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DEVELOPMENT OF A STABLE BETA TITANIUM ALLOY

by

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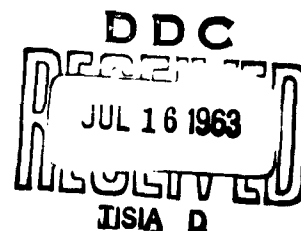
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DEVELOPMENT OF A STABLE-BETA TITANIUM ALLOY

ABSTRACT

Research has progressed toward the objective of developing stable-beta titanium sheet alloys capable of being precipitation hardened to high strengths. Screening criteria are rollability, and response to heat treatment. In this phase of the contract, additions of 12.5 and 15% Cr and Mn were made to the three base alloys, Ti-17V-3Al, Ti-8V-8Mo-3Al, and Ti-15Mo-3Al. All such alloys were found to have poor fabrication characteristics, and also low ductilities in both the solution treated and aged conditions. The maximum amounts of Cr, Mn, and Fe which can be added to the three groups of alloys therefore are:

<u>Base Alloy</u>	<u>Maximum % of Cr, Mn or Fe Tolerable For Fabrication (to nearest 2.5%)</u>
Ti-17V-3Al	10.0 Cr 10.0 Mn 7.5 Fe
Ti-8Mo-8V-3Al	10.0 Cr 10.0 Mn 7.5 Fe(borderline)
Ti-15Mo-3Al	10.0 Cr 10.0 Mn 5.0 Fe

In this report the response of alloys to heat treatment was evaluated by means of tensile testing and metallographic examination. Oxidation tests revealed that the Ti-17V-3Al alloy group had the poorest oxidation resistance, and the Ti-15Mo-3Al group the highest, as measured by loss in weight/square cm. Densities of Phase I alloys were found to vary from 0.175 to 0.189 pound/cubic inch. Based on the above studies, three base alloys were selected for Phase II.

INTRODUCTION

The object of this contract is to develop stable-beta titanium alloys capable of being hardened to useful combinations of strength and ductility. Toward this end, in the First Quarterly Report, the minimum amounts of Cr, Mn, and Fe which could be added to the three base alloys of Ti-8Mo-3V-3Al, Ti-15Mo-3Al, and Ti-17V-3Al, while retaining the beta phase upon plate cooling, were determined. For the first base alloy they were 7.5% each of either Cr and Mn, and 5% Fe; for the second, 5% of Cr, Mn, and Fe; and for the third, 10% Cr, and 7.5% Mn or Fe. Alloys containing less than these amounts were unstable, undergoing a loss in ductility and an increase in strength when aged. While it was found that the amount of iron needed for stabilization of the beta phase was close to the amount found to cause cracking during rolling and embrittlement on aging, no fabrication trouble was experienced with up to 10% of either Cr or Mn. With the prospect of being able to obtain bases of higher strength than those so far discovered, it was decided to make additions of 12.5 and 15% Cr and Mn only to each of the three base alloys and to investigate fabrication and tensile properties of the resulting compositions.

Density and oxidation determinations were made on selected Phase I alloys and analytical determinations made of those alloys thought most suitable as Phase II bases.

MATERIALS AND PROCEDURES

Materials used in the period covered in this report were the same as those described in the First Quarterly Report; like melting, processing methods and testing procedures were also used. Oxidation tests were performed in air using 1-inch square pieces of 0.050-inch gauge sheet in open crucibles. The specimens were weighed before and after scale removal by sandblasting. Density determinations were made by the usual method of weighing in air and then in deaerated water using a precision balance.

DISCUSSION AND RESULTS

In work reported in the First Quarterly Report, it was found that additions of up to 10% Cr and Mn to the base alloys resulted in alloys having good fabrication properties. It was, therefore, decided to try two further increments of Cr and Mn in steps of 2.5% each, to see if new alloys combining both good fabrication properties and higher annealed strengths could be obtained.

Rolling Results

Sheet was processed as in Phase I, being hot rolled to 0.080-inch gauge, then being cleaned by sandblasting and pickling, before being cold rolled to 0.050-inch gauge. Figures 1 to 3 are typical of the severe cracking taking place in all of the "new" alloys, mainly during the cold rolling operation. It may be concluded that, without modification of rolling procedure, it will not be possible to produce good strip from alloys of these compositions. However, production of sheet which is hot rolled to final gauge may be possible with some of the alloys; Ti-17V-12.5Mn-3Al may be considered marginal in this respect.

Tensile Results

Tensile tests were performed upon all except two alloys, Ti-8Mo-8V-15Mn-3Al and Ti-15Mo-15Mn-3Al, where the sheet produced was so badly cracked that no specimens could be obtained. Tests were performed on the other alloys after solution treatment at 1350F for 15 minutes, followed by slow cooling, and after this treatment plus an age of 8 hours at 900F. With the exception of Ti-17V-12.5Cr-3Al and Ti-15Mo-12.5Cr-3Al, which were marginal, the results shown in Table I indicate that little or no ductility existed in these alloys. Strengths and moduli increased with replacement of vanadium by molybdenum, and to a lesser extent with additions of manganese and chromium. It will be observed that aging this group of alloys caused little change in tensile properties, which demonstrates that it would be impractical to attempt precipitation hardening through further additions of Cr or Mn.

Based on these results and those previously reported,⁽¹⁾ it appears that a limited range of useful composition exists in each base alloy: with insufficient beta stabilizing elements present, the alloys are unstable; with too much, they cannot be fabricated.

1. Rosenberg, H. W., "Development Of A Stable-Beta Titanium Alloy," Contract No. DA-30-069-ORD-3743, First Quarterly Report, June 29 through September 30, 1962.

Limits of acceptability therefore are:

<u>Base Material</u>	<u>Min. % Alloying Element Needed For Stability (to nearest 2.5%)</u>	<u>Max. % Alloying Element Tolerable For Fabrication (to nearest 2.5%)</u>
Ti-17V-3Al	10.0 Cr 7.5 Mn 7.5 Fe	12.5 Cr 10.0 Mn 7.5 Fe
Ti-8Mo-8V-3Al	7.5 Cr 7.5 Mn 5.0 Fe	10.0 Cr 10.0 Mn 7.5 Fe
Ti-15Mo-3Al	5.0 Cr 5.0 Mn 5.0 Fe	10.0 Cr 10.0 Mn 5.0 Fe

The limits for iron are rather narrower than for the other two elements. The onset of brittleness with increasing alloy content may be related to the body-centered-cubic beta structure. By analogy with other alloy systems, the transition temperature of beta might be expected to be sensitive to total alloy content, particularly as the alloying element and the titanium base become further separated in the periodic table. Warm rolling the commercial Ti-13V-11Cr-3Al at 200 - 300F improves its rollability. Because a similar effect should be observable in the alloys of this contract, marginal rolling alloys were considered for Phase II materials. However, there is a possibility that this may result in certain undesirable aspects from the standpoints of producibility and quality.

Metallography

Metallographic examination of the alloys was undertaken to determine the recrystallization temperature and beta transes. Figure 4 illustrates extensive slip and elongated grains typical of the as-rolled condition. An innocuous precipitate was also usually present in the as-rolled structure. Occasionally extensive cracking was observed (as seen in Figure 6). After heat treating for $\frac{1}{2}$ -hour at 1250F, a precipitate was present in the grain boundaries, as in Figure 5, which gradually dissolved on heating at temperatures of 1450F and above, as in Figure 6. Recrystallization began at 1350F. Some grain growth was evident in many alloys at 1550F (Figure 7 is an example). Exceptions to these generalizations were the 15% Mo alloys containing 12.5 and

15% Mn, in which little precipitate appeared at any temperature. This lack of precipitate may be due to the sluggish eutectoid reaction of Mn which in binary Ti-Mn alloys often promotes the beta-isomorphous type of phase equilibrium. That very little precipitate appeared in the alloys containing Mo and Mn may also be due to the purities of the alloy constituents used in melting. For example, small amounts of precipitate accompany the use of an 85% V - 15% Al master alloy in Ti-13V-11Cr-3Al, whereas precipitate is absent when high purity elemental constituents are used. These particles remain undissolved even at temperatures at which intermetallic compounds are in solution in the beta phase. No significant effect of this precipitate on tensile properties has been shown.

In alloys whose tensile properties were detailed in the First Quarterly Report, subsequent metallographic examination revealed that a good correlation existed between tensile properties and microstructures. Figure 8 shows Ti-15Mo-7.5Mn-3Al after solution treatment, and Figure 9 shows the same alloy solution treated and aged. This alloy had 25% total elongation before and after aging, and may therefore be regarded as stable for the purposes of this program. The two figures indicate that little precipitate appeared upon aging. By way of contrast, Ti-17V-2.5Fe-3Al, an unstable alloy, had 15% elongation before aging and 6% afterwards, concomitant with considerable strengthening. Comparison of Figures 10 and 11 show a darkening of grains after aging that is typical of alpha precipitation. Figures 12 and 13 show the microstructural response to aging of Ti-15Mo-5Cr-3Al, an alloy which may be regarded as being semi-stable. Note that alpha precipitate darkened certain grains only, others being unaffected. This is caused by polygonization of certain grains preliminary to full recrystallization, whereby such beta grains are subject to rapid aging. This polygonization is characteristic of partially recrystallized beta alloys and makes it difficult to tell whether true recrystallization has in fact occurred, unless the alloy is aged after annealing.

Oxidation Studies

In order to round out Phase I, it was necessary to determine the oxidation resistance of selected alloys.

These tests were carried out at five temperatures, samples being given 2 hours exposure. Tests were carried out in open crucibles, using a still air electric furnace and samples 1-inch square, and results determined as grams of weight lost per square centimeter after sandblasting. This method will be replaced in future tests by a method using a covered crucible, as observations showed that the oxides tended to spall on removal from the furnace, and as sandblasting is not uniform.

Results are given in Table II and shown in Figures 14 to 16. It will be seen that the oxidation resistance of the 15% Mo group of alloys was superior, the least oxidation resistance being found among the 17% V group. Of the other alloying elements added, in all cases Cr conferred the greatest oxidation resistance with little to choose between Fe and Mn.

Density Determinations

Density determinations were carried out upon sandblasted and pickled portions of sheet, which were weighed in air and then in deaerated water and the apparent loss in weight determined. A precision balance capable of weighing 0.1 milligram was employed. Densities so calculated varied from 0.175 lbs/cu.in. to 0.189 lbs/cu.in. and are shown in Table III. This was an 8% range, meaning that an alloy of yield strength of 165,000 psi and a density of 0.189 lbs/cu.in. would only be as strong, on a strength-to-weight basis, as one of density 0.175 lbs/cu.in. and 153,000 psi yield strength. On this strength-to-weight basis, the base alloys selected for Phase II have ratios exceeding 850,000, whereas that of Ti-13V-11Cr-3Al is 715,000 to 750,000 in the annealed condition.

Analytical Determinations

In order to check that no formulation difficulties would be encountered in melting of alloys selected for Phase II, it was necessary to analyze the candidate alloys. Standard ASTM procedures, where applicable, were used for the analyses. Results in Table IV show that with exceptions of the Fe and Mn contents all values are close to those calculated. Some loss of Mn is normal due to its volatility; the higher Fe contents may be due to partitioning during melting.

SELECTION OF BASE ALLOYS FOR PHASE II

The criteria for selection of Phase II alloys for use as bases to which the hardening agents will be added are:

1. Stability
2. Fabrication Properties
3. Yield Strength, Total and Uniform Elongation
4. Ease of Melting
5. Density
6. Oxidation Resistance
7. Cost

These will now be dealt with in turn, and the final choices for Phase II discussed in detail.

Stability

This was defined as the ability of the alloy to undergo an aging treatment of 900F for 8 hours without increasing in strength more than 10% of the annealed value. When given this test, 11 of the 39 alloys under consideration did not meet this criterion and were therefore discarded. In all instances, these were alloys that had insufficient quantities of beta-stabilizing elements added. Increases in strength were accompanied by losses in ductility.

Ease of Fabrication

An important property of the alloys was their ability to be rolled into sheet without cracking. This was determined by hot rolling down to 0.080-inch, sandblasting and pickling, then cold rolling down to 0.050-inch. Thirteen of the thirty-nine alloys cracked up so badly in cold rolling that they could not be considered. This lack of rollability was associated with the higher percentages of beta-eutectoid stabilizers added and was aggravated to some extent when Mo was present in the base. It was found that good fabrication properties could not be retained when more than 7.5% Fe, 10% Cr or 10% Mn were added to the base alloys. Borderline cases, such as Ti-8Mo-8V-7.5Fe-3Al, which developed some cracks in rolling, were retained for further consideration as it was felt that with care good sheet could be produced. Fifteen alloys were thus candidates and the final choices were governed more by tensile properties than the other factors.

High Yield Strength and Uniform Elongation

In order to develop a commercially useful alloy capable of being hardened to high strengths, it may be advantageous to start with an alloy which possesses a good annealed strength along with adequate ductility. Furthermore, it is important that the alloy should have a good uniform elongation, as this is an index of drawability. Ten alloys, listed in Table V, met these selection criteria and were therefore selected as showing most promise in this project. These will be discussed later and the reasons for choice given. All ten of these alloys had a UTS over 140,000 psi and a yield strength of at least 135,000 psi, yet still possessed over 10% total elongation.

Ease of Melting

No difficulty in melting the alloys selected for Phase II from the viewpoint of loss of alloying elements was expected. By analogy with present commercial alloys, it would be expected that the manufacturers could melt the base alloys without trouble. Certain of the proposed additions, such as Be, might require study. A decision on one alloy, Ti-15Mo-7.5Mn-3Al, was deferred pending an investigation of master alloys that could be used.

Density

Density determinations showed that all the alloys fell within the range of 0.175 to 0.189 lbs/cu.in. Experimental values found are detailed in Table III. Although strength/density ratio may be a factor in application of the developmental alloys, no maximum density is presently stipulated.

Oxidation Resistance

Tests for oxidation resistance showed that this was higher with increasing Mo content, and lowest in those alloys containing 17% V. These data are of questionable significance in predicting behaviour in fabrication or service.

Cost

Estimations of the cost of preparing any of the ten alloys considered as candidates for Phase II showed that differences in material costs among them were negligible.

FINAL SELECTIONS FOR PHASE II

Three of the ten base alloys considered suitable for modification under Phase II were selected. These were:

1. Ti-8Mo-8V-7.5Fe-3Al
2. Ti-17V-10Cr-3Al
3. Ti-15Mo-5Fe-3Al

A fourth alloy, Ti-15Mo-7.5Mn-3Al, was deferred pending a decision on ease of melting.

It will be seen that one alloy from each type of base was chosen. This should uncover the relative advantages of each type and thus provide a comparison among them.

The Ti-8Mo-8V-7.5Fe-3Al alloy had a yield strength of over 160,000 psi, total elongation over 20%, a uniform elongation of 10%, a density of 0.180 lbs/cu.in., fair oxidation resistance, but borderline fabrication properties. Although cracking of the material did occur during rolling, this was not too serious and the sheet had the smoothly rounded outline characteristic of those alloys with good fabricability. During analysis of the material, it was found that the Fe content was $\frac{1}{2}\%$ higher than planned, which no doubt contributed to the rolling problem. This alloy from the all around point of view was the best developed during Phase I.

Ti-17V-10Cr-3Al had a yield strength of 135,000 psi, total elongation over 14%, uniform elongation of 10%, good rolling properties, a density similar to Ti-13V-11Cr-3Al (0.175 lbs/cu.in.), and fair oxidation resistance. This alloy, though not having the strength of some of the other alloys in the 17% V group, had greater ductility.

Ti-15Mo-5Fe-3Al had a yield strength of 140,000 psi, total elongation of almost 20%, uniform elongation of at least 10%, a density of 0.183 lbs/cu.in., and good oxidation resistance. This last might be due to the preferential oxidation of Mo in the alloy, to MoO₃ which may cause a depletion of Mo in the surface. This postulate is supported by the observation that during hot rolling there was a progressive decrease in the amount of fume emitted.

Ti-15Mo-5Fe-3Al was judged to be the best representative from the 15% Mo group, having a good combination of strength and uniform elongation (though a lower strength/weight ratio than alloys in the other two groups).

Ti-15Mo-7.5Mn-3Al had properties similar to the above alloy. With this alloy, however, melting of homogeneous ingots might prove to be a problem because of the volatility of Mn. It might become necessary to prepare a master alloy of Mo-Mn-Al first to minimize this.

SELECTION OF PHASE II PRECIPITATION-HARDENING ELEMENTS

General Considerations

Precipitation-hardening as a means of strengthening titanium alloys is presently used in several commercial alloys; well known amongst them are Ti-6Al-4V and Ti-13V-11Cr-3Al. The former hardens by precipitation of beta from martensitic alpha, and the other by way of a complex reaction that essentially consists of rejection of alpha from metastable beta. The purpose of this contract was to explore a third method of precipitation-hardening: precipitation of an intermetallic compound from a stable beta solution.

Conditions are favorable for precipitation of compounds in an alloy system when certain conditions are fulfilled. The solid solubility curve must be retrograde; examples of this are the Ti-Cu and Ti-Ge systems. Theoretically, in such systems precipitation-hardening can be accomplished by quenching from the beta field (or even from the beta plus compound field), and aging in the alpha plus compound region. Under such conditions, hardening may occur through formation of clusters of atoms which are initially coherent with the parent lattice, which thereby lead to high strengths in the alloy through coherency strains. Alternatively, if the alloy content of a beta eutectoid element is increased so that it exceeds the eutectoid composition, aging above the eutectoid temperature after sufficiently rapid cooling to retain beta could allow dispersion hardening by means of compound formation. Preferably, either of these two mechanisms should be fairly rapid at low temperatures so that hardening may be accomplished within reasonable times. Also, the compound precipitation kinetics should be such that significant hardening occurs before alpha begins to form. However, it is probably not possible to suppress alpha precipitation completely at reasonable levels of alloy content.

A literature survey was therefore undertaken of the alloying element characteristics pertinent to precipitation-hardening of the beta matrix. A list of the elements considered, with their atomic size factors, solubilities in alpha and beta titanium, electronegativity, valency, and type of intermetallic compound formed is given in Table VI. The relationship of the atomic size factors and electronegativity to the solubilities in alpha and beta titanium is shown graphically in Figures 17 and 18. Also included in the table is the respective cost (in dollars/pound) of each element and, based on this figure, the maximum amount which could be added without raising the cost of the alloy more than 10%. Other factors, such as volatility and density, may become of major importance when selecting an element for use as a precipitation-hardening agent.

Alloying elements may be divided into beta-eutectoid, compound formers, and peritectoid categories. They are considered next under these headings and the reasons for acceptance or rejection given. Table VII summarizes the reasons for rejection of elements found unsuitable.

Beta-Eutectoid Elements

A group of beta-eutectoid elements which have extensive solubility in the beta phase, but a restricted one in alpha titanium, are: Cr, Co, Cu, Fe, Mn, Ni, U, Ag, Au, Bi, Pb, and Tl. One of these elements, Cu, has already been the subject of a patent as a precipitation-hardening agent in a binary Ti-Cu alloy. It also has the largest solubility in alpha titanium of any of the above elements and thus was a natural choice for this project. Of the other agents, sufficiently large amounts of Fe, Cr, and Mn have been found to cause poor fabrication properties in Phase I; in addition, their sluggish eutectoid reactions would necessitate unduly long aging times. Uranium has a high density and is costly. Nickel and cobalt, however, are cheap, readily available, have medium densities, and should produce hardening when added in relatively small amounts. Cu, Ni, and Co were therefore selected from this group for use in Phase II.

With all the above elements, hardening is accomplished by precipitation of compounds listed in Table VI. Of these compounds, Ti_2Cu , Ti_2Co , and Ti_2Ni are isomorphous and therefore should be miscible in all proportions because of the similar size factors and electronegativity of their constituent atoms. Hence, it might be possible to substitute one element for another and achieve even greater hardening with mixtures of the compounds.

The other elements in the above group had various drawbacks which led to their rejection. For example, silver and gold have favorable atomic size factors with respect to titanium, and have also been the subject of investigation as precipitation-hardening agents. However, silver has extensive solid solubility in beta and the solvus slope is almost vertical. Because it was felt that excessive quantities of silver would be needed for hardening, the element was rejected. Gold has a favorable solvus slope, but is heavy and costly; it was, therefore, also discarded.

Other beta-eutectoid elements considered were Bi, Pb, and Tl. However, Bi boils below the melting point of titanium and was considered unsuitable; Pb and Tl also have high volatility. Although the Ti-Tl system is unknown, on a basis of size factor, electronegativity and position in the periodic table, its alloying effects are probably similar to those of Pb. Unreported TMCA work shows that upon melting a

Ti-5Tl binary alloy, only 0.5% Tl was retained and that was in solution. As lead and thallium have in addition high densities, they were dropped from consideration for use in Phase II.

Compound Formers

These are elements with limited solubility in both alpha and beta phases and which form compounds. The rather similar elements Ge and Si form isomorphous compounds Ti_5Ge_3 and Ti_5Si_3 . Germanium is an alpha stabilizer, with a favorable size factor and hence appreciable solubility in alpha titanium. However, germanium was not considered as it is currently priced at \$145.00 per pound. Attention was therefore shifted to silicon. Because of its unfavorable size factor, the solid solubility is restricted; however, it has a maximum solubility of over 0.5% in alpha and about 3% in beta. Silicon has also been used as an addition to promote high strength in other alloys. It is fairly cheap and of low density and was therefore included in Phase II.

Indium was also considered, but its high price plus the large amount needed for hardening ruled it out.

The Ti-Be system is largely unknown. However, Be lowers the melting point of titanium; a binary Ti-3Be alloy melts below 1800F. On the possibility that a favorable retrograde solvus may exist, Be was included in the list of hardening agents for Phase II. No problem with toxicity in button melting is anticipated as massive Be will be used and also because melting will be under atmospheric pressure with furnace and exhaust train scrubbing after melting.

Peritectoid Elements

Use of the rare earths as hardening agents was considered. Because of their low electronegativity values and relative size factor, solid solubility is limited. There is an apparent lack of chemical affinity between titanium and the rare earths because the precipitate is the rare earth itself. Therefore, if the binary alloys are fabricated at temperatures above the melting point of the rare earth and below their solvus temperatures, they are likely to be hot short. However, the complexity of the present alloys may result in the formation of compounds of the rare earths, such as with aluminum, which have high melting points and so overcome the difficulty. On these grounds it was decided to employ misch metal as a hardening agent. Substitution of Y, Nd or Gd might overcome the hot-shortness problem, but these elements cost at least \$100.00 per pound at current market prices.

Selection of Levels

The elements selected for use as hardening agents in Phase II were Cu, Ni, Co, Si, Be, and misch metal. It was decided to add the first three in quantities of 1, 3 and 5%, which should bracket the eutectoid compositions. If necessary, the total beta-stabilizing content of the alloy may be reduced to overcome embrittlement. As Si and Be have a low solubility in alpha, it was decided to add 0.5, 1, and 2% of each. Addition of the rare earths as misch metal in amounts of 1, 2, and 3% were decided upon, as these bracket the maximum binary solubilities. It is not known, however, what the effects of addition of the large quantities of beta-stabilizers will have on the precipitation equilibria or kinetics.

SUMMARY OF PHASE II PLANS

On the basis of the Phase I results, three alloys were selected as bases for Phase II: Ti-8Mo-8V-7.5Fe-3Al, Ti-17V-10Cr-3Al, and Ti-15Mo-5Fe-3Al. Selection of a fourth alloy, Ti-15Mo-7.5Mn-3Al, was deferred pending a decision of its melting feasibility.

Of the precipitation-hardening elements considered for Phase II, it was decided to use 1, 3 and 5% Cu, Ni and Co; 0.5, 1 and 2% of Si and Be; and 1, 2 and 3% of misch metal. Alloys containing these hardening elements will be prepared as 250 gram ingots and processed as before, the resulting 0.05-inch sheet being evaluated by means of hardness tests following solution heat treatment and aging.

CONCLUSIONS

1. Addition of 12.5 and 15% Cr or Mn to Ti-8Mo-8V-3Al, Ti-17V-3Al, or Ti-15Mo-3Al alloys resulted in poor fabrication properties and low ductility. On the basis of the present work, 10% Cr or Mn is the maximum amount of Cr or Mn that can be added to these bases consistent with good cold rolling properties.
2. Strength and moduli of the base alloys increased with replacement of V by Mo, and to a lesser extent with addition of Mn and Cr alloying elements.
3. Oxidation resistance was found to be lowest in alloys containing 17% V and highest in those with 15% Mo. Of the other alloying elements added, Cr conferred the best oxidation resistance.

TABLE I
TENSILE RESULTS ON ALLOYS CONTAINING 12.5 AND 15% Cr OR Mn

TMCA Lot No.	Composition	Heat Treatment	Tensile Properties				Elastic Modulus 10 ⁶ psi	Comments
			UTS psi	YS psi	Local El. %	Total El. %		
T3495	Ti-17V-12.5Cr-3Al	1350F-15SC	158,000	149,000	20	12.5	14	16.4
"	"	"	152,000	148,000	5	2.5	3	16.0
"	"	"	157,000	150,000	15	2.5	7	16.8
"	"	"	161,000	148,000	25	12.5	15	16.5
T3496	Ti-17V-15Cr-3Al	1350F-15SC	167,000	160,000	5	2.5	3	16.9
"	"	"	166,000	159,000	5	2.5	3	16.5
"	"	"	168,000	161,000	5	0	2	16.6
"	"	"	169,000	158,000	10	5	7	16.9
T3497	Ti-8V-8Mo-12.5Cr-3Al	1350F-15SC	159,000	153,000	15	5	7	16.5
"	"	"	160,000	154,000	15	5	7	17.2
"	"	"	160,000	155,000	10	2.5	5	17.1
"	"	"	162,000	156,000	5	0	2	17.1
T3498	Ti-8V-8Mo-15Cr-3Al	1350F-15SC	166,000	158,000	5	2.5	3	17.4
"	"	"	167,000	158,000	5	0	2	17.2
"	"	"	166,000	166,000	0	0	0	17.5
"	"	"	170,000	163,000	5	2.5	3	17.1
T3499	Ti-15Mo-12.5Cr-3Al	1350F-15SC	158,000	150,000	15	7.5	10	16.8
"	"	"	161,000	155,000	15	10	11	17.1
"	"	"	159,000	153,000	5	2.5	4	17.0
"	"	"	159,000	152,000	10	7.5	8	17.0
T3500	Ti-15Mo-15Cr-3Al	1350F-15SC	165,000	158,000	5	2.5	3	17.7
"	"	"	167,000	160,000	5	2.5	4	17.7
"	"	"	165,000	159,000	5	0	2	17.7
"	"	"	165,000	160,000	5	2.5	3	17.9
T3501	Ti-17V-12.5Mo-3Al	1350F-15SC	171,000	168,000	10	5	6	17.6
"	"	"	171,000	168,000	10	2.5	9	17.8
"	"	"	179,000	168,000	10	5	7	16.9
"	"	"	180,000	172,000	10	5	7	17.2
T3502	Ti-17V-15Mo-3Al	1350F-15SC	128,000	---	0	0	0	17.9
"	"	"	113,000	---	0	0	0	19.0
"	"	"	74,000	---	0	0	0	17.9
"	"	"	86,000	---	0	0	0	---

TABLE I (continued)
TENSILE RESULTS ON ALLOYS CONTAINING 12.5 and 15% Cr OR Mo

Composition	TMCA Lot No.	Heat Treatment	Tensile Properties				Elastic Modulus 10 ⁶ psi	Comments
			UTS psi	YS psi	Local El. %	Total El. %		
Ti-8V-8Mo-12.5Nb-3Al	T3503	1350F-158C	114,000	---	0	0	---	
"	"	"	110,000	---	0	0	17.4	
"	"	"	135,000	---	0	0	18.0	
"	"	"	128,000	---	0	0		
Ti-8V-8Mo-15Nb-3Al	T3504	1350F-158C						Sheet no good
"	"	"						"
"	"	"						"
"	"	"						"
Ti-15Mo-12.5Nb-3Al	T3505	1350F-158C	62,300	---	0	0	18.5	
"	"	"	11,600	---	0	0	---	
"	"	"	92,800	---	0	0	---	
"	"	"	133,000	---	0	0	18.5	
Ti-15Mo-15Nb-3Al	T3506	1350F-158C						Sheet no good
"	"	"						"
"	"	"						"
"	"	"						"

TABLE II
OXIDATION RESISTANCE OF SELECTED PHASE I ALLOYS

Heat No.	Alloy	Exposure Temp (°F)	Time (Hrs)	Starting Wt. (Gms)	Wt. After Exposure	Gain (Gms)	Wt. After Sand-Blast (Gms)	Total Wt. Loss (Gms)	% Loss	Wt. Loss Gms/sq cm
T3303	Ti-8V-8Mo-7.5Cr-3Al	1200	2	4.3582	4.3653	.0071	4.3467	.0115	.266	.00044
"	"	1400	"	4.2386	4.2581	.0195	4.2093	.0293	.69	.00113
"	"	1600	"	4.3287	4.4006	.0719	4.2124	.1163	2.7	.0046
"	"	1800	"	4.2453	4.3810	.1357	4.0073	.2380	5.6	.0092
"	"	2000	"	4.3792	4.6167	.2375	3.9729	.4063	9.3	.0157
T3304	Ti-8V-8Mo-10Cr-3Al	1200	"	4.3396	4.3464	.0068	4.3122	.0274	.63	.00106
"	"	1400	"	4.4532	4.4667	.0135	4.3608	.0924	2.07	.00347
"	"	1600	"	4.4732	4.5248	.0516	4.3778	.0954	2.14	.00368
"	"	1800	"	4.4904	4.5558	.0654	4.2571	.2333	5.2	.0090
"	"	2000	"	4.3787	4.4987	.1200	3.9726	.4061	9.3	.0157
T3306	Ti-8V-8Mo-5Fe-3Al	1200	"	4.4675	4.4771	.0096	4.4427	.0248	.56	.00096
"	"	1400	"	4.3893	4.4256	.0363	4.3350	.0543	1.23	.0021
"	"	1600	"	4.3917	4.5293	.1376	4.1075	.2842	6.5	.0110
"	"	1800	"	4.3697	4.5438	.1741	3.7446	.6251	14.4	.0242
"	"	2000	"	4.3636	4.6808	.3172	1/	--	--	--
T3307	Ti-8V-8Mo-7.5Fe-3Al	1200	"	4.2410	4.2513	.0103	4.2152	.0258	.61	.0010
"	"	1400	"	4.4261	4.4735	.0474	4.3488	.0773	1.75	.00298
"	"	1600	"	4.4692	4.6845	.2153	4.0235	.4457	10.0	.0172
"	"	1800	"	4.5596	4.7262	.1666	3.9734	.5862	12.8	.0277
"	"	2000	"	4.4539	4.9704	.5165	1/	--	--	--
T3310	Ti-8V-8Mo-10Mn-3Al	1200	"	4.5269	4.5330	.0061	4.5193	.0076	.17	.00029
"	"	1400	"	4.6384	4.6735	.0351	4.5822	.0562	1.2	.00217
"	"	1600	"	4.5414	4.5474	.0060	4.5736	.1678	3.65	.0065
"	"	1800	"	4.6594	5.4270	.7676	1/	--	--	--
"	"	2000	"	4.5950	6.2034	1.6084	1/	--	--	--
T3311	Ti-15Mo-5Cr-3Al	1200	"	4.2948	4.3005	.0057	4.2717	.0231	.54	.00089
"	"	1400	"	4.2814	4.2944	.0130	4.2209	.0605	1.4	.00234
"	"	1600	"	4.1950	4.2217	.0267	4.1536	.0414	.99	.0016
"	"	1800	"	4.2061	4.2446	.0385	3.8847	.3214	7.6	.0124
"	"	2000	"	4.3446	4.3670	.0244	4.0270	.3176	7.3	.0123

1/ Sample disintegrated in blast.

TABLE II (continued)
OXIDATION RESISTANCE OF SELECTED PHASE I ALLOYS

Heat No.	Alloy	Exposure Temp. (°F)	Time (Hrs)	Starting Wt. (Gms)	Wt. After Exposure	Gain (Gms)	Wt. After Sand-Blast (Gms)	Total Wt. Loss (Gms)	% Loss	Wt. Loss Gms/sq cm
T3313	Ti-15Mo-10Cr-3Al	1200	2	4.3939	4.3988	.0049	4.3523	.0416	.95	.00162
"	"	1400	"	4.5629	4.5726	.0097	4.5342	.0287	.61	.00111
"	"	1600	"	4.4090	4.4345	.0255	4.3285	.0805	1.8	.00312
"	"	1800	"	4.4949	4.4829	2/	4.3179	.1770	3.9	.0068
"	"	2000	"	4.4818	4.3802	2/	4.1211	.3607	8.0	.0140
T3316	Ti-15Mo-7.5Fe-3Al	1200	"	4.6139	4.6186	.0047	4.5648	.0491	1.03	.00195
"	"	1400	"	4.5952	4.6144	.0192	4.4894	.1058	2.3	.0042
"	"	1600	"	4.5306	4.5805	.0499	4.3637	.1669	3.7	.0069
"	"	1800	"	4.8792	4.9325	.0533	1/	--	--	--
"	"	2000	"	4.6644	5.1011	.4367	1/	--	--	--
T3317	Ti-15Mo-5Nb-3Al	1200	"	4.3180	4.3241	.0061	4.3045	.0135	.31	.00052
"	"	1400	"	4.2465	4.2679	.0214	4.2153	.0312	.73	.00121
"	"	1600	"	4.3009	4.3587	.0578	4.2048	.0961	2.24	.00372
"	"	1800	"	4.2779	4.2752	2/	4.0701	.2078	4.85	.0080
"	"	2000	"	4.1796	4.3467	.1671	1/	--	--	--
T3319	Ti-15Mo-10Nb-3Al	1200	"	4.6372	4.6423	.0051	4.6088	.0284	.61	.00110
"	"	1400	"	4.5986	4.6136	.0150	4.5604	.0382	.83	.00148
"	"	1600	"	4.6246	4.6311	.0065	4.5664	.0582	1.25	.00225
"	"	1800	"	4.5457	4.5917	.0460	4.3972	.1945	4.3	.0075
"	"	2000	"	4.5692	4.8703	.3011	1/	--	--	--
T3321	Ti-17V-7.5Cr-3Al	1200	"	4.0367	4.0517	.0150	4.0066	.0301	.75	.00116
"	"	1400	"	4.1451	4.1920	.0469	4.0527	.0924	2.2	.00357
"	"	1600	"	4.1901	4.3277	.1376	3.9690	.2211	5.3	.0085
"	"	1800	"	4.0351	4.2509	.2158	3.5392	.4959	12.3	.0182
"	"	2000	"	4.1486	--	--	3.7615	.3871	9.4	.0150
T3322	Ti-17V-10Cr-3Al	1200	"	4.2962	4.3051	.0089	4.2411	.0551	1.28	.00213
"	"	1400	"	4.3923	4.4212	.0289	4.3142	.0781	1.78	.00302
"	"	1600	"	4.3789	4.4754	.0965	4.1977	.1812	4.1	.0070
"	"	1800	"	4.3314	4.4466	.1152	3.8948	.4366	10.0	.0169
"	"	2000	"	4.1727	4.0098	2/	3.8915	.2812	6.7	.0158

1/ Sample disintegrated in blast.
2/ Oxide lost due to draft.

TABLE II (continued)
OXIDATION RESISTANCE OF SELECTED PHASE I ALLOYS

Heat No.	Alloy	Exposure Temp (°F)	Time (Hrs)	Starting Wt. (Gms)	Wt. After Exposure	Gain (Gms)	Wt. After Sand-Blast (Gms)	Total Wt. Loss (Gms)	% Loss	Wt. Loss Gms/sq cm
T3325	Ti-17V-7.5Fe-3Al	1200	2	4.3503	4.3635	.0132	4.3296	.0721	1.66	.00279
"	"	1400	"	4.3270	4.3991	.0721	4.1160	.2110	4.9	.0082
"	"	1600	"	4.2825	4.7492	.4667	1/	--	--	--
"	"	1800	"	4.3445	5.3271	.9826	1/	--	--	--
"	"	2000	"	4.2360	4.7785	.5425	3.5003	.7357	17.4	.0285
T3327	Ti-17V-7.5Fe-3Al	1200	"	4.1412	4.1509	.0097	4.1321	.0091	.22	.00035
"	"	1400	"	4.1911	4.2937	.1026	4.0701	.1210	2.9	.00468
"	"	1600	"	4.0510	4.3271	.2761	3.5961	.4549	11.2	.0176
"	"	1800	"	4.1770	4.4970	.3200	3.3327	.9443	22.7	.0365
"	"	2000	"	4.0500	4.5795	.5295	1/	--	--	--
T3328	Ti-17V-10Fe-3Al	1200	"	4.2501	4.2683	.0182	4.2421	.0160	.38	.00062
"	"	1400	"	4.3394	4.3206	2/	4.2651	.0743	1.7	.00277
"	"	1600	"	4.1649	4.3525	.1876	3.7224	.4425	10.6	.0171
"	"	1800	"	4.2477	4.4156	.1679	3.6146	.6331	14.9	.0245
"	"	2000	"	4.1564	4.3130	.1566	1/	--	--	--

1/ Sample disintegrated in blast.
2/ Oxide lost due to draft.

TABLE III
DENSITIES OF SELECTED PHASE I ALLOYS

<u>Heat No.</u>	<u>Alloy</u>	<u>Densities g/cc</u>		<u>Average Lbs/In³</u>
		<u>No. 1</u>	<u>No. 2</u>	
T3302	Ti-8Mo-8V-5Cr-3Al	4.879	4.879	0.1763
T3498	" " " 15Cr "	5.100	5.101	0.1843
T3305	" " " 2.5Fe "	4.853	4.851	0.1754
T3307	" " " 7.5Fe "	5.005	5.003	0.1807
T3308	" " " 5Mn "	4.907	4.907	0.1774
T3504	" " " 15Mn "	5.192	5.191	0.1876
T3320	Ti-17V-5Cr-3Al	4.751	4.749	0.1717
T3496	" " " 15Cr "	4.974	4.970	0.1794
T3323	" " " 2.5Fe "	4.724	4.724	0.1707
T3325	" " " 7.5Fe "	4.870	4.869	0.1760
T3326	" " " 5Mn "	4.774	4.776	0.1726
T3502	" " " 15Mn "	5.048	5.046	0.1823
T3311	Ti-15Mo-5Cr-3Al	5.023	5.023	0.1815
T3500	" " " 15Cr "	5.228	5.227	0.1888
T3314	" " " 2.5Fe "	4.976	4.974	0.1795
T3316	" " " 7.5Fe "	5.143	5.141	0.1858
T3317	" " " 5Mn "	5.025	5.023	0.1817
T3502	" " " 15Mn "	5.288	5.288	0.1912

Ti-13V-11Cr-3Al - 4.80

Pure Ti - 4.51

TABLE IV

ANALYSIS OF PHASE I BUTTONS

<u>Alloy</u>	<u>Heat No.</u>	<u>Cr</u>	<u>Mo</u>	<u>V</u>	<u>Fe</u>	<u>Mn</u>	<u>Al</u>	<u>C</u>	<u>O</u>	<u>N</u>	<u>H</u>
Ti-17V-10Cr-3Al	T3322	9.75		16.9			3.27	.026	0.119	.024	.0133
Ti-8V-8Mo-7.5Fe-3Al	T3307		7.84	8.06	8.0		3.27	.022	0.100	.012	.0089
Ti-15Mo-5Fe-3Al	T3315		15.0		5.8		3.06	.026	0.117	.011	.0057
Ti-15Mo-7.5Mn-3Al	T3318		15.2			6.98	2.89	.026	0.144	.011	.0070

TABLE V
A COMPARISON OF THE TEN MOST PROMISING ALLOYS FOUND IN PHASE I

<u>Alloy</u>	<u>UTS</u> <u>(psi)</u>	<u>Yield</u> <u>(psi)</u>	<u>Uniform</u> <u>Elong.%</u>	<u>Total</u> <u>Elong.%</u>	<u>Rolling</u> <u>Properties</u>	<u>Density</u> <u>Lbs/In³</u>	<u>Oxidation</u> <u>Resistance</u>
Ti-17V-10Mn-3Al	155,000+	150,000	10+	12+	Good	0.1774 ₁ /	Fair
Ti-17V-7.5Fe-3Al	160,000	150,000	6	12	Fair	0.1760	Fair
Ti-17V-10Cr-3Al	145,000	135,000	10	14+	Good	0.1755 ₁ /	Fair
Ti-8Mo-8V-7.5Fe-3Al	160,000+	160,000	17	20+	Fair	0.1807	Fair
Ti-8Mo-8V-5Fe-3Al	150,000	145,000	10	12+	Good	0.1780 ₁ /	Fair
Ti-8Mo-8V-10Cr-3Al	145,000+	140,000	10	15	Fair	0.1803 ₁ /	Good
Ti-8Mo-8V-10Mn-3Al	150,000	150,000	2.5	11+	Good	0.1825 ₁ /	Fair
Ti-15Mo-7.5Mn-3Al	140,000+	140,000+	17	20+	Good	0.1864 ₁ /	Good
Ti-15Mo-5Fe-3Al	145,000	142,000	10	21	Good	0.1826 ₁ /	Good
Ti-15Mo-10Cr-3Al	144,000+	139,000	12	20	Fair	0.1852 ₁ /	Good

Ti-13V-11Cr-3Al							
(annealed)	125,000	120,000	-	10	Fair	0.175	Fair
(aged)	190,000	170,000	-	4	-	-	Fair

All above Phase I alloys tested after aging for 8 hours at 900F.

1/ Calculated density.

TABLE VI

PROPERTIES OF POTENTIAL PHASE II HARDENING ELEMENTS

Element	Crystal Structure	Size Factor	Type Alloy System	Chemical Valency	Electro-Negativity	Max. Solubility in		Formula	Compound Type	Hardener Cost (\$/lb)	Max. % Addition Consistent With Not Raising Cost of Alloy More Than 10%
						Alpha	Beta				
Ag	F.C.C.	-2.0	B	1	1.9	14	24	Ti ₃ Ag	F.C.C.	13.3	3.5
Au	F.C.C.	-2.0	B	3	2.4	6.6	42	Ti ₃ Au	Tetragonal	510.0	0.09
Be	C.P.H.	-23.0	B*	2	1.5	17	?	TiBe	Tetragonal	62.0	0.82
Bi	Rhombohedral	+23.8	B	3, 5	1.9	1.5	33	Ti ₃ Bi	Tetragonal	2.25	23.1
Ce	F.C.C.	+22.0	D	3	1.1	0.37	4	None	-	25.0	2.1
Co	C.P.H.	-15.0	B	2	1.8	1	17	Ti ₃ Co	F.C.C.	1.57	33.6
Cr	B.C.C.	-13.0	B	3	1.6	0.5	100	TiCr ₂	F.C.C.	1.19	43.8
Cu	F.C.C.	-12.7	B	2	1.9	2.1	17	Ti ₃ Cu	F.C.C.	0.47	100
Fe	B.C.C.	-13.0	B	3	1.8	0.2	25	TiFe	B.C.C.	0.30	100
Gd	C.P.H.	+22.5	D	3	1.1	?	?	None	-	275.0	0.12
Ge	Diamond Cubic	-5.4	C	4	1.8	4	12	Ti ₃ Ge ₃	Hexagonal	145.0	0.30
In	F.C.T.	+7.0	A	3	1.7	22	?	Ti ₃ In	D019	18.25	1.56
La	C.P.H.	+28.0	D	3	1.1	47	?	None	-	270.0	0.19
Mn	Complex cubic	-7 to -24	B	2, 4	1.5	0.4	35	TiMn	Tetragonal	0.37	100
Nd	C.P.H.	+24.5	D	3	1.2	?	?	None	-	360.0	0.14
Ni	F.C.C.	-15.0	B	2	1.8	0.2	12	Ti ₃ Ni	F.C.C.	1.00	52.2
Pb	F.C.C.	+19.0	B	2, 4	1.8	167	45	Ti ₃ Pb	D019	0.09	100
Si	Diamond cubic	-20.5	B	4	1.8	0.5	3	Ti ₃ Si ₃	Hexagonal	1.60	32.6
Tl	C.P.H.	+17.0	B*	4	1.8	0.5	?	?	-	7.50	6.9
U	Orthorhombic	-7.5	B	4, 5, 6	1.7	3.8	100	TiU ₂	Hexagonal	50.00	1.04
Y	C.P.H.	+23.0	D	3	1.3	1.07	?	None	-	405.00	0.13

Key to types of alloy systems:

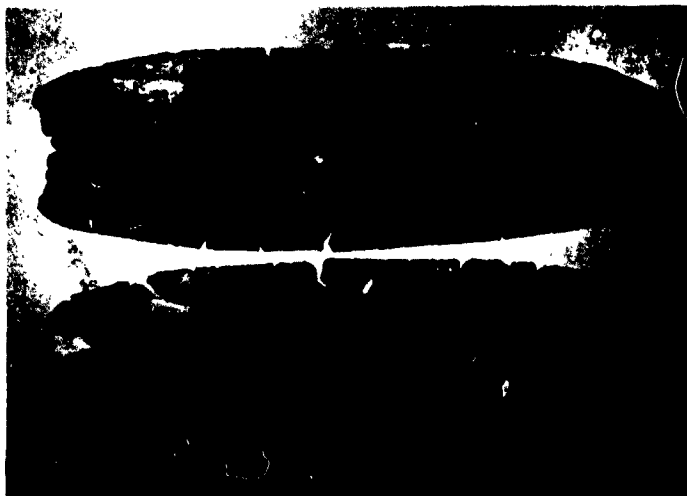
- A - Neutral
 B - Beta-eutectoid
 C - Alpha-compound
 D - Alpha-element
 * - System unknown; type thought most likely on basis of size factor and position in periodic table

- C.P.H. - Close-packed hexagonal
 F.C.C. - Face centered cubic
 B.C.C. - Body centered cubic
 F.C.T. - Face centered tetragonal

TABLE VII

REASONS FOR REJECTION OF PHASE II HARDENING AGENTS

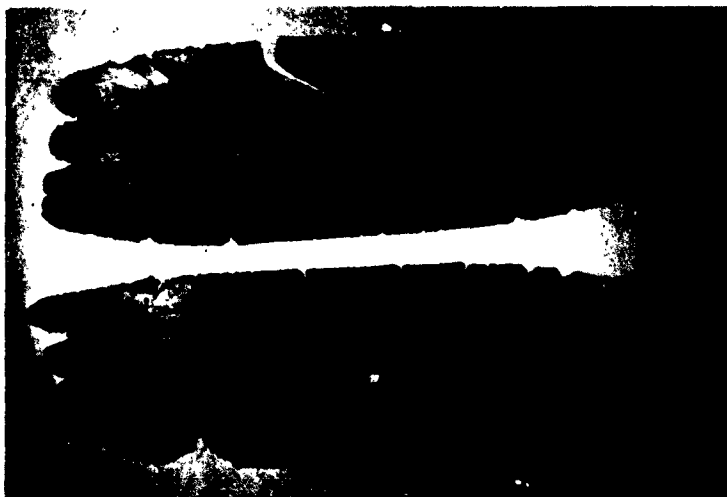
Element	Grounds of Rejection				Other Agent Considered Better
	Volatility	Density	Cost	Already Found to Cause Embrittlement in Phase I Alloys Excessive Amount Needed for Hardening	
Ag				X	(Cu)
Au		X	X		(Cu)
Bi	X	X			
Cr				X	(Ni)
Fe				X	(Ni)
Gd, La, Nd, Y			X		(Misch Metal)
Ce			X		(Si)
In			X	X	
Mn				X	
Pb	X	X			
Tl	X	X			
U		X			



A.

Ti-17V-12.5Cr-3Al

Ti-17V-15Cr-3Al

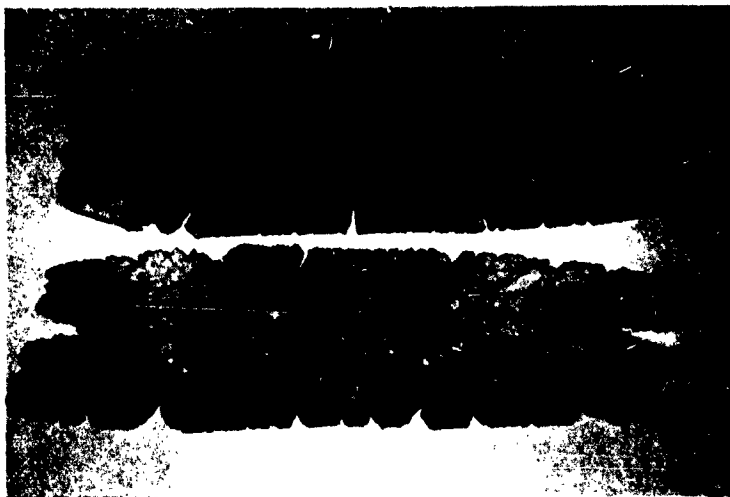


B.

Ti-8Mo-8V-12.5Cr-3Al

Ti-8Mo-8V-15Cr-3Al

Figure 1. Appearance Of Sheet After Cold Rolling.



A.

Ti-15Mo-12.5Cr-3Al

Ti-15Mo-15Cr-3Al

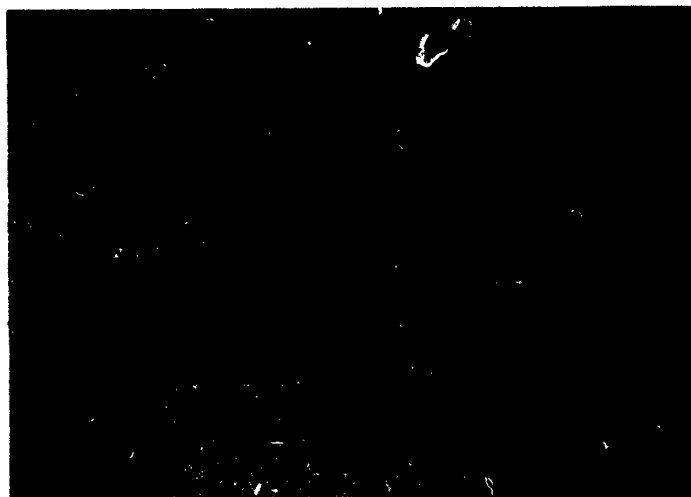


B.

Ti-17V-12.5Mn-3Al

Ti-17V-15Mn-3Al

Figure 2. Appearance Of Sheet After Cold Rolling.



A.

Ti-8Mo-8V-12.5Mn-3Al

Ti-8Mo-8V-15Mn-3Al



B.

Ti-15Mo-12.5Mn-3Al

Ti-15Mo-15Mn-3Al

Figure 3. Appearance Of Sheet After Cold Rolling.



5053A

Oxalic + Kroll Etch

250X

Figure 4. T-3498, Ti-8Mo-8V-15Cr-3Al As-Rolled. Slip Bands In Distorted Beta Grains. A Few Particles Of Second Phase Present.



5055B

Oxalic + Kroll Etch

250X

Figure 5. T-3500, Ti-15Mo-15Cr-3Al. Heat Treated $\frac{1}{2}$ -Hour At 1250F, Quenched. Unrecrystallized With Precipitate Mainly At Grain Boundaries.



5059D

250X

Oxalic + Kroll Etch

Figure 6. T-3504, Ti-8V-8Mo-15Mn-3Al, Heat Treated $\frac{1}{2}$ -Hour At 1450F, Quenched. Transgranular And Intergranular Cracking Produced In Cold Rolling. Recrystallized With Some General Precipitate.



5059E

250X

Oxalic + Kroll Etch

Figure 7. T-3504, Ti-8V-8Mo-15Mn-3Al. Heat Treated $\frac{1}{2}$ -Hour At 1550F, Quenched. Fully Recrystallized With Some Grain Growth And A Few Coarse Particles Of Second Phase.



4064D

250X

Oxalic + Kroll Etch

Figure 8. T-3318, Ti-15Mo-7.5Mn-3Al. Annealed 15 Minutes At 1350F, Slow Cooled. 60% Recrystallized Beta With Scattered Particles Of Second Phase. Uniform Elongation 25%; Total Elongation 30%; UTS 142,000 psi.



4064F

250X

Oxalic + Kroll Etch

Figure 9. T-3318, Ti-15Mo-7.5Mn-3Al. Heat Treated 15 Minutes At 1350F, Slow Cooled, Plus 8 Hours Age At 900F. Little Change in Microstructure. Uniform Elongation 15%; Total Elongation 22%; UTS 144,000 psi.



4066D

250X

Oxalic + Kroll Etch

Figure 10. T-3323, Ti-17V-2.5Fe-3Al. Annealed 15 Minutes At 1350F, Slow Cooled. Partly Recrystallized With Second Phase At Former Grain Boundaries. Uniform Elongation 7.5%; Total Elongation 15%; UTS 120,000 psi.



4066E

250X

Oxalic + Kroll Etch

Figure 11. T-3323, Ti-17V-2.5Fe-3Al. Heat Treated 15 Minutes At 1350F, Slow Cooled, Plus 8 Hours Age at 900F. Heavy Precipitate (alpha + TiFe?) Within Grains. Uniform Elongation 5%, Total Elongation 7%; UTS 197,000 psi.



4062A

250X

Oxalic + Kroll Etch

Figure 12. T-3311, Ti-15Mo-5Cr-3Al. Annealed 15 Minutes At 1350F, Slow Cooled. 50% Recrystallized Beta With Slight Traces Of Second Phase. Uniform Elongation 12.5%; Total Elongation 19%, UTS 126,000 psi.



4062B

250X

Oxalic + Kroll Etch

Figure 13. T-3311, Ti-15Mo-5Cr-3Al. Heat Treated 15 Minutes At 1350F, Slow Cooled Plus 8 Hour Age At 900F, Slow Cooled. Unrecrystallized Grains Darkened By Alpha With Others Unchanged. Uniform Elongation 10%; Total Elongation 18%, UTS 135,000 psi.

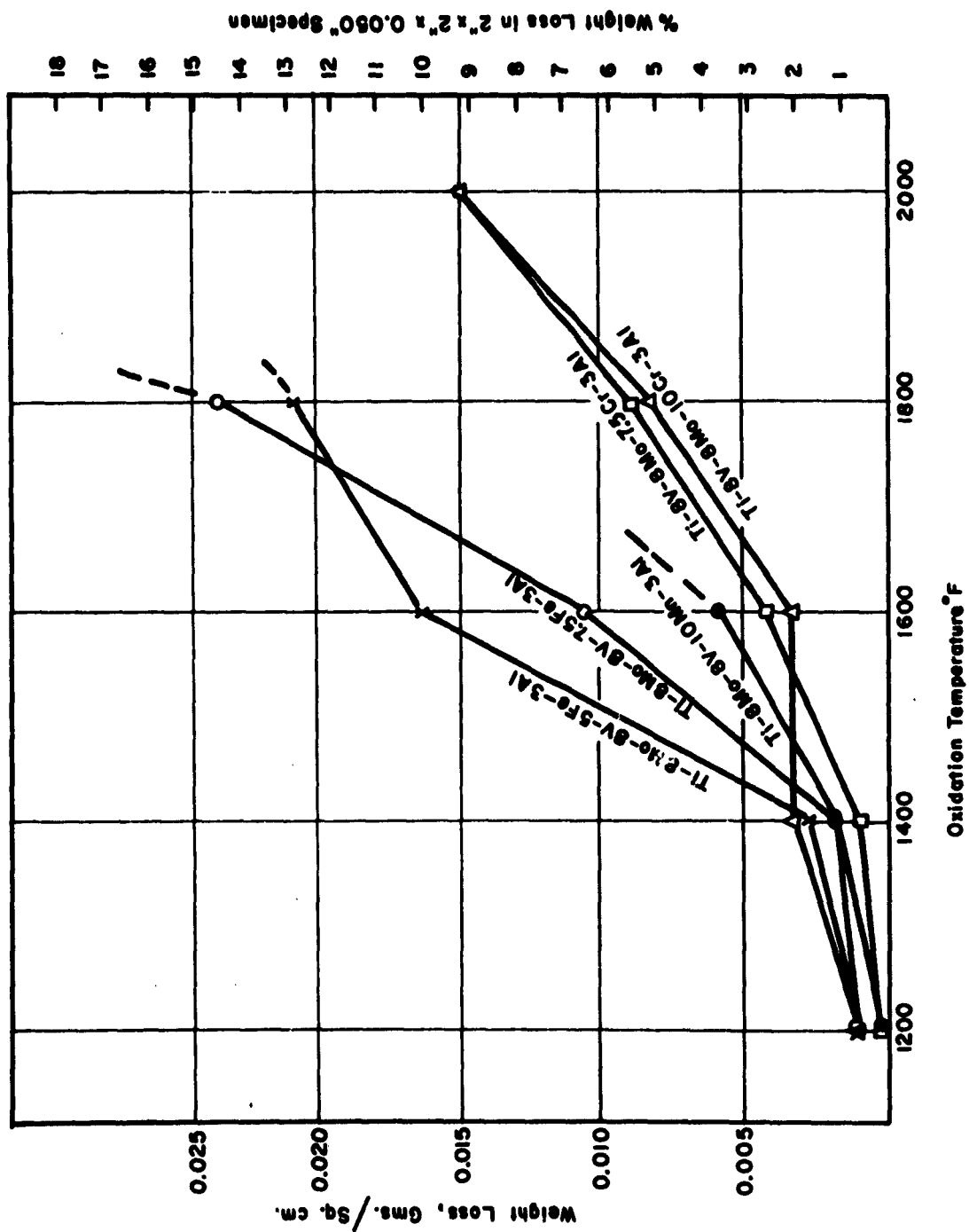


FIGURE 14. OXIDATION RATES FOR Ti-8V-8Mo-X-3Al ALLOYS AFTER 2 HOURS EXPOSURE.

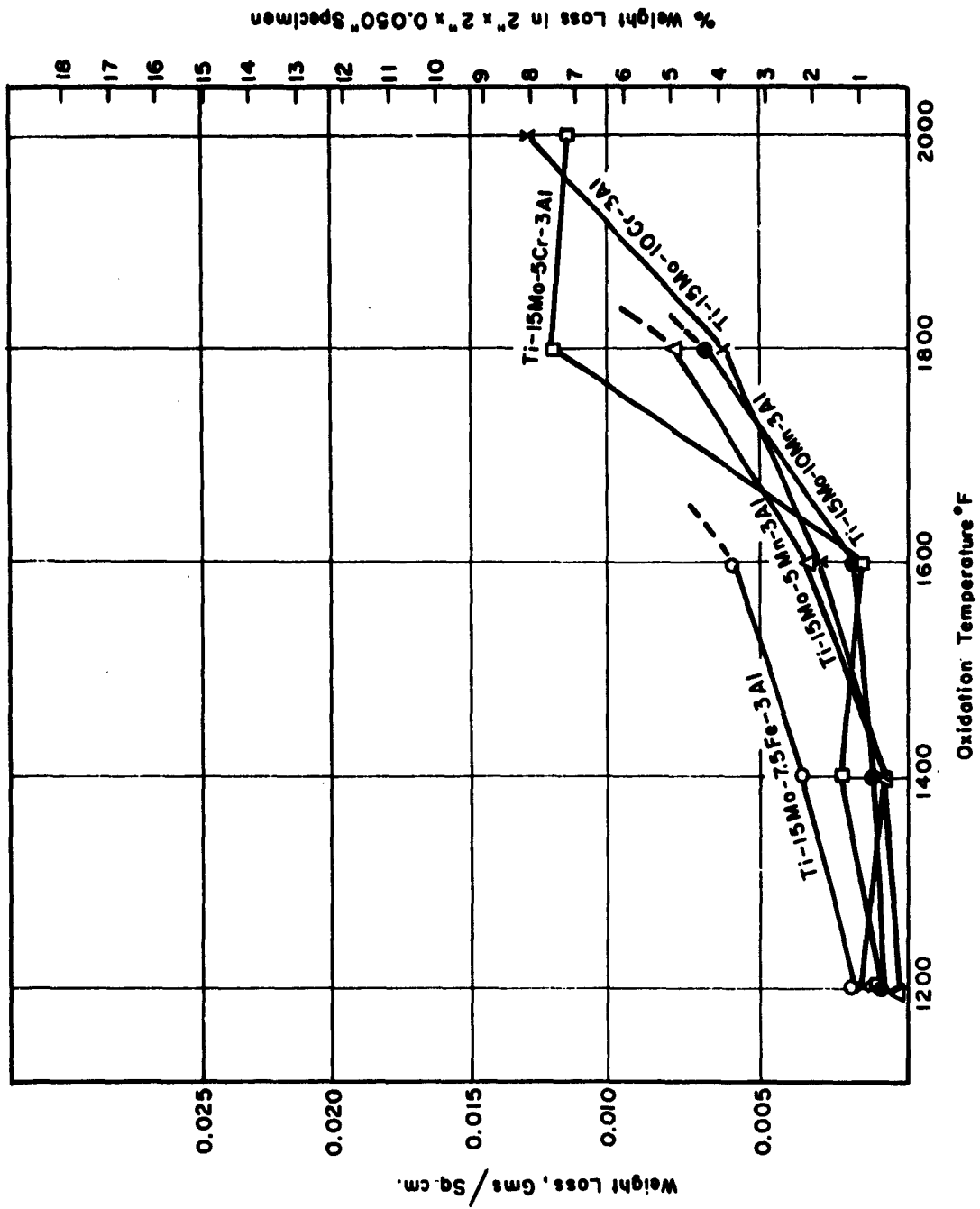


FIGURE 15. OXIDATION RATES FOR Ti-15Mo-X-3Al ALLOYS AFTER 2 HOURS EXPOSURE.

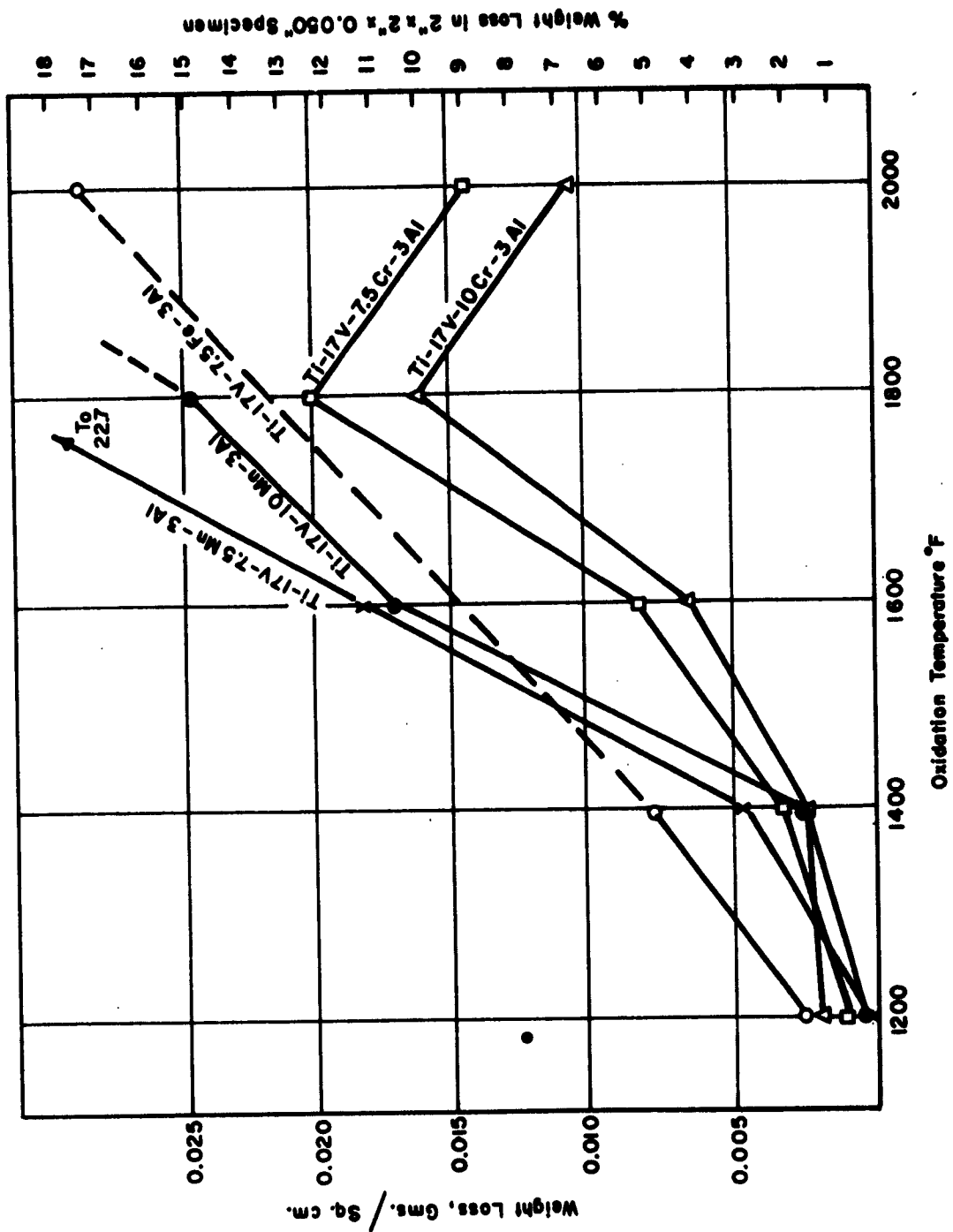


FIGURE 16. OXIDATION RATES FOR Ti-17V-X-3Al ALLOYS AFTER 2 HOURS EXPOSURE.

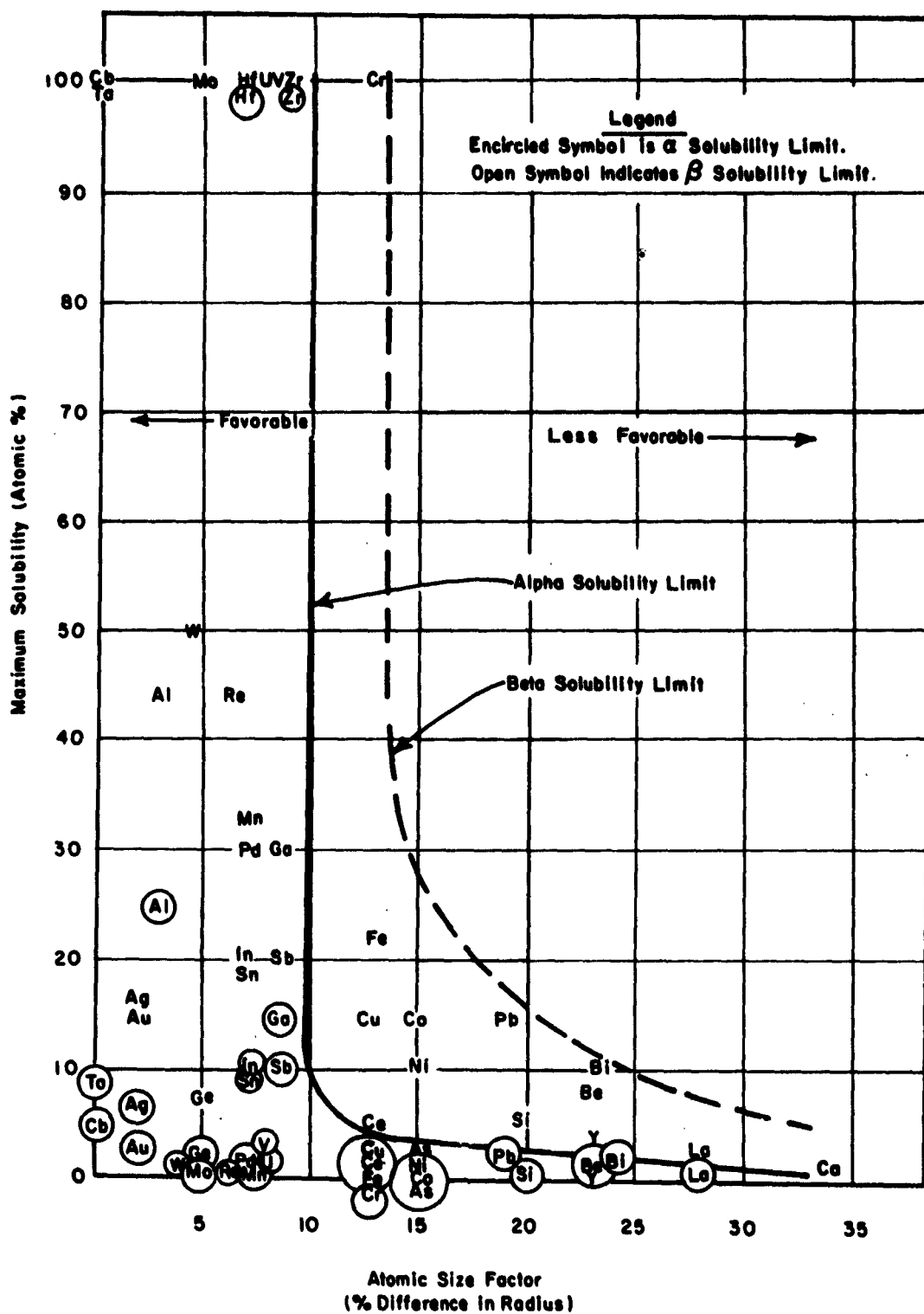


FIGURE 17. INFLUENCE OF SIZE FACTOR ON SOLID SOLUBILITY.(SUBSTITUTIONAL ALLOYING)



**FIGURE 18. INFLUENCE OF ELECTRONEGATIVITY ON SOLID SOLUBILITY.
(SUBSTITUTIONAL SOLUBILITY)**

AD <u>Watertown Arsenal Laboratories, Watertown 72, Mass.</u> DEVELOPMENT OF A STABLE BETA TITANIUM ALLOY - <u>D. B. Bunker</u> <u>Titanium Metals Corporation of America, New York, N. Y.</u> Report No. WAL TR 435/2-1, October 1 to December 31, 1943, 14 pp - tables - illus, (Contract DA-30-068-ORD-3743), DA Proj 58323008, Unclassified Report Phase I was completed by determining the hot rolling limits of the base alloys. Reliability became more difficult as the total eutectoid content increased, and as the Mo content increased in the base. Tensile tests were used to evaluate thermal stability of the alloys. Density measurements were made and oxidation tests carried out; alloys containing Mo and Cr exhibited the best oxidation resistance. Phase II base alloys were selected and various hardening agents discussed. NO DISTRIBUTION LIMITATIONS	UNCLASSIFIED 1. Alloys, titanium	AD <u>Watertown Arsenal Laboratories, Watertown 72, Mass.</u> DEVELOPMENT OF A STABLE BETA TITANIUM ALLOY - <u>D. B. Bunker</u> <u>Titanium Metals Corporation of America, New York, N. Y.</u> Report No. WAL TR 408/2-1, October 1 to December 31, 1943, 14 pp - tables - illus, (Contract DA-30-068-ORD-3743), DA Proj 58323008, Unclassified Report Phase I was completed by determining the hot rolling limits of the base alloys. Reliability became more difficult as the total eutectoid content increased, and as the Mo content increased in the base. Tensile tests were used to evaluate thermal stability of the alloys. Density measurements were made and oxidation tests carried out; alloys containing Mo and Cr exhibited the best oxidation resistance. Phase II base alloys were selected and various hardening agents discussed. NO DISTRIBUTION LIMITATIONS	UNCLASSIFIED 1. Alloys, titanium
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