variations in fabricating procedures can also cause poor reproducibility of properties. A major objective of the heat-treatment program described in the next section was the minimizing of these fabrication variables.

Heat Treatment

Process-annealing treatments were completed on the first ingot of each of the compositions shown in Table 28. Similar heat treatments on material from the second heats of these compositions are not complete. The annealing treatments were carried out on specimens of the alloys in two prior conditions: (1) as hot rolled, and (2) annealed and cold rolled to 20 per cent reduction in thickness. The annealing treatment used for specimens subsequently cold rolled consisted of heating for four hours at 1250 F in a dried-argon atmosphere and furnace cooling. The process-annealing treatments subsequently given to specimens of each alloy in each of the two prior conditions consisted of heating in an argon atmosphere for periods of 1/2 and 2 hours at 1400 or 1500 F, and air cooling. The results of tensile and hardness tests made on specimens so treated are presented in Table 29.

Good combinations of strength and ductility were produced in all of the alloys by one or more of the annealing treatments employed. In nearly every instance, the treatments at 1400 F were superior to those at 1500 F. The tensile strengths of the annealed specimens were, in all cases, lower than those obtained in the as-hot-rolled condition (Table 28). The decrease in tensile strength was accompanied, however, by an increase in ductility. Only the Ti-5%Mn-2.5%Cr and the Ti-5%Mn-2%Fe alloys developed tensile strengths greater than 180,000 psi in the annealed condition. Cold rolling 20 per cent prior to annealing increased the tensile strengths slightly and lowered ductility correspondingly. Increasing the annealing time from 1/2 to 2 hours at 1400 F had little effect on any of the alloys except the Ti-5%Mn-2%Fe composition. The ductility of the as-hot-rolled and annealed specimens of this alloy increased sharply for the longer annealing time. Increasing the annealing time at 1500 F generally was not beneficial.

TABLE 28. AS-HOT-ROLLED PROPERTIES OF DIFFERENT HEATS OF VARIOUS HIGH-STRENGTH ALLOYS(1)

Heat No.	Intended Composition, %	Actual Composition, %	Ultimate Tensile Strength, psi(2)	Elongation, % in 1 inch	VHN	Minimum Bend Radius, T(3)
WM24A(5)	3Mn–3.5Cr	–	213,400	5.0	397	24
WS289A	3Mn–3.5Cr	2.74Mn–3.09Cr	197,700(4)	5.0	418	–
WS305A	3Mn–3.5Cr	–	185,000	4.5	387	8
WM23A(5)	5Mn–2.5Cr	–	227,500	3.0	446	12
WS282A	5Mn–2.5Cr	5.02Mn–2.13Cr	211,400(4)	3.0	421	–
WS291A	5Mn–2.5Cr	–	186,000	8.0	390	4
WN18A(5)	5Mn–2Mo	–	215,800	3.5	421	24
WS271A(6)	5Mn–2Mo	–	200,000	6.0	437	4
WS286A	5Mn–2Mo	4.78Mn–1.87Mo	192,600(4)	4.0	373	–
WS299A	5Mn–2Mo	–	195,000	4.0	366	8
WN23A(5)	5Mn–2Fe	–	220,700	4.5	444	24
WS281A	5Mn–2Fe	4.46Mn–1.87Fe	(8)	(8)	459	–
WS294A	5Mn–2Fe	–	188,000	6.0	421	8
WM49A(7)	3.5Cr–3.5V	–	200,200	6.0	403	12
WS285A	3.5Cr–3.5V	3.37Cr–3.0V	176,500(4)	10.0	373	–
WS304A	3.5Cr–3.5V	–	186,100	5.5	373	8
WL82A(7)	5Cr–1.5Fe	–	218,500	3.0	431	> 24
WS296A	5Cr–1.5Fe	4.30Cr–1.43Fe	187,100(4)	6.5	421	–
WS301A	5Cr–1.5Fe	–	211,600	1.0	397	4
WN28A	5Mo–4Fe	–	214,500	3.0	411	24
WS287A	5Mo–4Fe	4.79Mo–3.43Fe	203,200(4)	6.0	413	–
WS290A	5Mo–4Fe	–	201,400	4.25	387	8

(1) All alloys were melted as 3/4- or 1-lb ingots and tested as 14-gage (0.064 inch) sheet.

(2) Substandard specimen 0.375 × 0.064 × 5 inches with a reduced section 0.250 × 0.064 × 1-1/4 inches.

(3) Specimen size 0.375 × 0.064 × 4 inches.

(4) Results of only one specimen. All other values are the average of two specimens.

(5) See Reference (3), Table 9.

(6) See Table 7, this report.

(7) See Reference (3), Table 9.

(8) Defective specimen.

TABLE 29. PROPERTIES OF SELECTED HIGH-STRENGTH ALLOYS AFTER VARIOUS HEAT TREATMENTS(1)

				Prior Condition—As Hot Rolled			Prior Condition—Annealed and Cold Rolled 20%(3)		
Heat No.	Intended Composition, %	Annealing Temperature, F(2)	Annealing Time, hours	Ultimate Tensile Strength, psi(4)	Elongation, % in 1 inch	VHN	Ultimate Tensile Strength, psi(4)	Elongation, % in 1 inch	VHN
WS289A	3Mn–3.5Cr	1400	1/2	168,700	13.5	376	176,900	9.5	387
	(Actual 2.74Mn–3.09Cr)		2	160,500	15.0	363	180,600	10.0	399
		1500	1/2	163,200	8.5	366	173,600	7.5	388
			2	147,700	15.0	319	168,300	2.0	384
WS282A	5Mn–2.5Cr	1400	1/2	184,400	6.0	422	187,600	7.5	373
	(Actual 5.02Mn–2.13Cr)		2	195,600	3.5	417	187,700	5.5	396
		1500	1/2	188,900	0.0	401	195,900	2.0	376
			2	144,300	0.5	405	186,900	0.0(5)	405
WS286A	5Mn–2.0Mo	1400	1/2	172,200	9.0	362	180,400	10.0	349
	(Actual 4.78Mn–1.87Mo)		2	175,600	7.5	372	176,600	9.5	359
		1500	1/2	170,200	5.0	417	172,200	3.5	355
			2	168,800	2.0	362	131,900	1.0	370
WS281A	5Mn–2Fe	1400	1/2	183,800	2.5	391	186,300	5.0	361
	(Actual 4.46Mn–1.87Fe)		2	172,300	15.0	368	190,400	4.5	380
		1500	1/2	182,700	8.0	369	184,900	5.0	388
			2	156,700	14.0	355	165,500	9.0	396
WS285A	3.5Cr–3.5V	1400	1/2	158,900	15.5	367	164,100	10.0	376
	(Actual 3.37Cr–3.00V)		2	150,000	16.0	366	159,400	10.5	346
		1500	1/2	153,400	11.0	355	147,600	8.0	336
			2	155,700	9.5	358	156,500	8.0	351
WS296A	5Cr–1.5Fe	1400	1/2	165,900	12.0	370	171,800	12.0	386
	(Actual 4.30Cr–1.43Fe)		2	166,500	11.5	385	173,900	10.0	388
		1500	1/2	156,800	11.0	357	162,000	8.0	387
			2	142,700	3.0	363	167,800	7.0	398
WS287A	5Mo–4Fe	1400	1/2	181,000	11.0	376	178,500	10.0	374
	(Actual 4.79Mo–3.43Fe)		2	179,300	10.0	369	181,000	7.0	392
		1500	1/2	168,800	1.0	445	180,700	0.0(5)	455
			2	160,000	1.0(5)	449	154,500	0.5	452

(1) All alloys were melted as 1-pound ingots and tested as 14-gage (0.064-inch) sheet.

(2) Annealing treatments were carried out in an argon atmosphere followed by cooling in still air.

(3) This prior annealing treatment was 4 hours at 1250 F in dry argon followed by furnace cooling.

(4) Substandard specimen 0.375 × 0.064 × 5 inches with a reduced section 0.250 × 0.064 × 1-1/4 inches.

(5) Broke outside gage marks. Uniform elongation recorded.

The results of these tests are encouraging. They indicate that relatively high strengths may be retained in many alloys after a simple annealing treatment. Such a treatment is mandatory to eliminate the effects of fabrication variables.

Five-Pound Ingots

Fabrication

A somewhat different fabrication procedure was necessary for the five-pound ingots than is generally used for one-pound ingots. These particular ingots were 2-1/2 inches in diameter by about 6 inches long. To permit an upset-forging operation, they were cut into two sections approximately 3 inches long. The forged slabs were too large to be cross rolled into sheet. Therefore, the rolling procedure consisted of rolling in one direction until a width of about 6 inches was attained, at which point the slabs were about 1/4 inch thick. The slabs were then rolled at 90° to the original rolling direction to 14-gage (0.064 inch) sheet for testing. The final rolling of each ingot section was in a direction parallel to the long axis of the original ingot. Forging temperatures were 1750 to 1800 F and rolling was carried out at 1450 F.

Location of Test Specimens

A method of locating the position of test specimens in the sheet, relative to the original ingot, was devised in order to establish the effects of

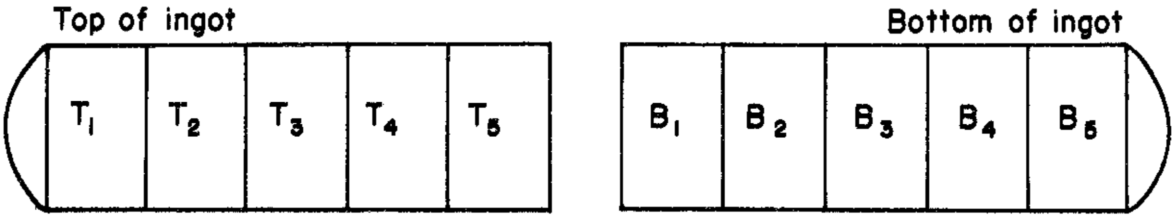


FIGURE 27. METHOD OF NUMBERING SHEET SECTIONS FOR APPROXIMATE LOCATION OF TEST SPECIMENS RELATIVE TO ORIGINAL INGOT

A-3265

any inhomogeneity on mechanical properties. This location system is illustrated in Figure 27. Sections five inches long were cut transverse to the direction of rolling and numbered consecutively from the end corresponding to the original top of the ingot. The letter "T" denotes sections cut from the top half of the original ingot, and the letter "B" denotes sections cut from the bottom half. After numbering, each sheet section was radiographed so that any areas of tungsten contamination could be avoided in the selection of test specimens.

As-Hot-Rolled Properties

Although it has been shown that the properties of hot-rolled material are not consistently reproducible, the large amount of data on hot-rolled alloys makes it desirable to continue to use this condition as a basis for comparison. Another reason for continuing this practice is that, eventually, adequate production control should permit the marketing of commercial titanium alloys in the hot-rolled condition.

Two tensile specimens and one bend specimen were taken from the top and bottom sections of each alloy and tested in the as-hot-rolled condition. Results of these tests are presented in Table 30. The actual chemical analyses shown in the table were obtained on shearings taken from the broken tensile specimens.

The Ti-3.5%Cr-3.5%V alloy had consistent tensile properties despite the large differences in chemical composition from top to bottom. The specimens taken from the top and bottom of the ingots of the Ti-5%Mn-3.5%Cr alloy showed a wide inhomogeneity in analysis and correspondingly large variations in properties. The chemical compositions of the top and bottom specimens of the 4%Fe complex alloy differed only slightly, but the properties were not uniform. Unless this alloy is extremely sensitive to composition, the large differences in tensile properties must have been due to fabrication variables. On the other hand, the 3%Mn complex alloy had fairly consistent properties.

The bend ductility of different specimens of the same heat varied widely for all alloys. Likewise, there seemed to be little correlation between the tensile ductility and the bend ductility. For instance, in the 3%Mn complex alloy, bend specimens taken from the same sections of sheet as the tensile specimens varied from 8T to 1.5T. This suggests the possibility that mechanical defects in the specimens or surface contamination might be a major factor in determining the bend properties of the sheet. However, the variables encountered during processing undoubtedly have a

TABLE 30. AS-HOT-ROLLED PROPERTIES OF SELECTED HIGH-STRENGTH ALLOYS IN THE FORM OF 14-GAGE SHEET[1]

Heat No.	Intended Composition, %	Specimen Location	Actual Composition, %	Ultimate Tensile Strength, psi[2]	Elongation, % in 1 inch	VHN	Minimum Bend Radius, T[3]
WT4A	3.5Cr-3.5V	T4	3.65Cr-5.06V	174,700	10.5	394	1
		T4		174,700	8.0		
		B2	3.65Cr-3.19V	179,700	9.0	376	3
		B2		182,300	9.0		
WT15A	5Mn-2.5Cr	T4	4.48Mn-2.00Cr	191,100	5.0	429	3
		T4		197,500	6.0		
		B4	3.12Mn-1.37Cr	178,600	8.0	405	8
		B4		167,700	12.0		
WT25A	1Cr-1Fe-3Mn-1Mo-1V	T3	0.90Cr-1.00Fe-3.74Mn-1.05Mo-1.01V	192,500	5.0	417	8
		T3		193,400	6.0		
		B4	1.20Cr-0.92Fe-3.48Mn-1.05Mo-0.98V	191,600	6.0	405	1.5
		B4		186,300	7.5		
WT32A	1Cr-4Fe-1Mn-1Mo-1V	T4	0.76Cr-3.24Fe-1.65Mn-1.05Mo-0.98V	223,800	2.0	442	24
		T4		220,900	2.5		
		B5	0.90Cr-3.15Fe-1.69Mn-1.12Mo-0.93V	182,100	5.0	366	3
		B5		176,000	8.0		

(1) Fabricated from 5-lb ingots.

(2) Substandard specimen 0.375 x 0.064 x 5 inches with a reduced section 0.250 x 0.064 x 1-1/4 inches.

(3) Specimen size 0.375 x 0.064 x 4 inches.

major effect on bend ductility as well as on the other properties. This experiment demonstrates further the necessity for heat treating or annealing rolled sheet to produce uniform properties.

Heat Treatment

Process Annealing. Tensile blanks were cut from sheet sections representing the top and bottom halves of the four alloy ingots. These were heated for 1/4 and 1 hour at 1300, 1400, or 1500 F and air cooled. Heating was done in a dried-argon atmosphere. Bend specimens were annealed for 1/4 hour at 1300 F and 1/4 and 1 hour at 1400 F. Complete results of tests made on the annealed specimens are given in Table 31.

The tensile strength of all four alloys annealed at 1300 F fell within the relatively narrow range of about 134,000 psi to 154,000 psi. The ductilities of all except the Ti-3.5%Cr-3.5%V alloy ranged from 17 to 21 per cent. The latter alloy had elongation values ranging from 11 to 15 per cent in 1 inch. Bend ductility for all alloys annealed for 1/4 hour at 1300 F was 1T or less. The maximum spread in tensile strengths for any one alloy was of the order of 10,000 psi. Thus, it appears that annealing at 1300 F tends to minimize variations in mechanical properties caused by differences in chemical composition, or uncontrollable fabrication variables. The tensile strengths produced by this treatment were relatively low, but the ductility was excellent. Increasing the annealing time from 1/4 to 1 hour at temperature seems to have little effect on the resultant properties. The 4%Fe complex alloy had the best combination of properties after the 1300 F annealing treatment.

Annealing at 1400 F caused wider spread of properties in the four alloys. Tensile strengths ranged from a low of about 150,000 psi for the Ti-3.5%Cr-3.5%V alloy to a high of about 190,000 psi for the 4%Fe complex alloy. The elongation for all of the alloys except the 4%Fe complex alloy generally ranged from 8 to 11 per cent in 1 inch. The latter alloy had somewhat lower ductility commensurate with its higher strength. Bend ductility of specimens annealed at 1400 F varied over a considerable range, and there was no consistent correlation with tensile properties. The tensile properties of the specimens from an individual heat were reasonably consistent after this annealing treatment.

The 1500 F annealing treatment resulted in generally lower tensile strengths than those obtained after the 1400 F anneal. The major exception to this was the T4 specimen of the 4%Fe complex alloy annealed for one hour at 1500 F. This specimen had the highest tensile strength encountered in these tests. The other three specimens of this alloy were too brittle to attain a true tensile strength. Variations in tensile properties of the specimens from a single heat were somewhat greater than those obtained on the specimens annealed at 1400 F. This was particularly true of the tensile ductility. It appears, then, that 1500 F is too high a temperature for annealing these alloys.

TABLE 31. MECHANICAL PROPERTIES OF SELECTED HIGH-STRENGTH ALLOYS(1) AFTER VARIOUS PROCESS-ANNEALING TREATMENTS(2)

Heat No.	Intended Composition, %	Annealing Temp, F	Annealing Time, hours	Specimen Location	Ultimate Tensile Strength, psi(3)	Elongation, % in 1 inch	VHN	Minimum Bend Radius, T(4)
WT4A	3.5Cr-3.5V	1300	1/4	T4	152,900	15.0	328	1
				B2	145,600	12.0	312	1
			1	T4	146,300	13.5	328	--
				B2	156,700	11.0	346	--
		1400	1/4	T4	161,100	9.5	351	3
				B2	160,300	11.0	344	3
			1	T5	160,000	9.0	359	3
				B2	180,500	4.0	371	8
		1500	1/4	T4	144,200	11.0	333	--
				B2	158,000	5.0	351	--
			1	T5	151,600	8.0	330	--
				B2	143,900	6.5	318	--
WT15A	5Mn-2.5Cr	1300	1/4	T3	136,600	21.0	309	0.25
				B4	135,500	19.0	306	0.25
			1	T3	140,100	21.0	319	--
				B4	134,600	20.0	306	--
		1400	1/4	T3	178,600	8.0	371	3
				B4	167,300	11.0	370	4
			1	T3	172,300	9.0	364	8
				B4	167,700	9.0	388	3
		1500	1/4	T3	152,100	11.0	354	--
				B4	157,400	7.5	358	--
			1	T3	163,600	9.0	363	--
				B4	148,400	12.0	333	--

TABLE 31. (Continued)

Heat No.	Intended Composition, %	Annealing Temp, F	Annealing Time, hours	Specimen Location	Ultimate Tensile Strength, psi(3)	Elongation, % in 1 inch	VHN	Minimum Bend Radius, T(4)
WT25A	1Cr-1Fe-3Mn-1Mo-1V	1300	1/4	T4	140,700	21.0	317	0.25
				B4	136,300	18.0	322	0.75
			1	T4	141,300	19.5	321	--
				B4	133,800	20.0	313	--
		1400	1/4	T4	174,100	11.0	381	3
				B4	164,500	11.5	372	3
			1	T4	177,900	6.0	373	12
				B4	168,300	10.0	365	8
WT25A	1Cr-1Fe 3Mn 1Mo-1V	1500	1/4	T4	162,000	9.0	377	--
				B4	149,400	12.0	354	--
			1	T4	175,900	6.0	390	--
				B4	143,300	10.0	354	--
WT32A	1Cr-4Fe-1Mn-1Mo-1V	1300	1/4	T4	154,300	18.0	342	1
				B3	149,400	19.0	333	0.5
			1	T4	151,700	17.0	336	--
				B4	143,700	18.0	327	--
		1400	1/4	T4	185,900	8.0	385	8
				B4	192,300	2.0(6)	404	8
			1	T4	190,400	4.0	405	8
				B4	193,700	2.0	425	12
		1500	1/4	T4	(5)	(5)	493	--
				B4	148,000	0.5	433	--
			1	T4	201,800	3.0	443	--
				B5	106,300	0.0	523	--

(1) All alloys were in the form of 14-gage (0.064") sheet rolled from 5-lb ingots.
(2) Specimens were heated in a dried argon atmosphere and cooled in still air.
(3) Substandard specimen 0.375 x 0.064 x 5 inches with a reduced section 0.250 x 0.064 x 1-1/4 inches.
(4) Specimen size 0.375 x 0.064 x 4 inches.
(5) Specimen too hard to grip.
(6) Specimen broke outside gage marks. Uniform elongation recorded.

From the above results, it appears that the effects of inhomogeneity of material and/or variations in fabricating procedure may be minimized in sections of this size by process-annealing treatments involving heating to 1300 or 1400 F and air cooling. A time at temperature of 1/4 hour appears to be sufficient. The 1300 F treatment produces very consistent tensile and bend properties, but the tensile strength is rather low. Annealing at 1400 F increases strength appreciably, but it is still considerably lower than that obtained in the as-hot-rolled condition in most of the alloys tested. An exception to this trend was in the specimens of the 4%Fe complex alloy taken from the bottom section of the ingot and annealed at 1400 F. These specimens had somewhat higher tensile strength than that obtained in the as-hot-rolled condition.

Age Hardening. The results of a limited investigation of the effects of age hardening on the mechanical properties of a series of Ti-Mn-Cr alloys were presented earlier in this report. These results were considered sufficiently encouraging to warrant further investigation of solution treating and aging and their effects on mechanical properties. Since the change in hardness is the most convenient means of measuring the effects of aging treatments, aging curves were determined for the four selected alloys described in the preceding sections, using Vickers hardness as a criterion.

Small specimens of the 14-gage sheet of each alloy were solution treated at 1300, 1400, and 1600 F by heating in an argon atmosphere for one hour and quenching in ice water (0 C). Preliminary work had shown these solution treatments to produce structures consisting of approximately 20 to 30 per cent, 50 per cent, and 100 per cent beta, respectively. Solution-treated specimens of each alloy were aged at 100 F intervals in the temperature range 200 to 900 F for periods of 1/2 to 24 hours and water quenched. All specimens were mounted in a room-temperature-setting resin to eliminate any reheating, given a metallographic polish, and subjected to Vickers hardness determinations. Hardness measurements were made on the cross section of the sheet specimens approximately midway between the two rolled surfaces. Average results of these tests are presented graphically in Figures 28 through 43. Note that Figures 28 to 31 do not contain hardness data for the Ti-3.5%Cr-3.5%V alloy solution treated at 1600 F. The data obtained after this treatment were too erratic to be of any use. The reason for the wide scatter in hardness of these specimens was found to be alloy segregation, indicated by gross differences in microstructure. Many of the specimens contained alternating bands of transformed and retained beta parallel to the rolled surfaces. Hardness determinations made on the two types of microstructure differed by as much as 100 VHN in the same specimen. Specimens of the same alloy solution treated at 1300 and 1400 F, in the alpha-beta range, showed very consistent results in the aged condition, indicating again that heat treatment in this temperature range tends to minimize the effects of inhomogeneity.

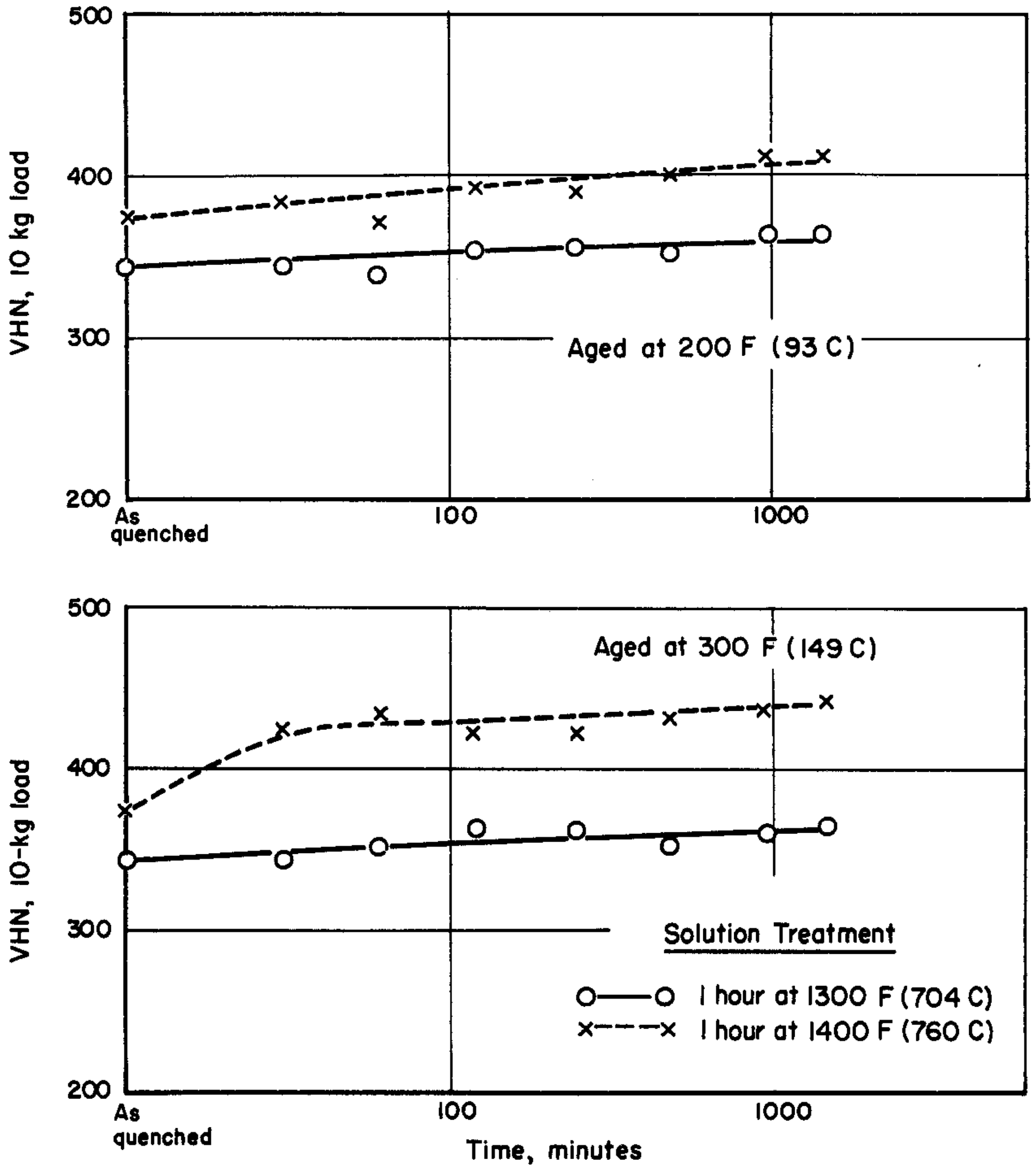


FIGURE 28. HARDNESS VERSUS AGING TIME FOR A Ti-3.5% Cr-3.5% V ALLOY SOLUTION TREATED AT VARIOUS TEMPERATURES

A-3266

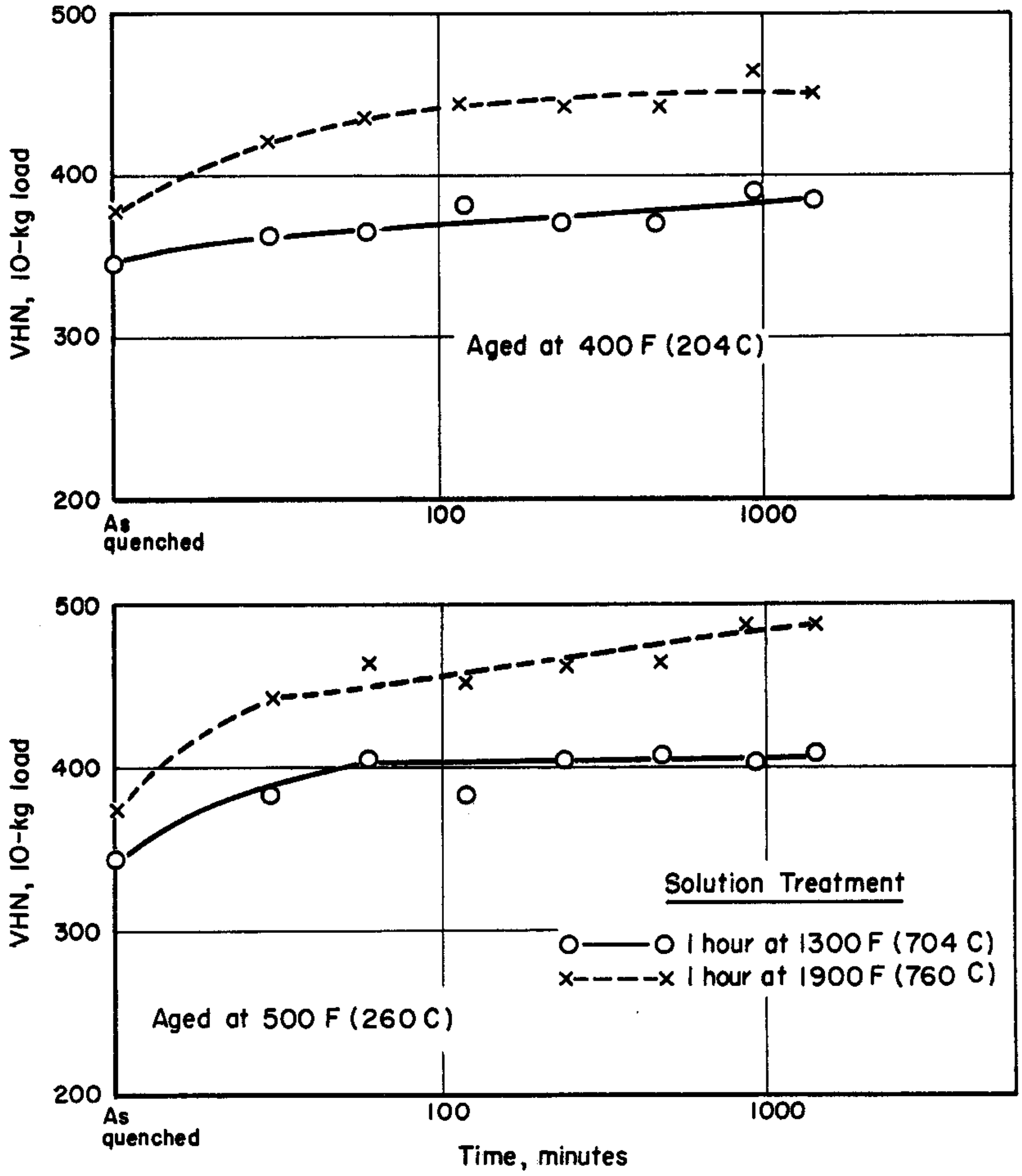


FIGURE 29. HARDNESS VERSUS AGING TIME FOR A Ti-3.5 % Cr-3.5 % V ALLOY SOLUTION TREATED AT VARIOUS TEMPERATURES

A-3267

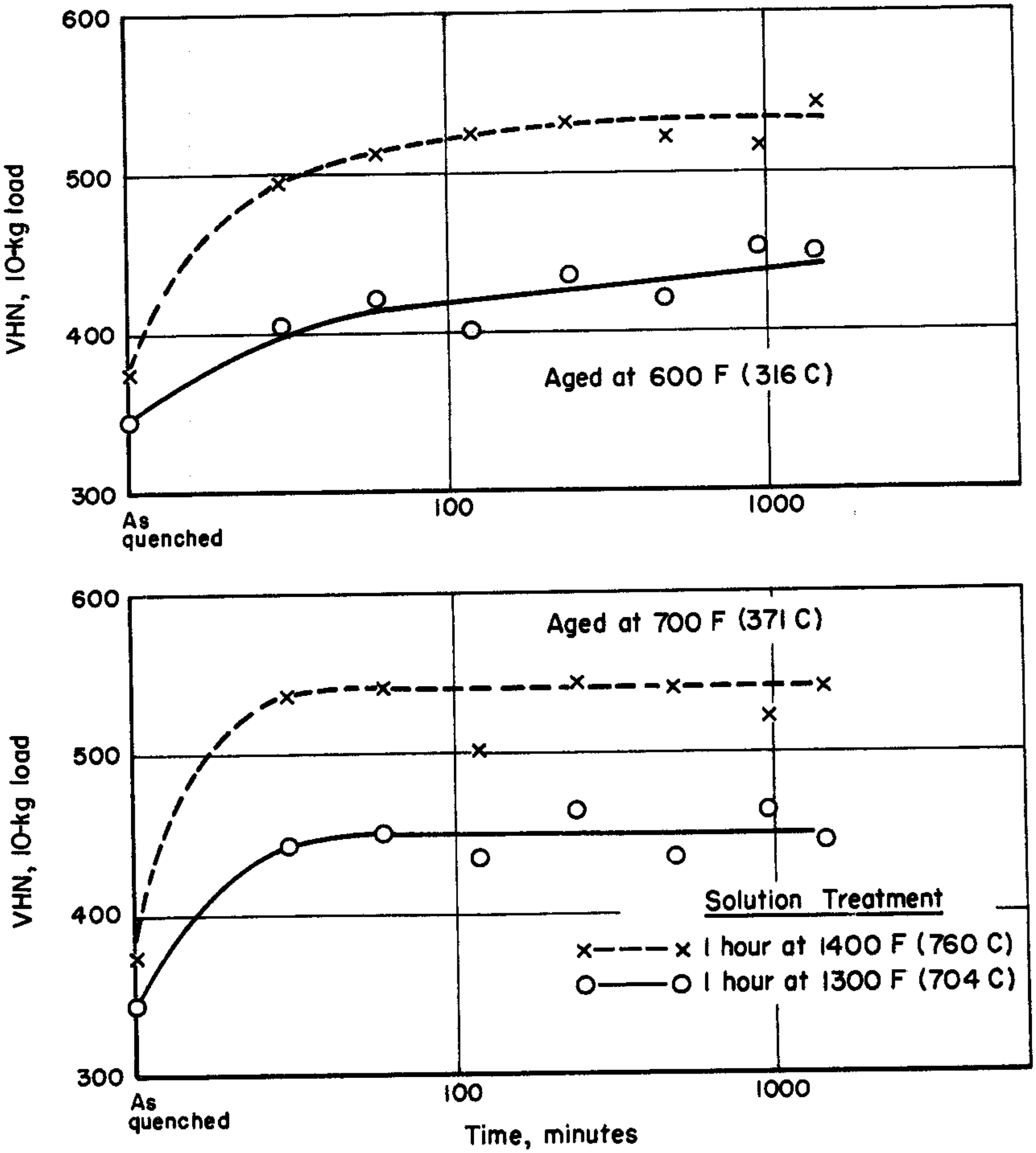


FIGURE 30. HARDNESS VERSUS AGING TIME FOR A Ti-3.5% Cr-3.5% V ALLOY SOLUTION TREATED AT VARIOUS TEMPERATURES

A-3268

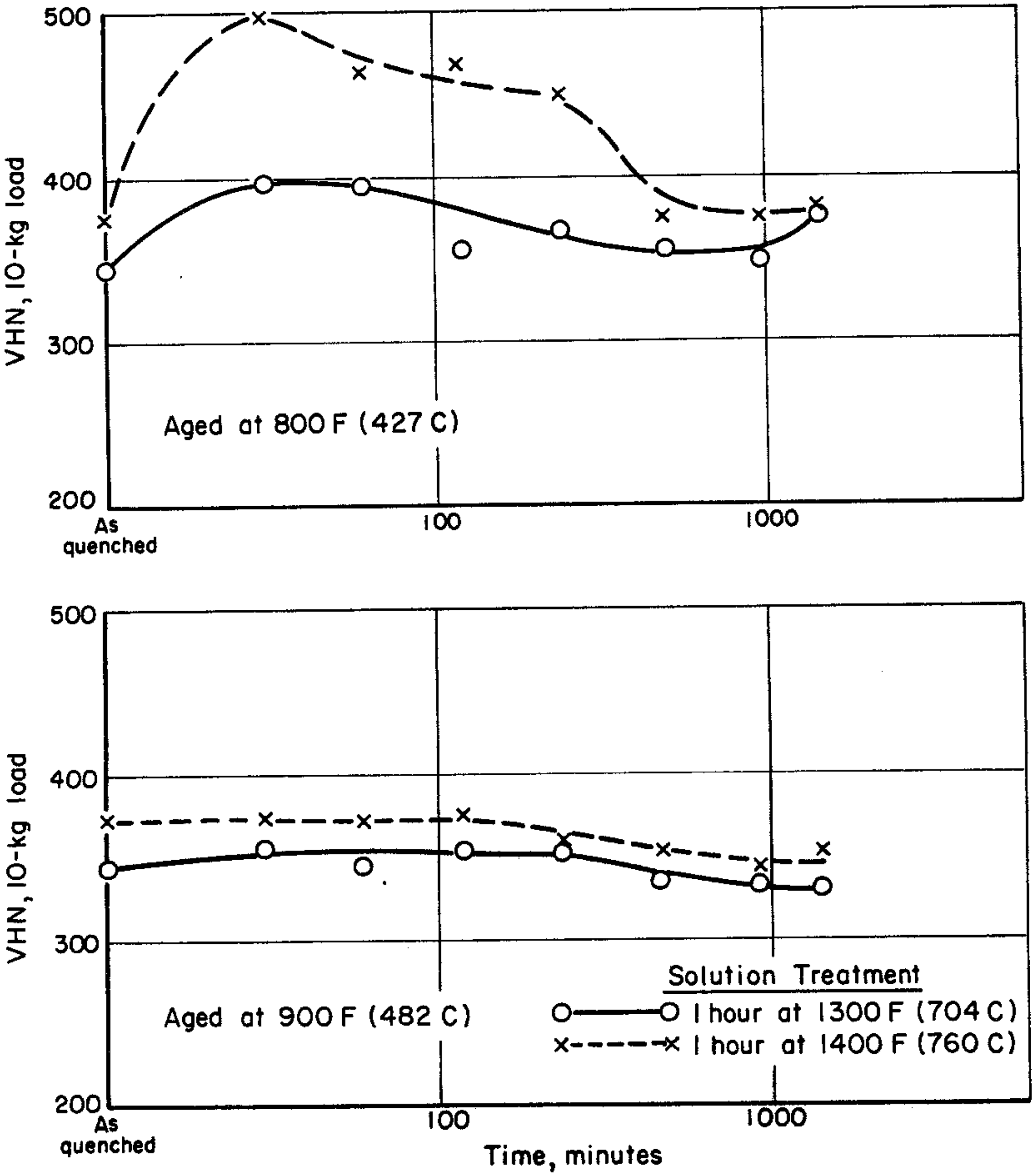


FIGURE 31. HARDNESS VERSUS AGING TIME FOR A Ti-3.5% Cr-3.5% V ALLOY SOLUTION TREATED AT VARIOUS TEMPERATURES

A-3269

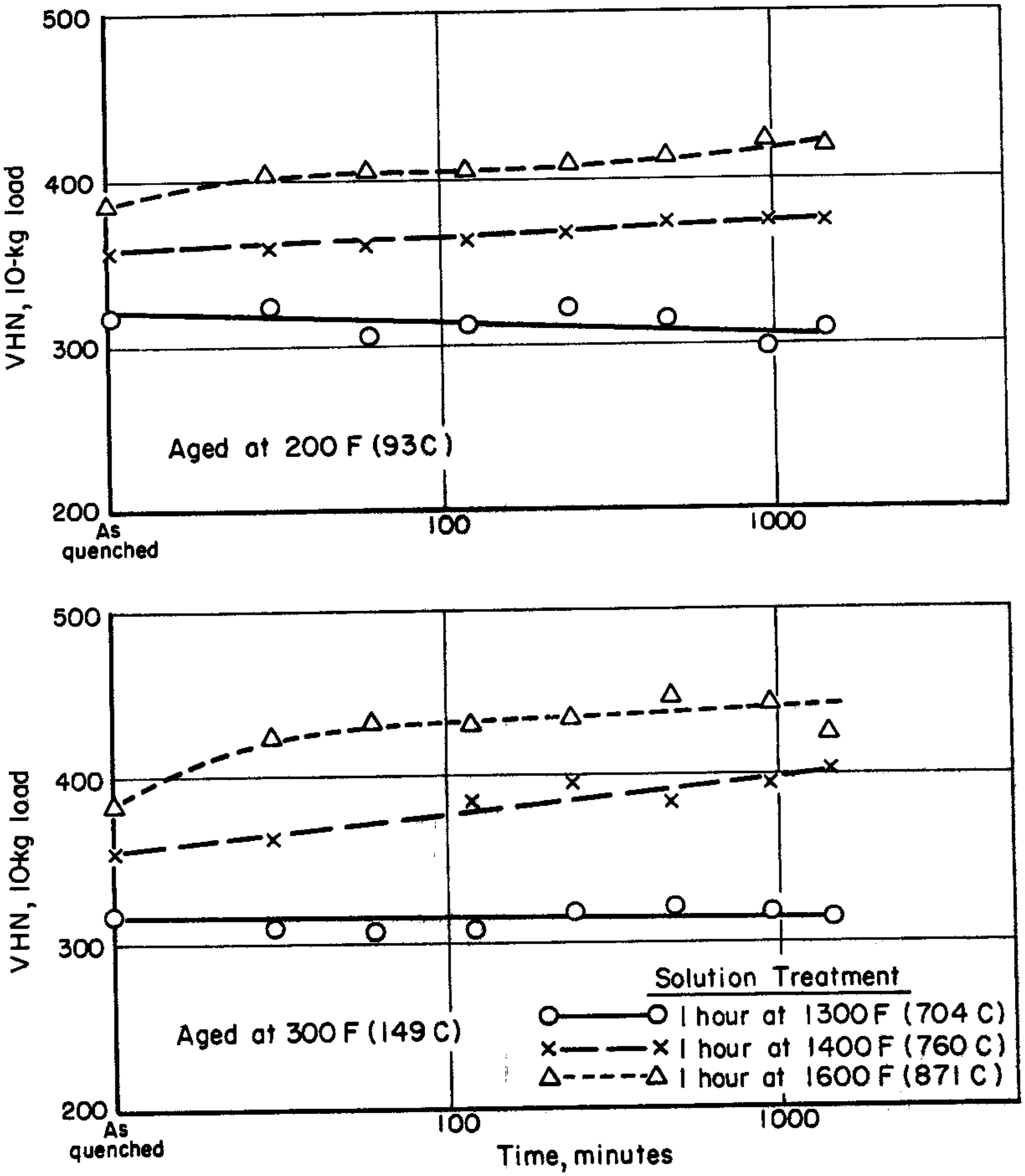


FIGURE 32. HARDNESS VERSUS AGING TIME FOR A Ti-5% Mn-2.5% Cr ALLOY SOLUTION TREATED AT VARIOUS TEMPERATURES

A-3270

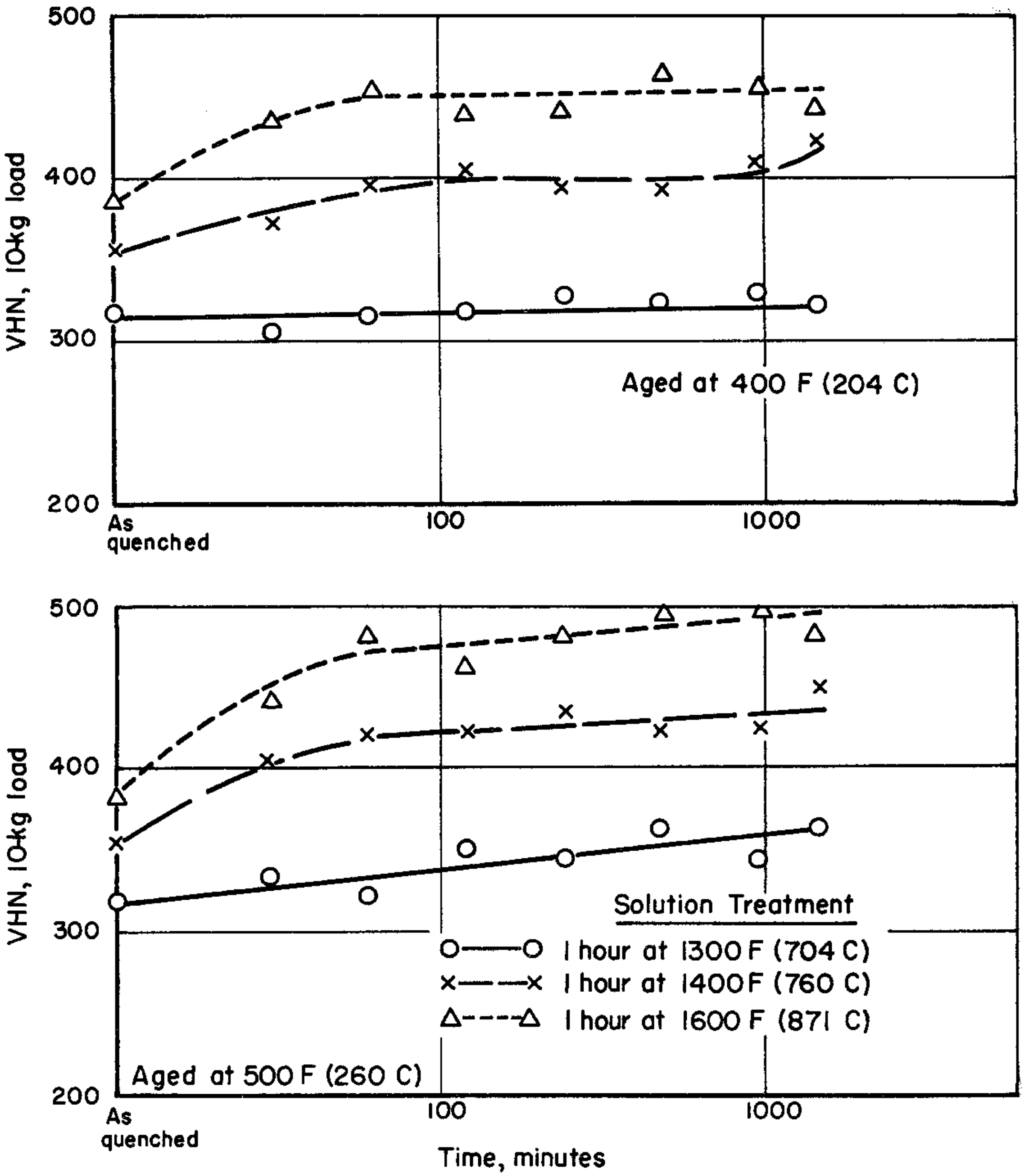


FIGURE 33. HARDNESS VERSUS AGING TIME FOR A Ti-5 % Mn-2.5 % Cr ALLOY SOLUTION TREATED AT VARIOUS TEMPERATURES

A-3271

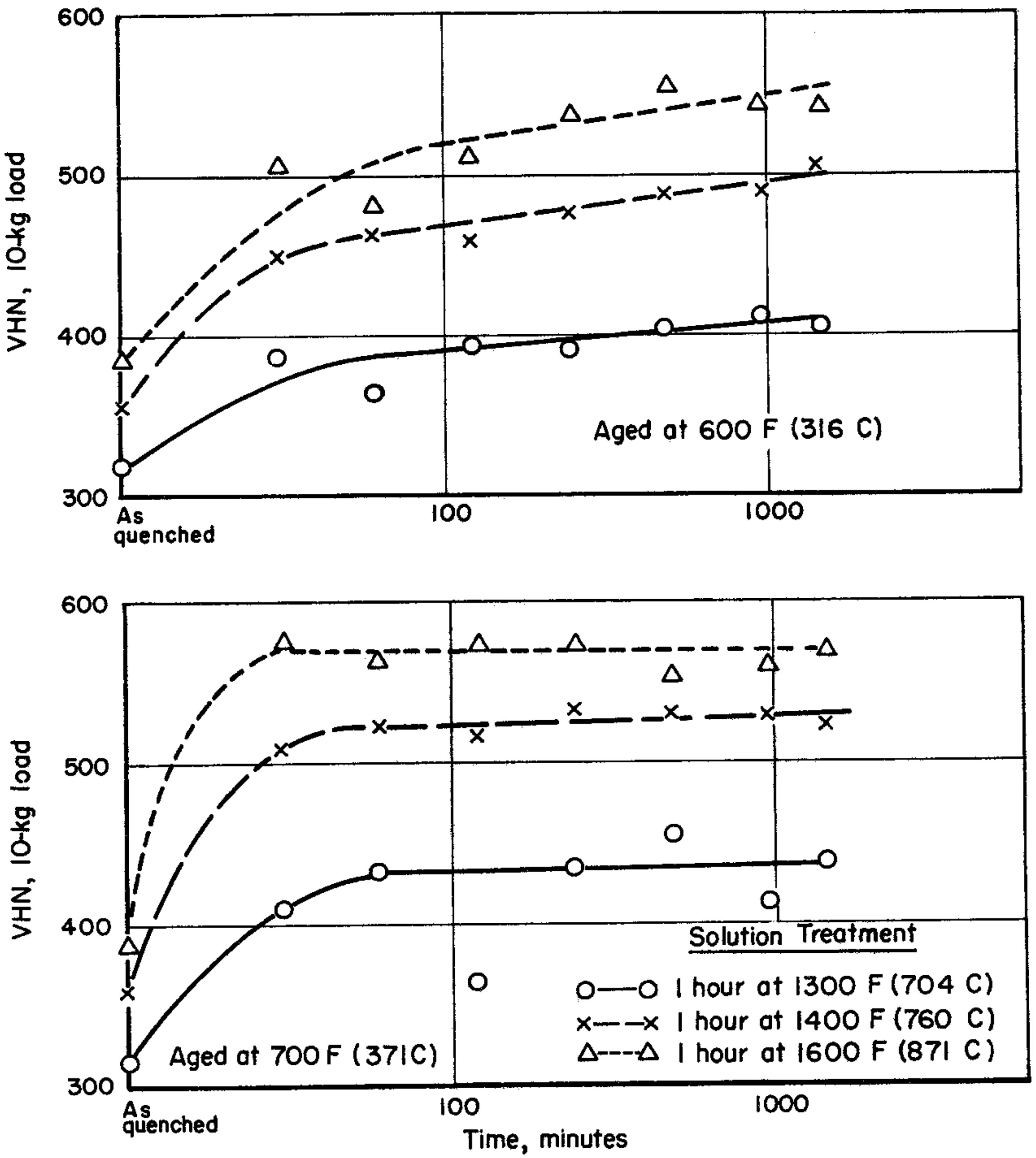


FIGURE 34. HARDNESS VERSUS AGING TIME FOR A Ti-5% Mn-2.5% Cr ALLOY SOLUTION TREATED AT VARIOUS TEMPERATURE

A-3272

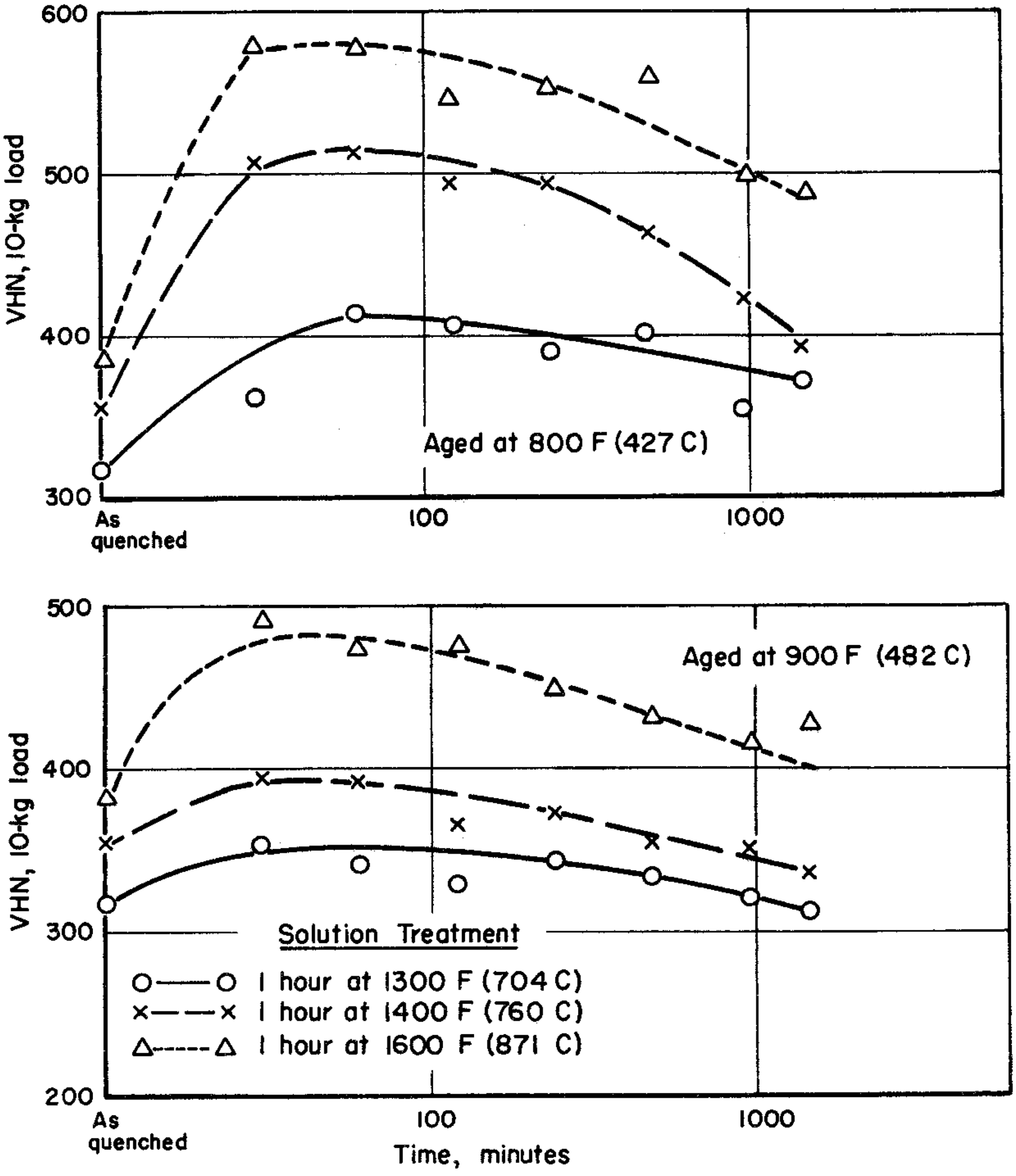


FIGURE 35. HARDNESS VERSUS AGING TIME FOR A Ti-5% Mn-2.5% Cr ALLOY SOLUTION TREATED AT VARIOUS TEMPERATURES

A-3273

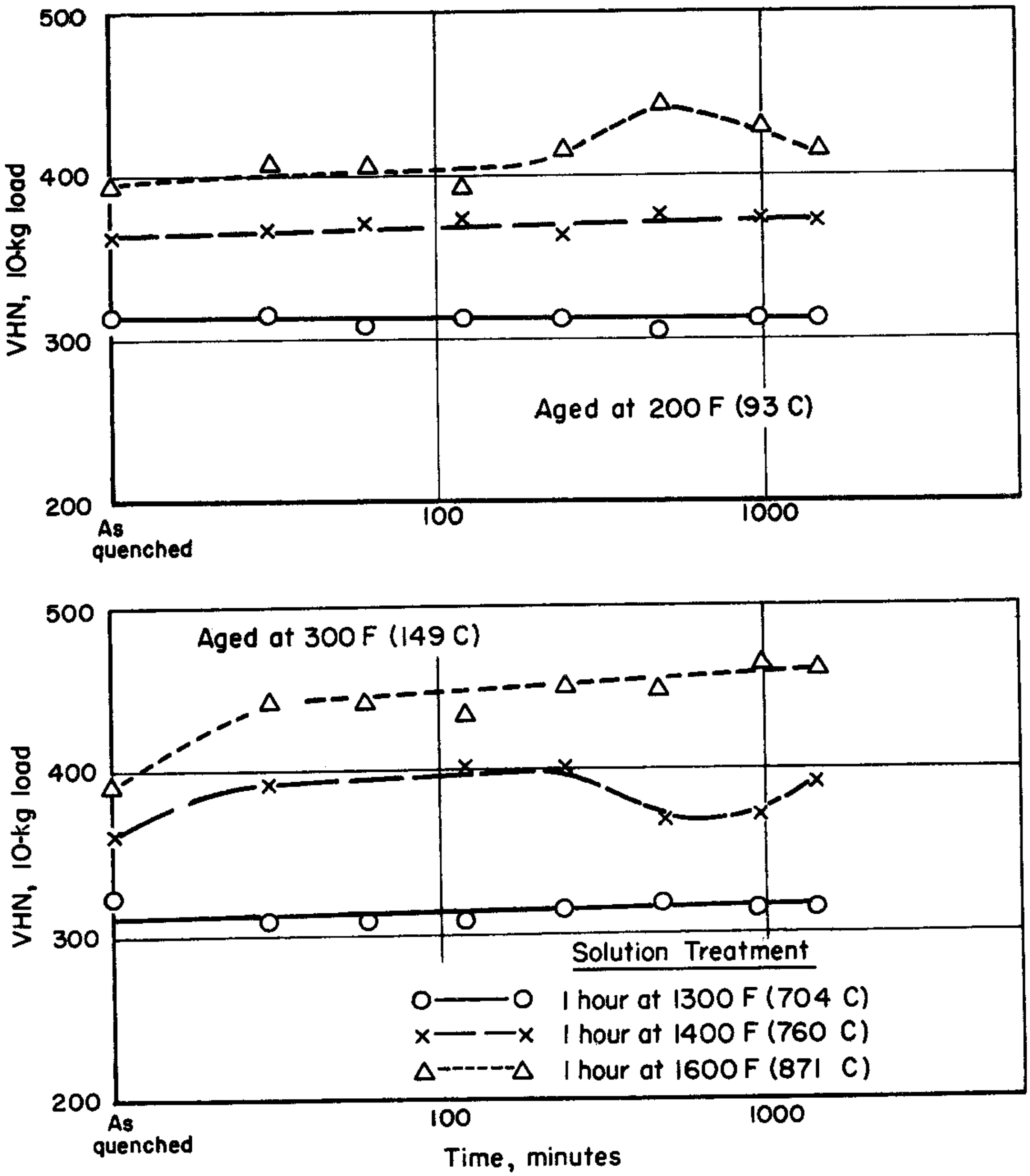


FIGURE 36. HARDNESS VERSUS AGING TIME FOR A Ti-1% Cr-1% Fe-3% Mn-1% Mo-1% V ALLOY SOLUTION TREATED AT VARIOUS TEMPERATURES

A-3274

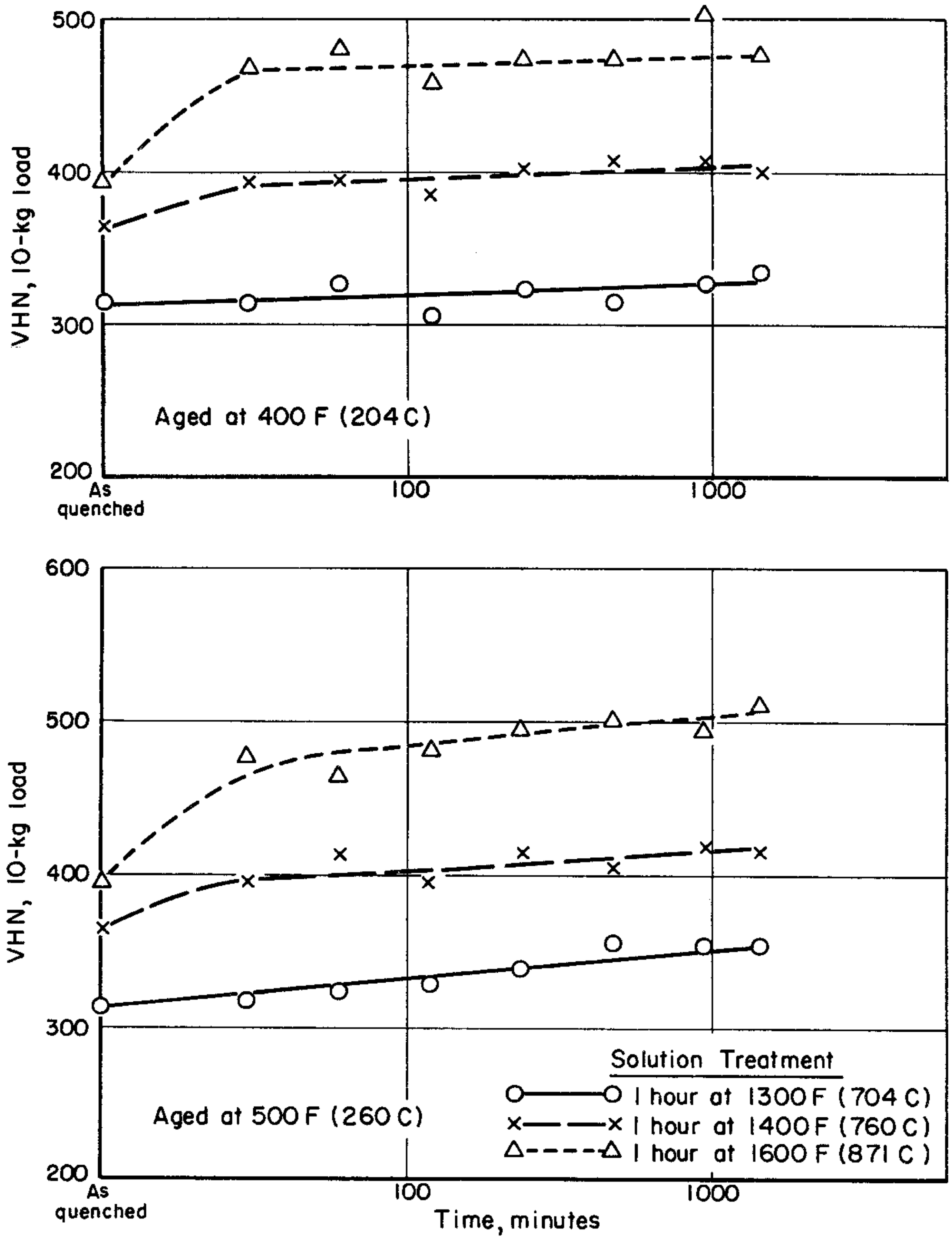


FIGURE 37. HARDNESS VERSUS AGING TIME FOR A Ti-1% Cr-1% Fe-3% Mn-1% Mo-1% V ALLOY SOLUTION TREATED AT VARIOUS TEMPERATURES

A-3275

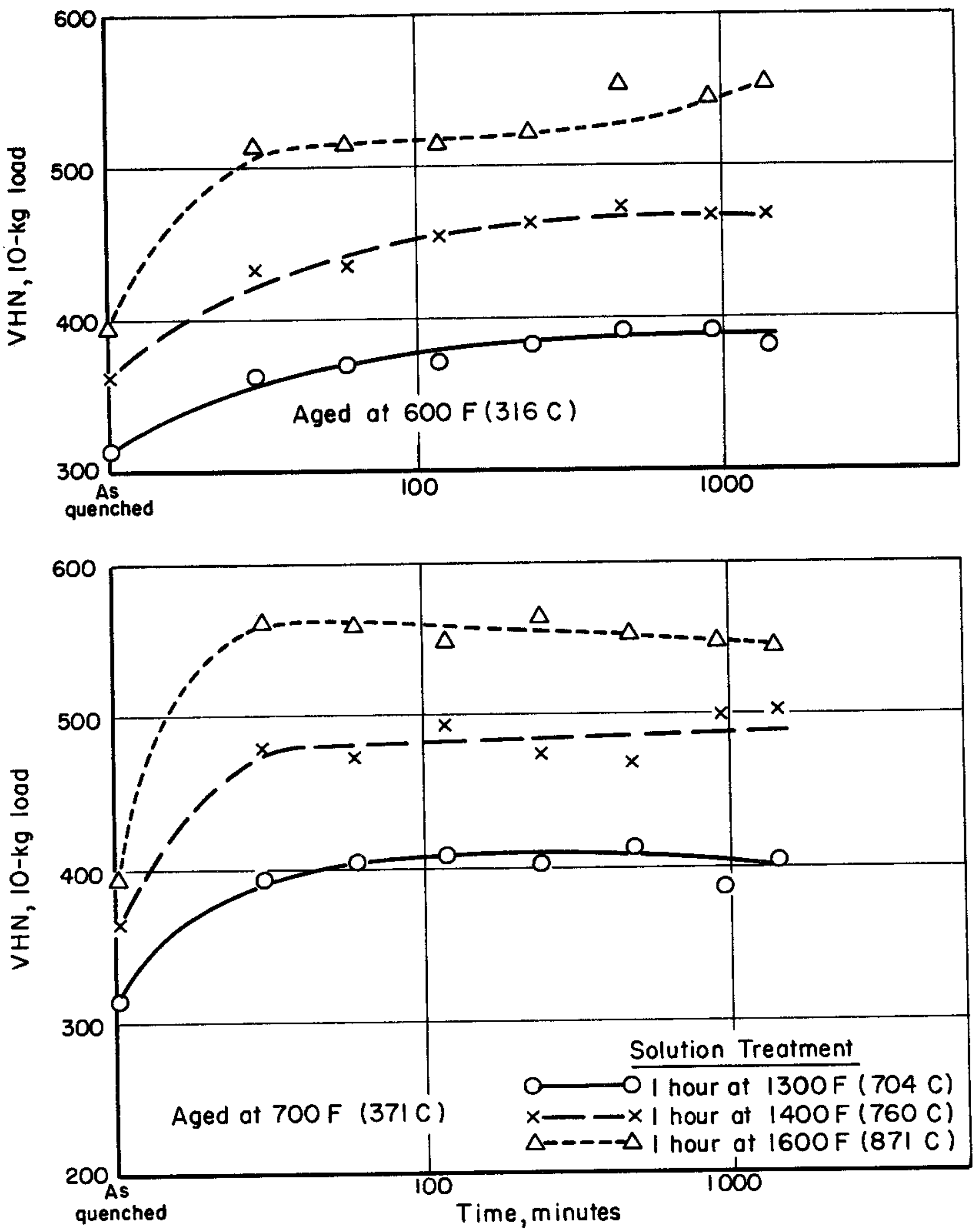


FIGURE 38. HARDNESS VERSUS AGING TIME FOR A Ti-1% Cr-1% Fe-3% Mn-1% Mo-1% V ALLOY SOLUTION TREATED AT VARIOUS TEMPERATURES

A-3276

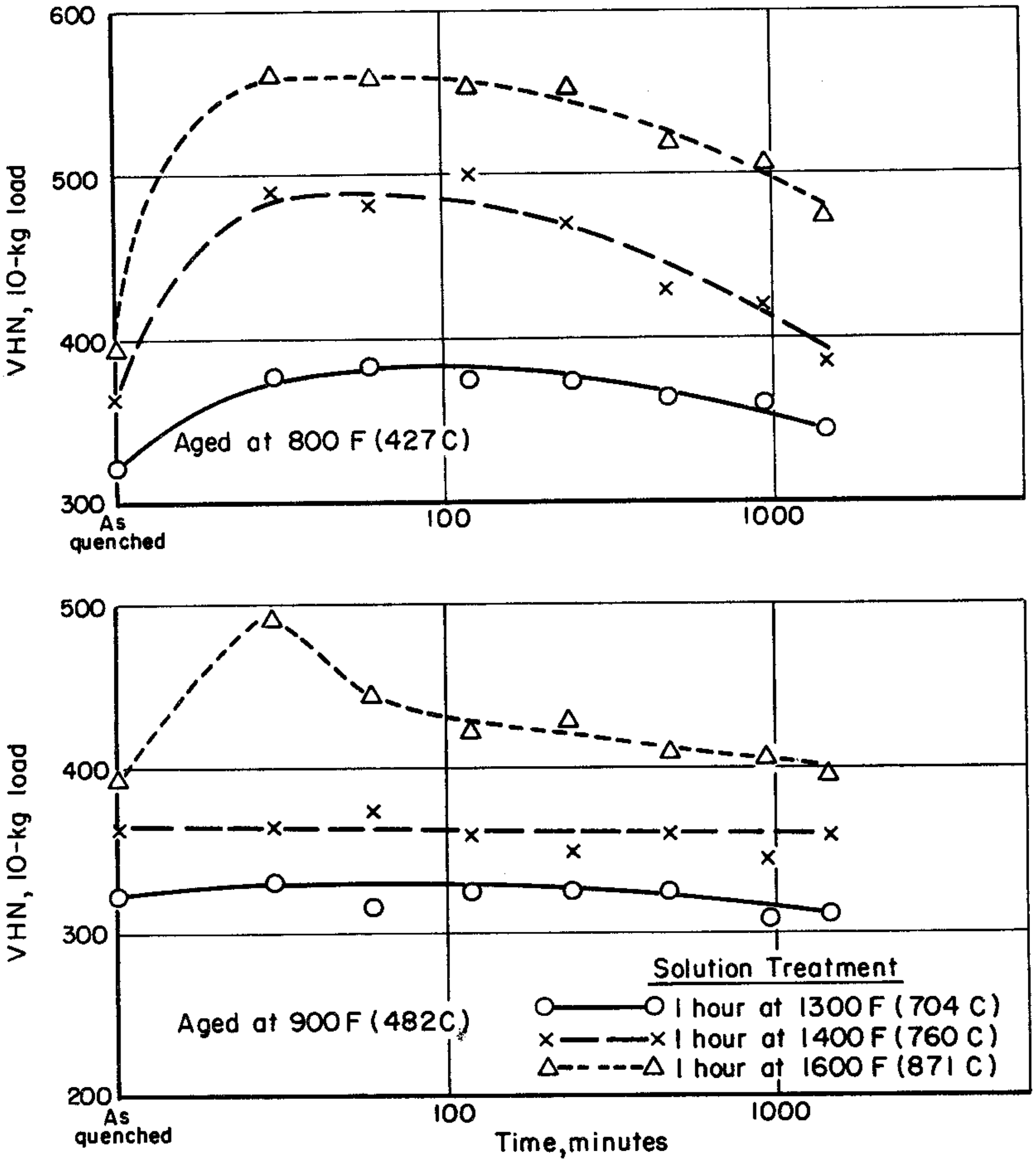


FIGURE 39. HARDNESS VERSUS AGING TIME FOR A Ti-1% Cr-1% Fe-3% Mn-1% Mo-1% V ALLOY SOLUTION TREATED AT VARIOUS TEMPERATURES

A-3277

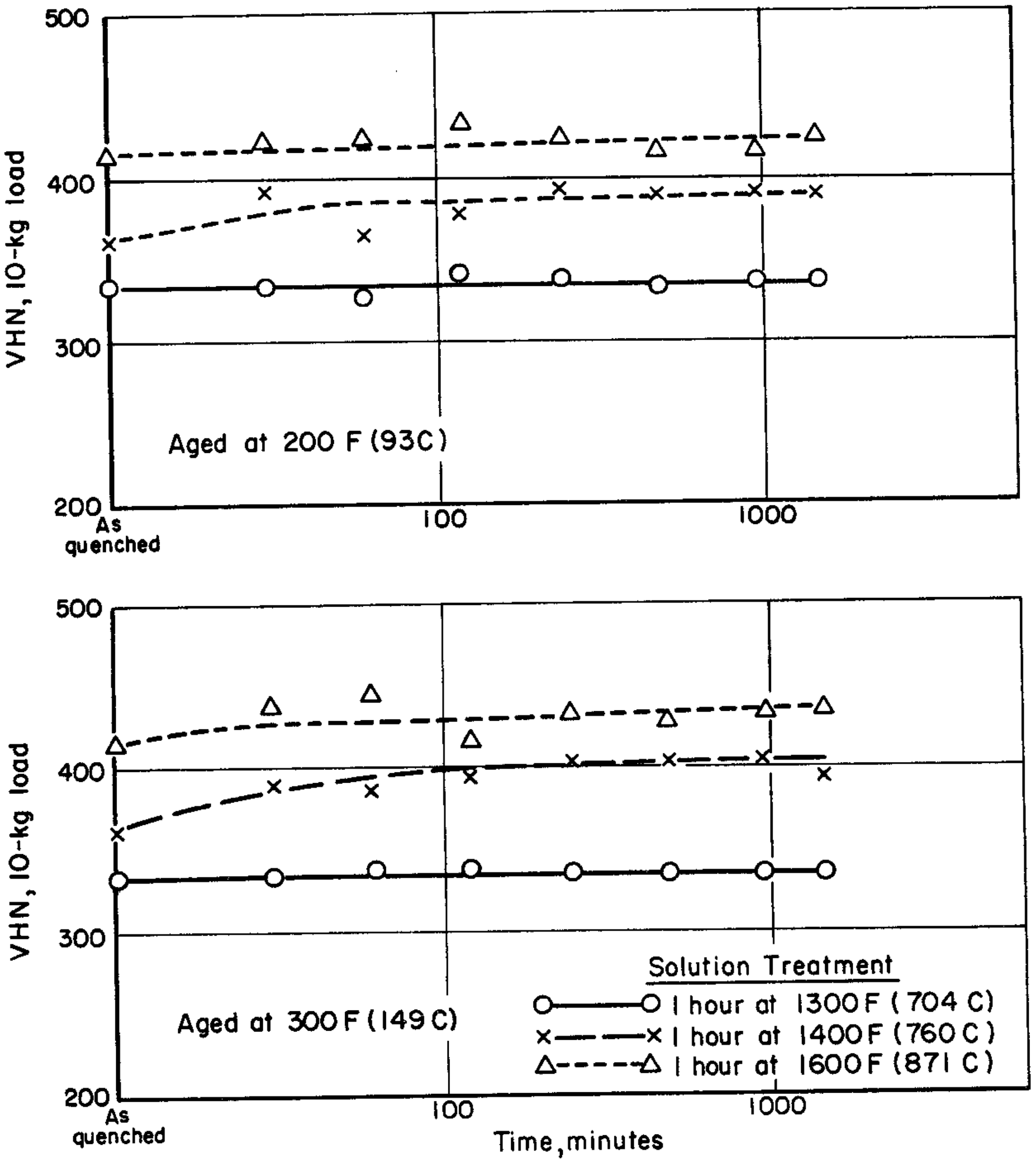


FIGURE 40. HARDNESS VERSUS AGING TIME FOR A Ti-1% Cr-4% Fe-1% Mn-1% Mo-1% V ALLOY SOLUTION TREATED AT VARIOUS TEMPERATURES

A-3278

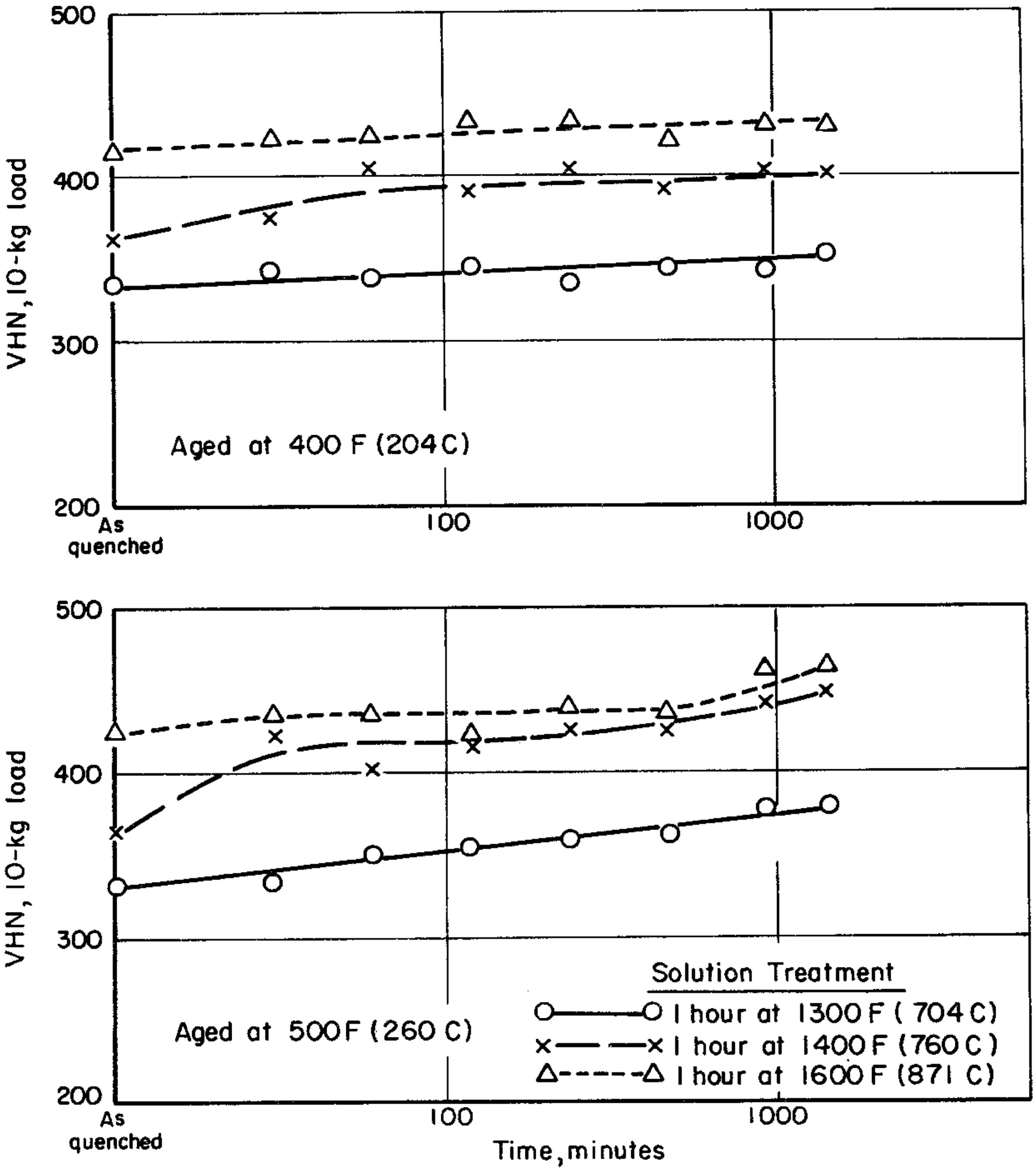


FIGURE 41. HARDNESS VERSUS AGING TIME FOR A Ti-1% Cr-4% Fe-1% Mn-1% Mo-1% V ALLOY SOLUTION TREATED AT VARIOUS TEMPERATURES

A-3279

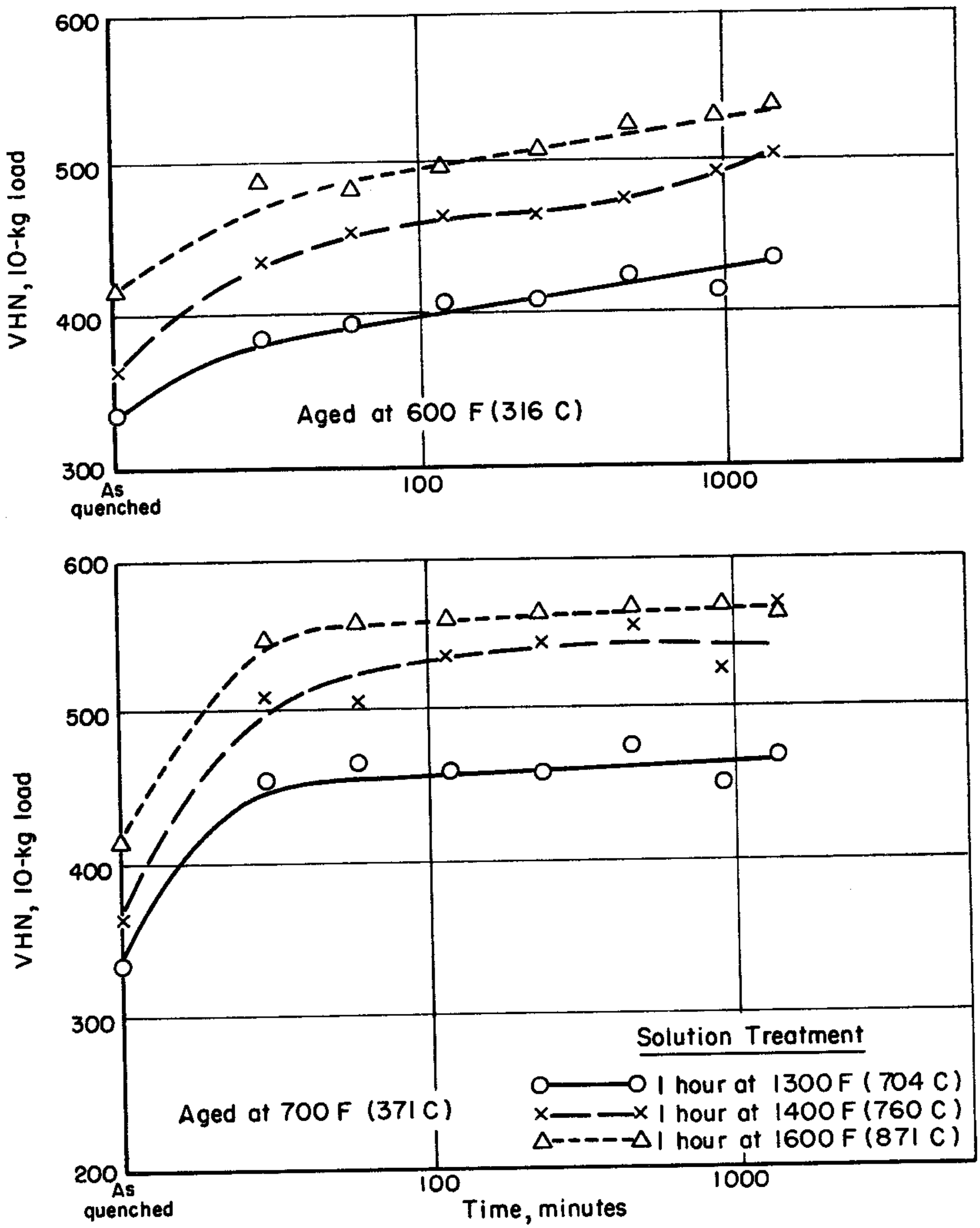


FIGURE 42. HARDNESS VERSUS AGING TIME FOR A Ti-1% Cr-4% Fe-1% Mn-1% Mo-1% V ALLOY SOLUTION TREATED AT VARIOUS TEMPERATURES

A-3280

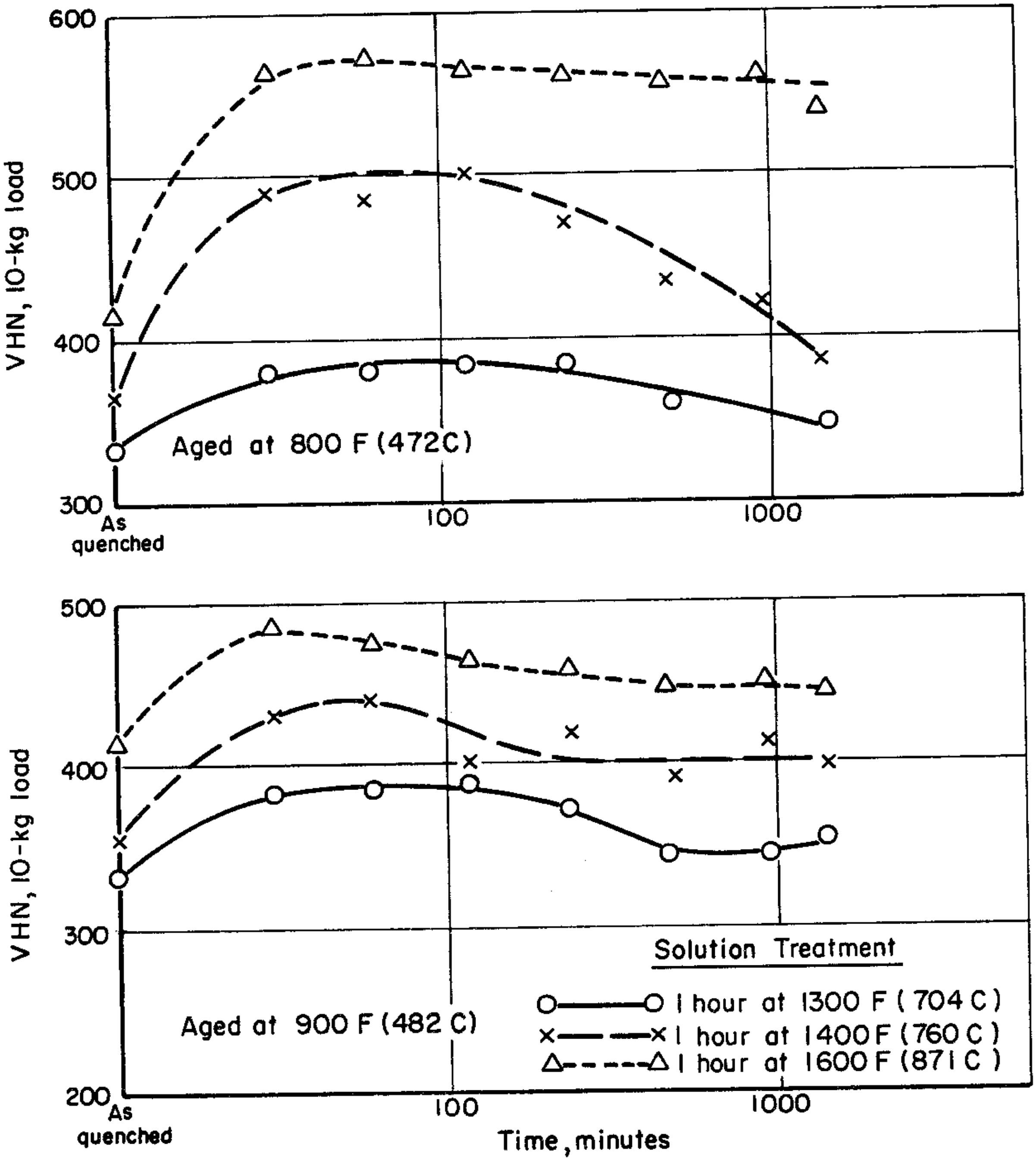


FIGURE 43. HARDNESS VERSUS AGING TIME FOR A Ti-1% Cr-4% Fe-1% Mn-1% Mo-1% V ALLOY SOLUTION TREATED AT VARIOUS TEMPERATURES

A-3281

All of the alloys showed essentially the same trends in the aged condition. Maximum hardness was attained at an aging temperature of 700 F for all of the solution treatments. The maximum hardness attained after the 1600 F solution treatment varied very little among the three alloys to which this treatment was applied. Hardness at a given aging temperature increased with increasing solution-treating temperature, and the rate of hardening followed the same trend. Hardness increases were detected in all alloys solution treated at 1400 or 1600 F and aged at temperatures as low as 200 F. This is in accordance with the results on low-temperature aging presented earlier in this report. All alloys overaged at a temperature of 800 F. From these aging curves, heat treatments will be selected which produce about the same hardness in the different alloys. These treatments will be applied to tensile and bend specimens of each of the alloys in order to determine the effects of the variables involved on the properties.

ISOTHERMAL-TRANSFORMATION STUDIES ON BINARY TITANIUM-CHROMIUM ALLOYS

Isothermal-transformation studies were made on 3, 5, and 8 per cent chromium alloys (nominal composition) to obtain fundamental information on the reactions occurring in beta-stabilized, titanium-base alloys. Such information will be useful in selecting treatments for promising high-strength alloys.

The equilibrium diagram of the Ti-Cr system is of the eutectoid type(4). The reported eutectoid composition is about 15 per cent chromium and the eutectoid temperature is 1265 F (685 C). Since the alloys used in this investigation were of the hypoeutectoid type, the expected microstructure for completely transformed specimens would be proeutectoid alpha and eutectoid.

Transformation characteristics of these alloys were studied by means of metallographic examination, X-ray diffraction, and Vickers hardness, and to a limited extent electron microscopy. The results have been summarized as time-temperature-transformation diagrams. A preliminary survey to determine the types of structures and hardnesses which could be developed by isothermal heat treatment was reported in Reference (2), pp 104-110.

Materials Used and Experimental Procedures

Three Ti-Cr alloys, with intended compositions of 3, 5, and 8 per cent chromium, were used. These were in the form of 14-gage (0.064 inch) hot-rolled sheet fabricated from arc-melted 3/4-pound ingots. Actual compositions for these materials are given in Table 32.

TABLE 32. ACTUAL COMPOSITIONS OF ALLOYS USED IN ISOTHERMAL-TRANSFORMATION STUDIES

Heat No.	Intended Chromium Content, %	Actual Chromium Content, %
WR17A	3	2.35
WR8A	5	4.64
WS152A	8	7.54

All heat treatments were carried out on small specimens approximately 3/8 by 1/2 by 0.064 inch. A standard beta-izing treatment of 1/2 hour at 1742 F (950 C) in a dried-argon atmosphere was applied to all specimens prior to isothermal transformation. Specimens were quenched from the beta-izing furnace into a well-deoxidized molten lead bath at the desired transformation temperature ±2 C, held for a predetermined time, and quenched in ice water.

The heat-treated specimens were mounted and polished for microscopic examination, using standard procedures. Final polishing was done on a high-speed wheel (1750 rpm), using jeweler's rouge as the abrasive. A 5-10 per cent aqueous solution of chromic acid was applied to the polishing wheel to assist in the removal of the disturbed layer on the surfaces of the specimens.

Two etching reagents were used on most of the specimens examined. The first, which was used on a majority of the specimens, consisted of a solution of 1 per cent HF and 1 per cent $Fe(NO_3)_3$ in water. The second, which was used largely to give better definition to the martensitic-type structures, consisted of 40 ml HNO_3, 15 ml HCl, and 5 ml HF diluted to 100 ml with water. On a few specimens having coarse alpha-beta structures, a staining etch of 2 per cent HF in water was used in conjunction with one of the above etchants to darken the alpha phase.

Surface contamination of the specimens during heat treatment was easily detected microscopically and contaminated areas were avoided in microstructural analysis.

Vickers hardness tests (10-kg load) were made on the polished metallographic specimens. Several impressions were made on each specimen in the center of the sheet cross section. All values given are the average of at least three readings. It must be pointed out that the hardness measurements were made on specimens mounted in Bakelite before it was known that aging induced by the heating of the Bakelite could cause hardening. Some of the hardness values may, therefore, be in error. In general, however, the high transformation temperatures should have eliminated the effects of the Bakelite curing temperature.

Samples for X-ray diffraction were prepared by grinding one end of the sheet specimens to a 30° wedge and etching about 0.008 inch off each surface in an NH_4F-H_2SO_4 water solution followed by an HF-HNO_3 glycerine solution. The latter treatment was used to remove etchant products left on the surface during the initial etch. Prepared specimens were analyzed by means of a Debye camera using vanadium radiation.

Determination of the Eutectoid Temperature

Since unavoidable variations in oxygen and nitrogen content of the alloys could affect the eutectoid temperature, it was desirable to check its approximate location. A simplified method was used based on the stability of the martensitic-type structure which forms when alloys containing less than about 6 per cent chromium are quenched from the beta-phase field. The 2.35 per cent and 4.64 per cent chromium alloys were water quenched from 1742 F (950 C) and reheated for one hour at temperatures of 1247 F (675 C), 1292 F (700 C), or 1337 F (725 C) and again quenched. Microscopic examination of these specimens revealed the following:

1. Specimens reheated to 1247 F were essentially the same as the specimens initially quenched from 1742 F.
2. Specimens reheated to 1292 F had small amounts of retained beta present in their structures.
3. Considerable amounts of retained beta were present in the specimens reheated to 1337 F.

Thus, it was indicated that the eutectoid temperature was between 1247 F (675 C) and 1292 F (700 C), and the reported value of 1265 F (685 C) (Reference 4) was used in this work.

The Ti-7.54%Cr Alloy

This alloy exhibits a retained beta structure on quenching from the beta region, as shown in Figure 44. Consequently, the structures developed by isothermal transformation are more easily interpreted than those of the two lower chromium alloys. The structures of the latter two alloys are complicated by the shear-type transformation which occurs at relatively high temperatures.

Isothermal-transformation studies of the 7.54%Cr alloy were made between 1202 F (650 C) and 842 F (450 C). Representative microstructures developed by the isothermal treatments are shown in Figures 45 through 59. Wherever possible, the following types of structure are illustrated for each transformation temperature:

1. Beginning of the alpha transformation.
2. Developed alpha precipitate prior to the beginning of $TiCr_2$ separation.
3. Beginning of $TiCr_2$ precipitation.
4. Structure at longest transformation time.

Proeutectoid alpha began to separate at holding times ranging from less than 10 seconds to 30 seconds. The appearance of the initial alpha precipitate at the several transformation temperatures is shown in Figures 45, 48, 51, 56, and 58. The Widmanstätten nature of the alpha precipitate, apparent at longer transformation times, is illustrated in Figures 46, 49, and 52. The alpha particles became increasingly finer with decreasing temperature until at 842 F the Widmanstätten pattern could not be resolved (Figure 58). Holding times of less than 10 seconds were not used, because of uncertainty about the time necessary for specimens to reach the temperature of the lead bath after quenching from 1742 F.

The rejection of the alpha phase below the eutectoid temperature causes enrichment of the beta phase in chromium until, at some concentration at or above the eutectoid composition, rejection of $TiCr_2$, presumably as a eutectoid with alpha, begins. X-ray diffraction data, given in Table 33, indicates the presence of $TiCr_2$ in considerable amounts in a specimen transformed for 48 hours at 1022 F. The microstructure of this specimen is shown in Figure 54. The dark etching constituent apparently contained the $TiCr_2$ phase. By comparing this structure with those obtained after shorter transformation times, the initial rejection of this dark etching phase was found at a transformation time of 1 hour at 1022 F (Figures 47, 50, 57, and 59). However, it was dark etching in all cases so that the progress of the transformation could be followed microscopically. At 1202 F (650 C), $TiCr_2$ appeared only after transformation times of the order of 200 hours.

The observation in Table 33 that the beta phase becomes enriched in chromium beyond the equilibrium value of 15 per cent, as alpha is rejected, is very interesting. This has been observed in other alloy systems, but it is believed that this is the first time it has been reported in a titanium alloy.

Transformation time for the 7.54%Cr alloy was not extended beyond 264 hours at any temperature. Since completion of transformation would be very difficult to detect microscopically, X-ray diffraction patterns were made of the specimens transformed for the longest time at each temperature. Results are included in Table 33. It is evident that transformation was completed only at 842 F (450 C), as indicated by the absence of the beta phase. The time-temperature-transformation curve for this alloy, based on light microscopy and X-ray evidence, is presented in Figure 60.

In general, the hardness of the 7.54%Cr alloy increased with decreasing transformation temperature, as shown in Figure 61. Hardnesses for the temperature range 1202 to 1022 F decreased as transformation progressed. At 842 F, the hardness increased with time.

TABLE 33. X-RAY DIFFRACTION DATA FOR ISOTHERMALLY TRANSFORMED SPECIMENS OF THE Ti-7.54% Cr ALLOY

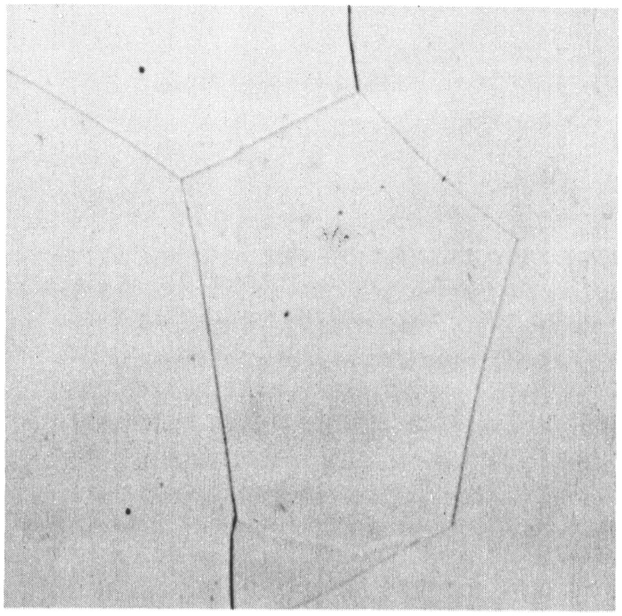
Heat No.	Isothermal Treatment(1)		Intensities of Phases(2)			Lattice Constant Retained β, (KX)	Cr Concentration Retained β, weight per cent(3)
	Temperature	Time	α	β	$TiCr_2$		
WS152A	As quenched in ice water		--	S	--	(4)	--
	1202 F (650 C)	10 minutes	S sld	VS d	--	3.226	14
	1202 F	264 hours	S sh	F sh	M sh long spots	3.233 ± .004	--
	1112 F (600 C)	187 hours	S d	VS vd	M cont. uneven	3.257 ± .006	--
	1022 F (550 C)	5 minutes	S sld	VS d	--	3.206	19
	1022 F	48 hours	S	MF d	M cont.	3.217 ± .005	16-1/2 ± 1-1/2
	1022 F	189 hours	S	M	M	3.215 ± .003	17 ± 1
	932 F (500 C)	30 minutes	S d	M vd	--	3.211	18
	842 F (450 C)	240 hours	S d	0(?)	0	--	--

(1) Specimens beta-ized for 30 minutes at 1742 F (950 C) in a dried-argon atmosphere prior to treatment. Wedge specimens prepared by grinding and etching.

(2) Intensity symbols: S = strong, M = medium, F = faint, V = very, d = diffuse, sld = slightly diffuse, sh = sharp, cont. = continuous.

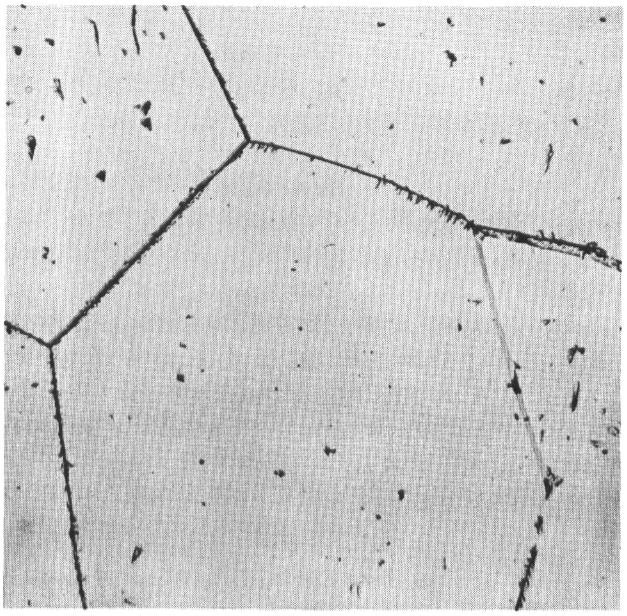
(3) Estimate of Cr content obtained from measurements of the lattice constant for β according to the data of Duwez and Taylor, Trans., ASM, Vol 44, 1952, p 495.

(4) Specimen was too coarse grained for determination.



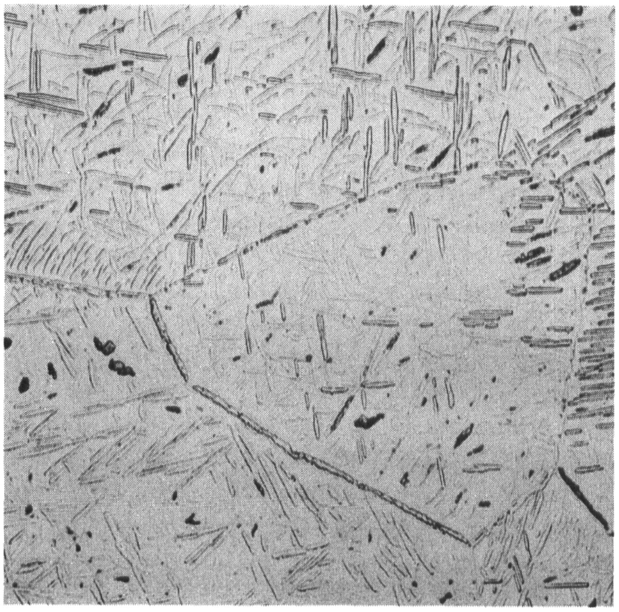
500X VHN 420 88158

FIGURE 44. 7.54 PER CENT Cr ALLOY AS QUENCHED IN ICE WATER FROM 1742 F (950 C)
Structure: Retained Beta



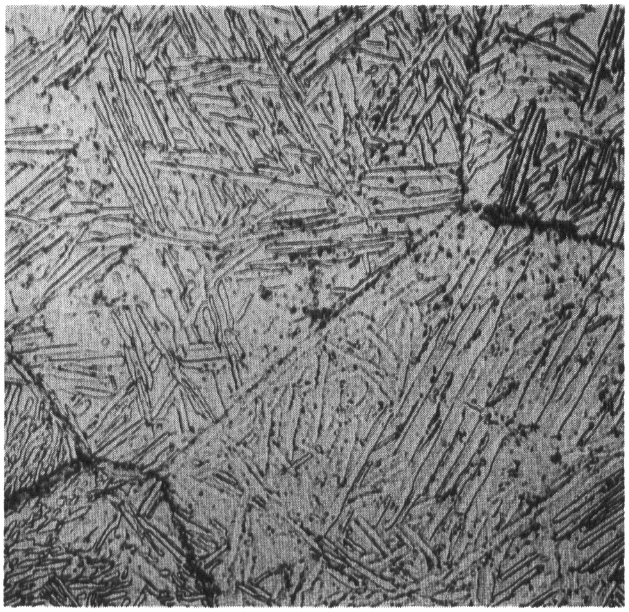
500X VHN 400 88159

FIGURE 45. 7.54 PER CENT Cr ALLOY ISOTHERMALLY TRANSFORMED FOR 30 SECONDS AT 1202 F (650 C)
Structure: Alpha in Beta Matrix; Beginning of Reaction



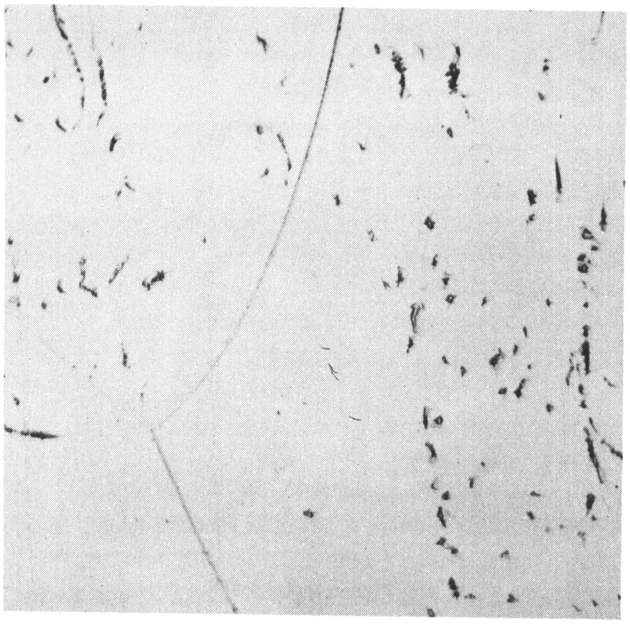
500X VHN 285 90364

FIGURE 46. 7.54 PER CENT Cr ALLOY ISOTHERMALLY TRANSFORMED FOR 72 HOURS AT 1202 F
Structure: Widmanstätten Alpha in Beta Matrix



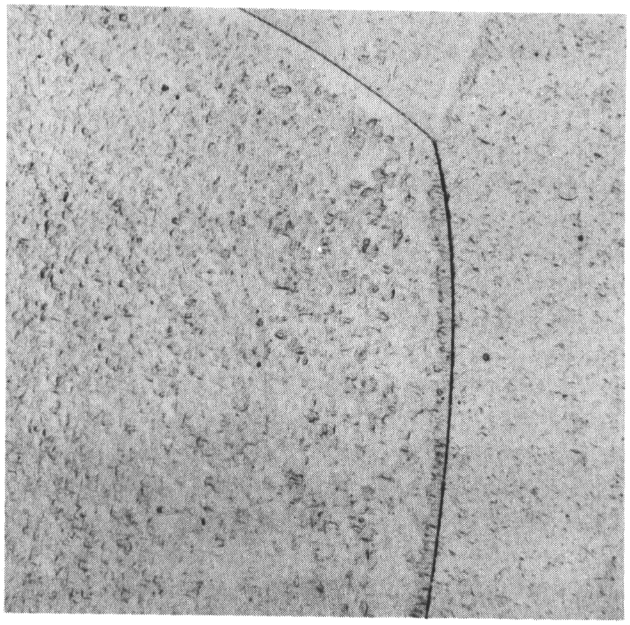
500X VHN 279 91438

FIGURE 47. 7.54 PER CENT Cr ALLOY ISOTHERMALLY TRANSFORMED FOR 264 HOURS AT 1202 F
Structure: Widmanstätten Alpha Plus $TiCr_2$ (Dark Etching) in a Beta Matrix



500X VHN 361 92328

FIGURE 48. 7.54 PER CENT Cr ALLOY ISOTHERMALLY TRANSFORMED FOR 10 SECONDS AT 1112 F (600 C)
Structure: Alpha in Beta Matrix; Beginning of Reaction

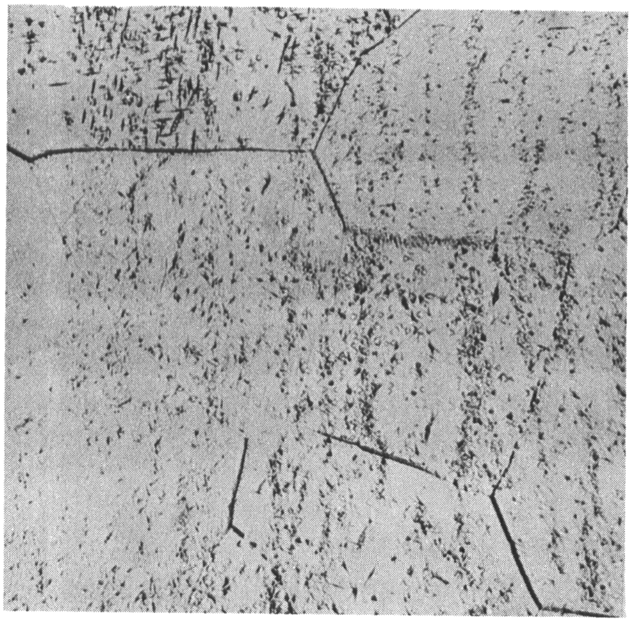


500X VHN 347 92324

FIGURE 49. 7.54 PER CENT Cr ALLOY ISOTHERMALLY TRANSFORMED FOR 1 HOUR AT 1112 F
Structure: Widmanstätten Alpha Plus $TiCr_2$ (At Grain Boundary) in Beta Matrix

500X VHN 293 92321

FIGURE 50. 7.54 PER CENT Cr ALLOY ISOTHERMALLY TRANSFORMED FOR 144 HOURS AT 1112 F
Structure: Widmanstätten Alpha and $TiCr_2$ in Beta Matrix



500X VHN 425 88163

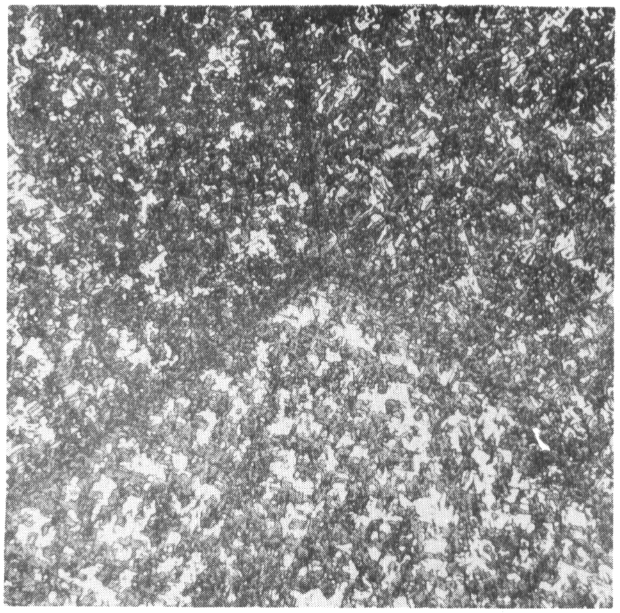
FIGURE 51. 7.54 PER CENT Cr ALLOY ISOTHERMALLY TRANSFORMED FOR 10 SECONDS AT 1022 F (550 C)
Structure: Alpha in Beta Matrix; Beginning of Reaction

500X VHN 354 88164

FIGURE 52. 7.54 PER CENT Cr ALLOY ISOTHERMALLY TRANSFORMED FOR 10 MINUTES AT 1022 F (550 C)
Structure: Widmanstätten Alpha in a Beta Matrix

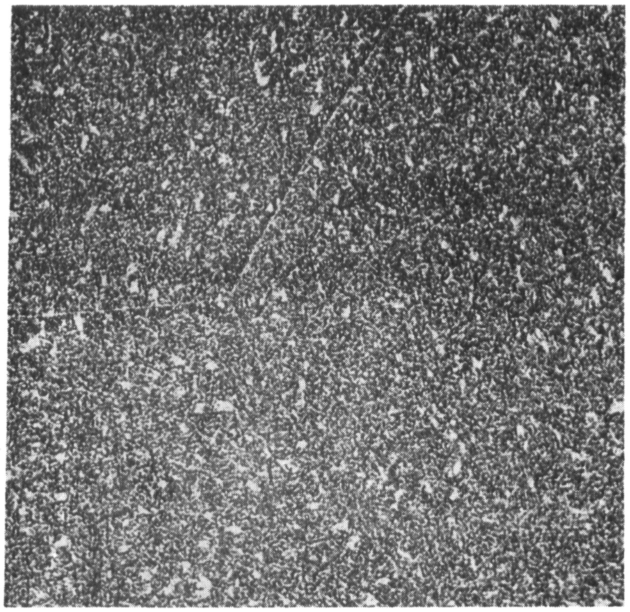
500X VHN 350 90357

FIGURE 53. 7.54 PER CENT Cr ALLOY ISOTHERMALLY TRANSFORMED FOR 1 HOUR AT 1022 F (550 C)
Structure: Widmanstätten Alpha Plus $TiCr_2$ in a Beta Matrix



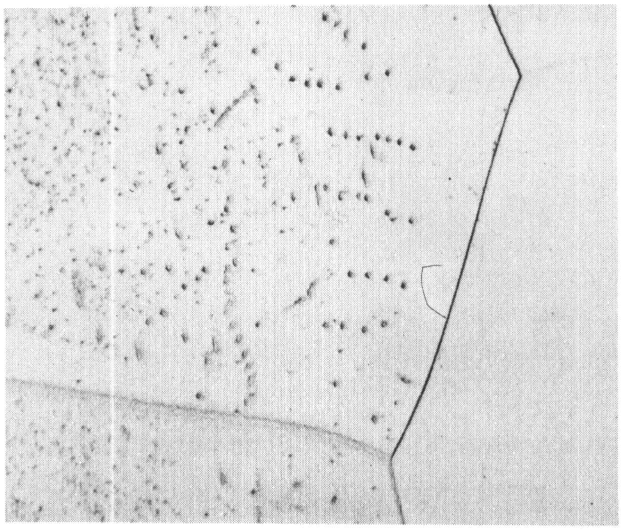
500X VHN 272 90807

FIGURE 54. 7.54 PER CENT Cr ALLOY ISOTHERMALLY TRANSFORMED FOR 48 HOURS AT 1022 F (550 C)
Structure: Alpha and $TiCr_2$ in a Beta Matrix



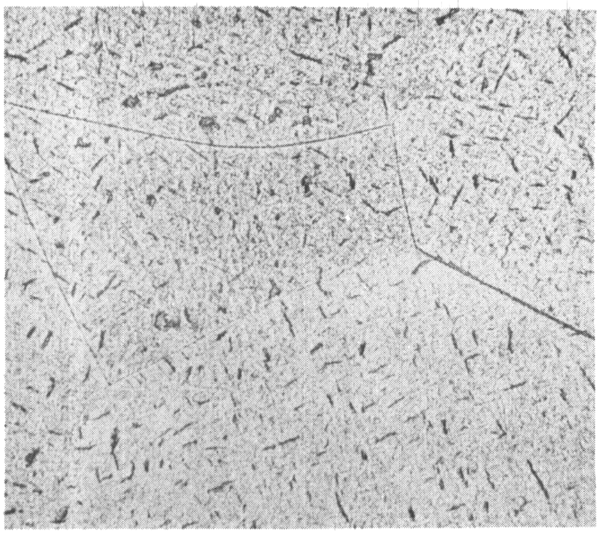
500X VHN 291 90370

FIGURE 55. 7.54 PER CENT Cr ALLOY ISOTHERMALLY TRANSFORMED FOR 189 HOURS AT 1022 F
Structure: Alpha, $TiCr_2$, and Beta



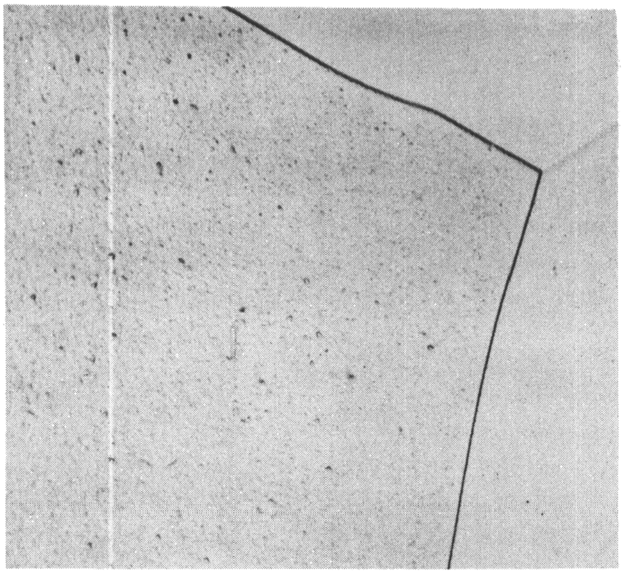
500X VHN 389 92329

FIGURE 56. 7.54 PER CENT Cr ALLOY ISOTHERMALLY TRANSFORMED FOR 10 SECONDS AT 932 F (500 C)
Structure: Alpha Precipitate in a Beta Matrix



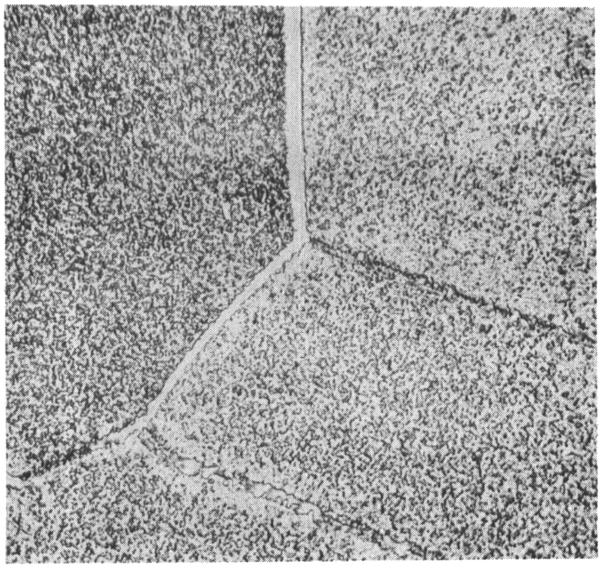
500X VHN 352 91435

FIGURE 57. 7.54 PER CENT Cr ALLOY ISOTHERMALLY TRANSFORMED FOR 192 HOURS AT 932 F
Structure: Alpha and $TiCr_2$ in a Beta Matrix



500X VHN 418 92330

FIGURE 58. 7.54 PER CENT Cr ALLOY ISOTHERMALLY TRANSFORMED FOR 30 SECONDS AT 842 F (450 C)
Structure: Alpha in a Beta Matrix



500X VHN 458 91431

FIGURE 59. 7.54 PER CENT Cr ALLOY ISOTHERMALLY TRANSFORMED FOR 192 HOURS AT 842 F
Structure: Alpha and $TiCr_2$ in a Beta Matrix

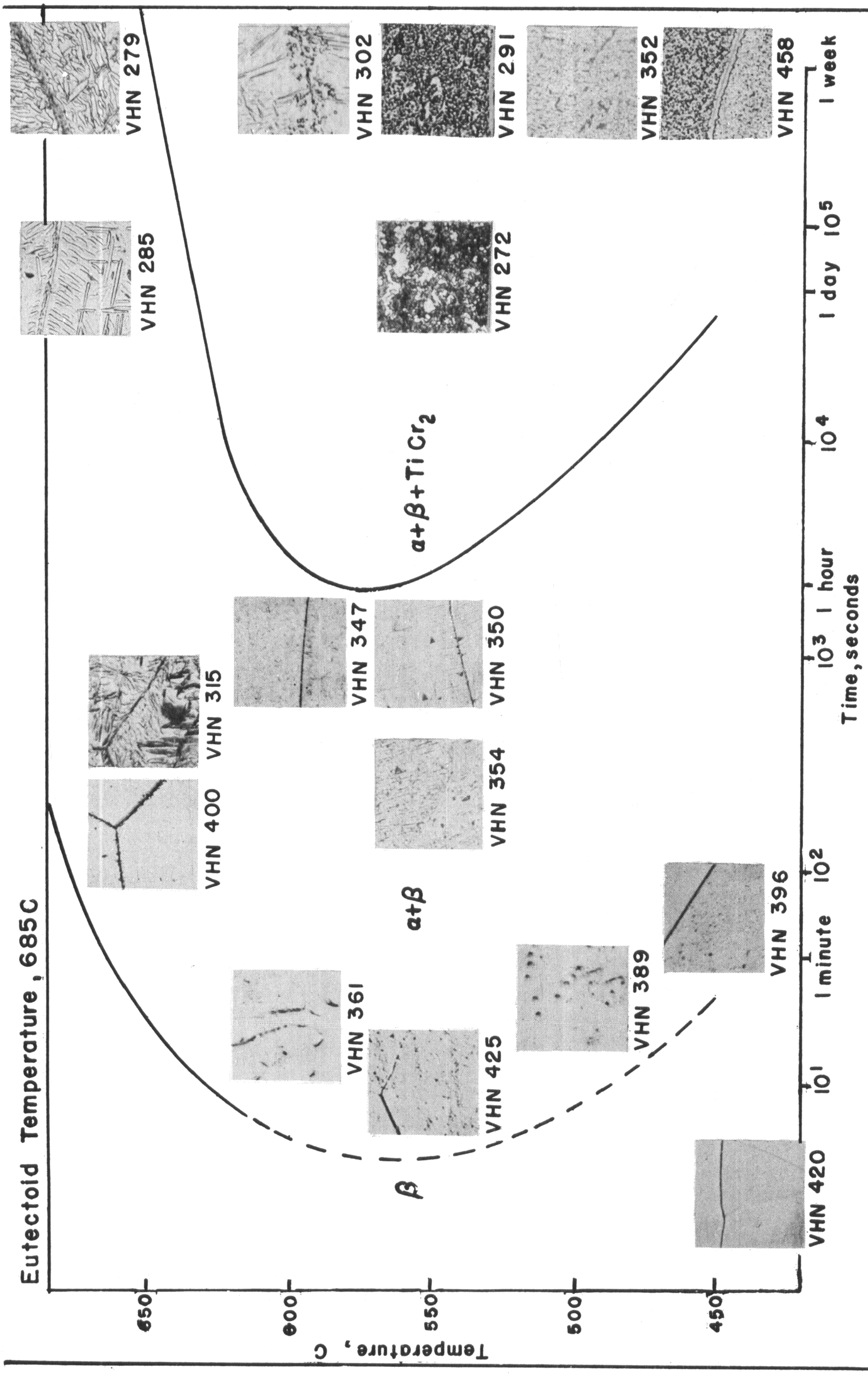


FIGURE 60. ISOTHERMAL-TRANSFORMATION DIAGRAM FOR THE Ti-7.54%Cr ALLOY

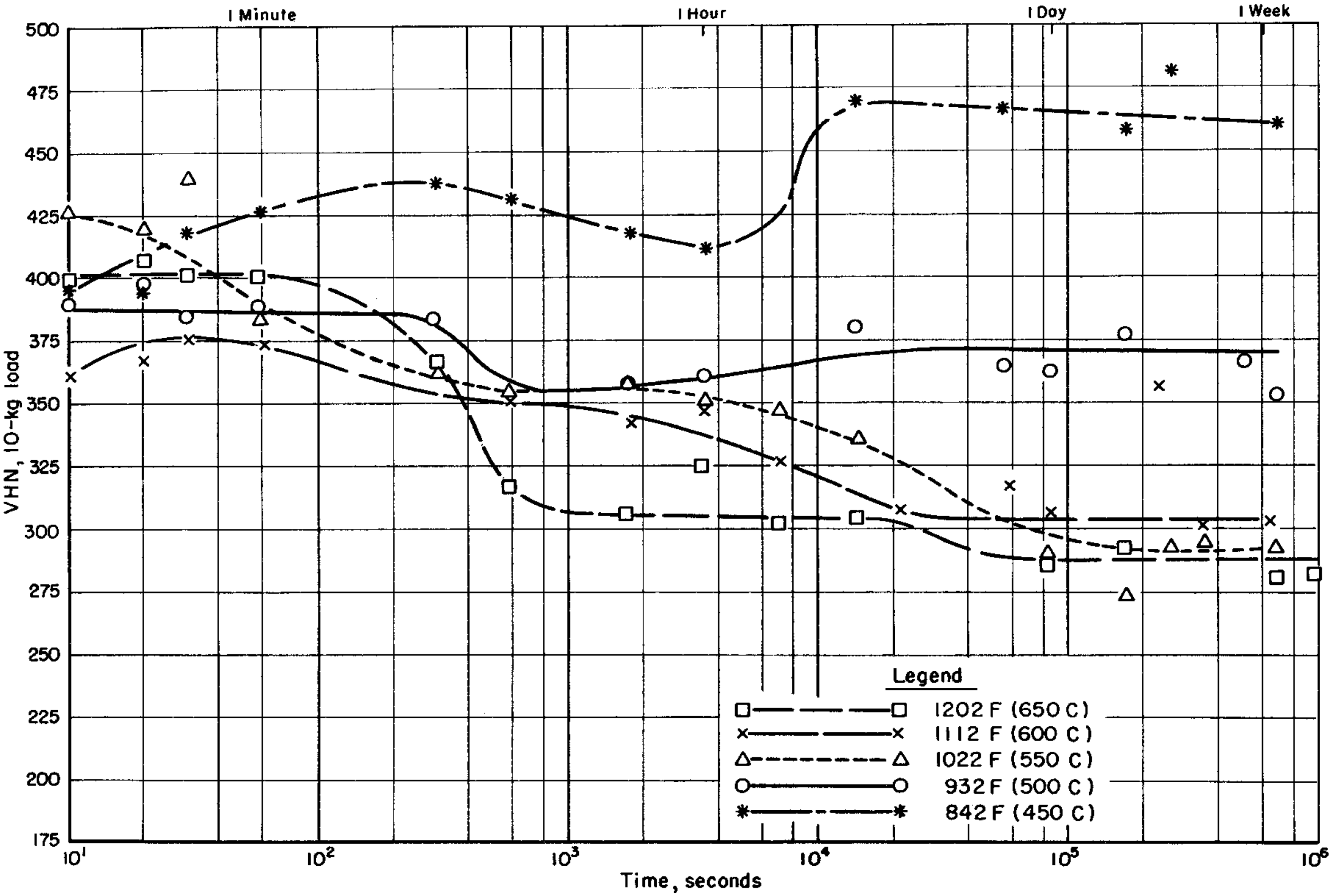


FIGURE 61. HARDNESS VERSUS ISOTHERMAL-TRANSFORMATION TIME FOR Ti-7.54 PER CENT Cr ALLOY
A-3282

The Ti-4.64%Cr Alloy

This alloy undergoes a diffusionless, shear-type transformation to a supersaturated alpha phase when quenched from the beta-phase field. It was desirable to know the temperature at which this reaction begins (M_s temperature), although it is not believed that the reaction is principally responsible for the hardening of titanium alloys. In this work, the M_s temperature has been considered the lower temperature limit for isothermal transformation.

A modification of the classical Greninger-Troiano technique similar to that used by DeLazaro, et al.,(5) on Ti-Mo alloys was used to determine the M_s temperature for this alloy. The method consists of quenching from a temperature in the beta field into a lead bath at various temperatures, holding for 10 seconds, transferring to another lead bath at 1202 F (650 C), holding 60 seconds, and water quenching. If the temperature (A) of the lead bath used for the first quench is below the M_s point, martensite needles (alpha prime) will form, the quantity of which will vary with the degree of under cooling. If the specimen is then quenched directly into cold water, the beta retained at temperature A will continue to transform to martensite. This is prevented by the transfer of the specimen from temperature A to 1202 F. At this latter temperature, the beta partially transforms to Widmanstätten alpha. This alpha-beta mixture was shown by experiment to be retained as such on quenching in water and it could be readily distinguished from the martensitic alpha prime.

A representative series of photomicrographs depicting the M_s determination for this alloy is shown in Figures 62 to 65. In Figure 62, the dark etching needles are martensitic plates, alpha prime. These become fewer at successively higher quenching temperature until, at 968 F (520 C), they completely disappeared. Thus, the M_s temperature for this alloy was established between 932 F (500 C) and 968 F (520 C).

The time-temperature relationships for the beginning of proeutectoid alpha separation were essentially the same as those for the 7.54 per cent chromium alloy. At 1202 F (650 C), alpha rejection began between 20 and 30 seconds with the development of the structure shown in Figure 66. At 1112 F (600 C) and 1022 F (550 C), the reaction began in less than 10 seconds (Figures 68 and 71). Continued transformation at these temperatures developed a well-defined Widmanstätten precipitate of the alpha phase, which became finer with decreasing transformation temperature (Figures 67, 69, and 72).

The initiation of the $TiCr_2$ reaction was determined by the same technique used for the 7.54 per cent alloy. X-ray diffraction data presented in Table 34 indicated this compound to be present in the specimen transformed 48 hours at 1022 F (550 C). The spheroidal particles shown in Figure 73 were believed to be eutectoid. The earliest microscopical evidence of this reaction was traced back to a transformation time of one hour. A photomicrograph of the specimen transformed for one hour was not made. The structure of this specimen was very similar, however, to that shown in Figure 72, which represents a two-hour transformation time.

The early stages of the $TiCr_2$ separation at 1112 F (600 C) is illustrated in Figure 69. X-ray evidence (Table 34) indicated that transformation was not complete after 264 and 187 hours at 1202 and 1112 F, respectively. The specimens transformed for longer periods than 48 hours at 1022 F were not examined by X-rays.

The time-temperature-transformation curve for this alloy, based on microscopic and X-ray data, is presented in Figure 75.

Hardness data for the isothermally transformed specimens are presented graphically in Figure 76. The hardness, in general, increased with decreasing transformation temperature and decreased with increasing time at a given temperature.

The Ti-2.35%Cr Alloy

The M_s temperature for this alloy was established between 1004 F (540 C) and 1040 F (560 C) by the method previously described.

Rejection of the proeutectoid alpha began in less than 10 seconds at 1202 F and 1112 F, as shown in Figures 77 and 80. The X-ray diffraction data shown in Table 35 confirm the presence of alpha after 10 seconds at 1202 F.

The beginning of the $TiCr_2$ reaction was determined wholly on the basis of microscopic examination (Figures 79 and 81). The presence of this phase after 187 hours at 1112 F (600 C) was confirmed by X-ray diffraction data. Transformation was not complete after 264 and 189 hours at 1202 and 1112 F, respectively, as shown by the presence of the beta phase (Table 35).

The TTT curve for this alloy is shown in Figure 83.

Hardness data are presented in Figure 84. As expected, the hardnesses of the specimens transformed at 1202 F were generally lower than those of the specimens transformed at 1112 F.

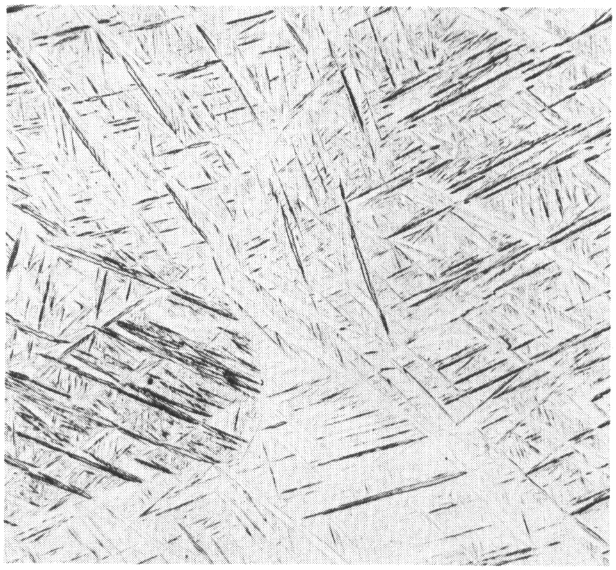
TABLE 34. X-RAY DIFFRACTION DATA FOR ISOTHERMALLY TRANSFORMED SPECIMENS OF THE Ti-4.64% Cr ALLOY

Heat No.	Isothermal Treatment(1)		Intensities of Phases(2)							Lattice Constant Retained β, (KX)	Cr Concentration Retained β, weight per cent(3)
	Temperature	Time	α		α'		β		$TiCr_2$		
WR8A	Ice Water Quench		F to M	d	S(?)	d	F to MF	d	--	--	--
	1202 F (650 C)	10 minutes	S	sh	0		S	sld	--	3.231	13
	1202 F	264 hours	S		--		S+	d	0	3.233 ± .004	12-1/2 ± 1
	1112 F (600 C)	187 hours	S		--		M	d	VF+ sh cont. uneven	3.219 ± .004	16 ± 1
	1022 F (550 C)	10 seconds	S	d	0		MS	d	--	3.246	8-1/2
	1022 F	48 hours	VS		--		MS	d	F cont.	3.209 ± .003	18-1/2 ± 1

(1) Specimens beta-ized for 30 minutes at 1742 F (950 C) in a dried-argon atmosphere prior to treatment. Wedge specimens prepared by grinding and etching.

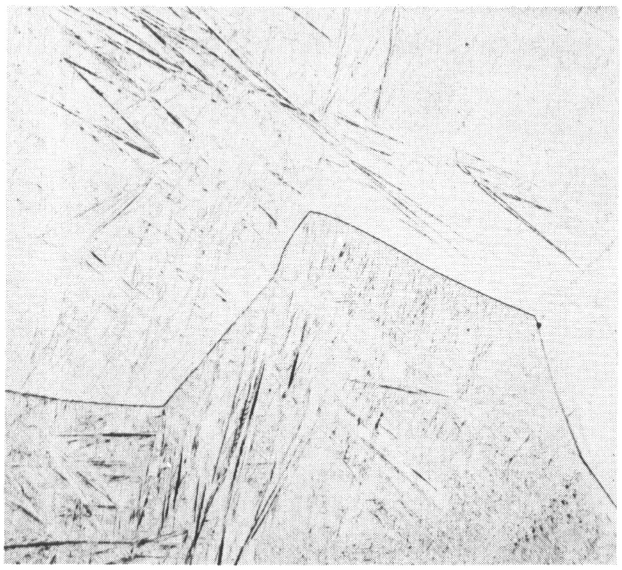
(2) Intensity symbols: S = strong, M = medium, F = faint, V = very, d = diffuse, sld = slightly diffuse, sh = sharp, cont. = continuous.

(3) Estimate of Cr content obtained from measurements of the lattice constant for β according to the data of Duwez and Taylor, Trans., ASM, Vol 44, 1952, p 495.



250X 91622

FIGURE 62. 4.64 PER CENT Cr ALLOY AS QUENCHED IN ICE WATER FROM 1742 F
Structure: Alpha Prime

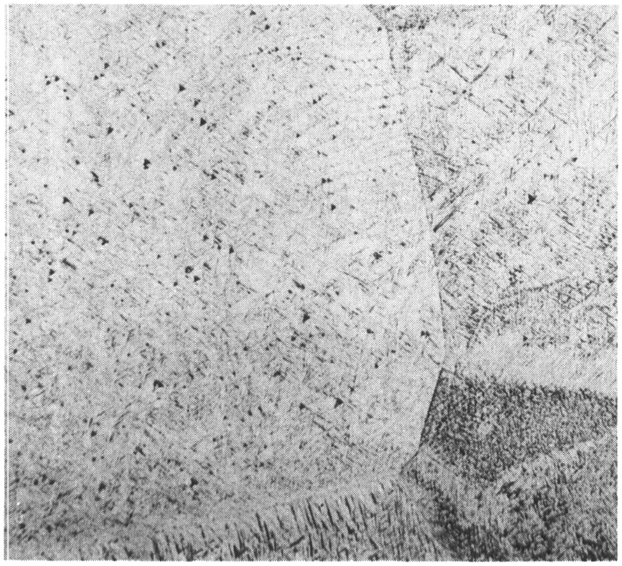


250X 91624

FIGURE 63. 4.64 PER CENT Cr ALLOY M_s DETERMINATION AT 842 F (450 C)
Structure: Alpha-Prime Plates, Alpha and Beta

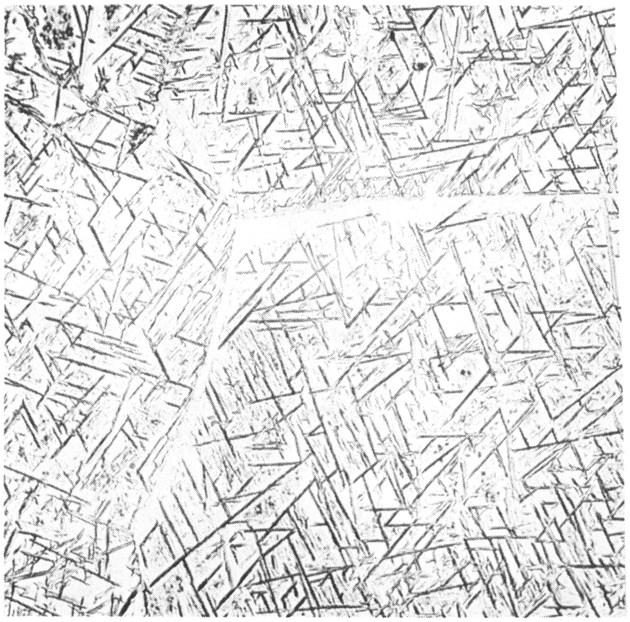
250X 91626

FIGURE 64. 4.64 PER CENT Cr ALLOY M_s DETERMINATION AT 932 F (500 C)
Structure: Alpha-Prime Plates, Alpha and Beta



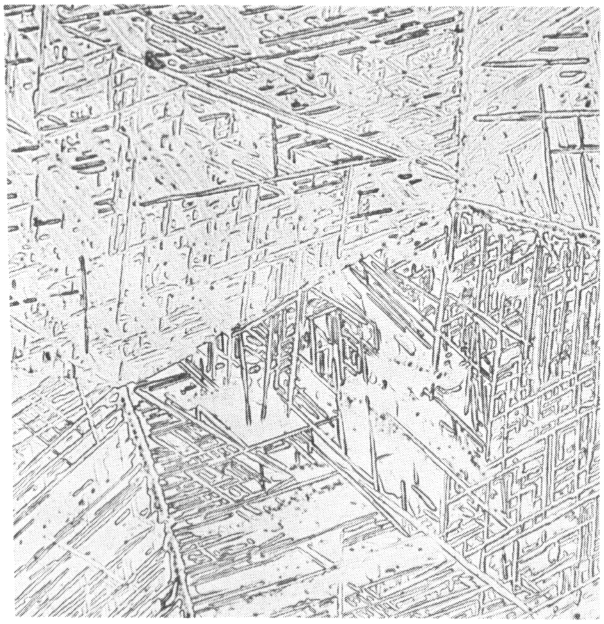
250X 91627

FIGURE 65. 4.64 PER CENT Cr ALLOY M_s DETERMINATION AT 968 F (520 C)
Structure: Alpha in Beta Matrix



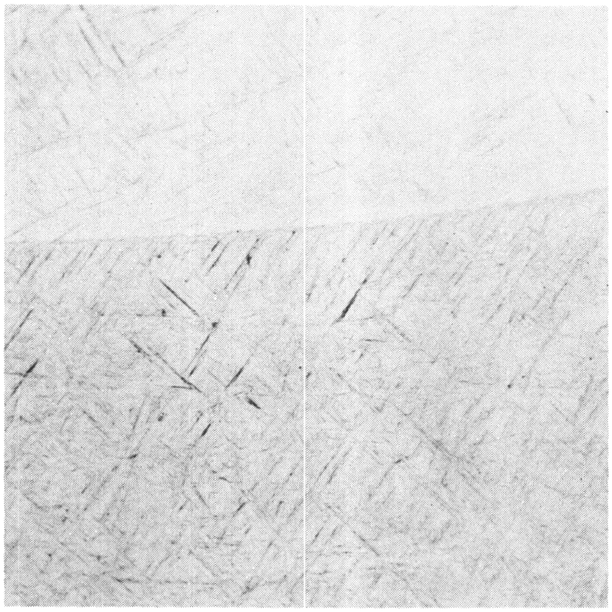
500X VHN 388 88156

FIGURE 66. 4.64 PER CENT Cr ALLOY ISOTHERMALLY TRANSFORMED FOR 30 SECONDS AT 1202 F (650 C)
Structure: Widmanstätten Alpha in a Beta Matrix



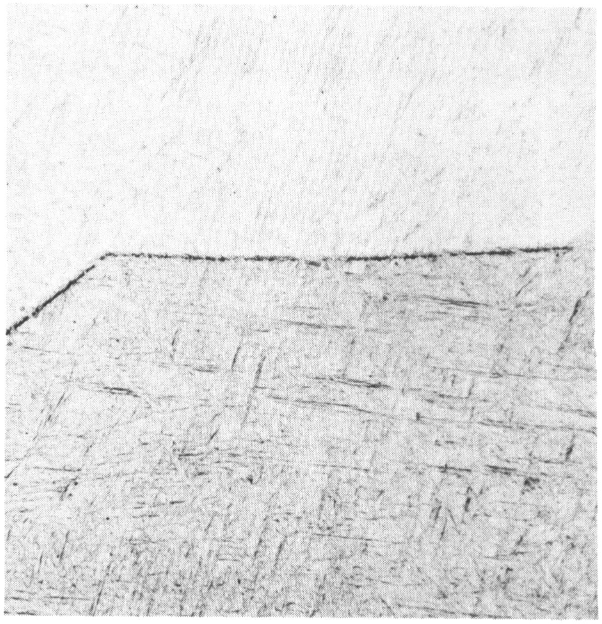
500X VHN 256 91437

FIGURE 67. 4.64 PER CENT Cr ALLOY ISOTHERMALLY TRANSFORMED FOR 264 HOURS AT 1202 F
Structure: Alpha and $TiCr_2$ in a Beta Matrix



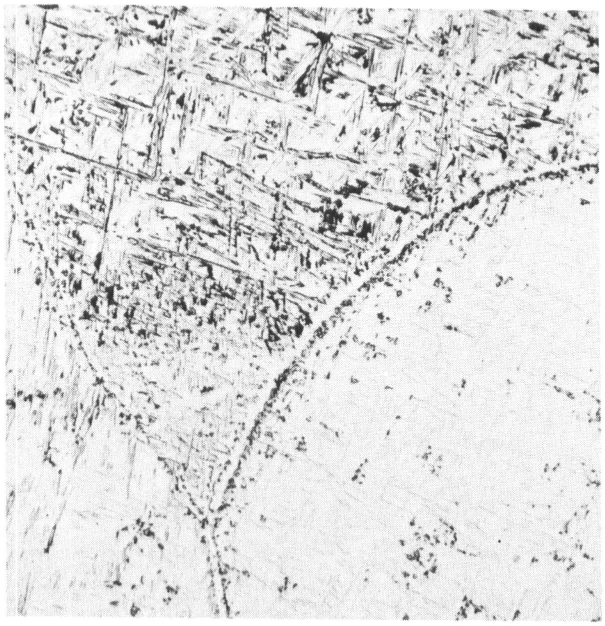
500X VHN 383 92327

FIGURE 68. 4.64 PER CENT Cr ALLOY ISOTHERMALLY TRANSFORMED FOR 10 SECONDS AT 1112 F (600 C)
Structure: Widmanstätten Alpha Precipitate in a Beta Matrix



500X VHN 92323

FIGURE 69. 4.64 PER CENT Cr ALLOY ISOTHERMALLY TRANSFORMED FOR 2 HOURS AT 1112 F
Structure: Widmanstätten Alpha Plus $TiCr_2$ in a Beta Matrix

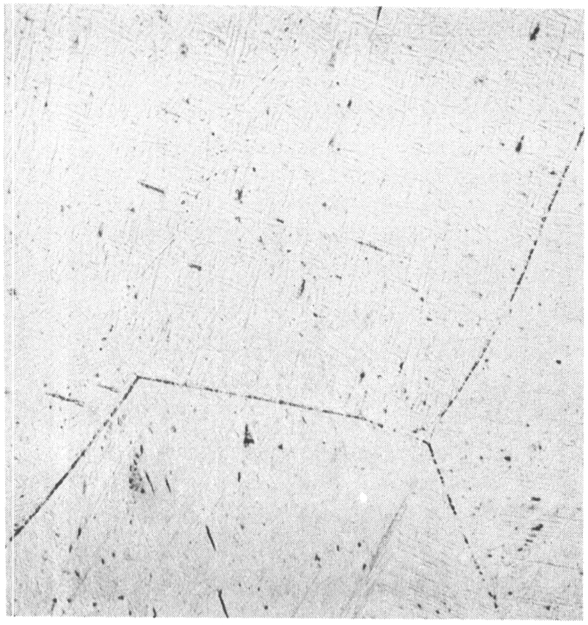


500X VHN 330 92320

FIGURE 70. 4.64 PER CENT Cr ALLOY ISOTHERMALLY TRANSFORMED FOR 144 HOURS AT 1112 F
Structure: Widmanstätten Alpha Plus $TiCr_2$ in a Beta Matrix

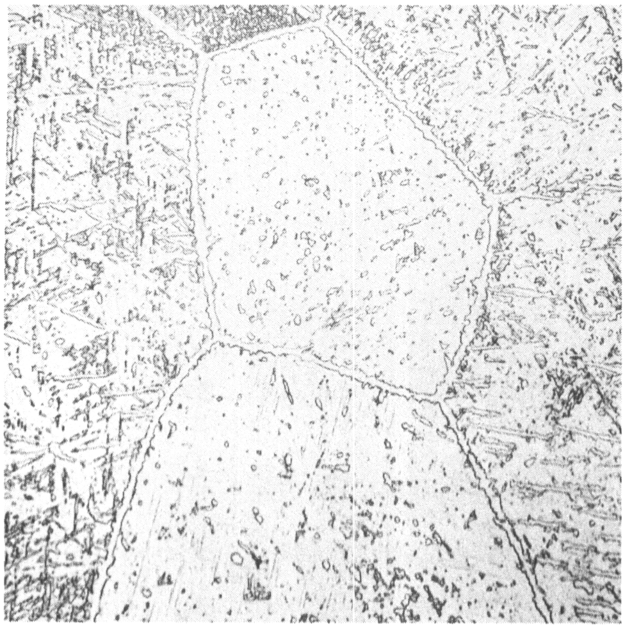
500X VHN 385 88161

FIGURE 71. 4.64 PER CENT Cr ALLOY ISOTHERMALLY TRANSFORMED FOR 10 SECONDS AT 1022 F (550 C)
Structure: Alpha in a Beta Matrix



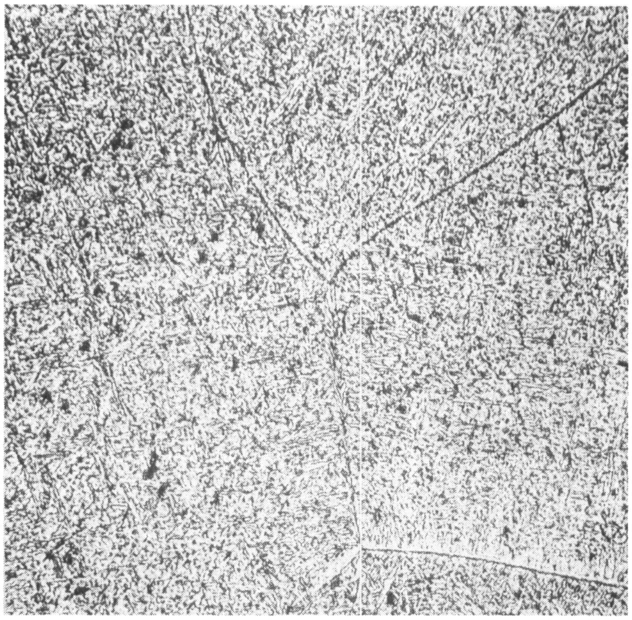
500X VHN 314 90356

FIGURE 72. 4.64 PER CENT Cr ALLOY ISOTHERMALLY TRANSFORMED FOR 2 HOURS AT 1022 F
Structure: Widmanstätten Alpha Plus $TiCr_2$ in a Beta Matrix



500X VHN 243 90803

FIGURE 73. 4.64 PER CENT Cr ALLOY ISOTHERMALLY TRANSFORMED FOR 48 HOURS AT 1022 F
Structure: Widmanstätten Alpha Plus $TiCr_2$ in a Beta Matrix



500X VHN 243 90366

FIGURE 74. 4.64 PER CENT Cr ALLOY ISOTHERMALLY TRANSFORMED FOR 189 HOURS AT 1022 F
Structure: Alpha and $TiCr_2$ in a Beta Matrix

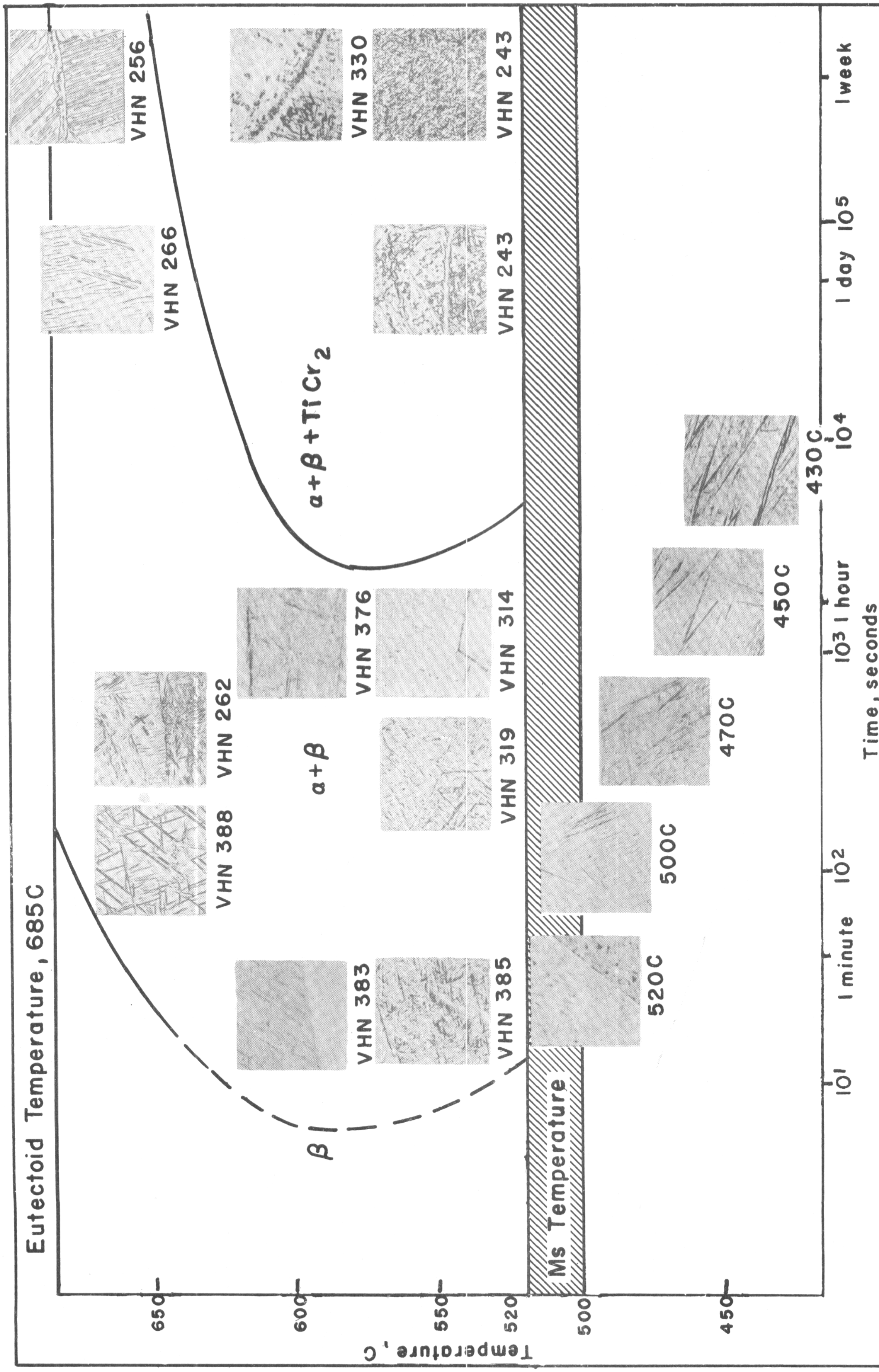


FIGURE 75. ISOTHERMAL-TRANSFORMATION DIAGRAM FOR THE Ti-4.64% Cr ALLOY

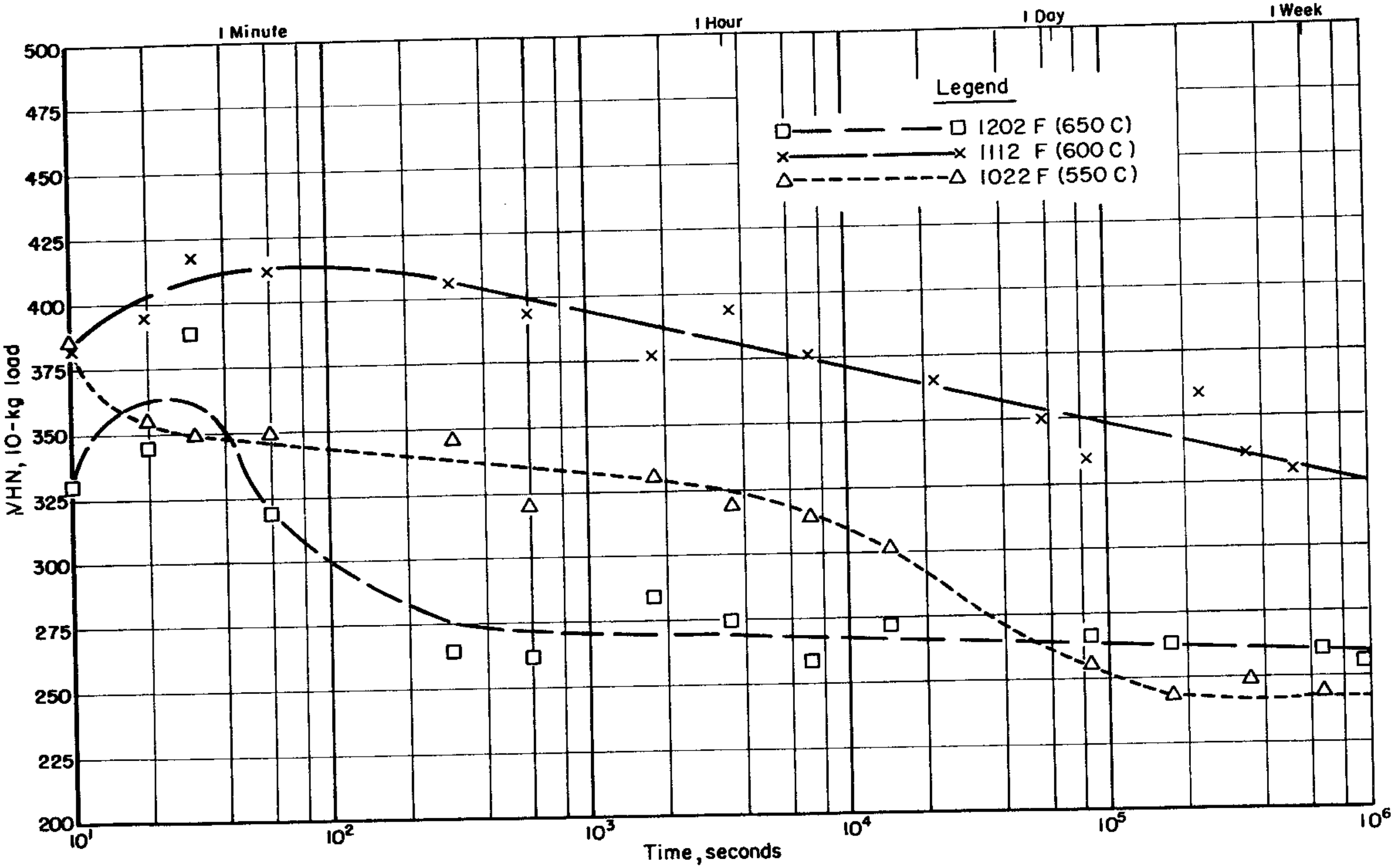


FIGURE 76. HARDNESS VERSUS ISOTHERMAL-TRANSFORMATION TIME FOR A Ti-4.64 PER CENT Cr ALLOY

A-3283

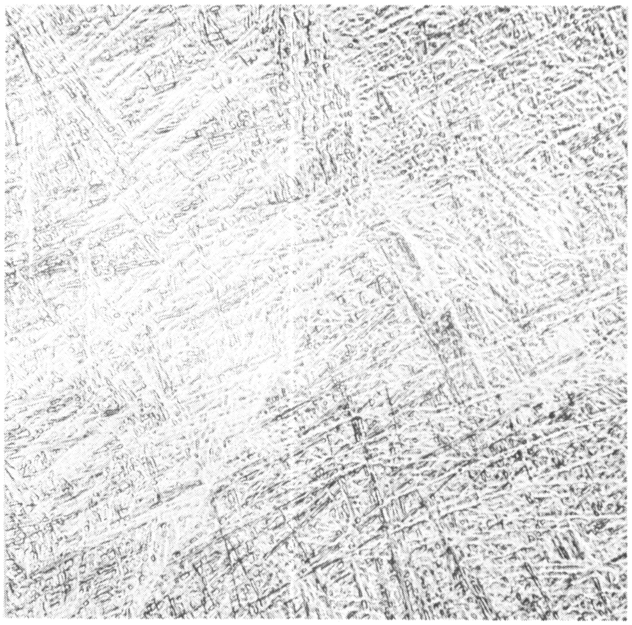
TABLE 35. X-RAY DIFFRACTION DATA FOR ISOTHERMALLY TRANSFORMED SPECIMENS OF THE Ti-2.35% Cr ALLOY

Heat No.	Isothermal Treatment(1)		Intensities of Phases(2)				Lattice Constant Retained β, (KX)	Cr Concentration Retained β, weight per cent(3)
	Temperature	Time	α	α'	β	$TiCr_2$		
WR17A	Ice Water Quench		0	S d	0 to VF	--	--	--
	1202 F (650 C)	10 seconds	S sld	0	S d	--	3.247	8
	1202 F	264 hours	S	--	MS d	0	3.239 ± .004	10 ± 1
	1112 F (600 C)	187 hours	S	--	MS d	Trace	3.222 ± .004	15 ± 1

(1) Specimens beta-ized for 30 minutes at 1742 F (950 C) in a dried-argon atmosphere prior to treatment. Wedge specimens prepared by grinding and etching.

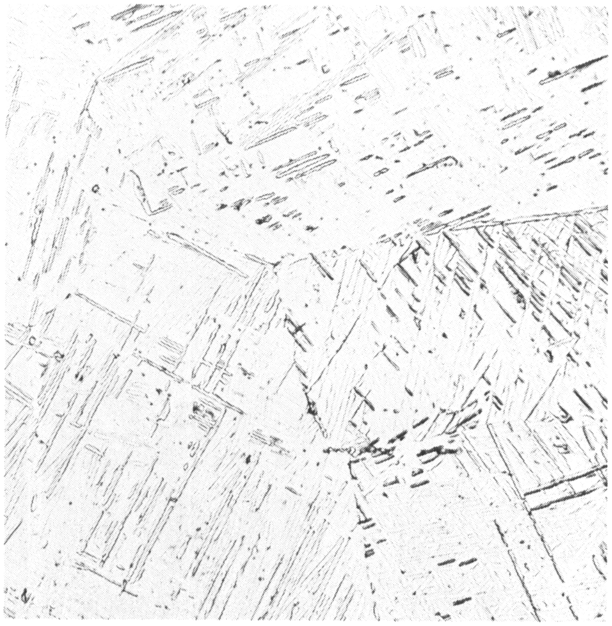
(2) Intensity symbols: S = strong, M = medium, F = faint, V = very, d = diffuse, sld = slightly diffuse, sh = sharp.

(3) Estimate of Cr content obtained from measurements of the lattice constant for β according to the data of Duwez and Taylor, Trans., ASM, Vol 44, 1952, p 495.



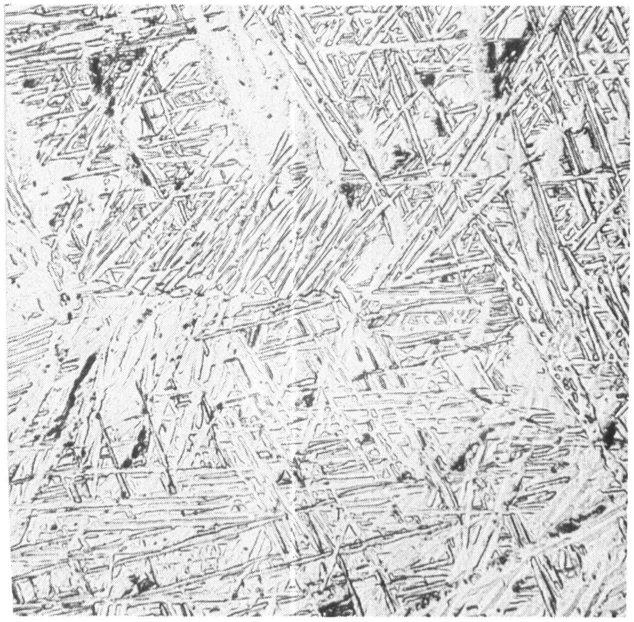
500X VHN 289 88153

FIGURE 77. 2.35 PER CENT Cr ALLOY ISOTHERMALLY TRANSFORMED FOR 10 SECONDS AT 1202 F (650 C)
Structure: Alpha in a Beta Matrix



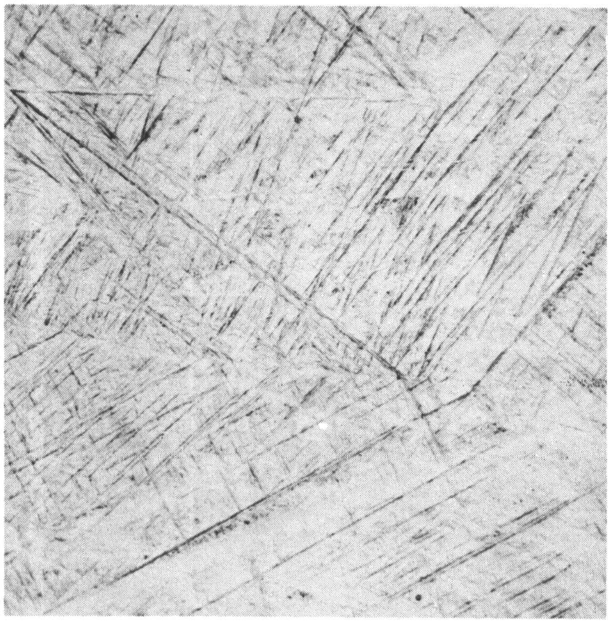
500X VHN 225 90360

FIGURE 78. 2.35 PER CENT Cr ALLOY ISOTHERMALLY TRANSFORMED FOR 48 HOURS AT 1202 F
Structure: Widmanstätten Alpha in a Beta Matrix



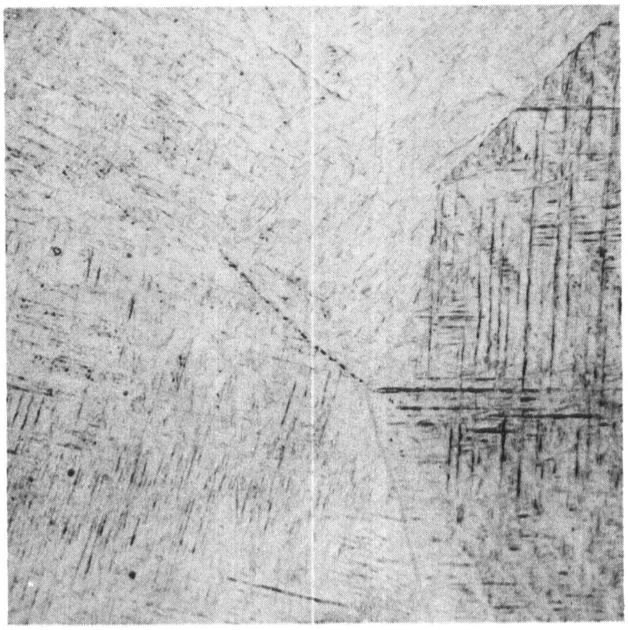
500X VHN 236 91436

FIGURE 79. 2.35 PER CENT Cr ALLOY ISOTHERMALLY TRANSFORMED FOR 264 HOURS AT 1202 F
Structure: Alpha and $TiCr_2$ in a Beta Matrix



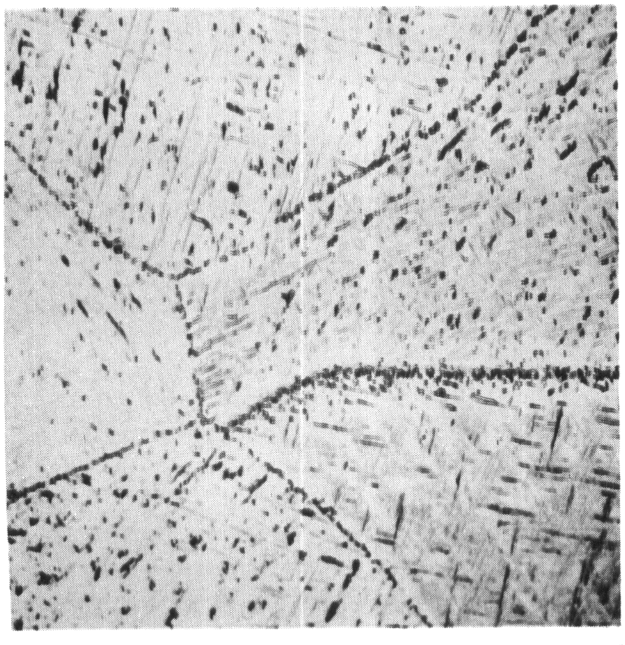
500X VHN 315 92325

FIGURE 80. 2.35 PER CENT Cr ALLOY ISOTHERMALLY TRANSFORMED FOR 10 SECONDS AT 1112 F (600 C)
Structure: Alpha in a Beta Matrix



500X VHN 359 92322

FIGURE 81. 2.35 PER CENT Cr ALLOY ISOTHERMALLY TRANSFORMED FOR 1 HOUR AT 1112 F
Structure: Widmanstätten Alpha Plus $TiCr_2$ in a Beta Matrix



500X VHN 247 92319

FIGURE 82. 2.35 PER CENT Cr ALLOY ISOTHERMALLY TRANSFORMED FOR 144 HOURS AT 1112 F
Structure: Alpha and $TiCr_2$ in a Beta Matrix

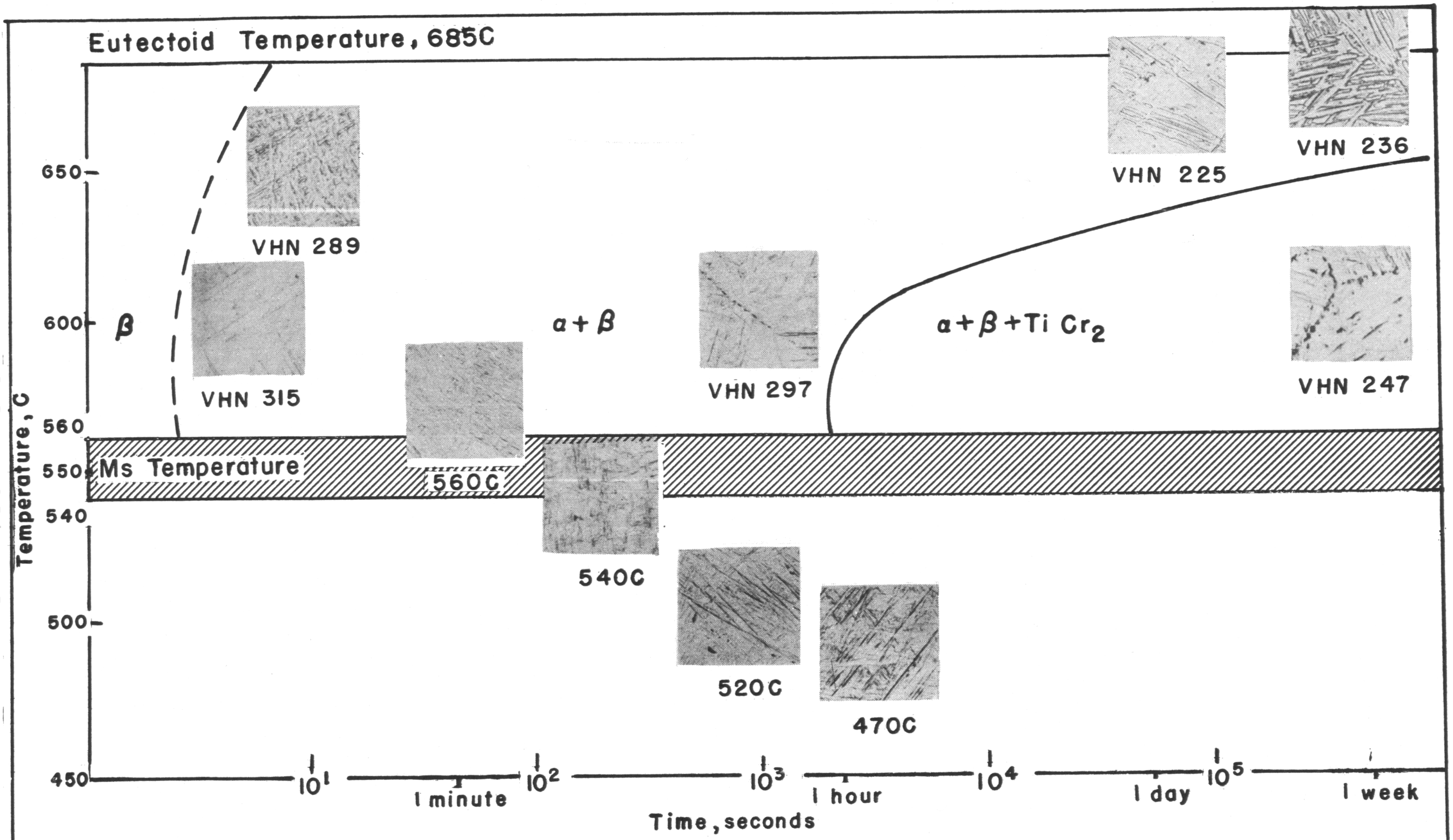


FIGURE 83. ISOTHERMAL-TRANSFORMATION DIAGRAM FOR THE Ti-2.35 % Cr ALLOY

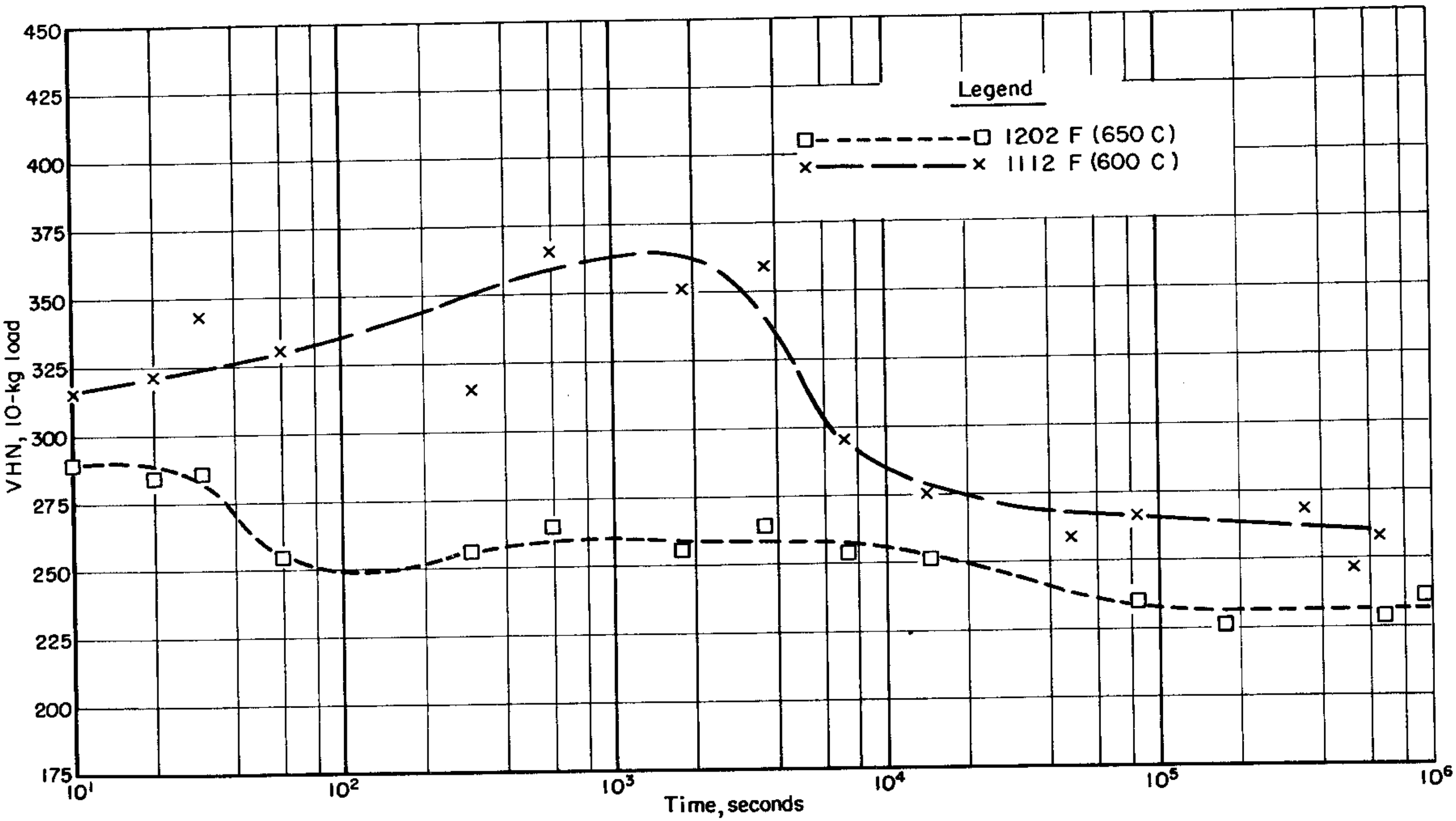


FIGURE 84. HARDNESS VERSUS ISOTHERMAL-TRANSFORMATION TIME FOR A Ti-2.35 PER CENT Cr ALLO

A-3284

ELECTRON-MICROSCOPE STUDIES

In connection with the isothermal transformation of the 7.54%Cr alloy, the electron microscope was used to examine some of the structures which could not be resolved at ordinary magnifications. The micrographs included here represent the first known attempts that anyone has made to apply this tool to titanium. It is hoped that, by the use of X-ray diffraction, the electron microscope, and electron diffraction, the mechanism of the beta to alpha and beta to eutectoid transformations may be better understood. It is believed by the authors that the beta to alpha transformation, as a nucleation and growth process, is responsible for the apparent age hardening in the retained beta structure at low temperatures. It is this hardening which is responsible for the high strength of the alloys now evolving from this research effort. Therefore, a reasonable effort to understand the hardening mechanism is necessary to the progress of the project.

An effort was made to observe closely the first alpha rejected in a specimen of the 7.54%Cr alloy transformed 30 seconds at 1022 F (550 C). A positive replica electron micrograph at 4000X is reproduced in Figure 85. The initial precipitation occurred primarily at grain boundaries and was obviously of a Widmanstätten nature. As alpha separation continued with increasing transformation time, the Widmanstätten pattern became increasingly clear (Figure 86). The reason for the mottled appearance of the alpha platelets shown in these two micrographs is not yet known. It is believed to be an artifact caused by an oxide film, which is invisible at lower magnifications.

An attempt to follow the transformation of the $TiCr_2$ phase and to resolve the dark-etching structure formed at 1022 F (Figure 54) was made. To provide a basis for comparison, a Ti-20%Cr alloy described earlier in this report (Figure 25) and known to contain the eutectoid was first photographed at 5000X. The structure of the 20%Cr alloy at this magnification is shown in Figure 87. The lighter etching, irregular lamellae are believed to be $TiCr_2$ in a matrix of dark-etching alpha. The large light areas are retained beta.

Electron micrographs at 4000X of specimens of the 7.54%Cr alloy transformed for 24 and 48 hours at 1022 F (550 C) are shown in Figures 88 and 89, respectively. These did not reveal any well-defined structure in the dark-etching phase. However, close examination of Figure 89 shows very faint traces of what may be $TiCr_2$ lamellae in the dark phase, indicating that this phase may have a eutectoid configuration. The lack of resolution of the eutectoid phase in Figure 89, as compared with the structure of Figure 87, may be attributable to the lower transformation temperature used.

4000X 3237C

FIGURE 85. 7.54 PER CENT Cr ALLOY ISOTHERMALLY TRANSFORMED FOR 30 SECONDS AT 1022 F (550 C); A POSITIVE REPLICA, PLATINUM-SHADOWED, ELECTRON MICROGRAPH

Structure: Widmanstätten Alpha Precipitate in a Beta Matrix

5000X

3235E

FIGURE 86. 7.54 PER CENT Cr ALLOY ISOTHERMALLY TRANSFORMED FOR 5 MINUTES AT 1022 F (550 C); A POSITIVE REPLICA, PLATINUM-SHADOWED ELECTRON MICROGRAPH

Structure: Widmanstätten Alpha Precipitate in a Beta Matrix

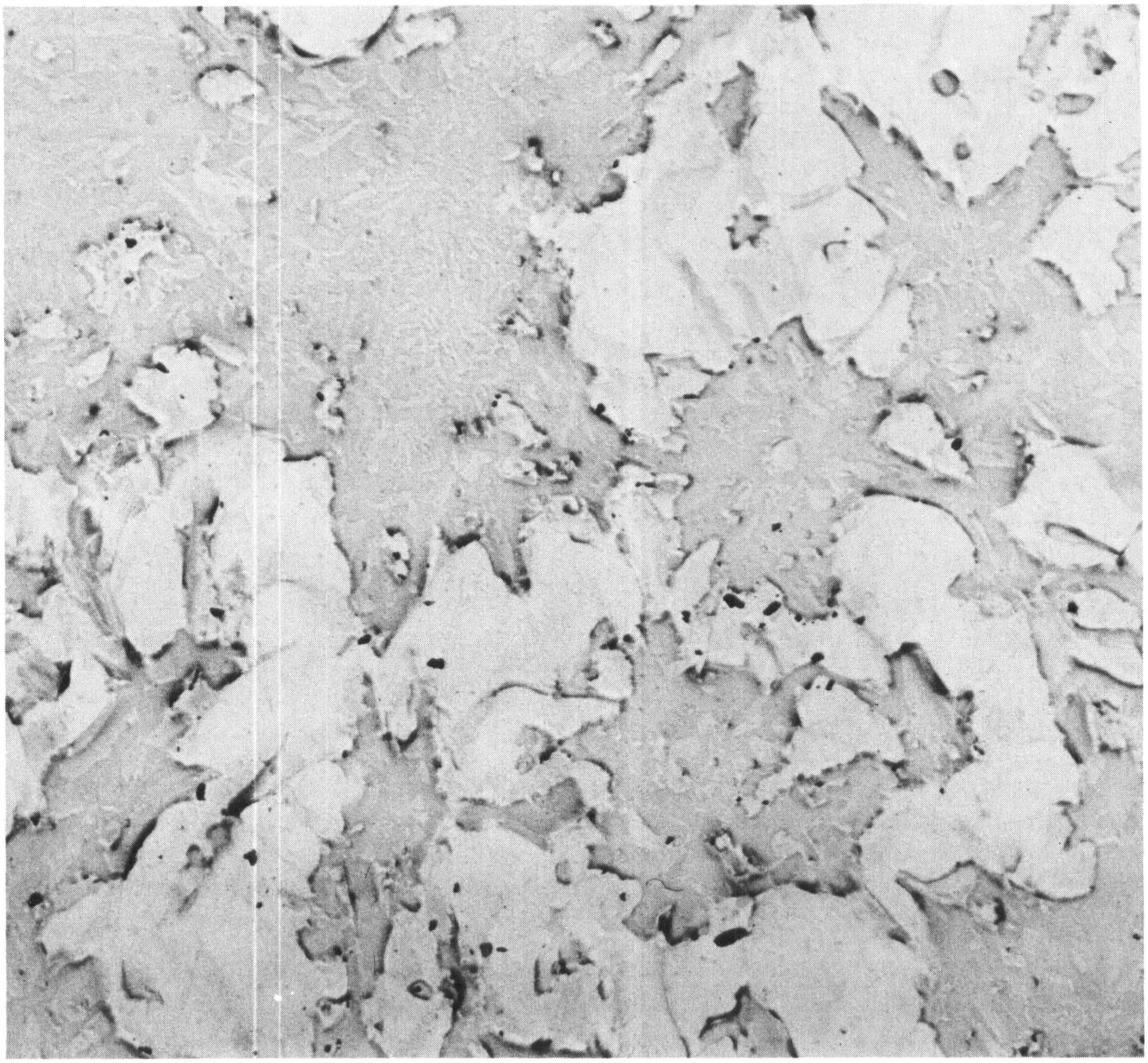
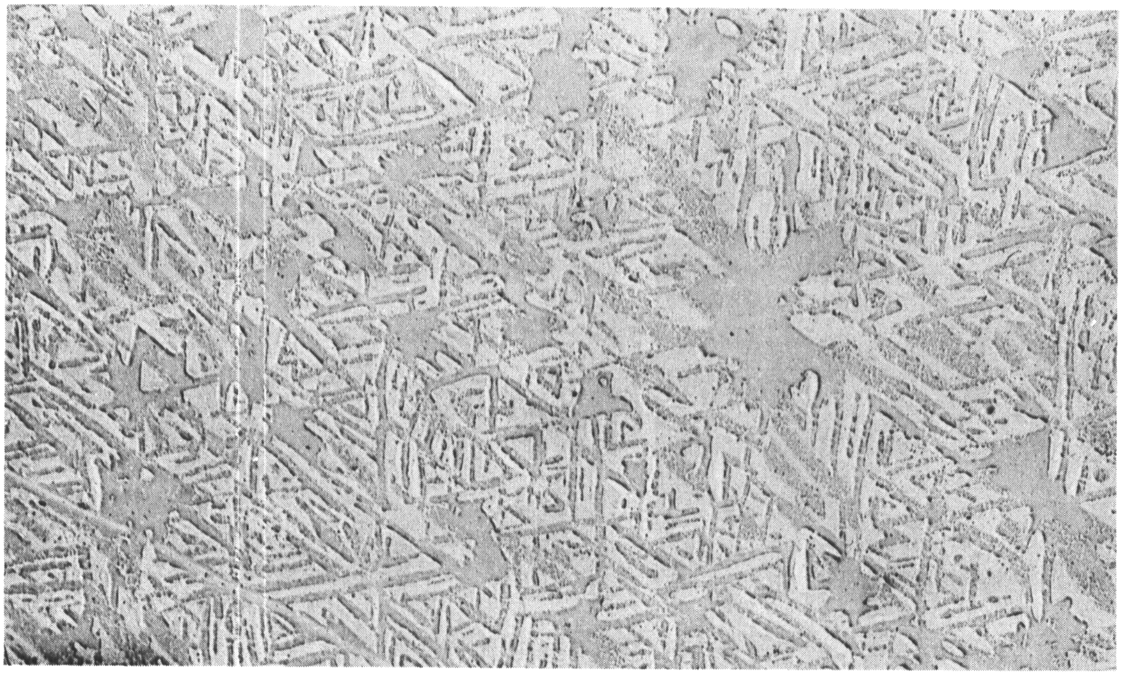


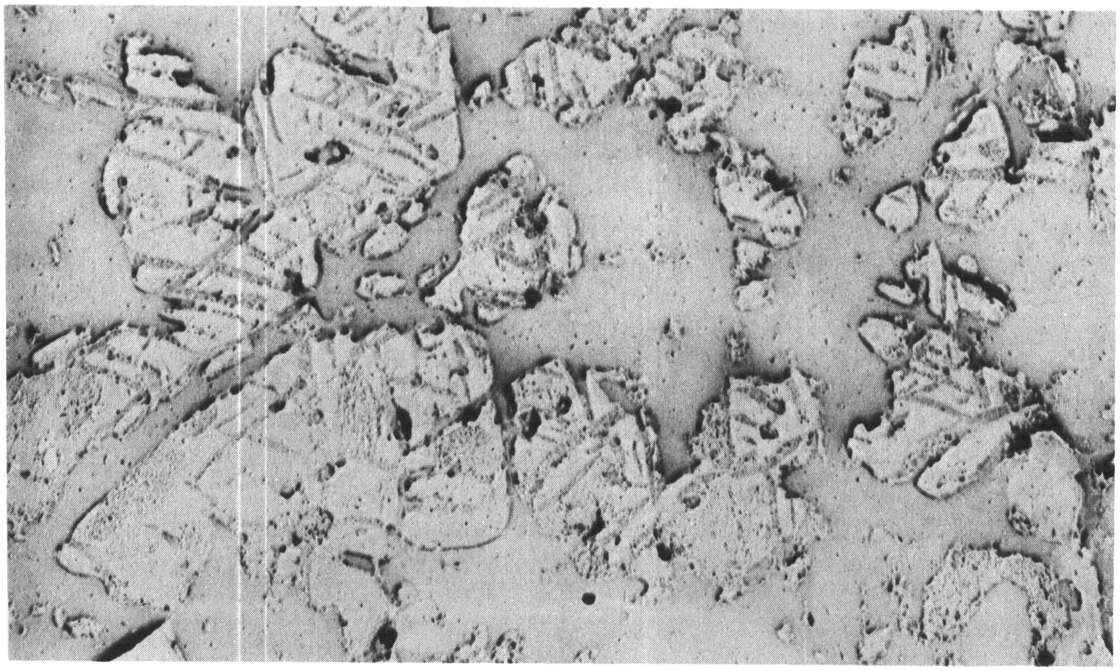
FIGURE 87. Ti-20 PER CENT Cr ALLOY ISOTHERMALLY TRANSFORMED FOR 6 HOURS AT 1112 F (600 C); A POSITIVE REPLICA, PLATINUM-SHADOWED ELECTRON MICROGRAPH

Structure: $TiCr_2$ Lamellae in Dark-Etching Alpha Fields; Retained Beta is Light-Etching Constituent



4000X 3212E

FIGURE 88. 7.54 PER CENT Cr ALLOY ISOTHERMALLY TRANSFORMED FOR 24 HOURS AT 1022 F (550 C)
Structure: Widmanstätten Alpha Plus $TiCr_2$ in a Beta Matrix



4000X 3211D

FIGURE 89. 7.54 PER CENT Cr ALLOY ISOTHERMALLY TRANSFORMED FOR 48 HOURS AT 1022 F
Structure: Widmanstätten Alpha Plus $TiCr_2$ in a Beta Matrix

An unusual feature of this transformation may be observed by comparing Figures 88 and 89. During its growth, the eutectoid has apparently absorbed previously existing plates of the alpha phase. In the well-known eutectoid systems, such as the iron-carbon system, the eutectoid forms only in those areas occupied by the high-temperature phase when the eutectoid reaction is initiated.

WELDING OF THE Ti-15%Cr ALLOY

One of the major objectives of current titanium alloy development is the production of ductile welds in intermediate strength sheet alloys (140,000-psi tensile strength). Consequently, exploratory welding tests on a few alloys of the intermediate-strength type were considered desirable. The first alloy selected for preliminary evaluation was the Ti-15%Cr alloy described earlier in this report (Table 1).

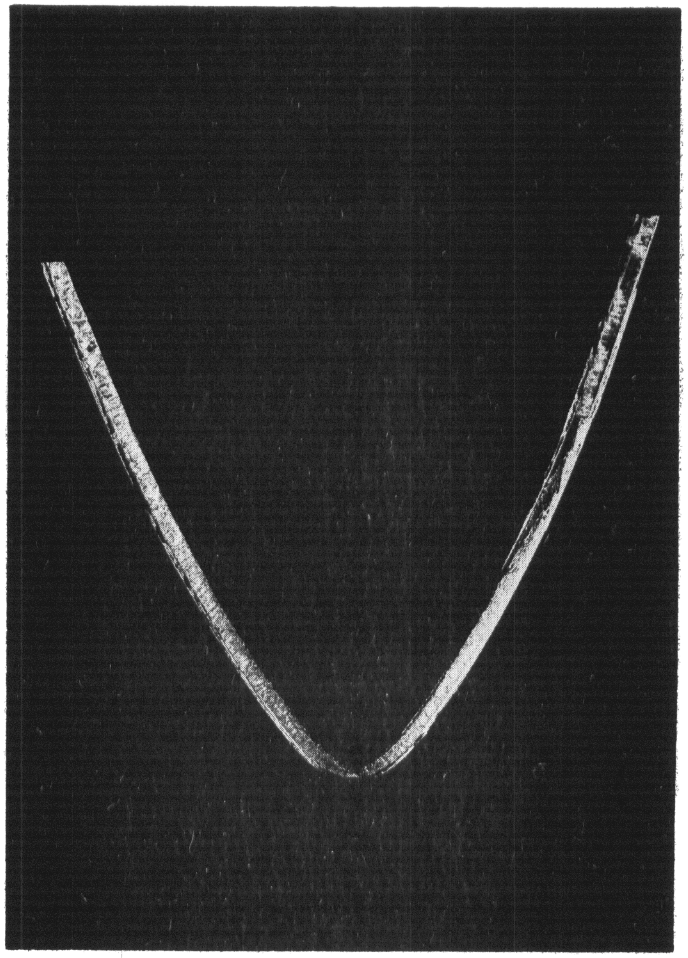
Two strips, 1 inch by 5 inches, of 0.064-inch sheet were used. These strips were joined along the 5-inch dimension by a single-pass weld. The strips were clamped to a steel back-up plate and welded in a controlled-atmosphere chamber filled with tank helium of 99.99+% purity. A Heliarc torch was employed with filler rod consisting of 1/8-inch strips sheared from the original sheet. The completed weld had a bright surface, and no cracking or porosity could be detected by visual examination.

The welded sheet was then sheared into a 1 x 4-inch specimen and tested for bend ductility using the standard bend test dies. The minimum bend radius was 3/16 inch or about 3T. An edgewise view of the specimen after bending over a 1/8-inch radius is presented in Figure 90. A transverse crack appeared in the weld metal after bending over the 1/8-inch radius, but it did not extend into the base metal.

A transverse section was cut from the end of the bend specimen for microscopic examination and hardness tests. A hardness survey was made on this specimen using a Knoop indenter with a 500-gram load. Hardness impressions were made at 0.1 mm (0.004 inch) intervals starting in the base metal and extending to the center of the weld. All hardnesses fell within the range 280 to 310 Knoop. The variations in hardness showed no consistent trend; thus, the average hardness of the base metal, heat-affected zone, and weld metal were essentially the same.

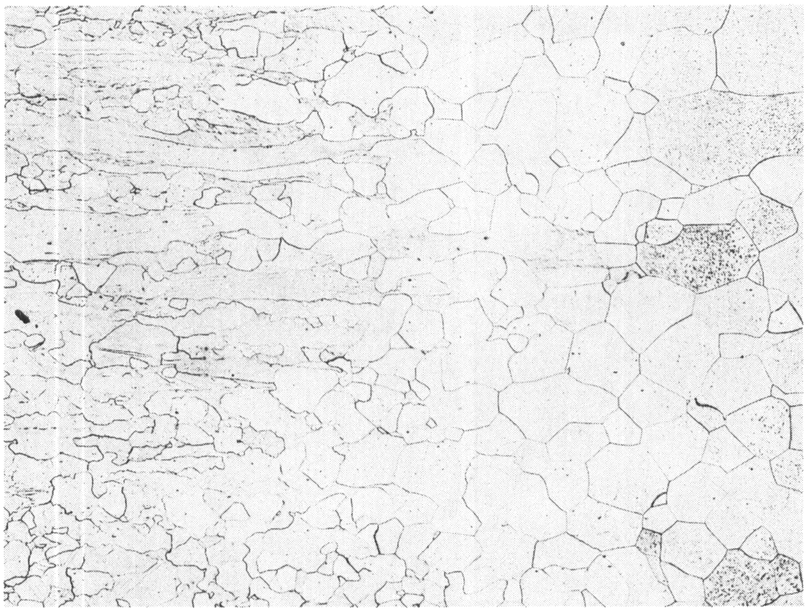
The microstructure of the base metal, shown at the left of Figure 91, is single-phase, elongated beta grains. The heat-affected zone immediately adjacent to the base metal (Figure 91) consists of single-phase, equiaxed beta grains. Nearer the weld, the grain size became larger and a precipitate of alpha and/or $TiCr_2$ appeared as shown in the right side of Figure 91 and in Figure 92. The weld metal was large grained and contained a heavy precipitate (Figures 93 and 94). Since the bend tests were made before the weld specimens were sectioned for metallographic examination, it is not known whether the crack shown in Figure 93 existed prior to bending. If such was the case, cracking may be prevented in future work by applying a suitable postweld heat treatment.

The success of these experiments in producing a ductile weld in an intermediate-strength titanium alloy has greatly stimulated interest in this field. Further tests are in progress on intermediate-strength alloys of both the metastable beta and the alpha-beta type.



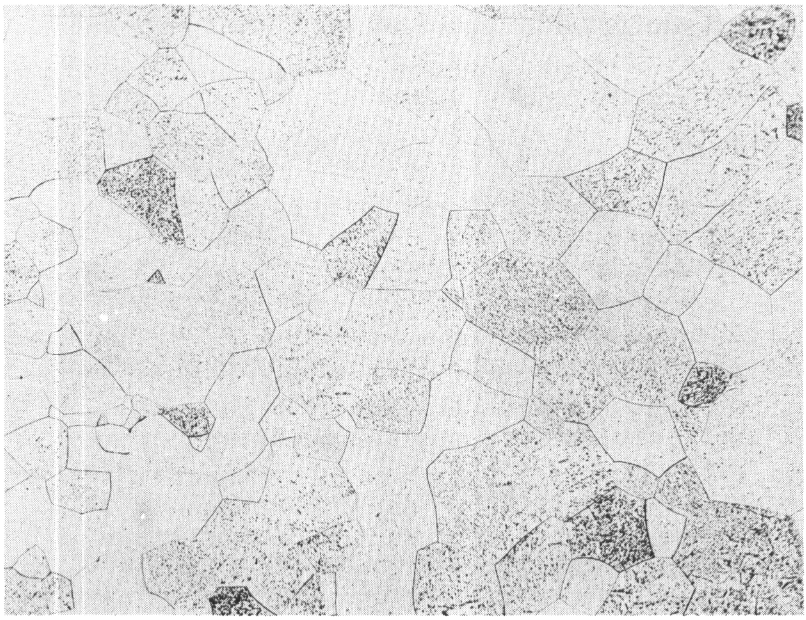
2X 93419

FIGURE 90. PHOTOGRAPH OF ARC-WELDED 0.064-INCH SHEET OF THE Ti-15 PER CENT Cr ALLOY AFTER BENDING OVER A 2T RADIUS



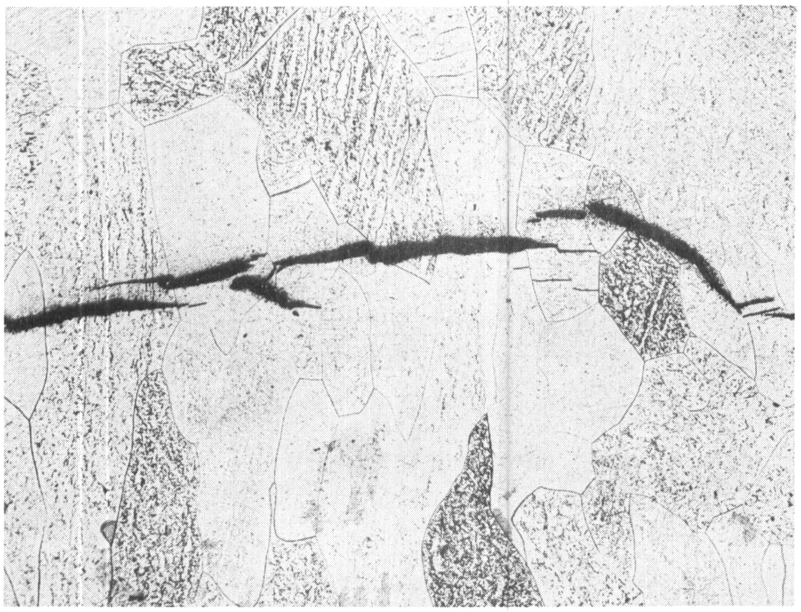
100X 93427

FIGURE 91. ARC-WELDED Ti-15 PER CENT Cr ALLOY, SHOWING SINGLE-PHASE BETA STRUCTURE OF BASE METAL (AT LEFT) AND BEGINNING OF HEAT-AFFECTED ZONE



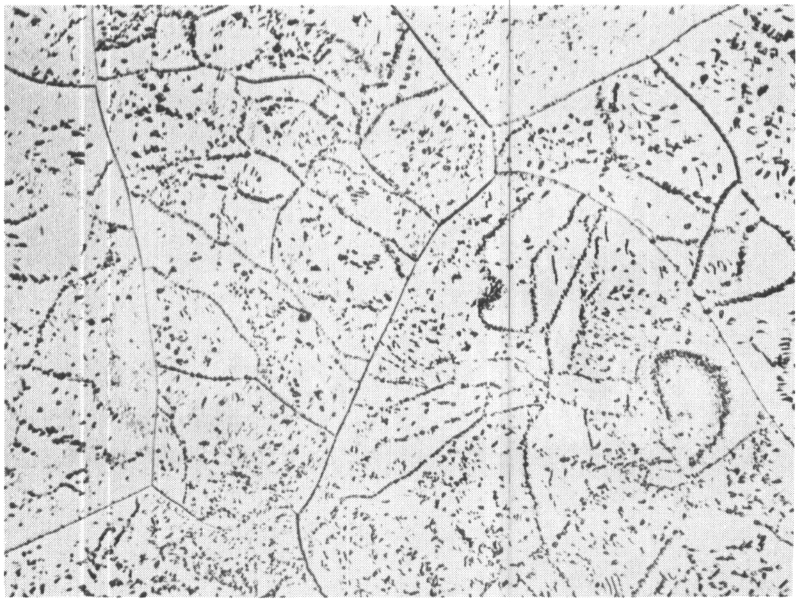
100X 93426

FIGURE 92. ARC-WELDED Ti-15 PER CENT Cr ALLOY, SHOWING HEAT-AFFECTED ZONE NEAR WELD AND EQUIAXED BETA GRAINS WITH ALPHA PRECIPITATE AND POSSIBLY $TiCr_2$



100X 93424

FIGURE 93. ARC-WELDED Ti-15 PER CENT Cr ALLOY SHOWING STRUCTURE OF WELD METAL AND INTERNAL CRACK



500X 93429

FIGURE 94. ARC-WELDED Ti-15 PER CENT Cr ALLOY SHOWING STRUCTURE OF WELD METAL AT HIGHER MAGNIFICATION

Structure: Beta Matrix With Alpha Precipitate and Possibly Some $TiCr_2$

EXTRUSION OF TITANIUM

Metal Trims, Inc., of Youngstown, Ohio, has contracted with the Material Laboratory, WPAFB, to develop a method for extruding titanium. This contract, signed in March, 1952, was undertaken as a result of preliminary extrusion tests on unalloyed titanium billets supplied to Metal Trims by Battelle. The extrusions were described briefly in Reference (2), pp 127-138.

Several American firms have obtained licenses, through the Vanadium Corporation of America, to use the Ugine Séjournet extrusion process. This method consists essentially of heating the billet in a salt bath to avoid oxidation; wrapping it in a blanket of woven glass and extruding at high temperatures. The glass, plastic or molten at extrusion temperatures, provides the necessary lubrication and protects the dies from direct contact with the metal. Several compositions may be used, one of which is shown below:

SiO_2	60%
Al_2O_3	5%
CaO	14%
Na_2O	14%
Other	7%

Although this process is apparently very successful, the high license costs may discourage many businesses from using it. Furthermore, it is not known that this method of providing lubrication is necessarily the easiest and most effective.

The objective of the current program is to develop a technique for extruding titanium which will be independent of the Ugine Séjournet process, but equally as effective. The extrusions will be limited to a maximum cross-section diameter of 2 inches and will comprise shapes of potential use in aircraft.

The extrusion process development will be carried on at Metal Trims. Under a separate contract, Battelle will melt and supply extrusion billets of unalloyed and alloyed titanium. In addition to supplying the billets, Battelle will examine and evaluate the extrusions, and serve in an advisory capacity.

Extrusion Billets

Because of a long delay in the processing of the Metal Trims contract, the production of extrusion billets was not started until May, 1952. Since May, ten extrusion billets have been prepared and shipped to Metal Trims.

It is expected that between 50 and 125 billets, 4-7/16 inches in diameter by about 6 inches long, will be required to establish the proper extrusion temperature, pressure, lubricants, and die materials. In order to conserve titanium, the scrap titanium which has accumulated during the past several years on the alloy development project at Battelle is being melted and fabricated into billets for some of the initial trials. This scrap consists of alloys and unalloyed titanium in the form of broken tensile specimens and discarded ingots. Another source of scrap titanium is sponge which has accumulated because it is too fine to melt successfully in the arc furnaces. Much of the sponge is too high in nitrogen to use in the alloy development program. The processing of this material for melting is described later. The titanium-alloy scrap mentioned above was melted into ingot form in a continuous 6-inch-diameter arc furnace, using a water-cooled tungsten-tipped electrode and a helium-argon furnace atmosphere. A total of 4 ingots, weighing 270 pounds, were made. Figure 95 is a photograph of the titanium-alloy ingots. The largest ingot melted weighed 103 pounds.

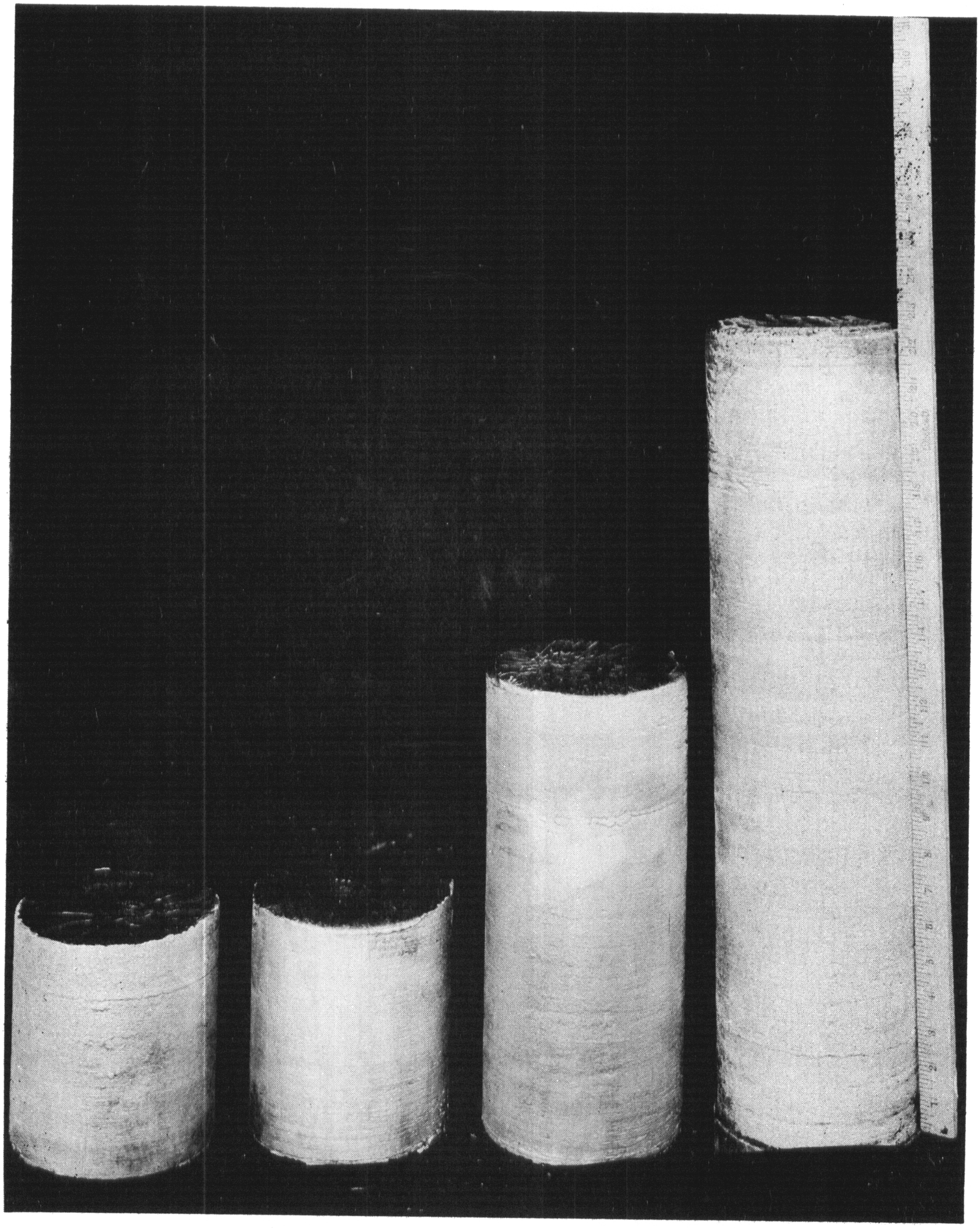
The ingots were hammer forged to approximately 4-3/4 inches in diameter at 1800 F and then machined to 4-7/16 inches in diameter by a minimum length of 6 inches for extrusion purposes. Table 36 is a compilation of data pertinent to the machined billets.

Reclamation of Fine Sponge

Titanium sponge, as purchased from the producer, runs approximately 30 per cent fines. At present, this material is not used for alloy development partly because it cannot be fed into the furnace without striking the electrode and partly because much of it is contaminated. Over a period of four years, approximately 700 pounds of titanium sponge of less than 6 mesh have been accumulated at Battelle. If it were usable, this would represent a value of $3,500.

It was believed that this sponge could be extruded into bars from which it could be rolled to sheet for remelting into ingots for extrusion purposes. This would effect a small saving on material cost and would also yield valuable information on the extrusion properties of titanium sponge.

It was learned that Revere Copper and Brass, Inc., at Detroit, had extruded solid titanium in copper cans. This work was done for a group at M. I. T., which provided detailed information on the process. A 6-1/4-inch-diameter die was designed and built for compacting the titanium sponge into briquettes. The briquettes were sealed under a vacuum in copper cans having a wall thickness of 1/16 inch. Figure 96 is a photograph of the compacting die mounted on the press, the scrap titanium sponge, the titanium sponge compacted into briquettes, the copper container, and a charged copper container ready for extruding.



91944

FIGURE 95. SIX-INCH-DIAMETER TITANIUM-ALLOY INGOTS FOR EXTRUSION BILLETS

Weight of Ingots From Left to Right: 30, 30, 60, and 103 Pounds, Respectively

TABLE 36. DATA ON SCRAP TITANIUM-ALLOY EXTRUSION SLUGS(1)

Identification(2)	Length Dimensions, in.(3)	Weight, lb	Hardness, BHN
10-1	6	15-1/2	277
10-2	6	15-1/2	289
10-3	6	15-1/2	296
10-4	6	15-1/2	294
10-5	5-3/4	14-1/2	305
11-1	6-1/4	16-3/8	301
11-2	6-1/4	16-1/8	309
11-3	6-3/8	16-1/4	306
11-4	6-1/4	16	289
12-1	6-1/8	15-1/2	289
	Total	156 3/4	

(1) Alloy composition not known.
(2) First number refers to ingot, dash number refers to extrusion billet.
(3) Diameter in all cases is 4-7/16 inches.

Inasmuch as the equipment at Metal Trims was not yet available and since Revere had previous experience, the sponge was extruded in the Revere plant at Detroit. The temperatures used for extruding the titanium sponge briquettes were 1500 and 1550 F. A 2-inch-diameter rod die was used. The die was preceded by a soft iron nosepiece to prevent galling of the titanium on the die. The extrusion ratio was 11 to 1, and an initial pressure of 2300 to 2800 psi was required to start the titanium sponge through the die. The runout pressure varied from 2200 to 2500 psi to complete the extrusion. Figure 97 is a photograph of a typical solid titanium bar extruded from the sponge.

It was found that the brass cutoffs, used to back up the copper extrusion cans, extruded to some depth into the center of the titanium bars. It is believed that the brass cutoffs were too hot during extrusion, and, if maintained at a lower temperature, this difficulty would have been avoided.

The extruded rods were forged to flat stock and rolled to sheet. The copper jackets were removed by grit blasting followed by pickling. The rolled sheet is now being melted into extrusion billets which will be shipped to Metal Trims, Inc.

Several of the extruded rods were forged to 1-inch-square or round sections from which tensile specimens were removed. This enabled a comparison to be made between the tensile properties of as-extruded and extruded and forged sponge titanium. Table 37 is a compilation of the tensile data obtained on the specimens.

TABLE 37. TENSILE DATA(1) OBTAINED ON THESE SPECIMENS

Specimen Condition	Sponge Mesh Size	Yield Strength at 0.2% Offset, psi	Ultimate Strength, psi	Elongation in 2 inches, %	Reduction of Area, %
As extruded	-20	63,000	82,100	9.5	9.5
As extruded	-6 +20	41,950	58,800	17.5	26.0
Annealed(2)	-20	69,500	81,700	2.5(5)	1.1
Annealed(2)	-6 +20	43,700	57,600	16.0	24.6
As forged(3)	-20	73,000(4)	86,500	8.5	10.4
As forged(3)	-6 +20	41,200(4)	51,900	23.0	34.1
Forged(3) and annealed(2)	-6 +20	37,500(3)	48,400	31.0	40.1

(1) Specimens were 0.505-inch-diameter tensile bars.
(2) Annealed at 1700 F and furnace cooled.
(3) Forged at 1700 F and air cooled.
(4) Single-value determination.
(5) Brittle fraction.

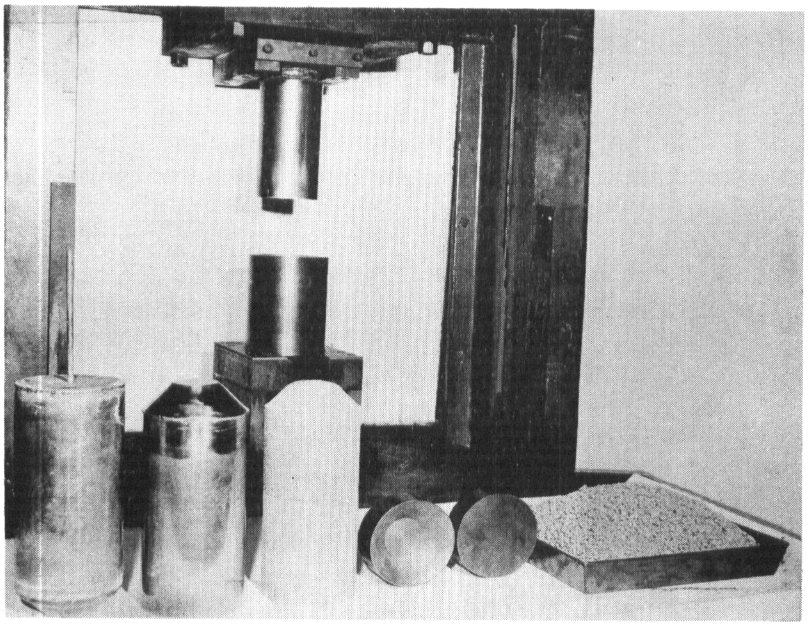
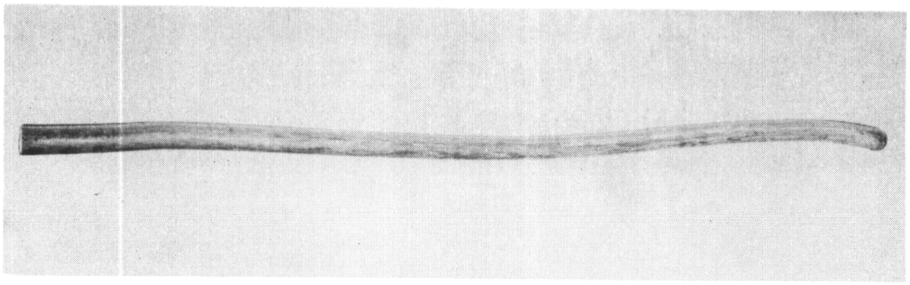


FIGURE 96. COMPACTING DIE, TITANIUM SPONGE, TITANIUM-SPONGE BRIQUETTE, COPPER CONTAINER, AND ASSEMBLED COPPER CAN CONTAINING TITANIUM SPONGE READY FOR EXTRUSION

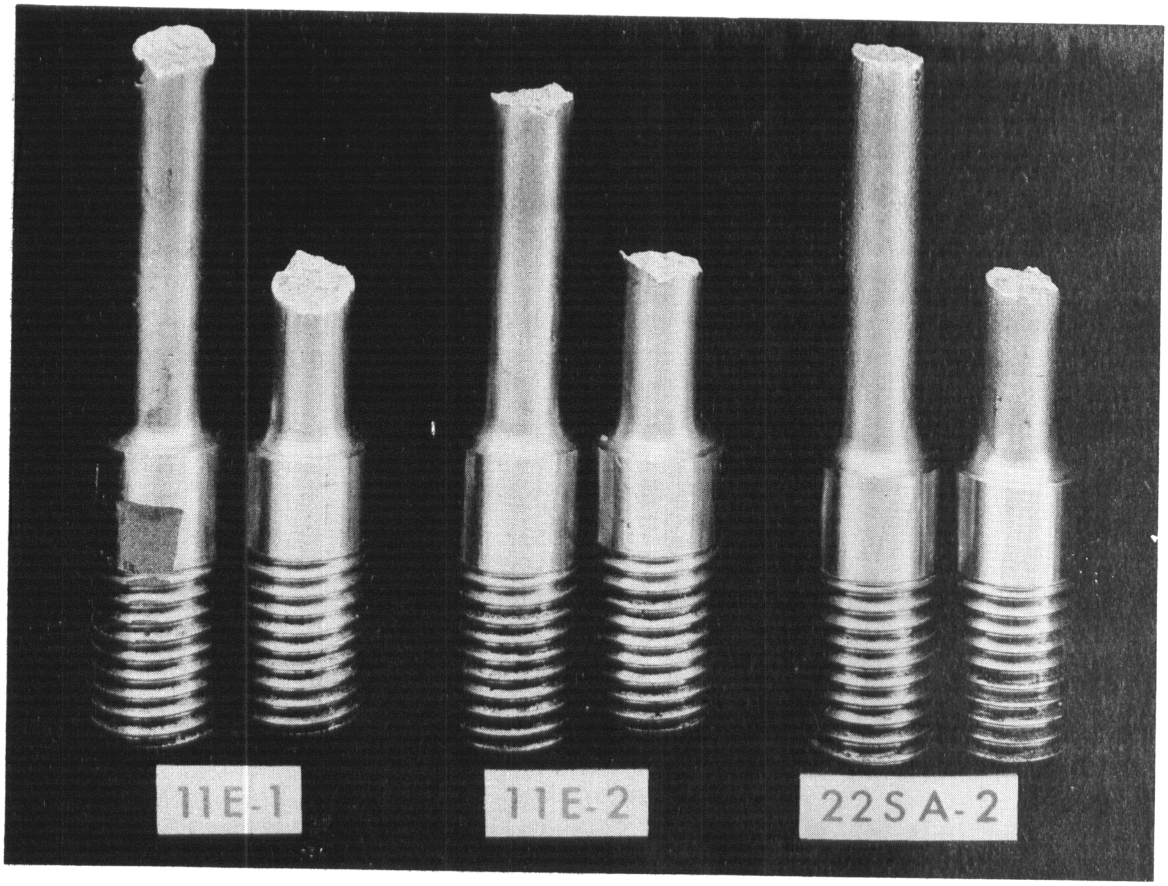


91649

FIGURE 97. SOLID 2-INCH-DIAMETER BAR EXTRUDED FROM TITANIUM SPONGE
Length of Bar: Approximately 6 Feet

The titanium extrusions prepared from minus 20-mesh titanium sponge had higher strengths and less ductility than the minus 6 to plus 20-mesh titanium sponge extrusions. This was expected because of the greater surface area of titanium available for contamination. Annealing of the extruded material did not have any appreciable effect on the tensile properties. However, forging, followed by annealing, decreased the yield and ultimate strengths and increased the ductility of minus 6 to plus 20-mesh titanium sponge extrusions. These tests are quite limited in number and the results do not, therefore, represent properties which should be expected if the process became commercial. Figure 98 is a photograph of several of the fractured tensile specimens. Figure 99 is a photomicrograph of a bar extruded from sponge which had a mesh size of minus 6 to plus 20. Figure 100 shows the structure of a similar bar except that the extruded bar was forged to a 1-inch square and annealed. The grain structure in both instances are equiaxed, but, in the annealed specimen, the grain size is considerably larger. Both photomicrographs exhibit what appear to be either voids or inclusions.

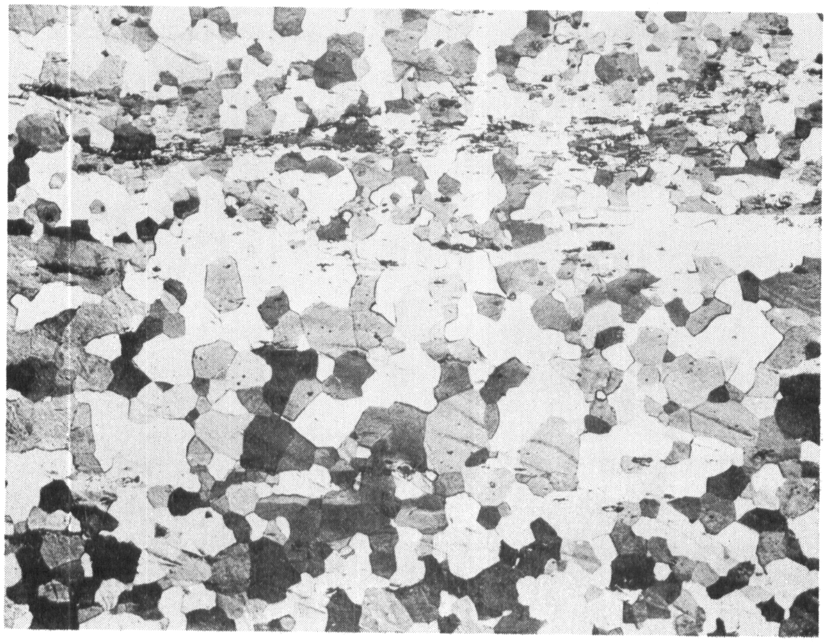
Data from which this report was prepared are contained in Battelle Memorial Institute Laboratory Notebooks Nos. 5948, 6326, 6700, 6722, and 6741.



93395

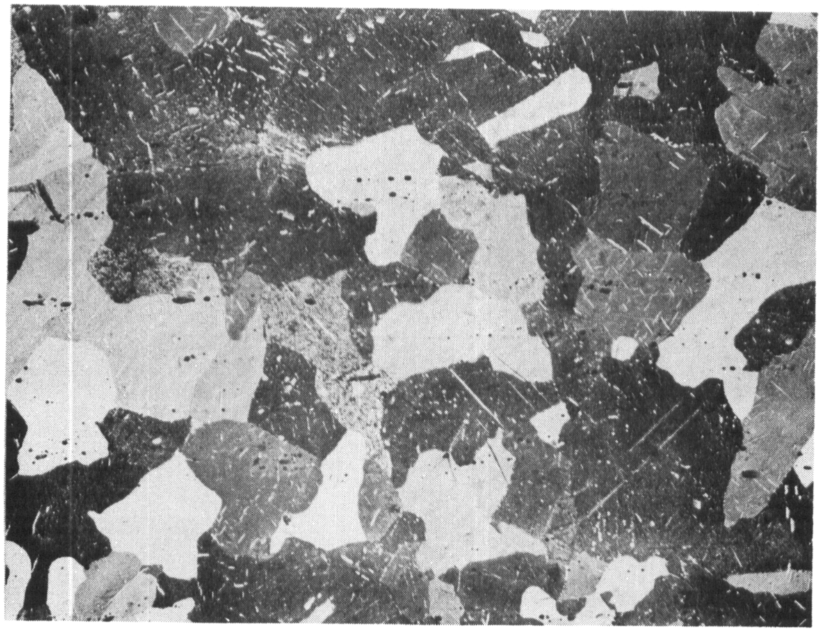
FIGURE 98. TYPICAL TENSILE FRACTURES OF BARS EXTRUDED FROM TITANIUM SPONGE

Specimens 11E-1 and 11E-2 are of As-Extruded Titanium Sponge, -6 to +20-Mesh Size; Specimen 22SA-2, Prepared From Same Mesh-Size Sponge, was Forged to a 1-Inch Square and Annealed at 1700 F



100X 93433

FIGURE 99. ROD EXTRUDED FROM -6 to +20-MESH TITANIUM SPONGE
Note Presence of Voids and Inclusions



100X 93435

FIGURE 100. ROD EXTRUDED FROM -6 to +20-MESH TITANIUM SPONGE, FORGED TO 1-INCH SQUARE, ANNEALED AT 1700 F, AND FURNACE COOLED

REFERENCES

1. Summary Report - Part III, "Preparation and Evaluation of Titanium Alloys", July 30, 1949.

2. AFTR 6623, "Development of Titanium-Base Alloys", June, 1951.

3. AFTR 6218, Part 2, "Research and Development on Titanium Alloys", June, 1950.

4. Van Thyne, R. J., Kessler, H. D., and Hansen, M., "The Systems Titanium-Chromium and Titanium-Iron"; Trans. ASM, Vol 44, 1952, pp 974-989.

5. DeLazaro, D. J., Hansen, M., Riley, R. E., and Rostoker, W., "Time-Temperature Transformation Characteristics of Titanium-Molybdenum Alloys"; Journal of Metals, March, 1952, pp 265-269.

APPENDIX I

PREPARATION OF EXPERIMENTAL TITANIUM-BASE ALLOYS

APPENDIX I

PREPARATION OF EXPERIMENTAL TITANIUM-BASE ALLOYS

Arc-Melting Furnaces

The single-electrode furnace described in Reference (1), pp 24-31, inclusive, is still being used in the preparation of alloys. Ingots weighing 3/4 pound to 2 pounds have been prepared in this furnace for the study of titanium-base alloys. The arc is maintained between a water-cooled, tungsten-tipped electrode and the charge, which is contained in a water-cooled copper crucible. The positions of the electrode and, hence, the length of the arc are controlled by the operator.

A more durable crucible than the thin-walled spun crucible has been made for use in the small single-electrode furnace. The new crucible is made from copper tubing 2-1/4 inches ID by 2-3/4 inches OD. A machined flange is attached to the tubing with soft solder and a machined bottom is attached by brazing or with a high-melting-point silver solder. The use of a heavy-walled crucible has reduced puncturing of the crucible by the arc. However, the thin-walled, spun-copper crucible is still preferred for making most ingots smaller than one pound because the bottom is partially rounded and the small ingots made in it have a more satisfactory shape for forging.

A second 4-inch, single-electrode furnace of the type described in Reference 2 has been constructed. This is being used to produce 10- and 20-pound ingots of the various alloys. The crucible for this furnace is made of 4-inch ID by 4-1/2-inch OD copper tubing, having a plate brazed to the bottom and a machined flange brazed to the top.

Melting Procedure

Initial Melting

The melting procedure described in Reference (1), pp 21-32, was modified in an attempt to reduce or eliminate contamination of the ingots by tungsten from the electrode. The "long arc" melting technique in use at present is discussed here.

With the present furnaces, the best way to prevent tungsten contamination of the ingot is to melt the sponge in such a manner that spattering of molten titanium onto the hot tungsten electrode is minimized. Leaching and vacuum treatment of the sponge prior to melting also reduces spattering during melting.

For making 3/4-pound to 2-pound ingots, a starting charge of 70 grams is placed in the bottom of the crucible before the arc-melting furnace is assembled. After the furnace is assembled, it is evacuated and flushed with argon twice and then argon is allowed to flow through the furnace at a pressure of about 1-1/2 inches of mercury. Direct current from a 900-ampere welding generator is supplied to the electrode for melting. The open-circuit voltage of the generator is 75 volts. Melting is started by adjusting the generator to give its minimum current, about 250 amperes, and then lowering the electrode until the tip touches the titanium sponge and an arc is struck. The electrode is then immediately raised until the arc operates at 30 volts. During normal operation, the gap between the electrode and the titanium sponge is about 1-1/2 to 2 inches. This combination of arc length and voltage is maintained until the surface of the original 70-gram titanium sponge charge is almost completely melted. The electrode is then lowered until the arc voltage is 20 to 22 volts and the current 25 amperes. Melting is continued until as much as possible of the original charge has been melted under these conditions. The electrode is then held stationary and the current is increased to about 450 amperes. The arc voltage usually increases along with the arc current, even though the arc gap is not changed. After a short interval, the arc voltage is again reduced to 20 volts by gradually lowering the electrode. A current of 450 amperes and an arc voltage of 20 volts are maintained until as much of the original charge as possible has been melted. Without changing the electrode gap, the current is then increased to about 600 amperes. Finally, the current is increased to 750 amperes at 20 volts. Melting is continued until the arc has "broken through" the melt, that is, until the pool of molten metal in the crucible has become so fluid that the arc penetrates to the bottom of the crucible in the deepest portion of the molten pool. The current is then immediately reduced to 250 amperes in preparation for adding sponge titanium.

Before dropping in new sponge, the electrode is raised 2 to 3 inches and moved near the side of the crucible away from the charging hole. Enough sponge to cover the top of the melt is then dropped in gradually. The arc is rather unstable at this time, not only because of its length, but also because the surface of the melt is almost frozen. Care must be taken not to feed too much sponge titanium at one time or the arc will be extinguished. The freshly charged material is slowly melted by gradually lowering the electrode to decrease the arc gap and arc voltage, and by increasing the current. Melting then proceeds much as described for the initial 70-gram charge. When the new charge has melted satisfactorily next to the crucible wall, the current is lowered, the electrode raised, and a new charge introduced carefully as before. After five or six charges have been melted, it is possible to raise the current to 750 amperes in two steps after each addition. Seven to ten additions are usually used for a 3/4-pound ingot. On the last two or three charges, the current can often be raised directly from 250 to 750 amperes. This is possible because the molten pool is deeper toward the end of the melt and the melt does not freeze during charging. At this stage, a current of 250 amperes is sufficient to melt all but the extreme edges of the ingot. A current of 750 amperes is still required to melt the edges sufficiently to produce a smooth surface on the ingot.

The primary advantages of the present technique over the one formerly used are (1) much of the melting is done with a long arc gap, thereby minimizing titanium spatter onto the electrode tip; (2) the current is low during initial melting of the sponge, permitting slow gas evolution from the sponge, with a resulting reduction in spatter; (3) after the initial charge, the sponge rests on the hot ingot for a time after it is charged and evolves much of its gas before it becomes submerged, which minimizes spattering; and (4) charges are made in smaller increments and melting is done slowly so there is less likelihood of gas pockets being formed under the ingot surface. The disadvantages of the present technique are (1) the greater time required to melt an ingot (40 to 60 minutes for a 3/4-pound ingot, as compared with 10 to 15 minutes with the former practice); (2) greater likelihood of oxygen or nitrogen contamination because of the long melting time; and (3) greater danger of burning through a crucible because of the long arc lengths and long melting times used.

Two very recent changes have been made to improve the melting technique. One is the use of a magnesium ribbon to start the arc. This eliminates loss of tungsten from the electrode during striking of the arc. The magnesium starting ribbon is wrapped around the electrode tip with about 1/4-inch extending beyond the tip. On striking the arc, the magnesium volatilizes and does not enter the ingot. The other change in technique entails raising the current slowly, rather than suddenly. A sudden increase in current may cause a shallow pool of metal to splash onto the electrode tip.

The melting procedure for the 4-inch arc-melting furnace is quite similar to that for the smaller furnace. A starting charge of 300 grams is used. The arc is struck with an open-circuit voltage of 90 volts. The initial current is 500 amperes, and is slowly increased to approximately 1500 amperes in melting the initial charge. When an addition is made, the electrode is raised until a glow discharge is observed at the spatter line (i.e., the height to which titanium has spattered onto the crucible wall). The electrode is maintained in this position while the charge is added slowly. Care must be exercised to prevent feeding of a large amount of fine particles of sponge at one time. Too large an amount of fine particles may extinguish the arc or cause the arc to burn through the side of the crucible. The amount of material charged at any one time is determined by the operator. The newly charged sponge is permitted to rest on the hot ingot until the gas evolution from the sponge diminishes, after which the electrode is lowered and melting resumed. The current is not decreased during charging in the 4-inch furnace, as is done in the smaller furnace.

In an effort to reduce tungsten contamination, the use of a nickel-bonded titanium-carbide electrode tip was investigated in the small arc furnace. An initial charge of 100 grams of titanium sponge with a 300-gram feed charge was used. The total melting time was approximately thirty-five minutes. For five minutes of this period, a current of 850 amperes was used. Chemical analysis of the 400-gram button showed 0.34 carbon and 0.06 nickel. The use of the titanium carbide electrode tip, therefore, was unsatisfactory because of the high contamination of the ingots with carbon and nickel.

Remelting

All ingots are forged, rolled, and sheared for remelting. This procedure is used in preference to machining chips because it is more rapid and it permits removal of tungsten-contaminated areas. Furthermore, the previous practice of crushing chips machined from alloy ingots often produced a large amount of very fine particles. These were so light that when they were charged into the furnace, they floated onto the electrode tip, resulting in severe tungsten contamination of the melt.

The present method of preparing the ingots for remelting is as follows. The ingots are forged and rolled to a thickness of approximately 1/8 inch. The plates are sand blasted, radiographed, and the tungsten-contaminated areas marked on the film. The contaminated areas are then traced on the plates directly from the radiograph film and are removed by shearing. The uncontaminated portions of the plates are then sheared into pieces approximately 1/4 inch square, which are then pickled in the following solution:

200 ml	H_2O
100 ml	H_2SO_4
15 g	NH_4F

The pickled, washed, and dried sheared pieces are mixed thoroughly and remelted into the final ingot.

Titanium Melting Stock

The processing of titanium sponge prior to melting is now as follows. The titanium sponge, with minus 8-mesh particles removed, is leached in methanol three times, each time for about 24 hours. Leaching is done in approximately 25-pound lots. The sponge is then oven dried to drive off any residual methanol(1), after which it is vacuum treated at

---

(1) Most of the sponge was dried by evacuating to about 1 mm Hg for periods of 3 to 5 days. It was later found that oven drying was more effective.

700 F. When a pressure of 10 microns is attained, the treatment is complete. The time required for treatment is determined by the amount of magnesium chloride in the sponge. After treatment, the sponge is stored in air-tight cans.

The sponge is screened to a minus 1/2-inch plus 1/8-inch size as charge material for the small furnace. Sponge for the large furnace is screened to minus 3/4 inch, plus 1/4 inch. The oversize material, which is approximately plus 1-1/2 inches, is used as initial charge material.(1) The undersize sponge is used for initial charge material and for making master-alloy compacts.

Method of Making Alloy Additions

Methods of making alloy additions were described in Reference (2), Section I. The only change is that all additions of chromium, manganese, and iron are now made as master alloys. Chromium is added as a Ti-65% Cr master alloy. Manganese and iron additions are made as Ti-50%Mn and Ti-50%Fe master alloys, respectively. These master alloys are arc melted in the same manner as regular ingots. They are crushed in an iron mortar to the size required for alloy additions.

Alloy ingots produced during the period covered by this report are reported in Table 38.

---

(1) Some of the oversize is broken to the feed size in a small hammer mill.

TABLE 38. INGOTS PRODUCED, MAY 19, 1951, to MAY 18, 1952

3/4-Pound Ingots

2V-2.0Mo	3.5V-2.0Mo	5V-2.0Mo
2V-3.5Mo	3.5V-3.5Mo	5V-3.5Mo
2V.5.0Mo	3.5V-5.0Mo	5V-5.0Mo
2V-2Mo-1Cr	3.5V-2Mo-1Cr	5V-2Mo-1Cr
2V-2Mo-3Cr	3.5V-2Mo-3Cr	5V-2Mo-3Cr
2V-2Mo-1Mn	3.5V-2Mo-1Mn	5V-2Mo-1Mn
2V-2Mo-3Mn	3.5V-2Mo-3Mn	5V-2Mo-3Mn
2V-1.5Fe	3.5V-1.5Fe	5V-1.5Fe
2V-3.0Fe	3.5V-3.0Fe	5V-3.0Fe
2V-4.5Fe	3.5V-4.5Fe	5V-4.5Fe
2V-1.5Fe-1Mn	3.5V-1.5Fe-1Mn	5V-1.5Fe-1Mn
2V-1.5Fe-3Mn	3.5V-1.5Fe-3Mn	5V-1.5Fe-3Mn
2V-2.0Mn	3.5V-2.0Mn	5V-2.0Mn
2V-3.5Mn	3.5V-3.5Mn	5V-3.5Mn
2V-5.0Mn	3.5V-5.0Mn	5V-5.0Mn
2V-2Mn-1Cr	3.5V-2Mn-1Cr	5V-2Mn-1Cr
2V-2Mn-3Cr	3.5V-2Mn-3Cr	5V-2Mn-3Cr

2.75Mn-2Fe-1Mo	4.25Mn-2Fe-1Mo
2.75Mn-2Fe-2Mo	4.25Mn-2Fe-2Mo
2.75Mn-2Fe-3Mo	4.25Mn-2Fe-3Mo
2.75Mn-2Cr-1Mo	4Mn-2Cr-1Mo
2.75Mn-2Cr-2Mo	4Mn-2Cr-2Mo
2.75Mn-2Cr-3Mo	4Mn-2Cr-3Mo
2.5Cr-2Fe-1Mo	4Cr-2Fe-1Mo
2.5Cr-2Fe-2Mo	4Cr-2Fe-2Mo
2.5Cr-2Fe-3Mo	4Cr-2Fe-3Mo

2Mn-1.5Co	3.5Mn-1.5Co	5Mn-1.5Co
2Mn-3.0Co	3.5Mn-3.0Co	5Mn-3.0Co
2Mn-4.5Co	3.5Mn-4.5Co	5Mn-4.5Co

0.5Mo-0.5Mn	2Mo-0.5Fe
0.5Mo-9.0Mn	2Mo-7.0Fe
8.0Mo-8.0Mn	8Mo-6.0Fe
9.0Mo-0.5Mn	9Mo-0.5Fe
10.0Mo-10.0Mn	10Mo-8.0Fe

TABLE 38. (Continued)

3.5Cr-3.5V
5.0Cr-1.5Fe

5.0Mo-4.0Fe

3Mn-3.5Cr
5Mn-2.5Cr
5Mn-2.0Fe
5Mn-2.0Mo

1-Pound Ingots

3Cr
5Cr
8Cr

5Mo
10Mo

1.0Fe
2.0Fe
3.0Fe
4.0Fe
5.0Fe
7.5Fe
10.0Fe
15.0Fe

2-1/2-Pound Ingots

15Cr

5-Pound Ingots

3.5Cr-3.5V
5.0Mn-2.5Cr
1Fe-1Cr-3Mn-1Mo-1V
4Fe-1Cr-1Mn-1Mo-1V

10-Pound Ingots

3.5Cr-3.5V
5.0Mn-2.5Cr
4Fe-1Cr-1Mn-1Mo-1V

In Process

10 lb
1Fe-1Cr-3Mn-1Mo-1V

TABLE 38. (Continued)

Ingots Now in Preparation

5-Pound Ingots

5.0Mn-2.0Mo
3.5Mn-3.5Cr
4.5Mo-3.5Fe
12.0Cr
4.0Fe
7.5Fe
10.0Fe

10-Pound Ingots

4.5Mo-3.5Fe
5.0Mn-2.5Cr
5.0Mn-2.0Mo
3.5Mn-3.5Cr

4Fe-1Cr-1Mn-1Mo-1V

APPENDIX II

FABRICATION OF EXPLORATORY ALLOYS

APPENDIX II

FABRICATION OF EXPLORATORY ALLOYS

The fabrication procedure used on most of the alloys described in this report was developed during the previous year's work on this contract, Reference (2), page 15. The procedure consists of upsetting the 2-1/2-inch-diameter ingots on edge to form a cube. The cube is then elongated by forging to a 1-inch-square rod approximately 3 inches long, which is cubed again, and finally flattened into a plate approximately 3/8 inch thick. The plates are scalped on the flat surfaces to a depth of 0.010 inch and hot rolled at 1450 F to 14-gage sheet, using equal rolling reductions in the transverse and longitudinal directions. The cross-rolling procedure was introduced to minimize anisotropy in the resultant sheet.

It is realized that this procedure probably does not produce optimum properties in all alloys. The large number of alloys tested under this contract necessitated the use of a standardized fabrication technique, however.

APPENDIX III

SUPPLEMENTARY AGE-HARDENING DATA ON EXPLORATORY ALLOYS

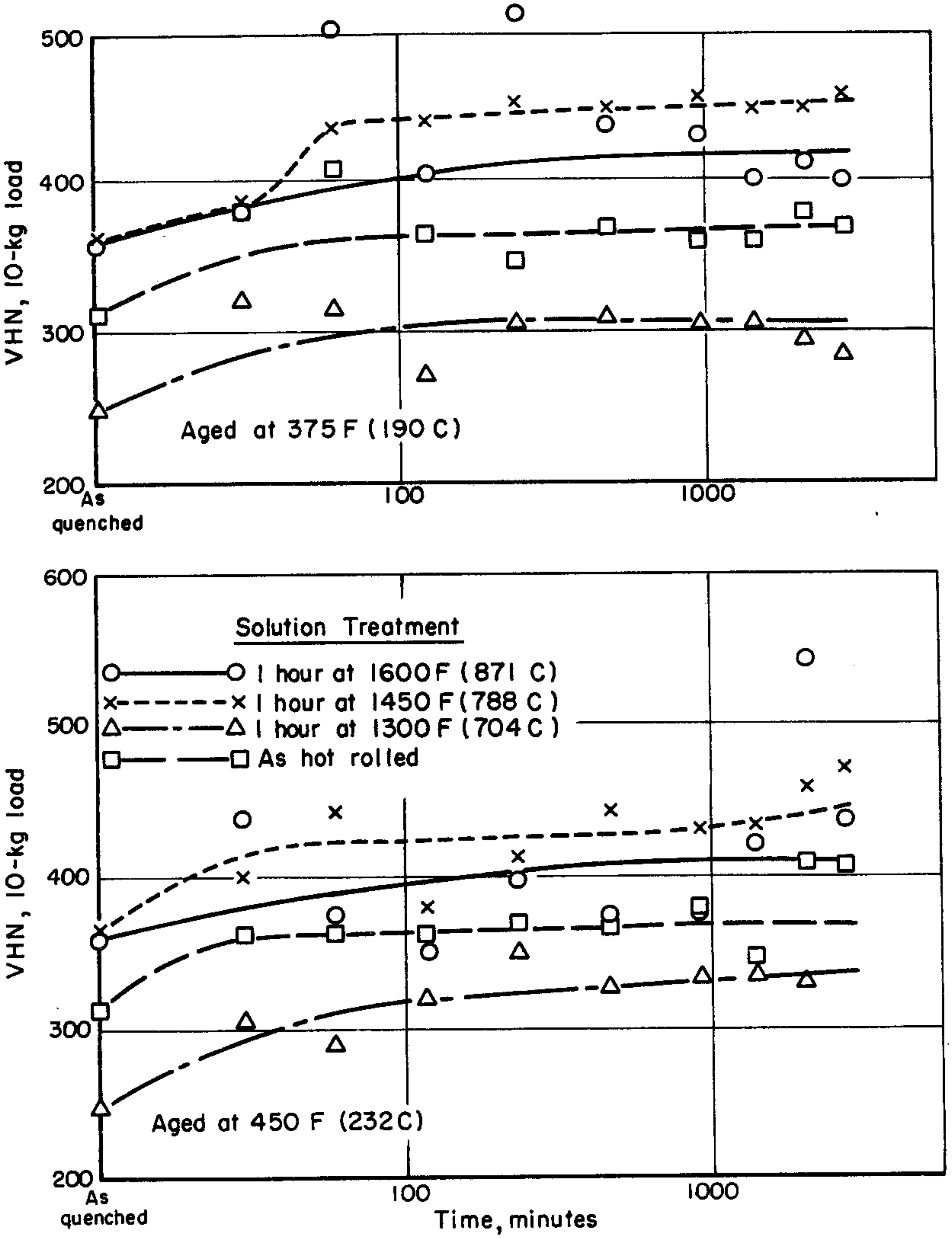


FIGURE 101. HARDNESS VERSUS AGING TIME FOR A Ti-1% Mn-3.5% Cr ALLOY SOLUTION TREATED AT VARIOUS TEMPERATURES

A-3285

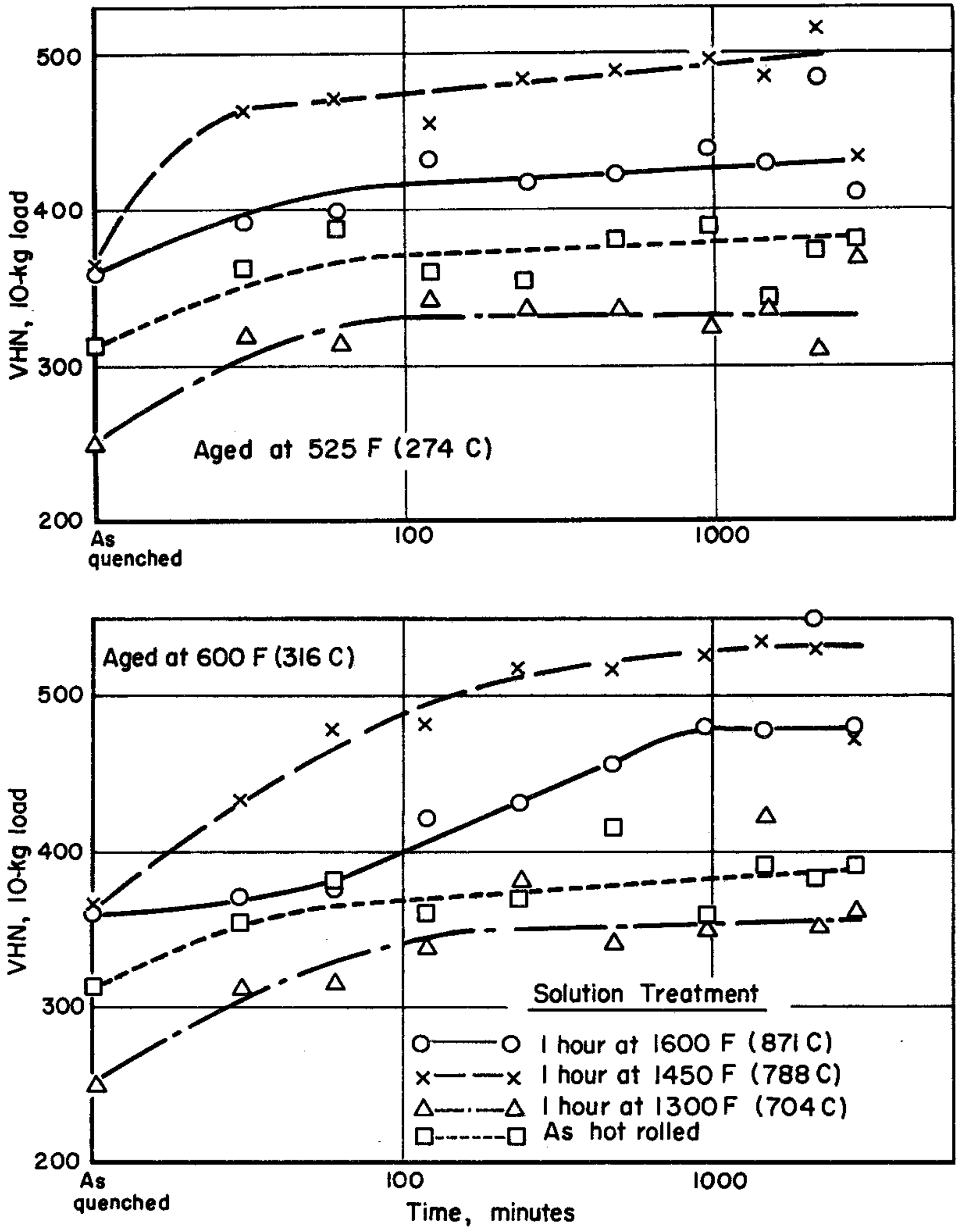


FIGURE 102. HARDNESS VERSUS AGING TIME FOR A Ti-1% Mn-3.5% Cr ALLOY SOLUTION TREATED AT VARIOUS TEMPERATURES

A-3286

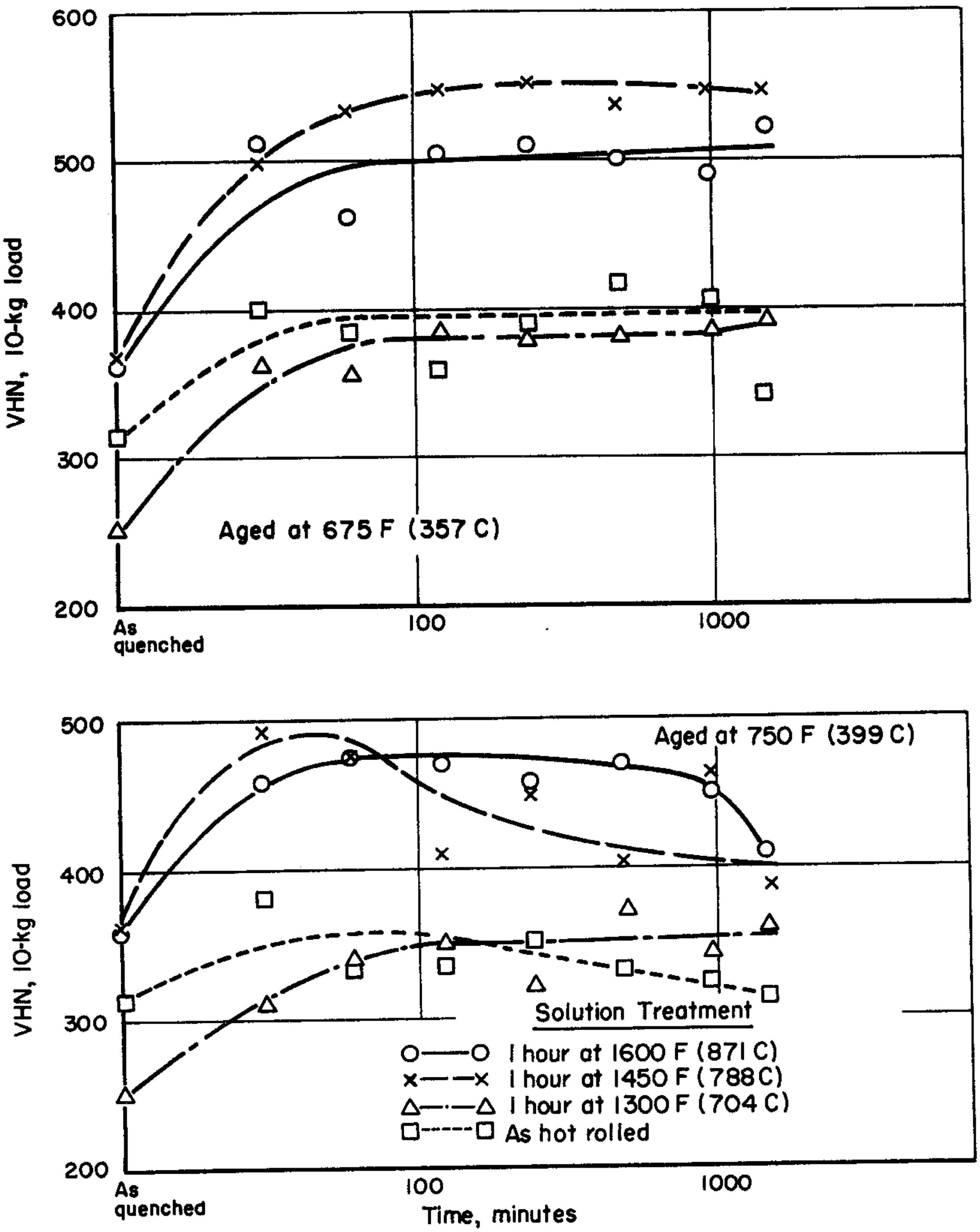


FIGURE 103. HARDNESS VERSUS AGING TIME FOR A Ti-1% Mn-3.5% Cr ALLOY SOLUTION TREATED AT VARIOUS TEMPERATURES

A-3287

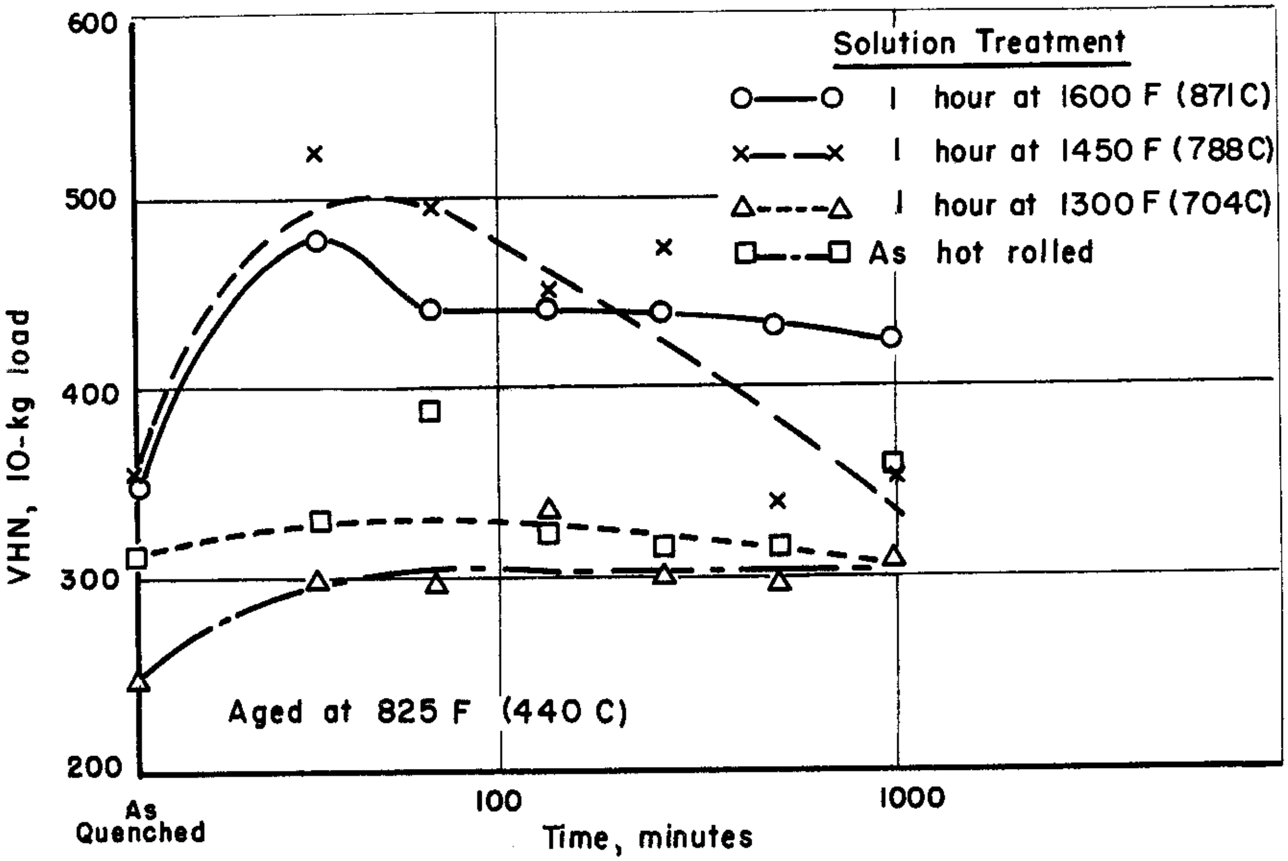


FIGURE 104. HARDNESS VERSUS AGING TIME FOR A Ti-1% Mn-3.5% Cr ALLOY SOLUTION TREATED AT VARIOUS TEMPERATURES

A-3288

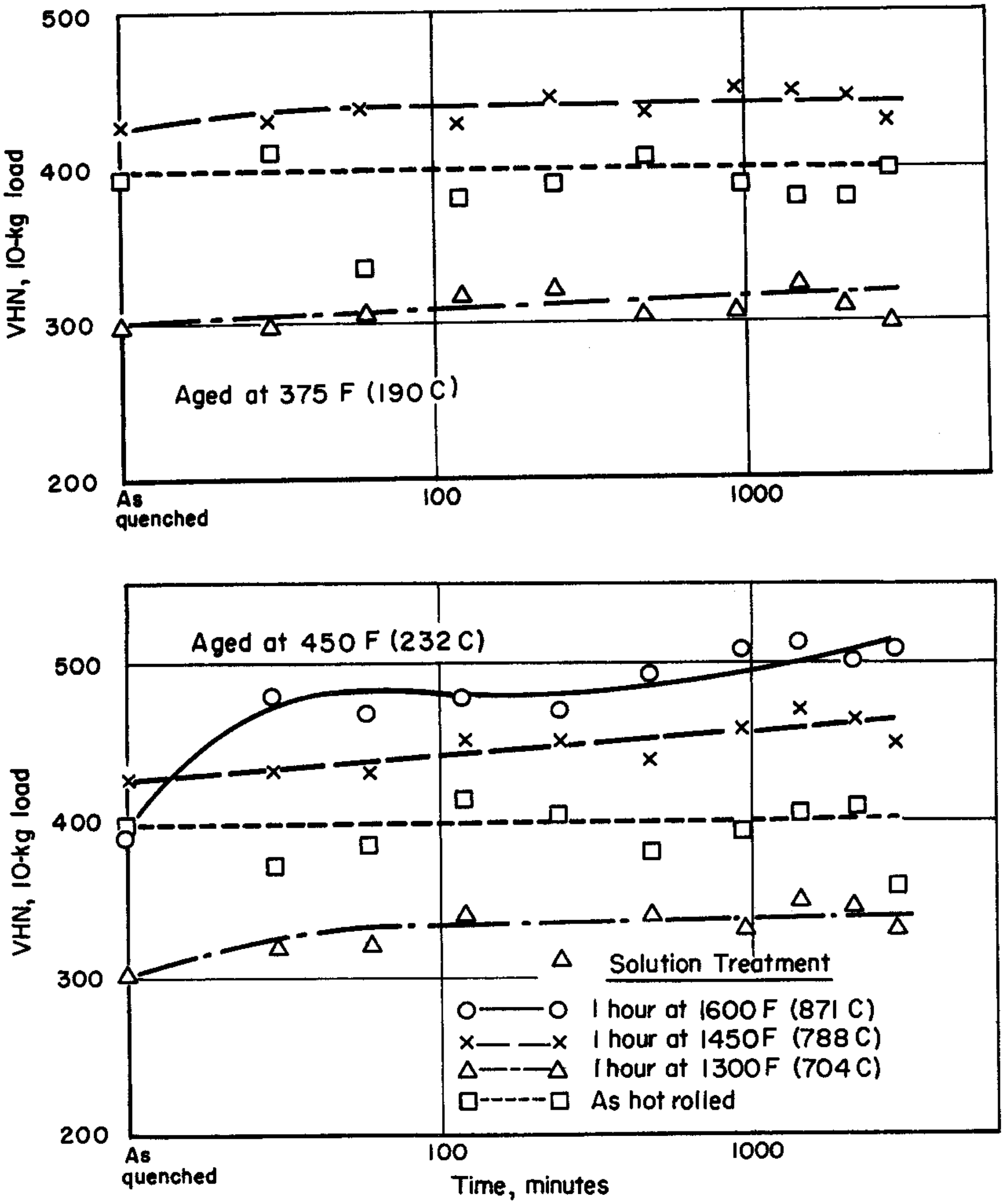


FIGURE 105. HARDNESS VERSUS AGING TIME FOR A Ti-2% Mn-3.5% Cr ALLOY SOLUTION TREATED AT VARIOUS TEMPERATURE

A-3289

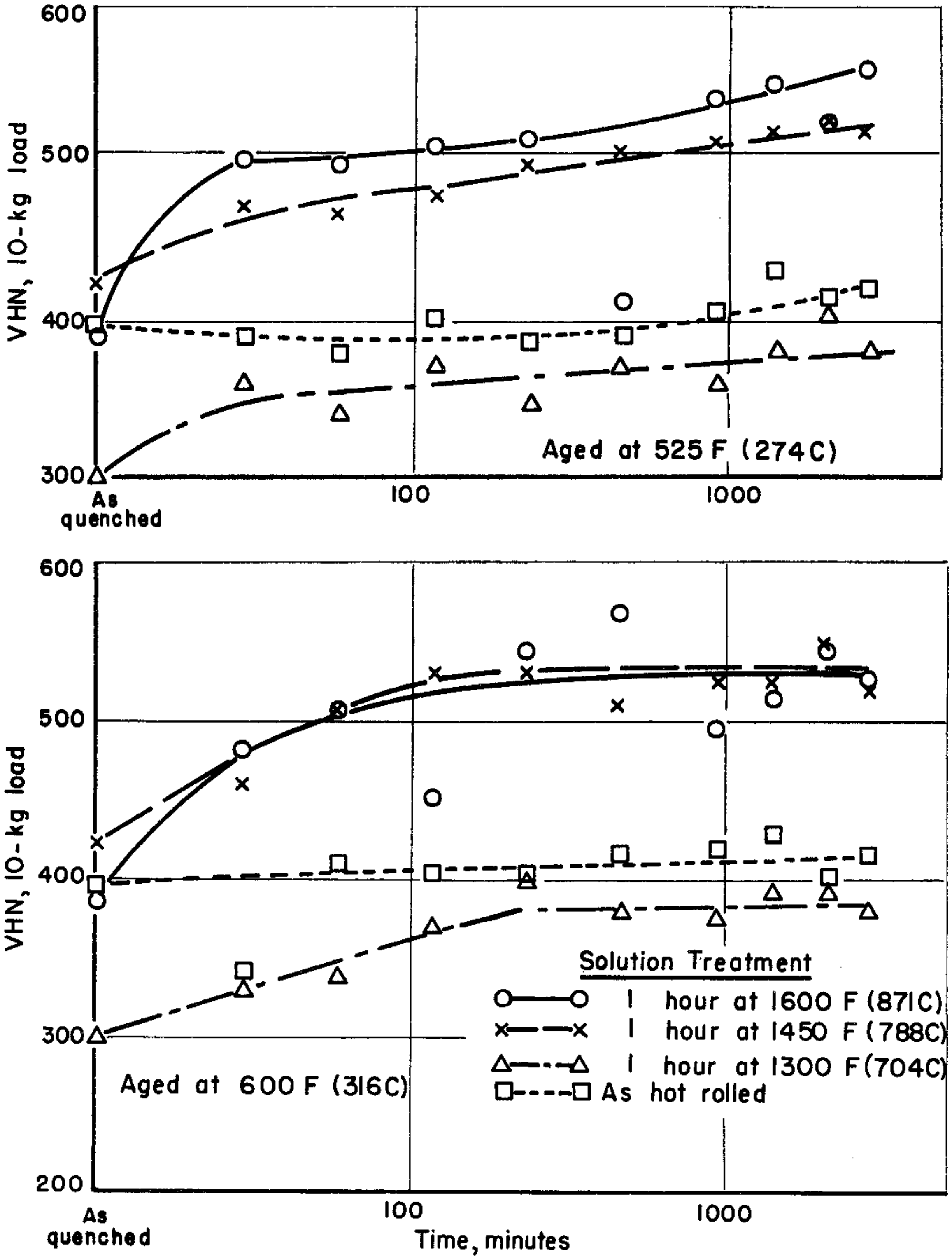


FIGURE 106. HARDNESS VERSUS AGING TIME FOR A Ti-2% Mn-3.5% Cr ALLOY SOLUTION TREATED AT VARIOUS TEMPERATURES

A-3290

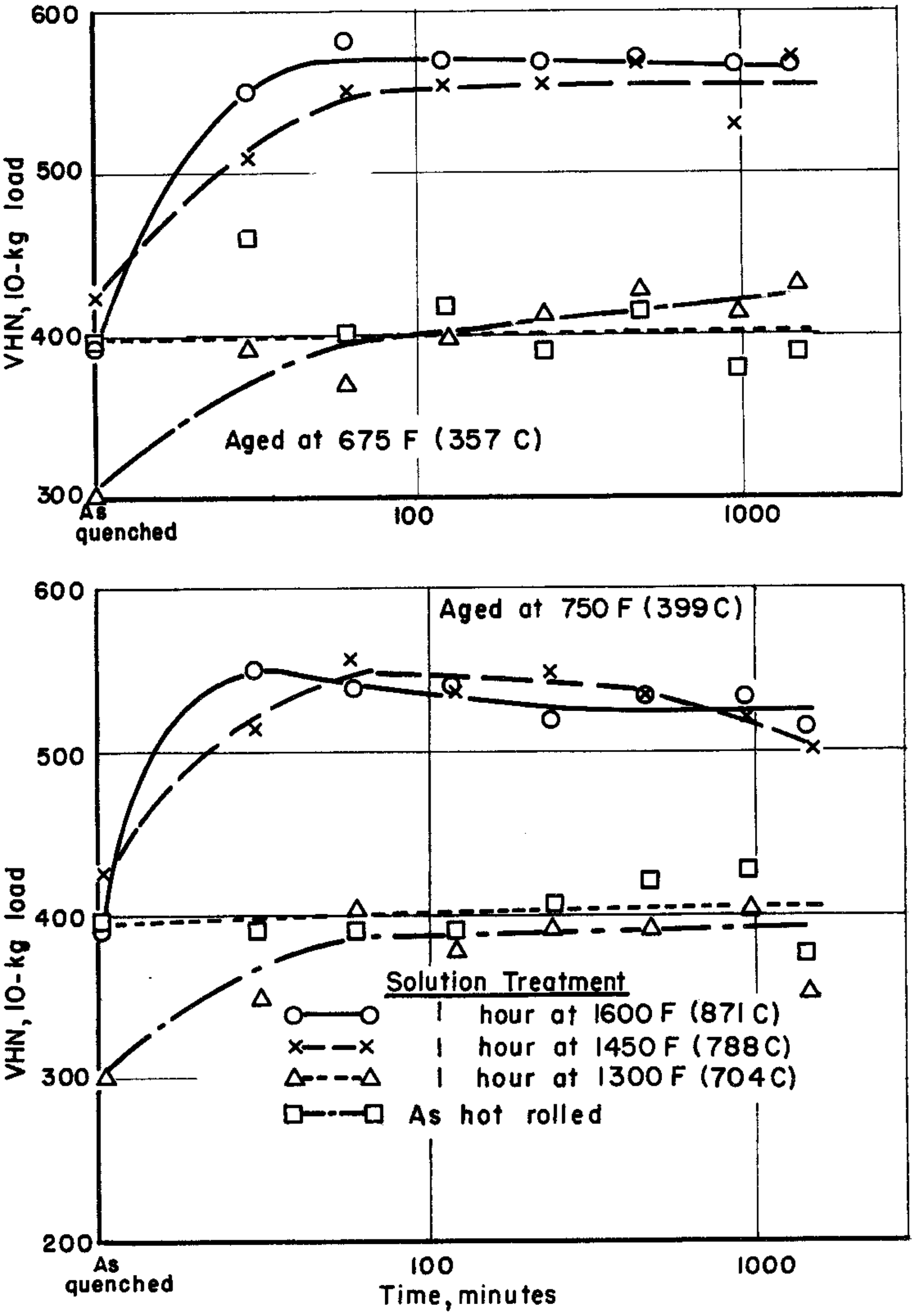


FIGURE 107. HARDNESS VERSUS AGING TIME FOR A Ti-2% Mn-3.5% Cr ALLOY SOLUTION TREATED AT VARIOUS TEMPERATURES

A-3291

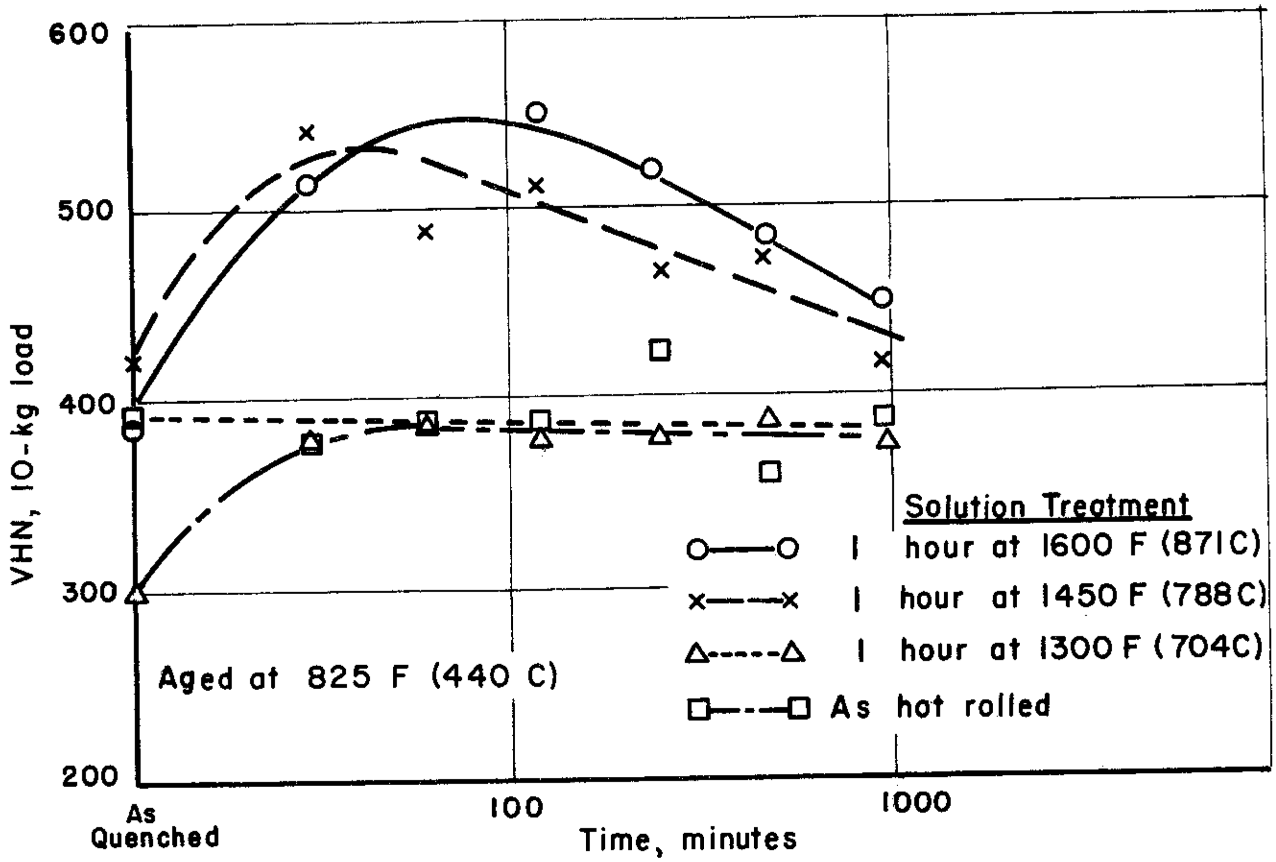


FIGURE 108. HARDNESS VERSUS AGING TIME FOR A Ti-2% Mn-3.5% ALLOY SOLUTION TREATED AT VARIOUS TEMPERATURES

A-3292

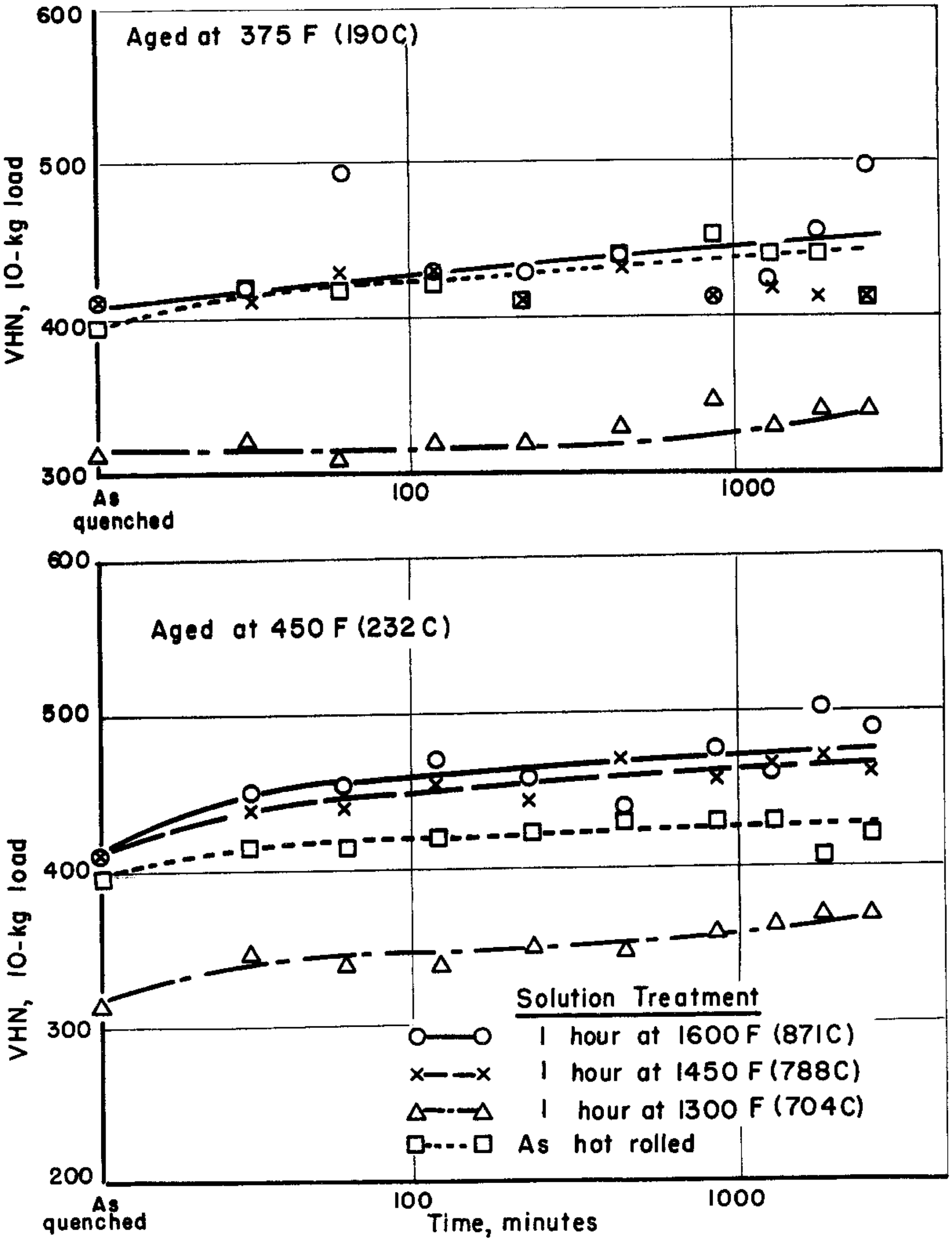


FIGURE 109. HARDNESS VERSUS AGING TIME FOR A Ti-3% Mn-3.5% Cr SOLUTION TREATED AT VARIOUS TEMPERATURE

A-3293

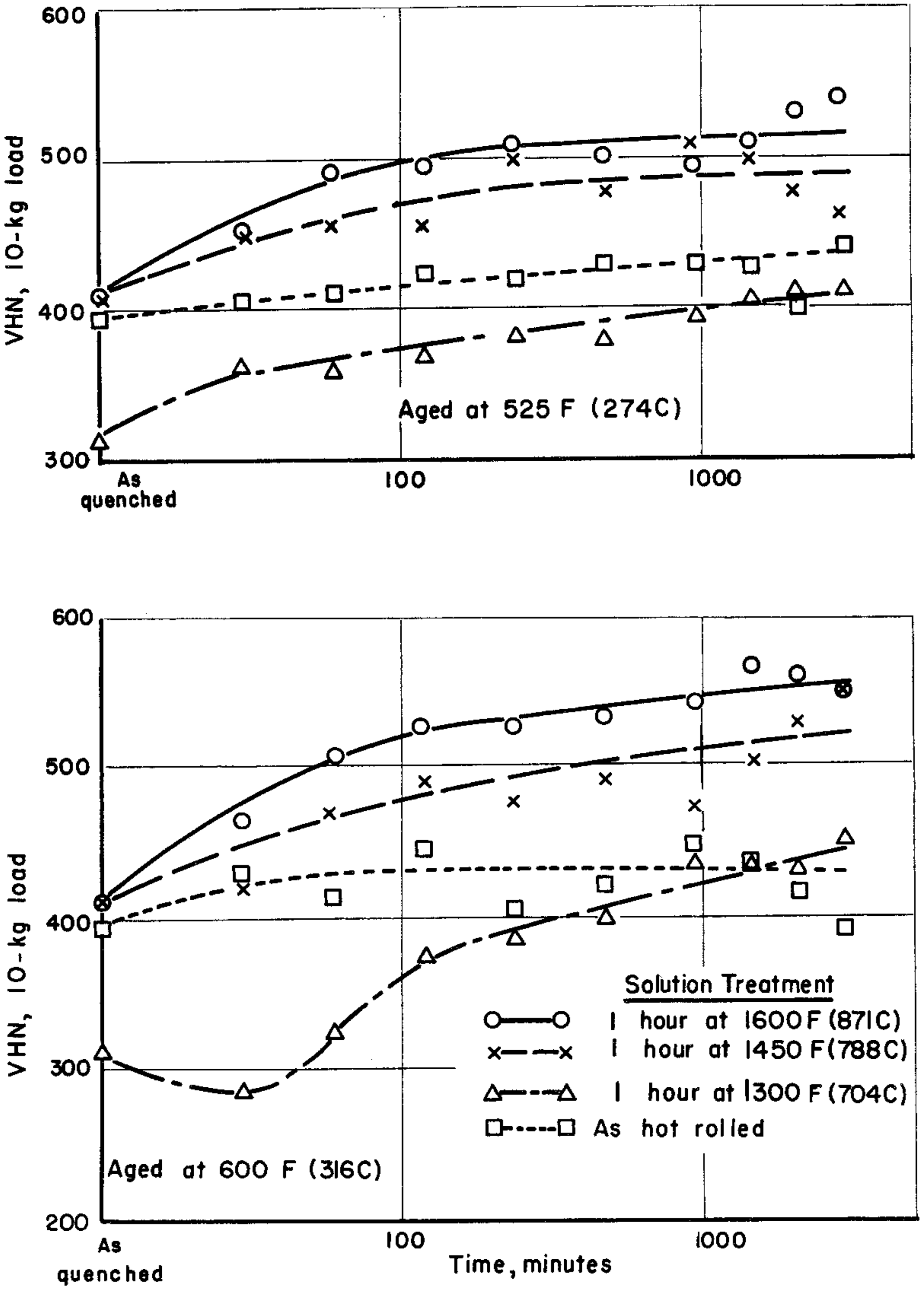


FIGURE 110. HARDNESS VERSUS AGING TIME FOR A Ti-3% Mn-3.5% Cr SOLUTION TREATED AT VARIOUS TEMPERATURES

A-3294

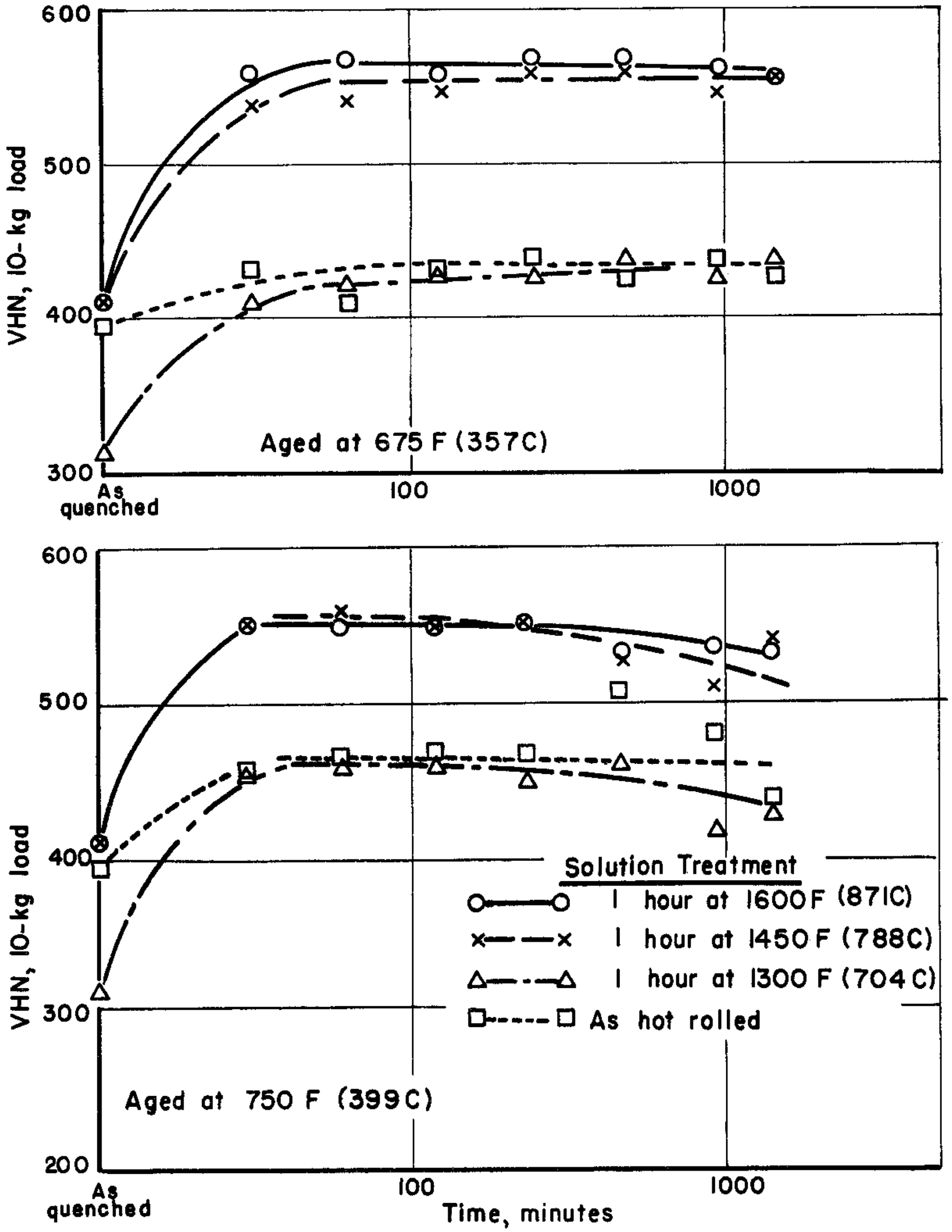


FIGURE III. HARDNESS VERSUS AGING TIME FOR A Ti-3% Mn-3.5% Cr ALLOY SOLUTION TREATED AT VARIOUS TEMPERATURES

A-3295

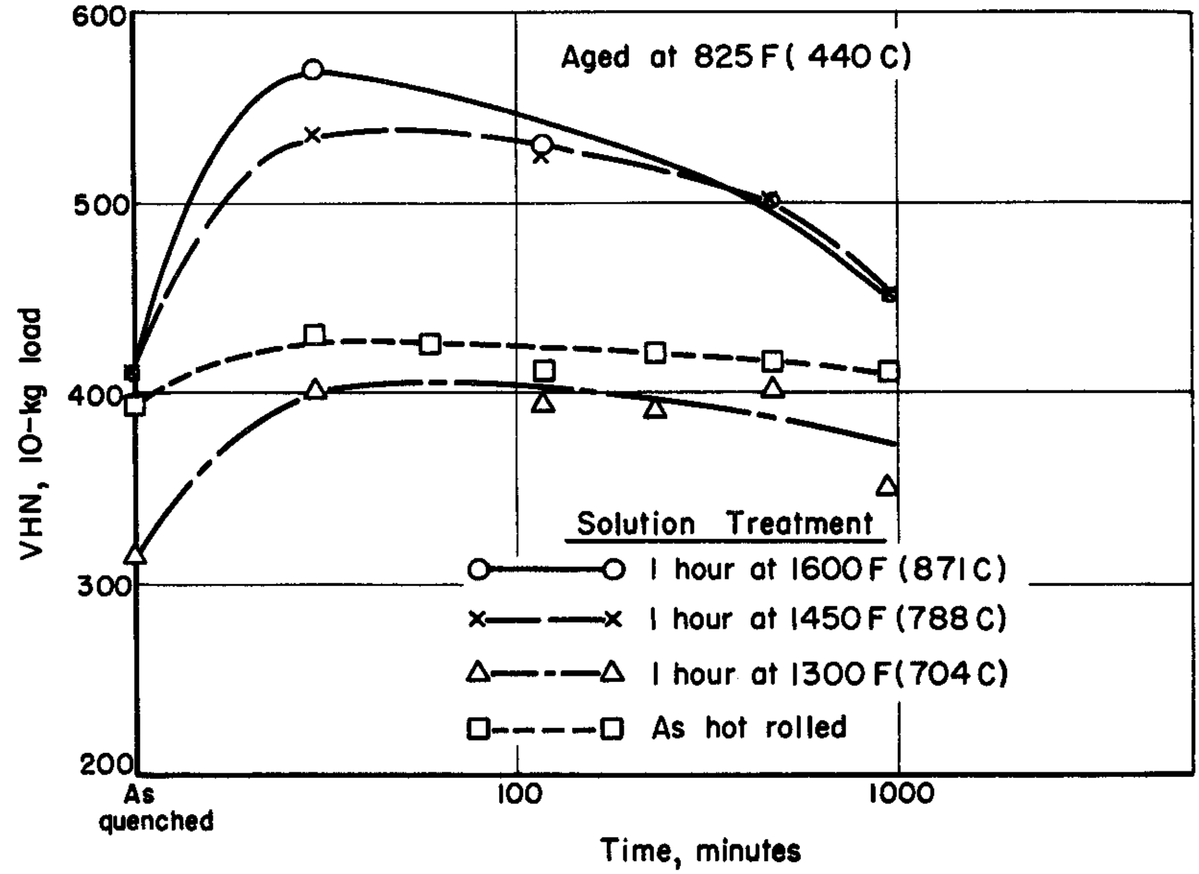


FIGURE 112. HARDNESS VERSUS AGING TIME FOR A Ti-3% Mn-3.5% Cr ALLOY SOLUTION TREATED AT VARIOUS TEMPERATURES

A-3296

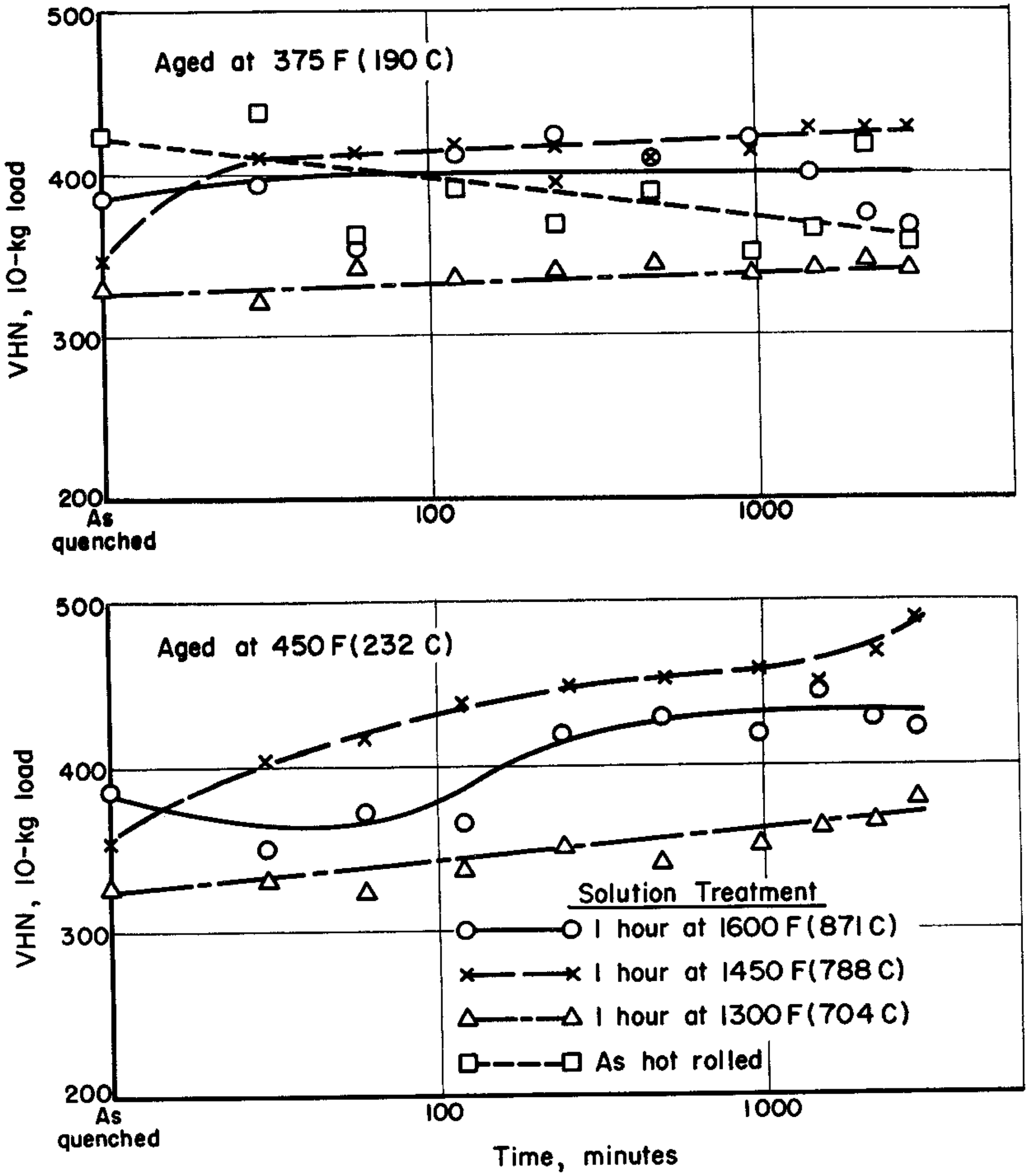


FIGURE 113. HARDNESS VERSUS AGING TIME FOR A Ti-5% Mn-3.5% Cr ALLOY SOLUTION TREATED AT VARIOUS TEMPERATURES

A-3297

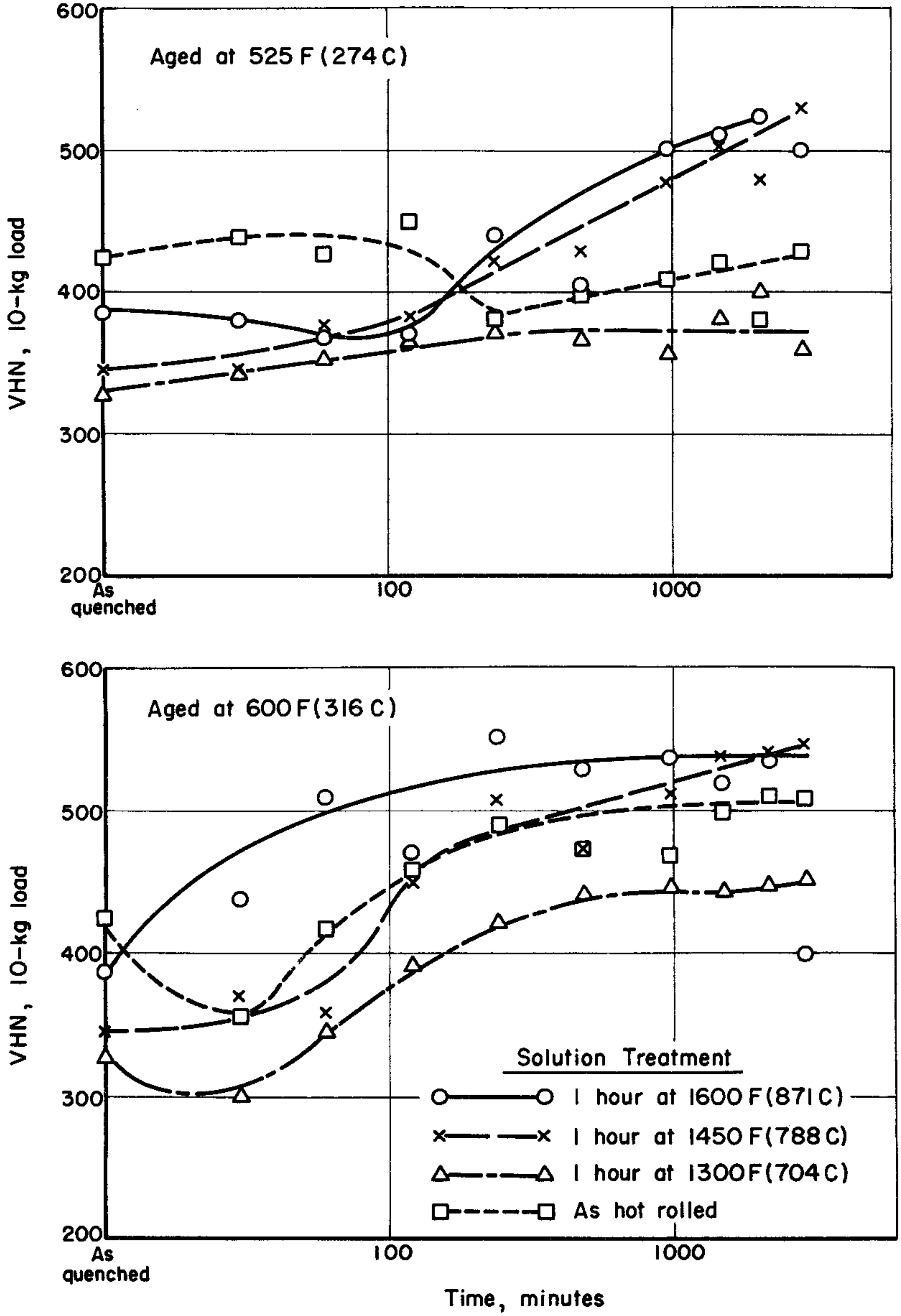


FIGURE 114. HARDNESS VERSUS AGING TIME FOR A Ti-5% Mn-3.5% Cr ALLOY SOLUTION TREATED AT VARIOUS TEMPERATURES

A-3298

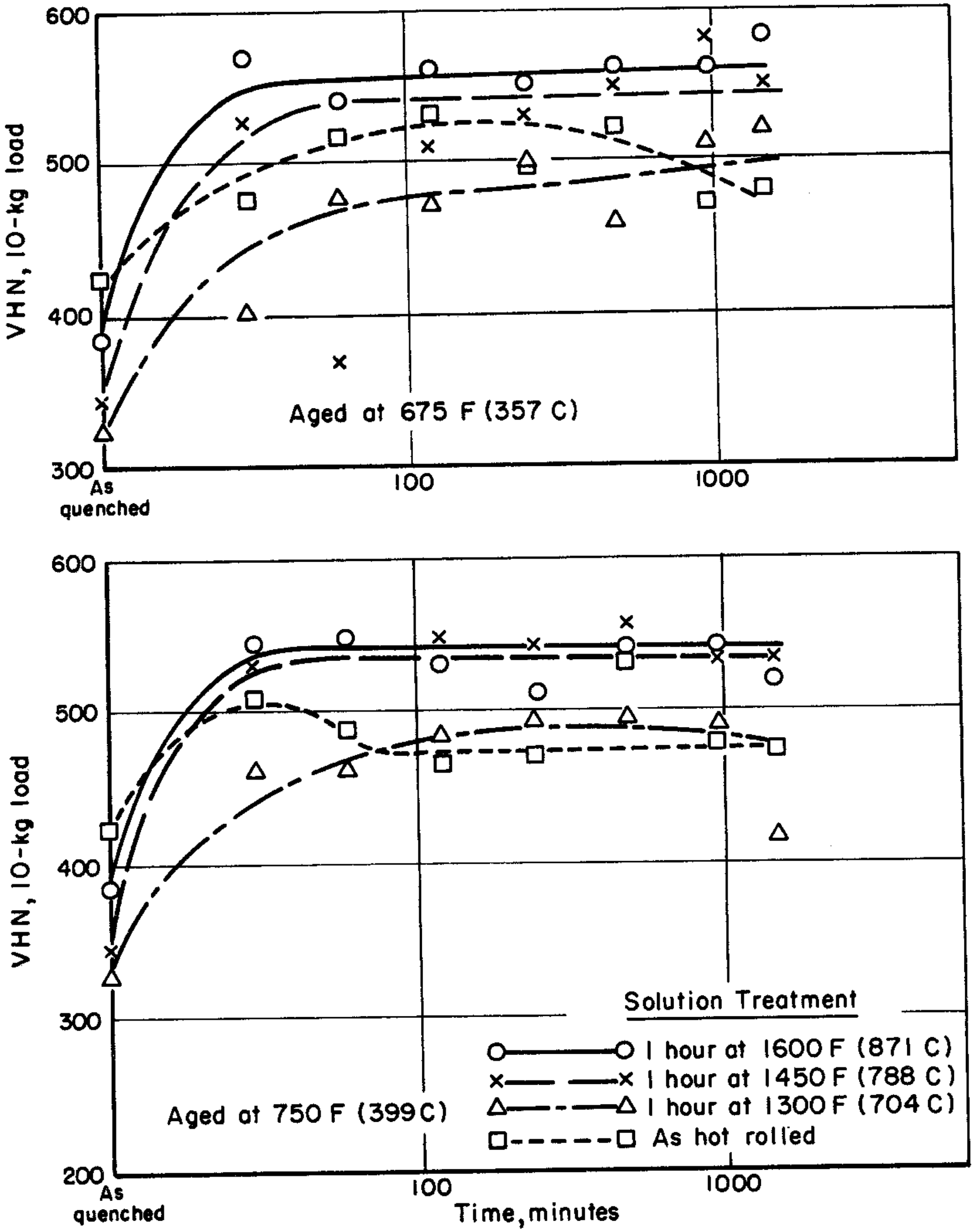


FIGURE 115. HARDNESS VERSUS AGING TIME FOR A Ti-5% Mn-3.5% Cr ALLOY SOLUTION TREATED AT VARIOUS TEMPERATURES

A-3299

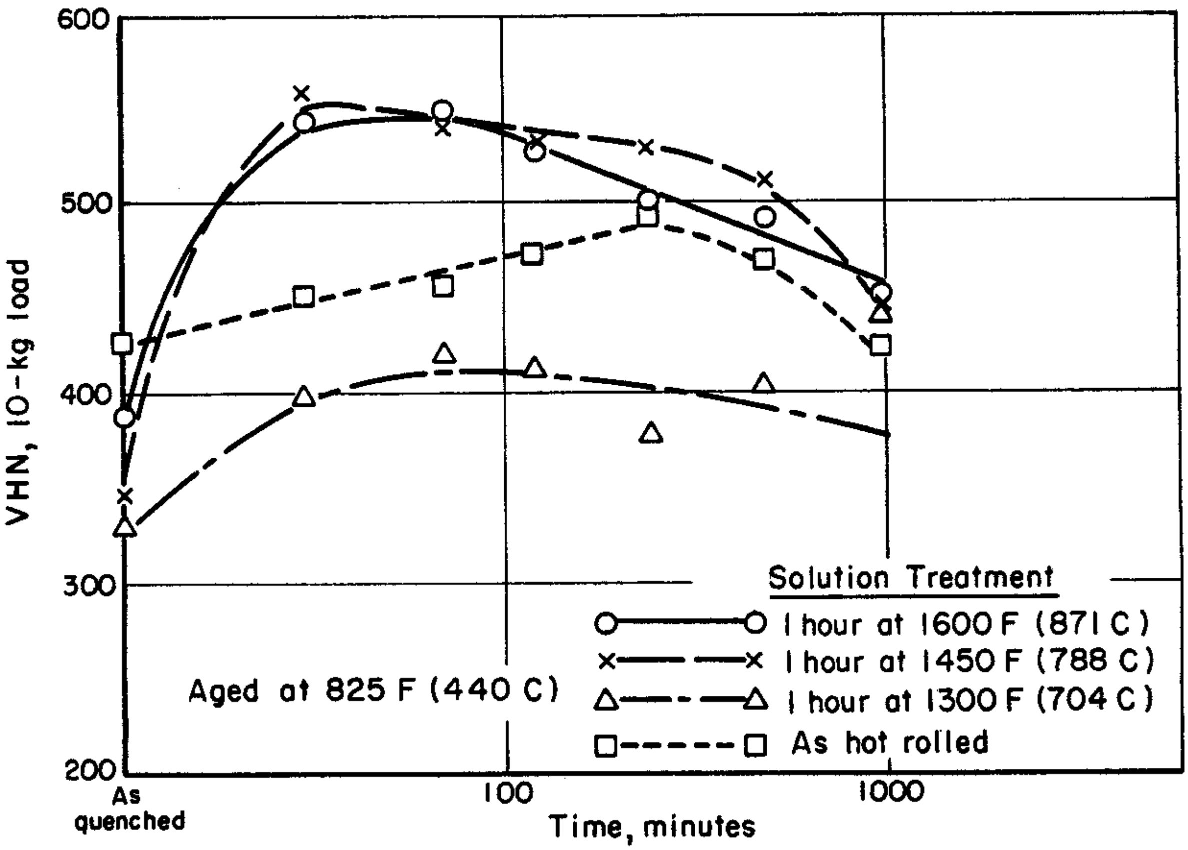


FIGURE 116. HARDNESS VERSUS AGING TIME FOR A Ti-5% Mn-3.5% Cr ALLOY SOLUTION TREATED AT VARIOUS TEMPERATURES

A-3300

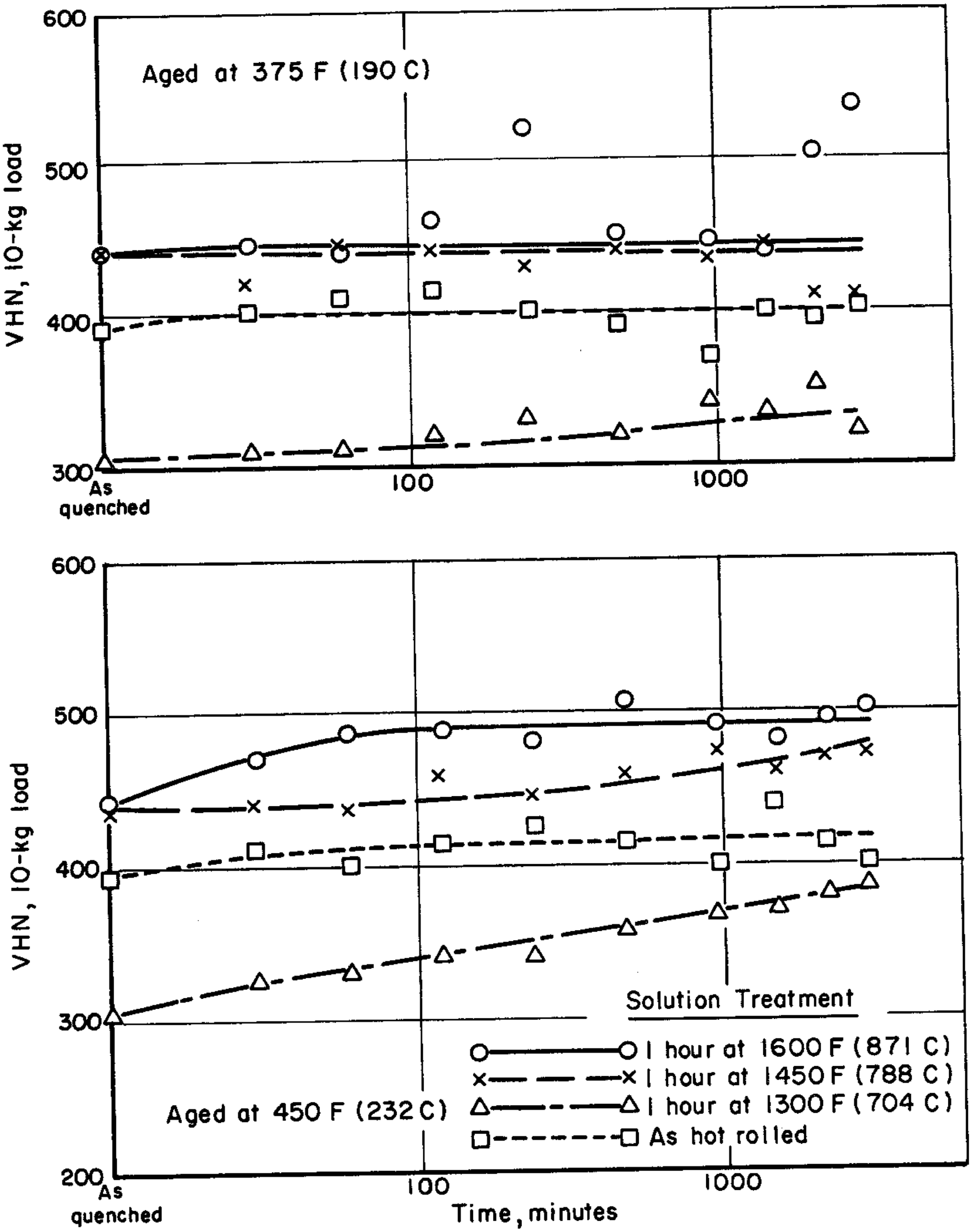


FIGURE 117. HARDNESS VERSUS AGING TIME FOR A Ti-1% Mn-5% Cr SOLUTION TREATED AT VARIOUS TEMPERATURES

A-3301

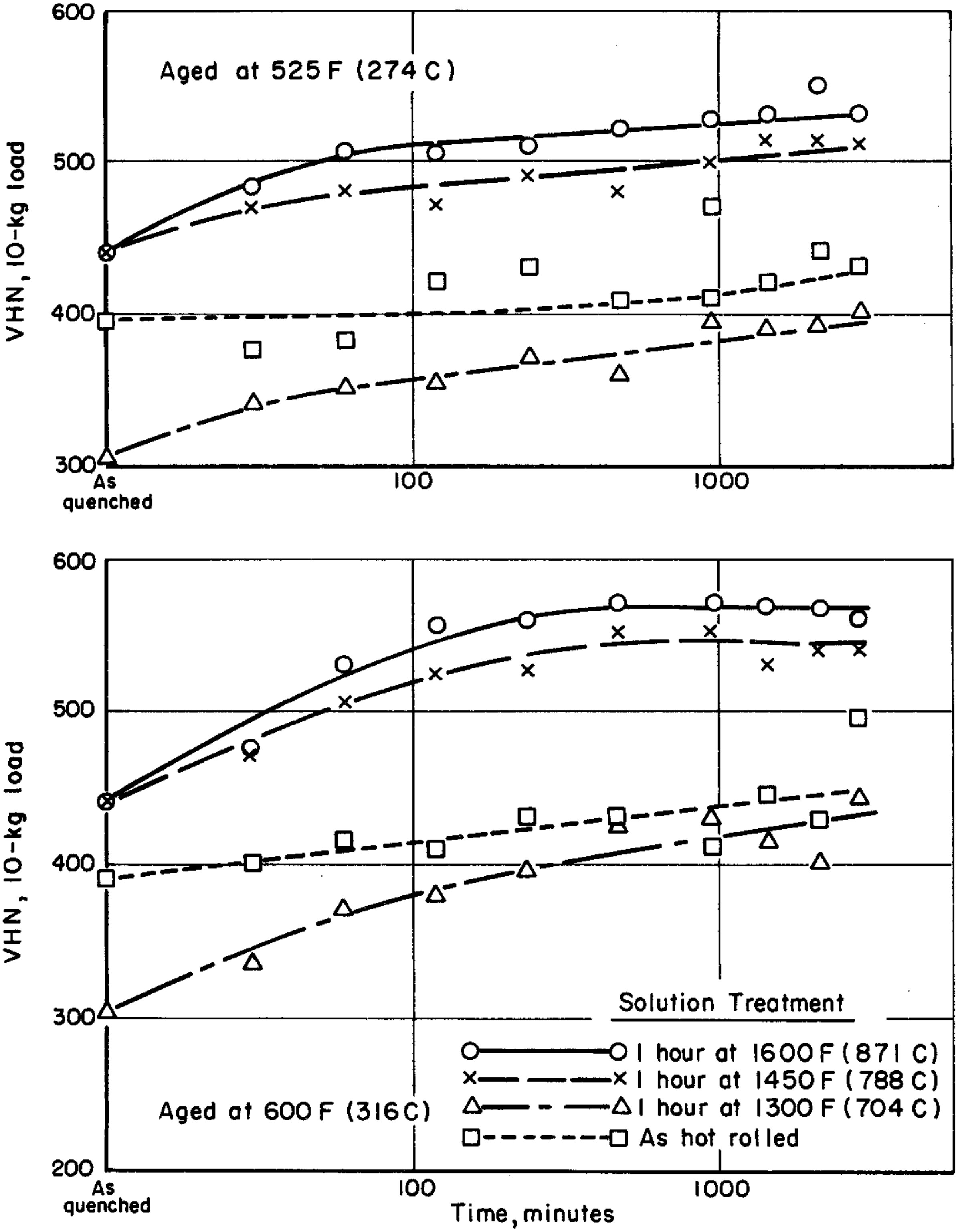


FIGURE 118. HARDNESS VERSUS AGING TIME FOR A Ti-1%Mn-5% Cr ALLOY SOLUTION TREATED AT VARIOUS TEMPERATURES

A-3302

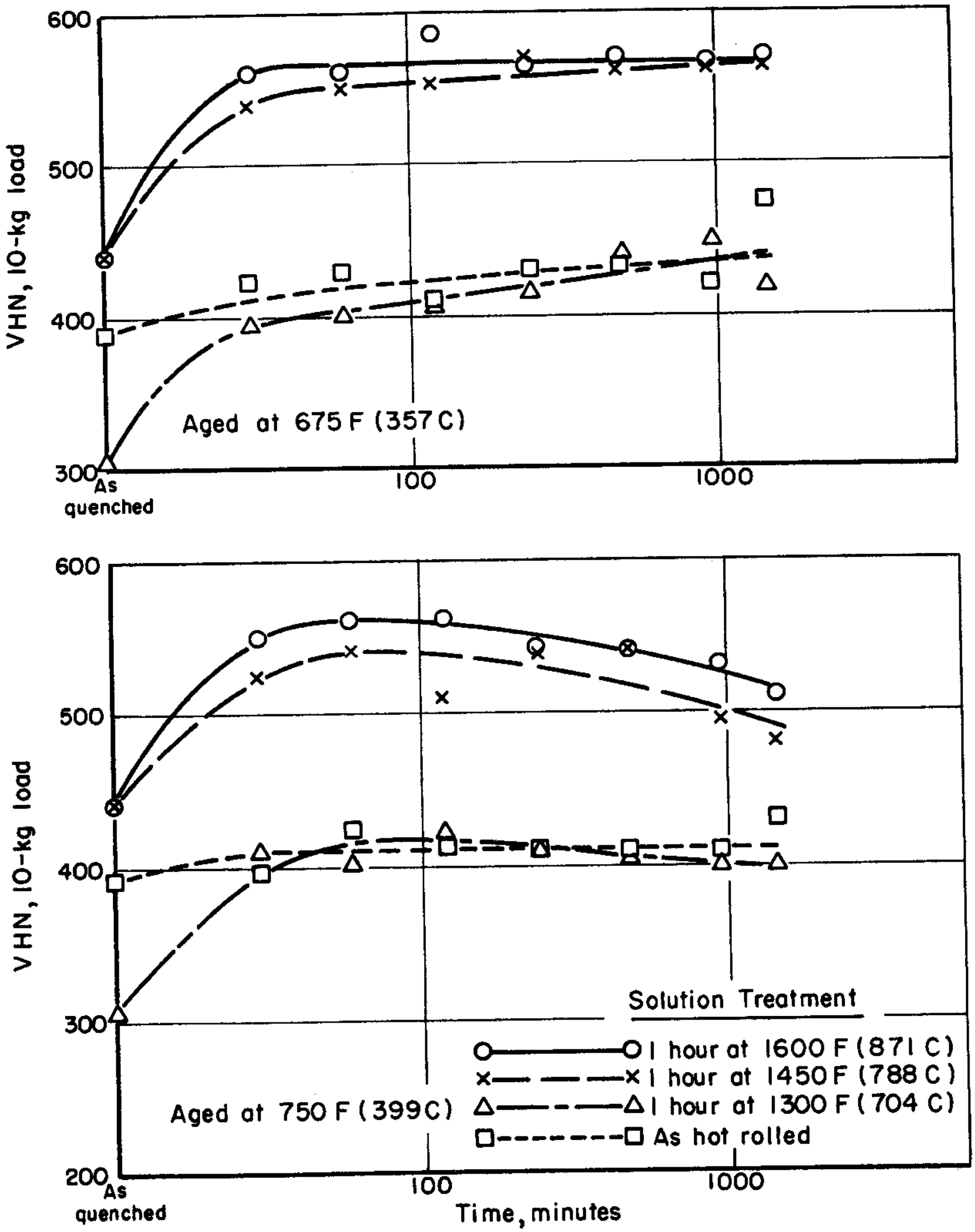


FIGURE 119. HARDNESS VERSUS AGING TIME FOR A Ti-1% Mn-5% Cr ALLOY SOLUTION TREATED AT VARIOUS TEMPERATURES

A-3303

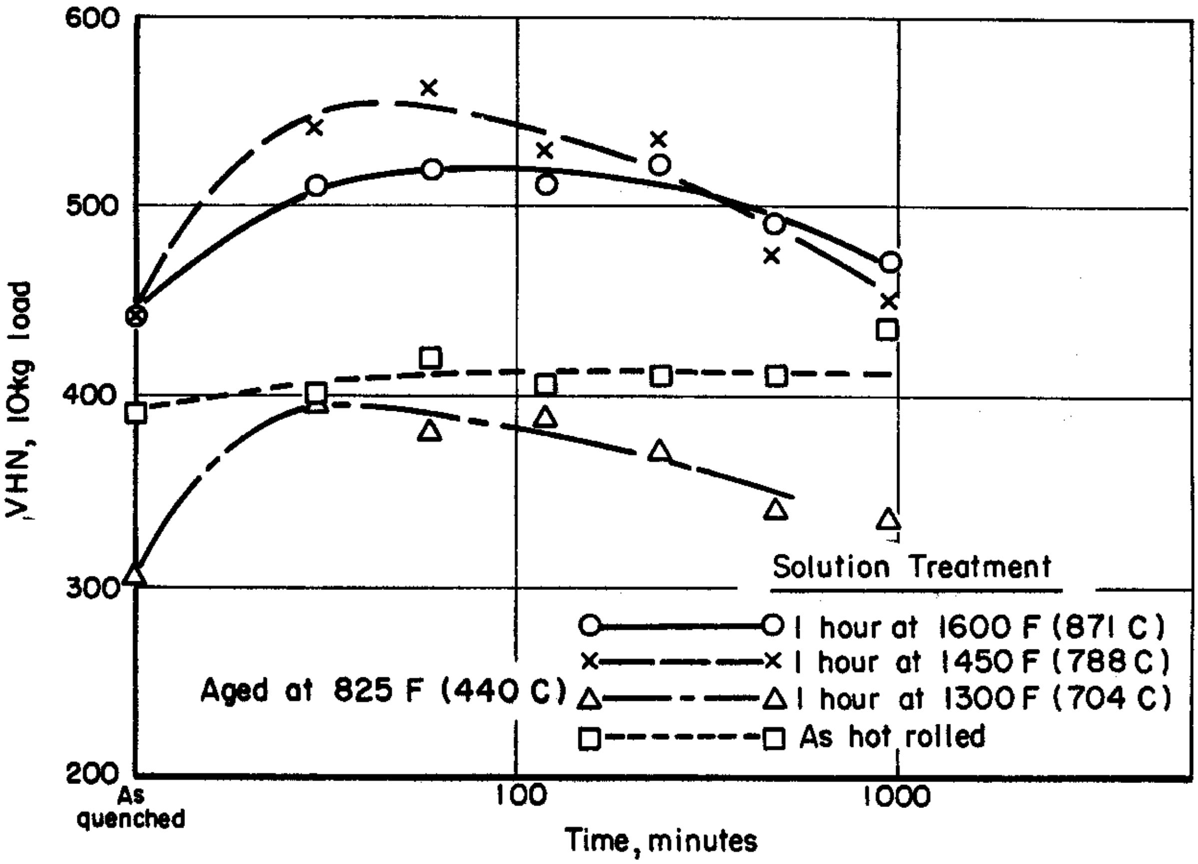


FIGURE 120. HARDNESS VERSUS AGING TIME FOR A Ti-1% Mn-5% Cr ALLOY SOLUTION TREATED AT VARIOUS TEMPERATURES

A-3304

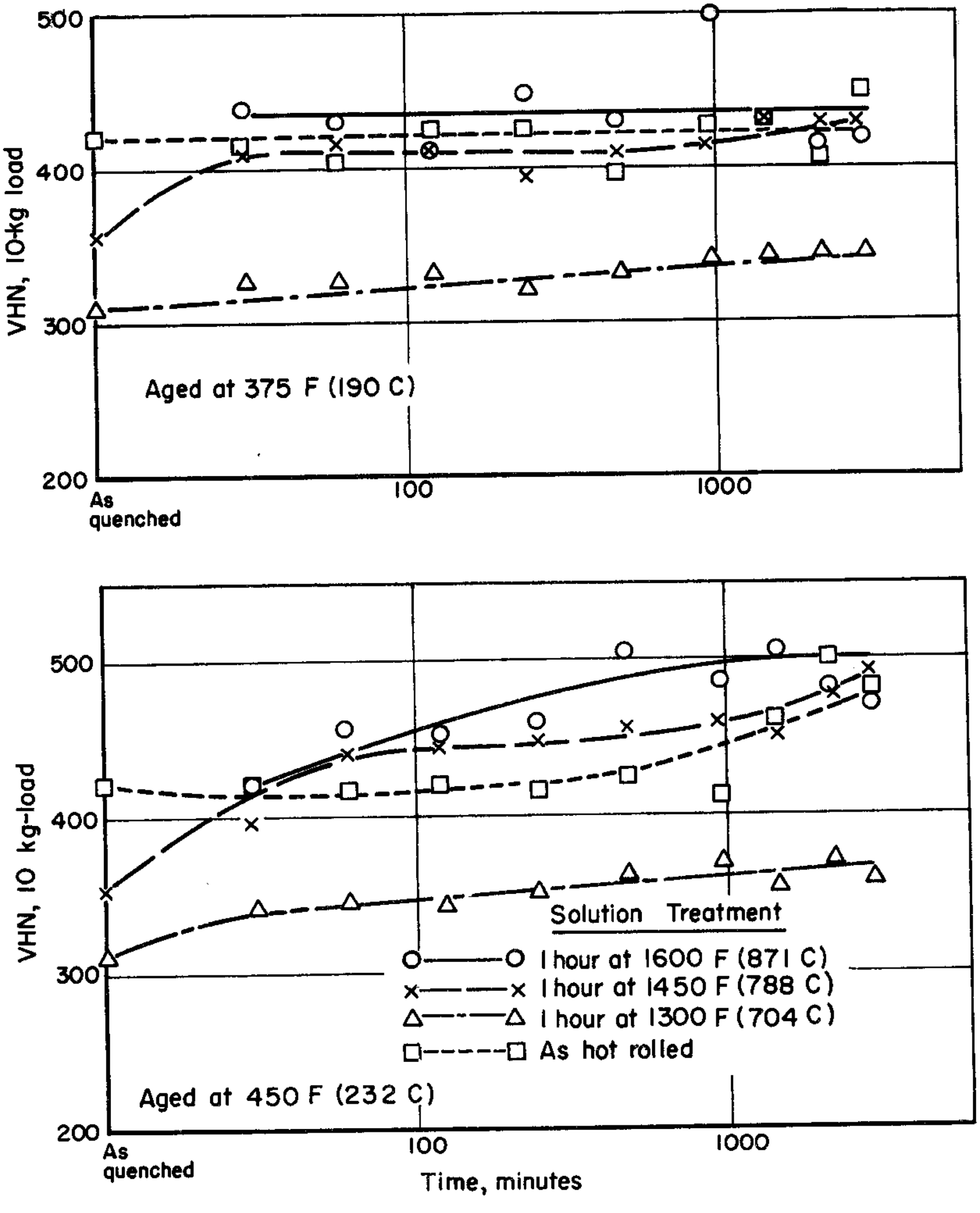


FIGURE 121. HARDNESS VERSUS AGING TIME FOR A Ti-2% Mn-5% Cr ALLOY SOLUTION TREATED AT VARIOUS TEMPERATURES

A-3305

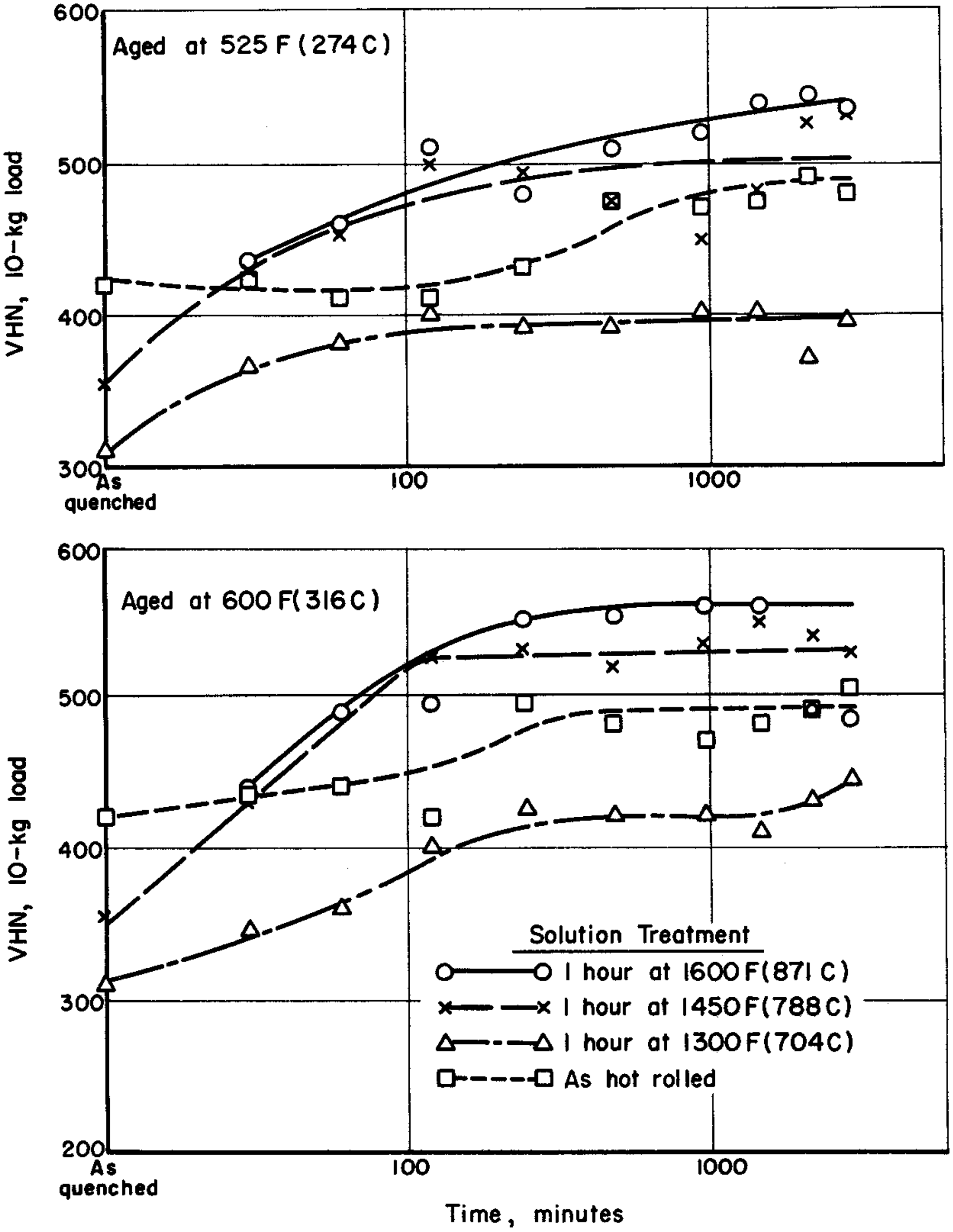


FIGURE 122. HARDNESS VERSUS AGING TIME FOR A Ti-2% Mn-5% Cr ALLOY SOLUTION TREATED AT VARIOUS TEMPERATURES

A-3306

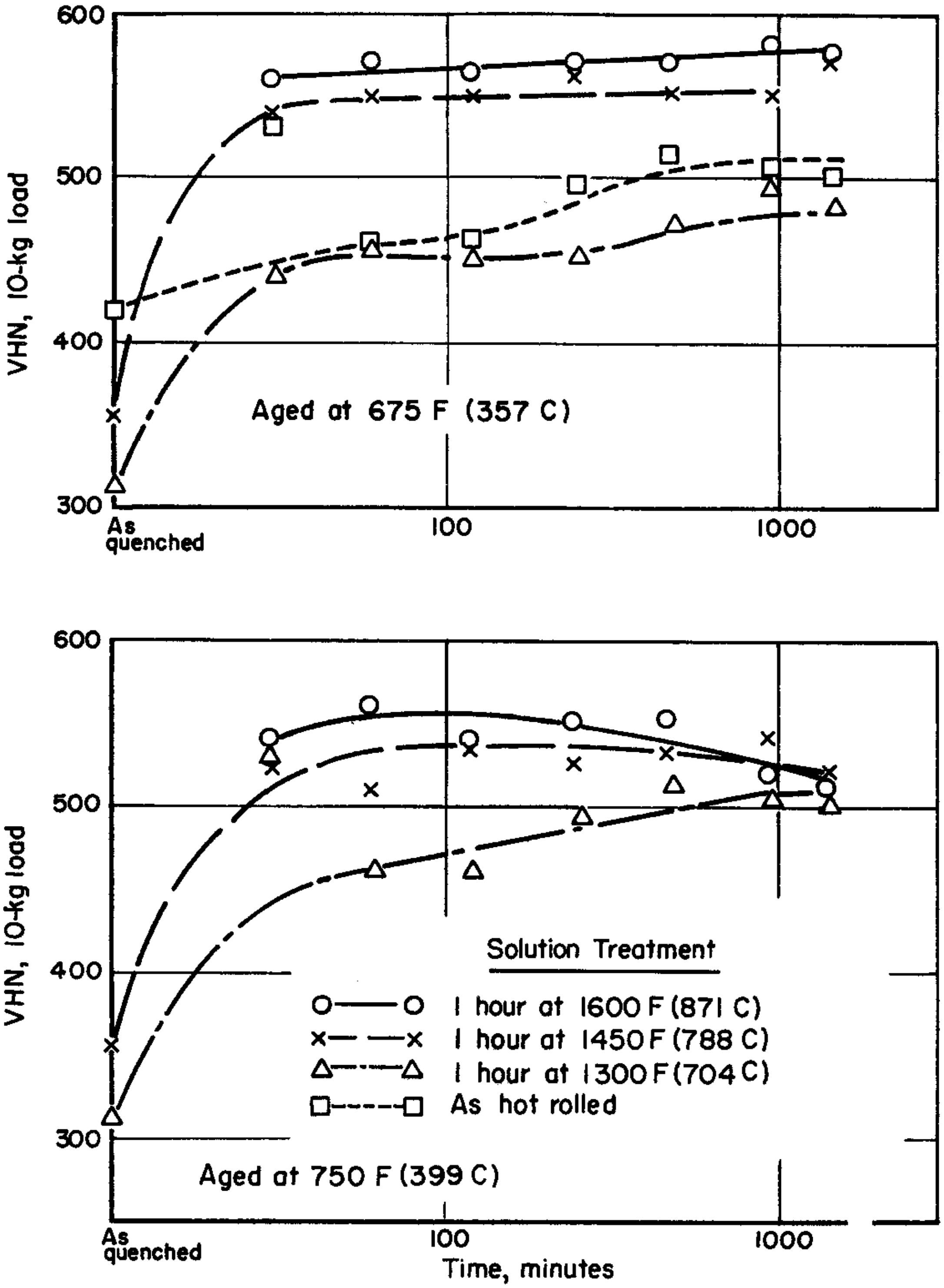


FIGURE 123. HARDNESS VERSUS AGING TIME FOR A Ti-2% Mn-5% Cr ALLOY SOLUTION TREATED AT VARIOUS TEMPERATURES

A-3307

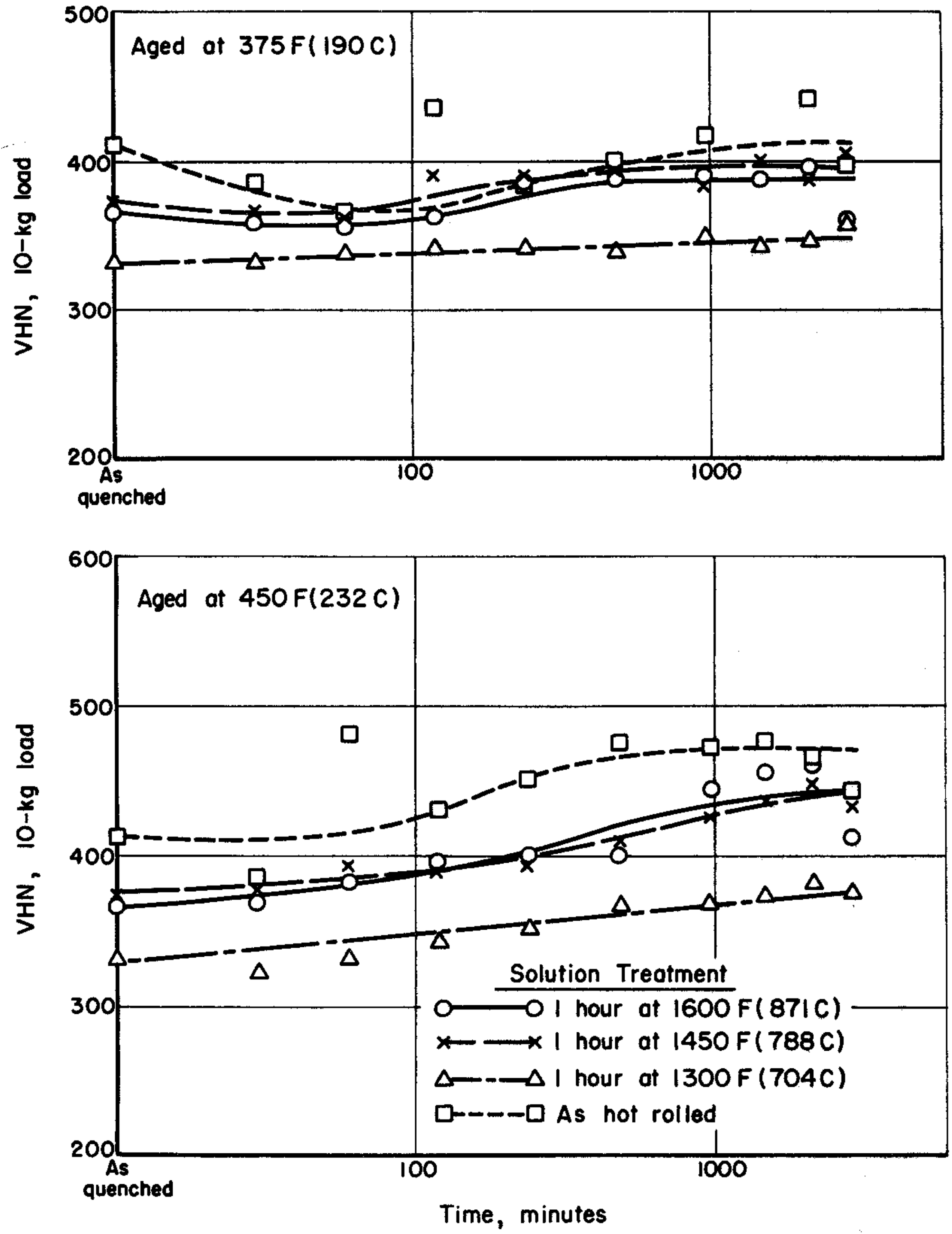


FIGURE 124. HARDNESS VERSUS AGING TIME FOR A Ti-3% Mn-5% Cr ALLOY SOLUTION TREATED AT VARIOUS TEMPERATURES

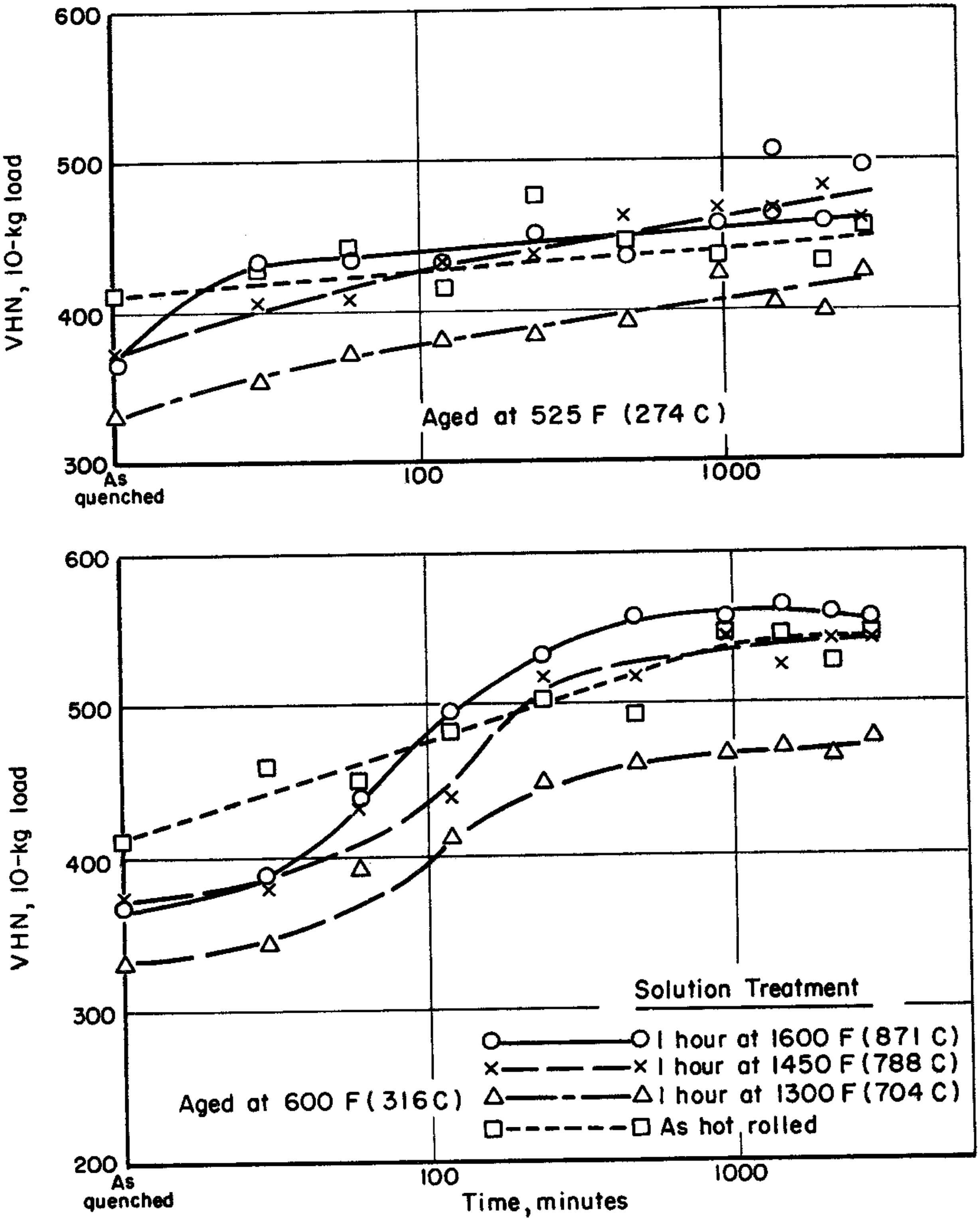


FIGURE 125. HARDNESS VERSUS AGING TIME FOR A Ti-3% Mn-5% Cr ALLOY SOLUTION TREATED AT VARIOUS TEMPERATURES

A-3309

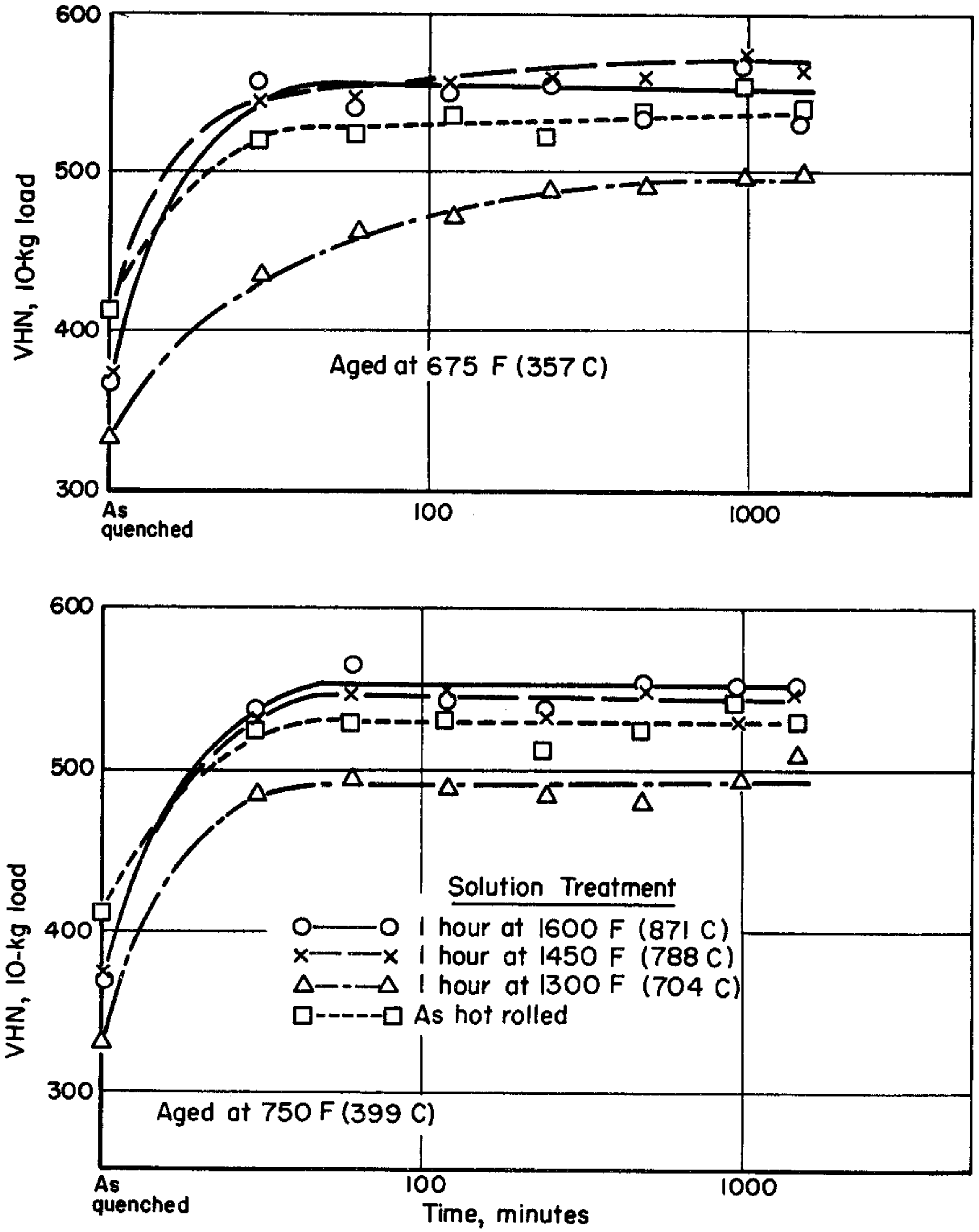


FIGURE 126. HARDNESS VERSUS AGING TIME FOR A Ti-3% Mn-5% Cr ALLOY SOLUTION TREATED AT VARIOUS TEMPERATURES

A-3310

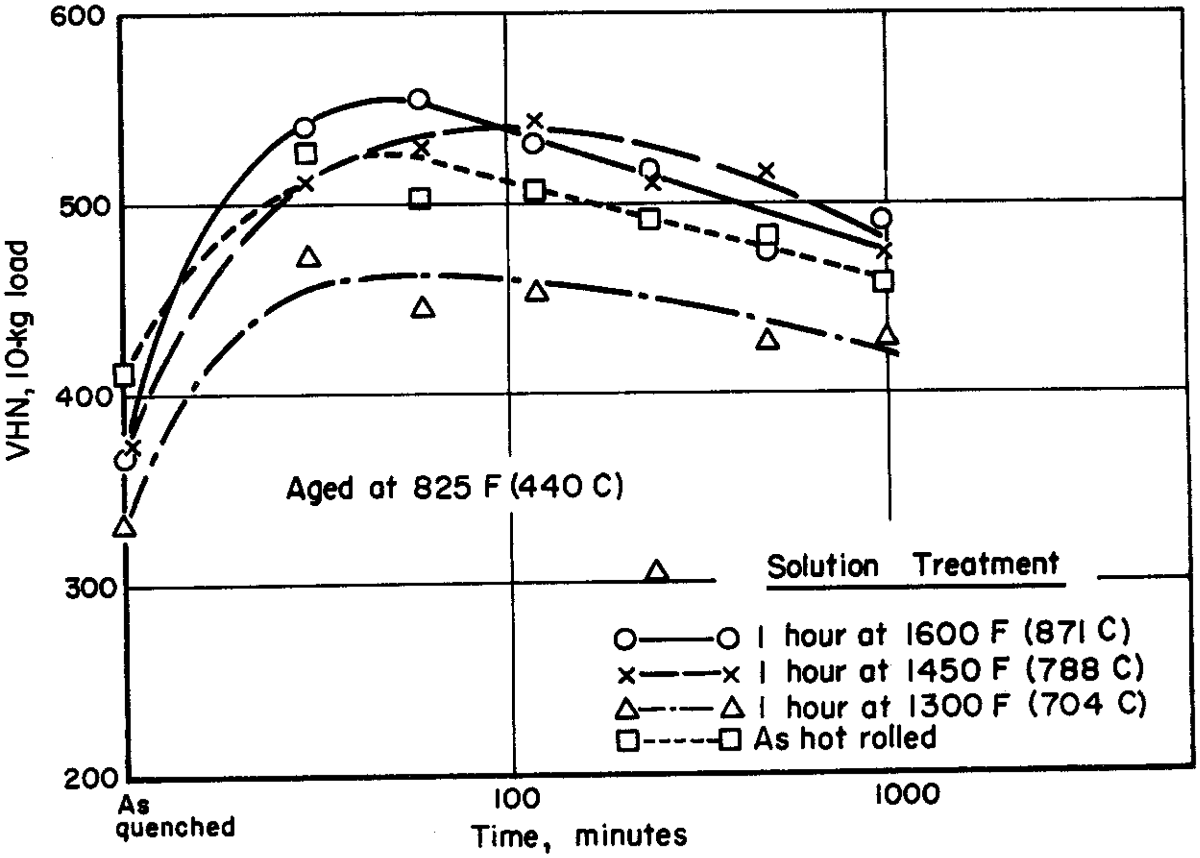


FIGURE 127. HARDNESS VERSUS AGING TIME FOR A Ti-3% Mn-5% Cr ALLOY SOLUTION TREATED AT VARIOUS TEMPERATURES

A-3311

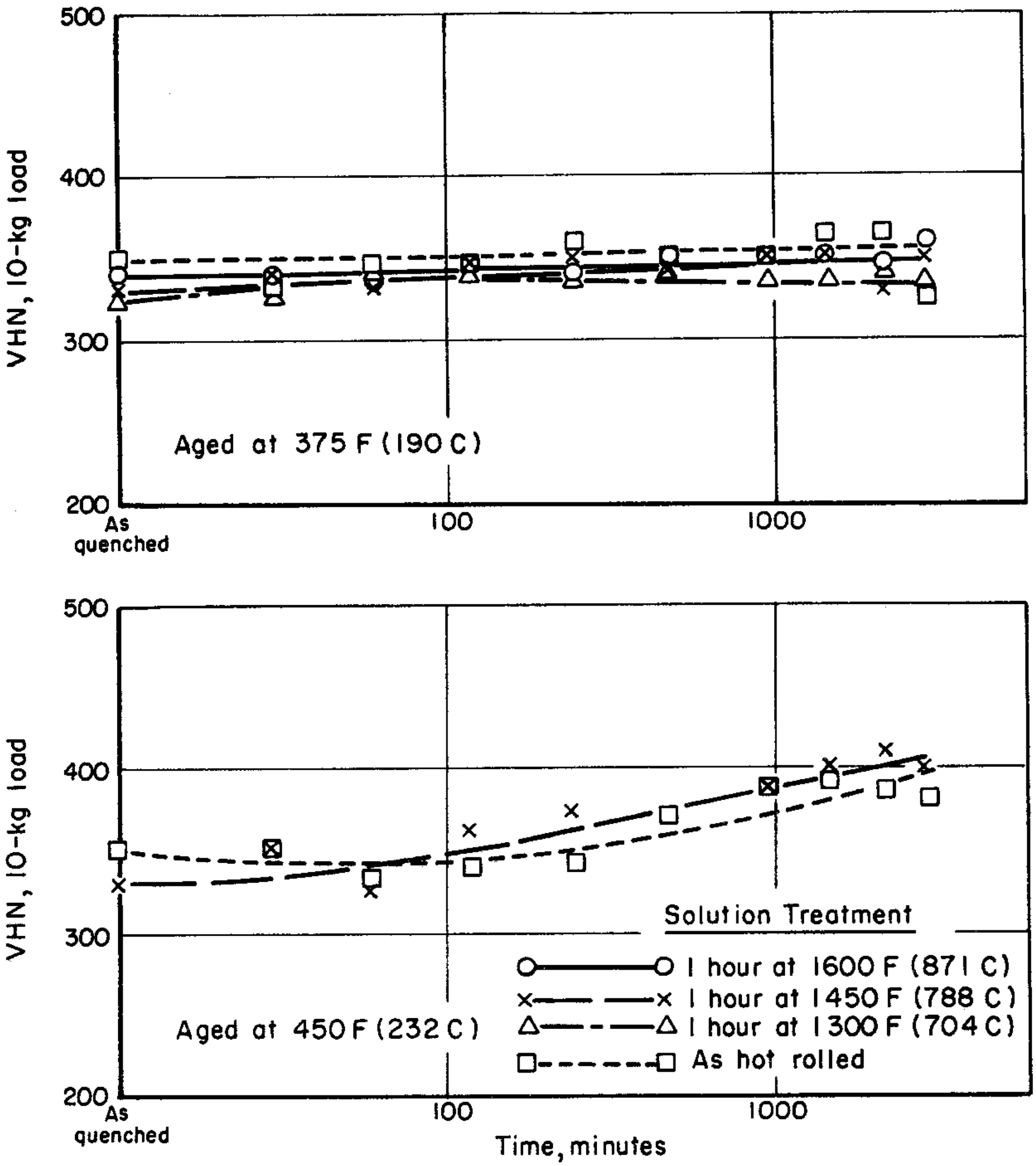


FIGURE 128. HARDNESS VERSUS AGING TIME FOR A Ti-5% Mn-5% Cr ALLOY SOLUTION TREATED AT VARIOUS TEMPERATURES

A-3312

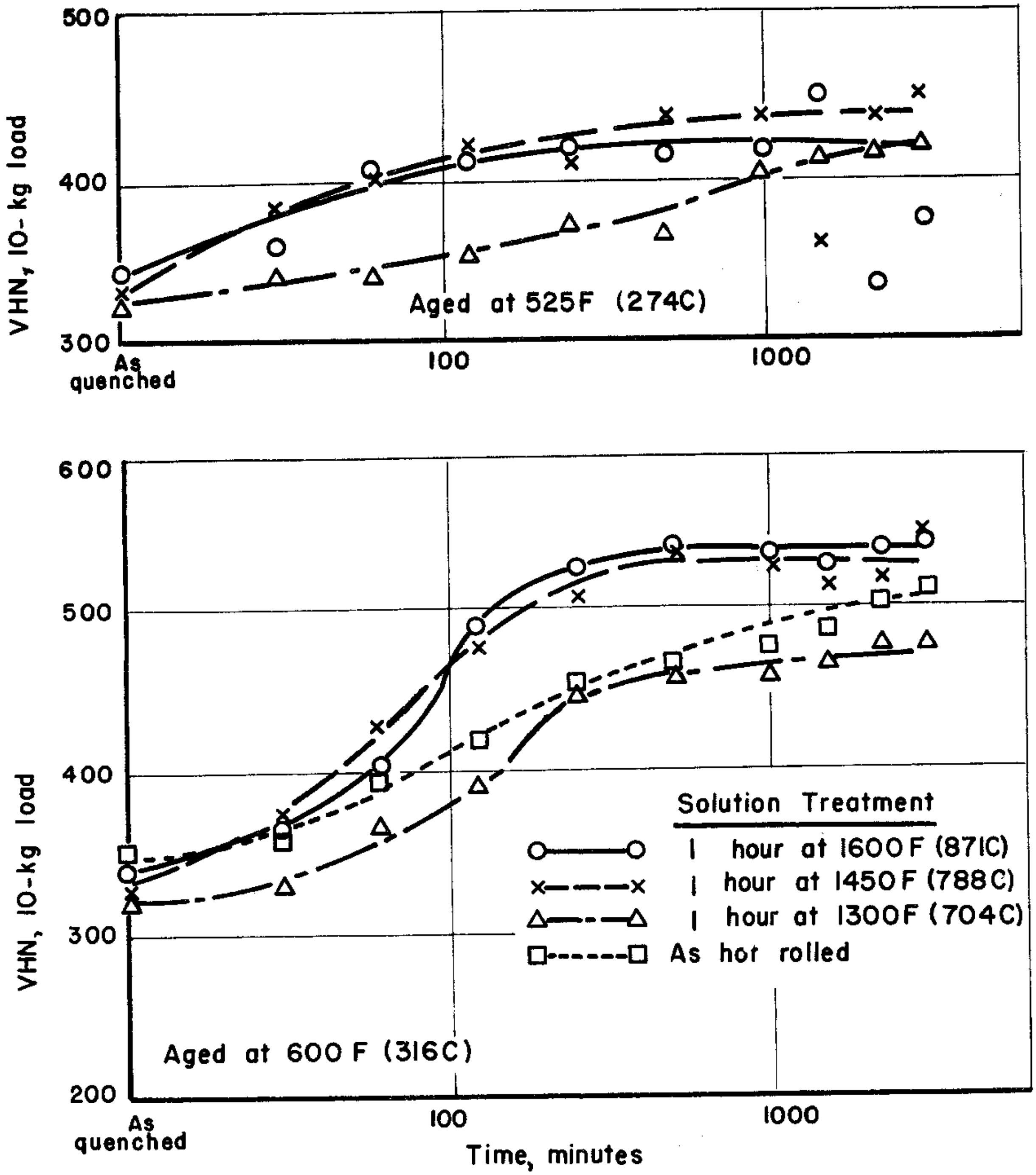


FIGURE 129. HARDNESS VERSUS AGING TIME FOR A Ti-5% Mn-5% Cr ALLOY SOLUTION TREATED AT VARIOUS TEMPERATURES

A-3313

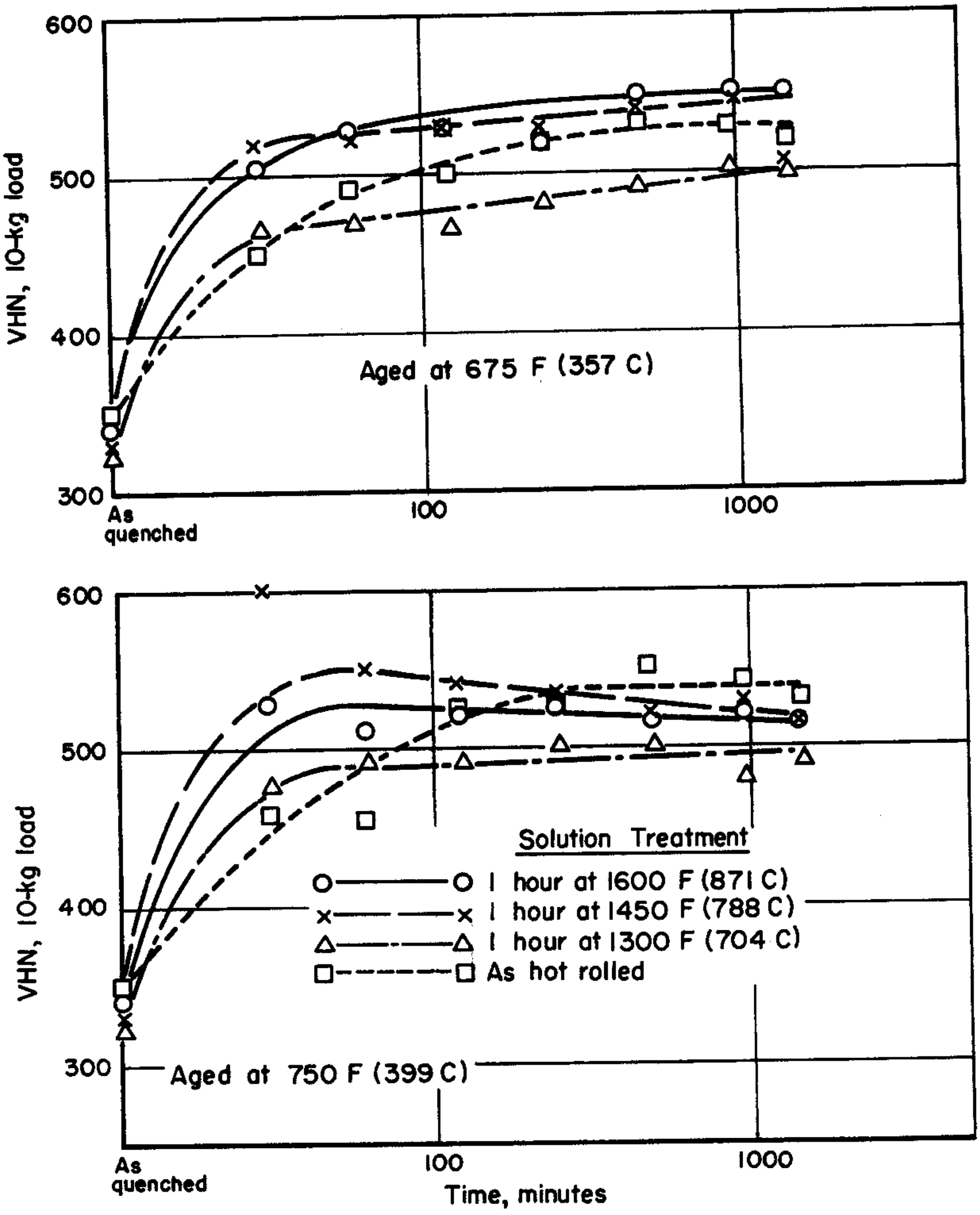


FIGURE 130. HARDNESS VERSUS AGING TIME FOR A Ti-5% Mn-5% Cr ALLOY SOLUTION TREATED AT VARIOUS TEMPERATURES

A-3314

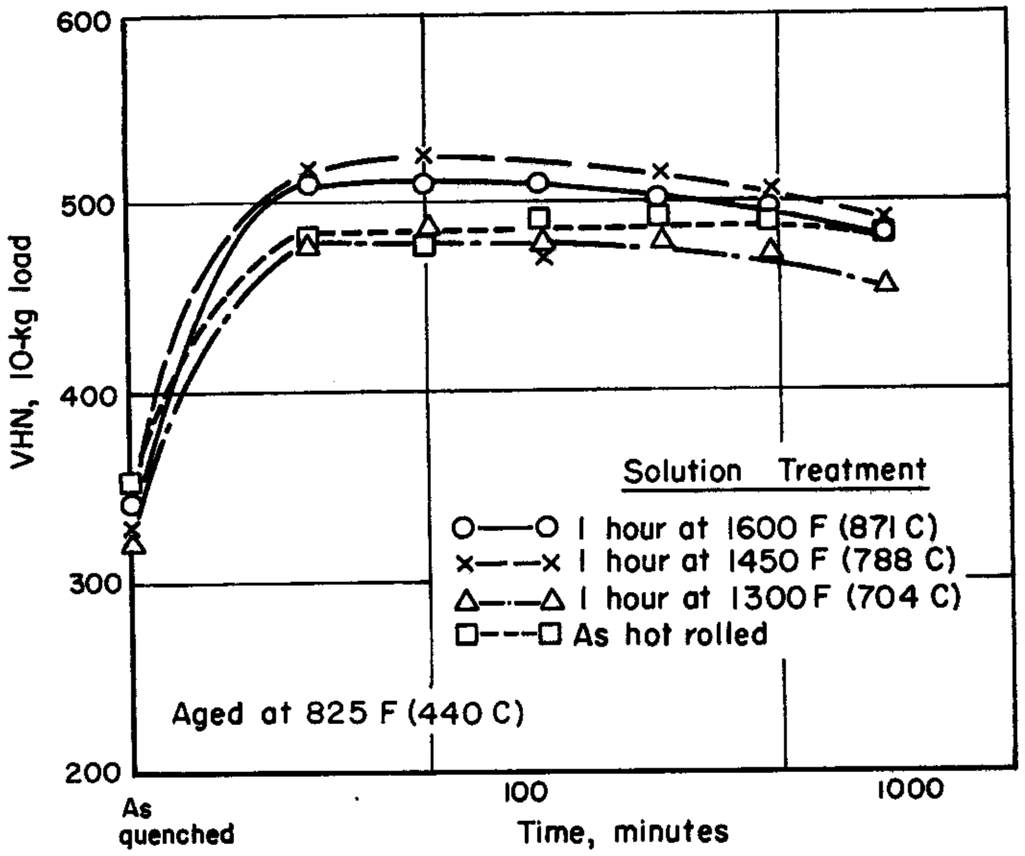


FIGURE 131. HARDNESS VERSUS AGING TIME FOR A Ti-5% Mn-5% Cr ALLOY SOLUTION TREATED AT VARIOUS TEMPERATURES

A-3315