

Feb. 26, 1974

L. A. ROSALES ET AL
THERMOMECHANICAL METHOD OF FORMING HIGH-STRENGTH
BETA-TITANIUM ALLOYS

3,794,528

Filed Aug. 17, 1972

4 Sheets-Sheet 1

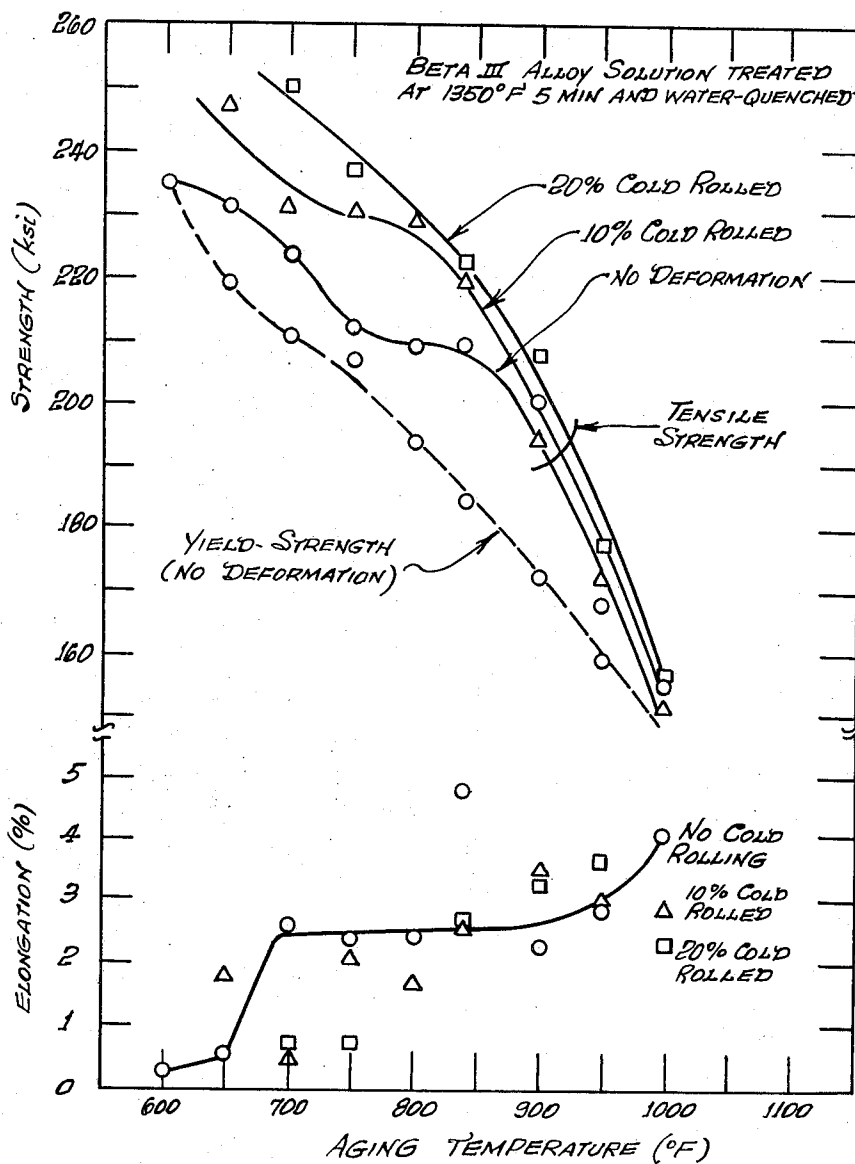


Fig. 1.

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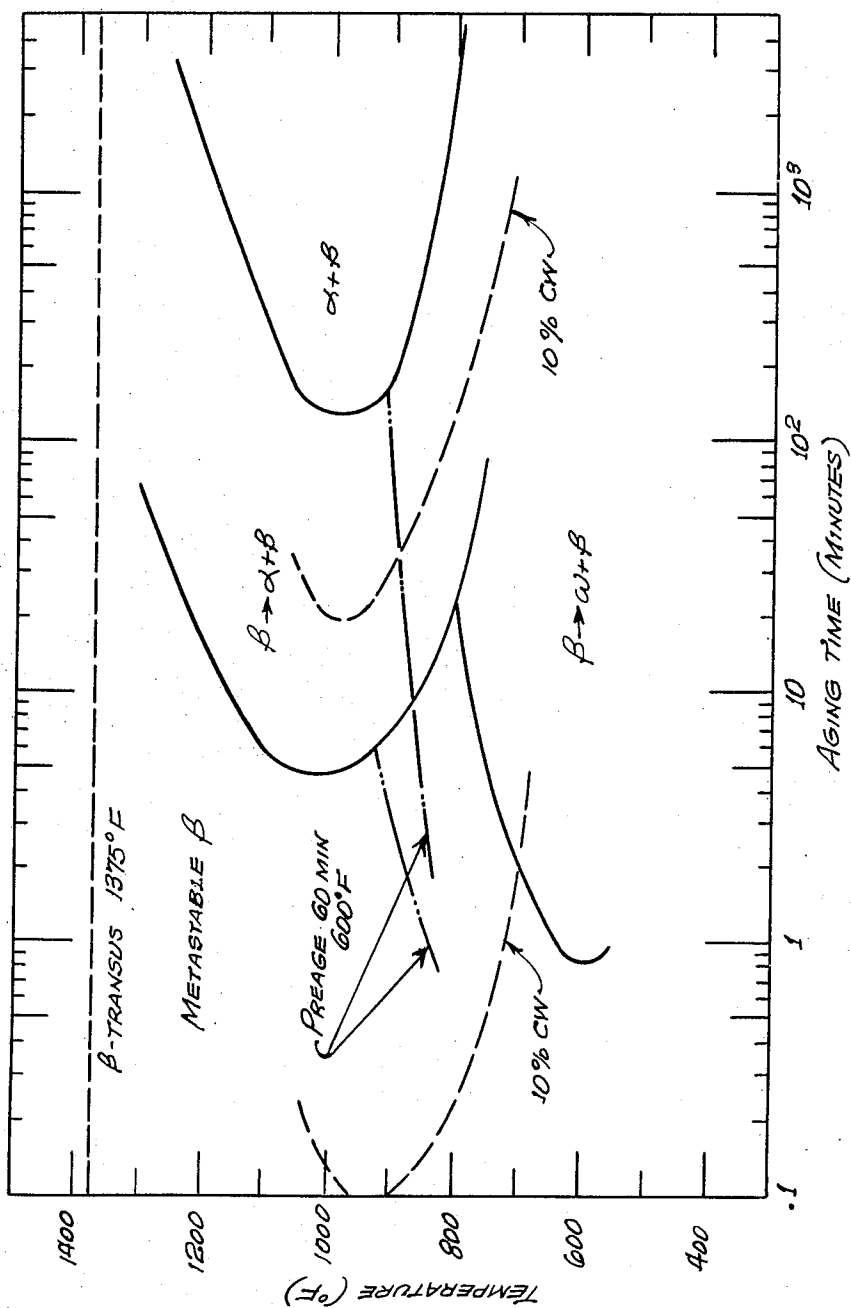


FIG. 2.

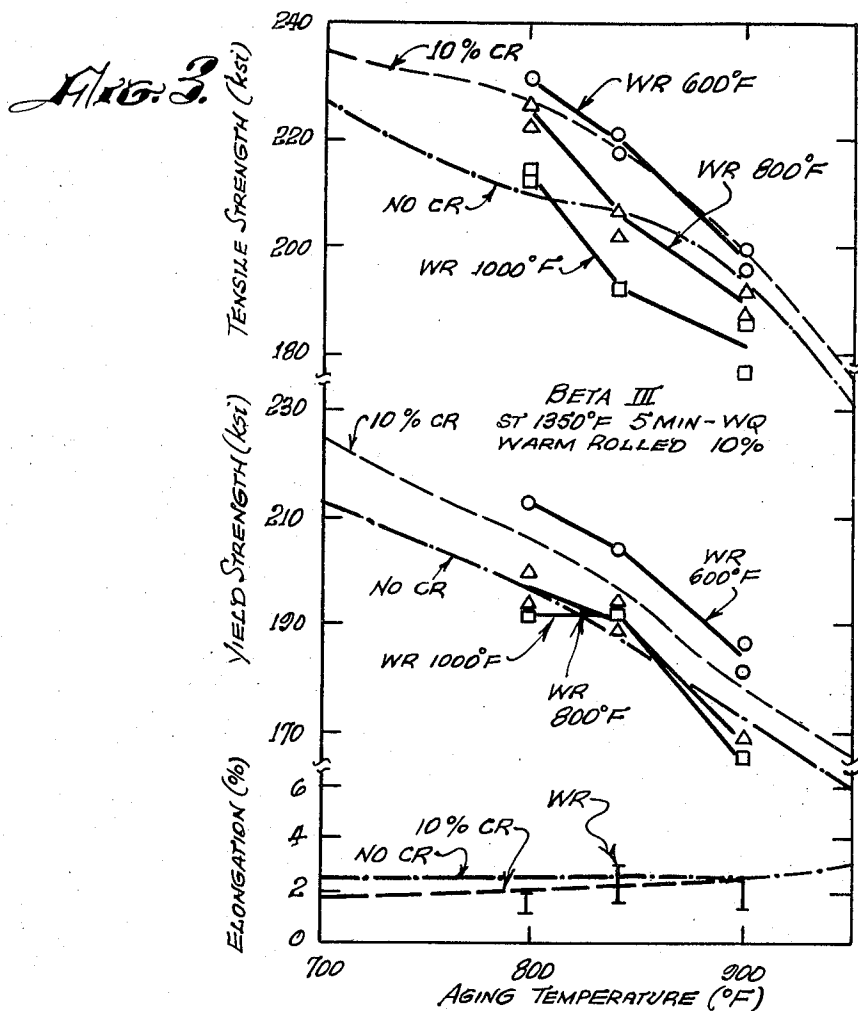
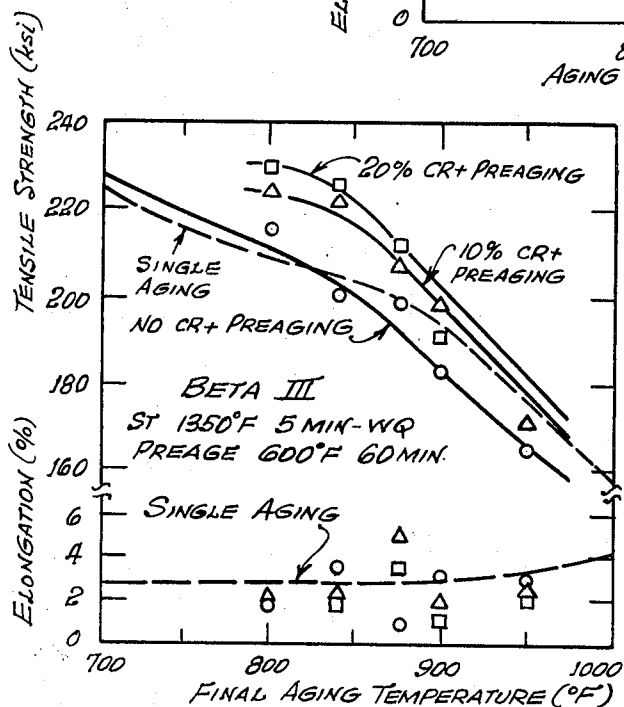
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Fig. 5.

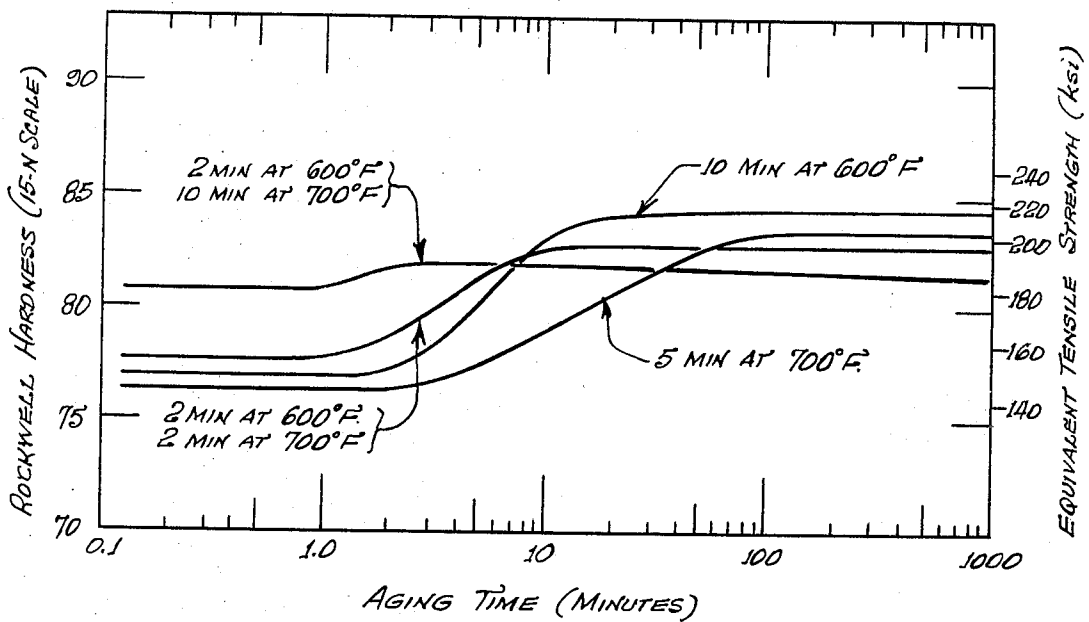
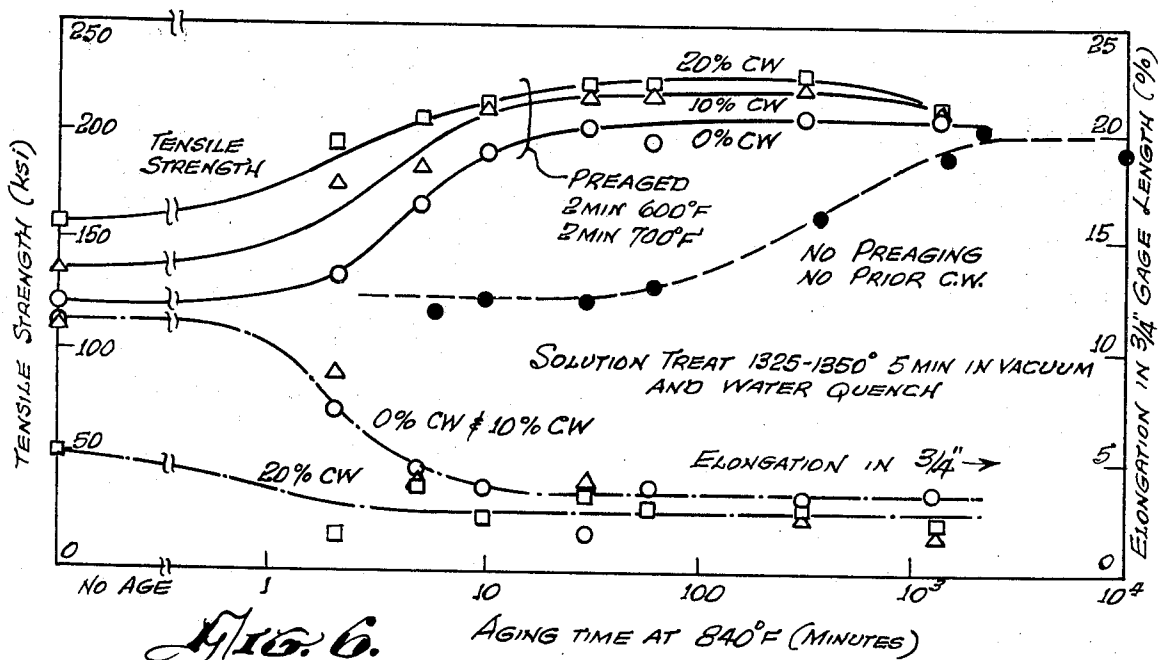


Fig. 6.

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THERMOMECHANICAL METHOD OF FORMING HIGH-STRENGTH BETA-TITANIUM ALLOYS

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U.S. Cl. 148—12.7

8 Claims

ABSTRACT OF THE DISCLOSURE

Beta-titanium alloys containing a relatively large frac-
tion of a beta-phase stabilizer are thermomechanically
treated to provide increased strength levels of greater than
220 k.s.i. with only a minimum loss of ductility, the treat-
ments involving heating the stock a point near its beta-
transus, rapid cooling, cold rolling and aging, the last
step including either a single final aging at a temperature
in the alpha-forming temperature range or a multi-step
aging process which includes pre-aging in the omega-form-
ing range prior to the final aging. Cold rolling plus final
aging improves strength and ductility, while cold-rolling
plus multi-step aging is very effective from a transforma-
tion time viewpoint, reducing the aging time needed to
obtain the optimum strength levels by factors of 10-100
over single aging.

BACKGROUND OF THE INVENTION

The present invention relates to beta-titanium alloys and,
particularly, to various thermomechanical treatments for
improving the strength and ductility of these alloys.

Titanium base alloys are extremely important structural
materials in the aerospace industries of the world where
high toughness and strength-to-weight ratios are of prime
concern. Increasing attention is being given to a class of
titanium alloys containing a large enough atom fraction of
beta-phase stabilizers so that, by rapidly cooling the metal
from the vicinity of its equilibrium beta-transus, a material
that is predominantly beta-phase can be produced. This
step of rapidly cooling from the vicinity of the beta-transus
generally is known as solution treatment and, following
such treatment, these alloys easily are formed into struc-
tural components at room temperature, the ease of forma-
tion resulting from the nature of the body-centered cubic
(BCC) structure of the beta-phased particles since such
structure can be deformed on numerous slip systems. Since
the beta-phase is metastable, subsequent aging treatments
produce microstructures exhibiting excellent tensile
strength at room temperatures and even higher strength
levels can be produced by aging these alloys at low tem-
peratures (250° F.-500° F.) to precipitate a large volume
fraction of uniformly distributed metastable, hexagonal,
closely-packed (HCP) omega-phase. However, such treat-
ments generally result in severe embrittlement which, of
course, severely restricts the utility of the treated material.
Alternatively, aging in the temperature range of alpha-
phase precipitation (850° F.-1000° F.) produces a stable
microstructure as contrasted with the metastable omega-
phase. However, even though these alpha-hardened micro-
structures exhibit good fracture toughness and thermal
stability, it has been found that, to avoid subsequent
omega-precipitation during service, extended aging times at
elevated temperatures are required. Such extended aging,
in turn, causes considerable coarsening of alpha-phase and
room temperature tensile strength levels of 160-180 k.s.i.
are common.

The present invention seeks to overcome these and other
difficulties by treating stock titanium alloys in a manner
that produces a thermally-stable, relatively high-strength

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alloy having a ductility rendering it acceptable for use for
a variety of purposes including its use as sheet metal for
air frames or rocket motor cases, plate and forged prod-
ucts, fastener stock, etc. In addition, the treatment ma-
terially reduces the aging times conventionally used in other
practices. These features, in general, are achieved by solu-
tion treating the stock alloy at a temperature in the
vicinity of but below the beta-transus of the alloy, rapid
cooling, cold-working prior to aging and then aging at
particular temperatures which will be defined in the en-
suing description.

BRIEF DESCRIPTION OF THE DRAWINGS

The invention is illustrated in the accompanying draw-
ings, FIGS. 1-6, which, as will be noted, are in the form
of plots or curves illustrating and comparing the effects
of certain thermomechanical treatments performed on a
typical alloy known as Beta-III. The nature of the draw-
ings will be clear from their legends and, of course, each
of the figures will be described in detail.

DETAILED DESCRIPTION OF THE INVENTION

The particular alloy used for the present study is avail-
able in sheet form from Crucible Materials Research Cen-
ter, Colt Industries, the alloy having a chemical analysis
as follows: 10.2 Mo, 7.4 Zr, 4.5 Sn, 0.11 Fe, 0.02 C, 0.012
N, 0.1 O, 0.0091 H. As has been indicated, Beta-III alloy
is typical of a series of beta-titanium alloys which have
been receiving increasing attention due to their toughness
and their strength-to-weight ratios. In general, these are
allotropic alloys characterized by the fact that a predom-
inance of the beta-phase can be retained by rapidly cooling
the metal from the vicinity of its equilibrium beta-transus
or, in other words, from the transitional temperature at
which a phase change occurs in the allotropic alloy. As
is well known, these alloys demonstrate three separate
phases known as the alpha, beta and omega phases the
omega phase being a mixture of the alpha and beta phases.
More specifically, the series of alloys which respond to the
present teachings are those which contain an atom frac-
tion of a beta-phase stabilizer sufficiently large to produce
upon quenching from its beta-phase transus a material
that contains the beta-phase particles as a major constituent
and, preferably, a constituent forming fifty percent or
more of the material. Other alloys similar to Beta-III which
have been tested in the manners to be described are a
Titanium Metals Corporation of America (TMCA) alloy
8823 and a Reactive Metal Co. (RMC) alloy 38644. These
tests have confirmed the results obtained with Beta-III as
a test specimen. Alloy 8823 has the following composi-
tion; 8 Mo, 8 V, 2 Fe, 3 Al and alloy 38644 has 4 Mo,
8 V, 6 Cr, 3 Al, 4 Zr.

As has been stated, previous treatments applied to these
alloys have resulted in an almost complete loss of ductility
when the aging is performed at relatively low temperatures
and a serious loss of strength when higher temperatures are
employed. Thus, as shown in FIG. 1 by the circle-con-
nected curves, the percentage elongation property of the
metal decreases very rapidly when the aging is accom-
plished at temperatures slightly under 700° F. For example,
aging at a temperature of 650° F. produces a high strength
level of about 230 k.s.i. but the high strength is achieved
at a great sacrifice of ductility. At higher temperatures,
such as 1000° F., percent elongation is retained and in
fact increased, but the tensile drops off at about 155 k.s.i.
At an aging temperature of 700° F., the resulting product
is quite attractive from a strength standpoint but, un-
fortunately the aging time to achieve such strength is in
the neighborhood of 300 hours, a factor that becomes very
significant when considered from an economic point of
view.

Before continuing, it should be noted that the 2% figure for elongation, as shown in FIG. 1 and other figures, represents an acceptable ductility, the reason being that since the present studies utilized relatively small specimens, the values of elongation are smaller by a factor of 2 to 3 compared to those obtained when larger sized test specimens are employed. Thus, in the present study, the specimens used were 0.75 inch gage. Aging of this small specimen at, for example, 900° F., produced an elongation of 2.2%. By way of comparison, a Beta-III specimen having a standard size of 2" gage showed an elongation of 7% when aged at 900° F. With further regard to the conditions employed in the present studies, the specimens used in deriving the data were machined from 0.02" thick sheets to produce a gage section 0.75" long and 0.88" wide. Tensile tests were performed using an Instron with a nominal strain rate of 4.44×10^{-4} sec.⁻¹ at room temperature. Transmission electron microscopy also was employed and, for this purpose, disc-shaped specimens first were trepanned from tensile specimens by way of electro-discharge machining, then ground to a thickness of .006", jet-polished to produce a "dish" and finally electro-polished in a solution of perchloric acid, methanol and n-butyl alcohol at 10 v. and at 40–90° C. A JEM-120 electron microscope operating at 120 kv. provided the results.

With exceptions that are specially noted, all specimens were subjected to a solution treatment at 1350° F. for 5 minutes, this treatment being carried out in an inert atmosphere or in a vacuum and the specimens subsequently were water quenched. This solution treatment is carried out below the beta-transus temperature of the Beta-III alloy, which, as shown in FIG. 2, is 1375° F. When the alloys are solution treated slightly above the beta-transus, the beta grain size increases rapidly resulting in brittleness for specimens aged at the lower alpha-forming range. For example, when Beta-III is solution treated at 1400° F., aging below 850° F. causes a total loss of ductility, a fact which should be contrasted with the ductility resulting when the same aging temperature is used for a specimen that is solution treated below the beta-transus point. Further, the grain size is over 10 microns in this case compared to that of about 0.8 micron for Beta-III solution treated at below the beta-transus point. As is known, the use of the inert atmosphere or vacuum in the solution treatment is a refinement which is not considered critical insofar as the results of the present study are concerned and it further should be noted that the rapid cooling of the alloy can be accomplished in manners other than by water-quenching.

The method of the present invention features a series of thermomechanical processing steps performed on a stock alloy or specimen solution treated in the manner which has been described, the steps resulting in the production of a metal material having higher strength, acceptable ductility and significant production economies due principally to the substantially reduced aging time needed to provide these desirable properties. Generally, the steps include a plastic deformation step performed on the quenched alloy by cold-rolling within a 5–20% reduction range, although other cold working using tension, compression or extrusion techniques can be substituted. Following the cold-working, the material is aged at a temperature preferably in the range of 700–850° F. for a period of time sufficient to develop the desired strength and ductility characteristics. Another featured processing step involves a pre-aging at a relatively low temperature which will be defined. Also, it has been found quite beneficial to utilize multi-step pre-aging techniques prior to the final aging. As will become apparent, the two separate steps of cold-working and pre-aging each has its independent advantages to the extent that, dependent upon the results that are to be obtained, each step advantageously can be employed independently of the other. Consequently, in the ensuing description, the re-

sults achieved by performing each of the steps independently first will be considered and then compared with the results achieved by combining the steps.

The results of cold-rolling on a solution treated Beta-III specimen are shown in FIG. 1 in which it will be seen that the triangle-connecting line or curve represents the results of the 10% cold-rolling plus aging at various temperatures, while the square-connecting line represents a comparable 20% cold-rolling. It will be apparent that either the 10% or 20% cold-rolling results in a material that is stronger than the single-aged material represented on FIG. 1 by the circle-connecting curves. Reduction ranging upwardly from about 5% appears beneficial, the precise amount of working being dependent upon the particular alloy being treated. The principal objective of the working appears to be the production of a permanent deformation in which dislocations similar to those to be described are demonstrated. Significant increases in tensile strength were found to occur when the aging temperature is in the range of 700°–850° F. Treatments of 10% cold-rolling and aging between 750° and 800° appeared to be optimum as the tensile strength increases by 20 k.s.i. whereas elongation remains about 2%. Although strength increases were greater at lower temperatures, the ductility decreases to an unacceptably low value or, as related to the present studies, to a value below 1%.

With further regard to the temperature range suitable for aging, it can be noted in FIG. 1 that a 20% cold-roll material aged at 700° F. has an unacceptable elongation of around .5%. The elongation percent curve for the 20% cold-roll material apparently has a rather sharp bend extending from about 750° F.–850° F., the strength at 850° F. being around 230 k.s.i. Thus, an acceptable temperature range for single aging the 20% cold-roll material would extend upwardly from about 850° F.

The actual temperature range for the particular alloy being aged can be determined from the temperature-time-transition (TTT) diagram for that alloy. FIG. 2, for example, is a TTT diagram for Beta-III. These diagrams are in rather common use for the purpose of showing the hardening reactions of the particular alloy or material being aged. The solid curves of the diagram define the start and finish of hardness changes during isothermal aging. Separate curves are shown for the separate phase reactions occurring as the specimen is aged for varying times at varying temperatures. The dashed curve at the left of the diagram indicates the effects of 10% cold-work, while the dash-dotted lines show the effects of pre-aging for 60 minutes at 600° F. All of the data shown in FIG. 2 was obtained using specimens which were water-quenched from 1350° F., the beta-transus of the Beta-III alloy being 1375° F.

For reasons which will become apparent, beneficial results insofar as strength and ductility are obtained when the particular material is aged at a temperature within the $\beta \rightarrow \alpha + \beta$ phase-reaction zone which, referring to FIG. 2, is the zone marked $\beta \rightarrow \alpha + \beta$ and lying within the solid curve which designates the start and finish of the hardness changes during aging. The $\alpha + \beta$ reaction zone for the 10% cold-worked Beta-III alloy is the area within the dashed curve of this particular figure. However, the present studies also have shown that the aging temperature should be within the lower portion of this dashed curve zone and, for this reason, the upper preferred limit of the temperature range suitable for single-aging purposes is a temperature below the nose of the dashed curve of the diagram. Comparing the TTT diagram of FIG. 2 with the data shown in FIG. 1 for the 10% cold-worked specimen, it will be seen that excellent strength and acceptable ductility for this cold-worked specimen is obtained when the aging is performed at a temperature range of about 700°–900° F. The dashed line curve of FIG. 2 shows that this range is a range lying within the $\alpha + \beta$ zone of the cold-worked material below the nose of the curve.

Although the discussion to this point has been con-

cerned with cold-working, this particular term is intended to include not only working at room temperature or below but, also working at temperatures which may extend up to 600° F. or above depending upon the particular alloy that is being treated. FIG. 3 shows the effects the so-called warm-rolling on the tensile properties of the Beta-III. To obtain this data, solution treated (1350° F.) Beta-III was reduced by 10% in thickness by rolling at temperatures of 600° F., 800° F. and 1000° F. The material then was aged for time sufficient to attain maximum hardness. Warm-working at 600° F. before aging increased the final tensile strength by about the same amount as 10% cold reduction prior to aging reaching a value of 227 k.s.i. for an 800° F. age. However, a small (approximately 3%-4%) increase in the value of the yield strength resulted for this warm-working temperature. Warm-rolling at 800° F. or 1000° F. prior to aging was ineffective for strength improvement. The ultimate strengths of warm-rolled specimens were found to be lower than those of single-aged materials for aging temperatures higher than 800° F.-850° F. The value for a 900° F. aging was about 180 k.s.i. for a 1000° F. rolling temperature and 188 k.s.i. for an 800° F. rolling temperature. This is below the strength level of 198 k.s.i. attained with cold-rolling or warm-rolling at 600° F. prior to aging at 900° F. Elongation data indicates that warm rolling has no beneficial effects on ductility relative to the cold-rolling and aging treatments. However, this data obviously may vary for other specimens. As stated, it has been found that the plastic deformation of the material accomplished by cold or warm working should be at a sufficiently low temperature to produce a non-recoverable plastic deformation. An acceptable amount of work reduction is found within a range of 5-20%.

Another advantageous effect of prior cold or warm working is the acceleration of the precipitation of the alpha-phase. Referring again to the TTT diagram of FIG. 2, it will be seen that the reaction start time is shortened by 30 times or more. The minimum temperature for direct alpha precipitation has been lowered from 800° F. to 700° F. As also is apparent in FIG. 2, the omega-reaction is not affected by cold-rolling.

Beneficial results also are obtained by multi-step aging treatments in which solution-treated alloys (1350° F., 5 minutes, water-quenched) are aged once or twice at the omega-forming temperature range. With regard to the particular specimen of the Beta-III metal, the TTT diagram of FIG. 2 shows the temperatures for these particular omega and alpha forming zones or ranges. In general it has been found that such pre-aging at the omega-forming temperature range produces, in a much shorter aging time, tensile strength levels comparable to those that can be obtained in single aging. FIG. 4 illustrates results that have been obtained for 10 and 20% cold-rolling plus pre-aging treatments, this figure also comparing these results with single aging as shown by the dotted line and another curve marked "no CR plus pre-aging" which, as will be apparent, indicates the fact that this curve represents data applicable to treatments in which the specimen material was not cold-worked but was subjected to pre-aging. The circles of FIG. 4 show that the maximum strengths reached for Beta-III when given a 60 minute pre-age at 600° F. and then aged at temperatures ranging from 800° F.-950° F. This data can be compared to the data obtained for single-aging (dotted line). Elongation data of single and multiple step aging materials are comparable. However, the significant consideration is that, to reach the maximum strength levels, multi-step aging required 10-100 times shorter aging times, a factor which provides obvious advantages at least from an economic viewpoint. FIG. 5 shows the hardness and equivalent tensile strength as a function of aging times for Beta-III given for various multi-step aging processes. In the FIG. 5 data, the final aging temperature was 840° F. in all cases. The maximum strength and time to attain the

strength were functions of the pre-age sequence. Three-step aging (2 minutes at 600° F., 10 minutes at 700° F. and a final age at 840° F.) produced a full strength condition after only two minutes of final aging at 840° F. compared to 2000 minutes required in a single aging at 840° F. Step-aging in ascending temperatures is more efficient in shortening the hardening time than a single pre-age temperature. Thus, for example a three step aging of two minutes at 600° F., two minutes at 700° F., and final age at 840° F. results in full strength in 15 minutes at 840° F. while pre-aging at 700° F. for five minutes results in full strength after about 200 minutes at 840° F.

As will be appreciated, the particular data which has been discussed with reference to FIGS. 4 and 5 relates only to a comparison of single step aging with multi-step aging so that, to this point, the effects of a thermochemical treatment including both cold-rolling and multi-step aging have not been considered or compared with other treatments. The data obtained for thermomechanical including both cold-rolling and pre-aging is shown in FIG. 4 by the square and triangle connecting lines. When cold-rolled Beta-III is pre-aged to produce omega-phase and finally aged at 840° F., the tensile strength is raised to a level equal to or higher than that obtained with cold-work plus aging. Also, this beneficial result is achieved with virtually no change in ductility in comparison to single aged Beta-III. However, as shown in FIG. 5, the kinetics of the hardening reaction during final aging were substantially accelerated to the extent that the end of hardening reaction can be reached in less than half of the time required for the cold work plus single aging treatment. FIG. 5 more specifically shows the effects of pre-aging in the $\beta+\omega$ region on the tensile strength and elongation of Beta-III aged in a lower temperature range of the $\alpha+\beta$ region (FIG. 2). The most significant aspect of the aging response compared to that of direct aging at the same temperature is the drastic acceleration of the hardening reaction. The maximum strength level of about 205 k.s.i. is reached within 20 minutes following the omega pre-aging rather than 2000 minutes required by direct aging. In the aged conditions, the ductility remains moderate. The pre-aging effects are observed in a narrow temperature range for the final aging treatment. When the pre-aging of 60 minutes at 600° F. was used, beneficial changes were found between 800 and 920° F. Start and finish of the hardening reaction during the final aging are shown by dash-dotted lines in FIG. 2. The upper bound corresponds to the reversion temperature of the omega-phase. The accelerated strengthening appears to be initially due to precipitation of omega-phase, while extended aging produces fine alpha-needles which contribute to the good tensile properties apparent when the material is subjected to omega-pre-aging plus final aging.

DISCUSSION

The data which has been presented in the foregoing description results principally from mechanical testing of the thermomechanically treated material. In addition, the studies have included extensive transmission electron microscopy to study the microstructure produced by the thermomechanical treatments and to ascertain to the maximum extent possible the fundamental reasons for the results which have obtained and are obtainable. A description and discussion of these microstructure studies now is presented to provide a clearer understanding of the basic principles upon which the present findings are premised.

Single aging of solution treated Beta-III at temperatures above 800° F. produces a microstructure consisting of needles of alpha-phase in a beta-matrix. These alpha-needles are oriented along a shear axis of $\langle 112 \rangle$ and sometimes result in a so-called basketweave structure because of all the possible variants. After aging at 1000° F., a typical diameter of these alpha-needles is in the order of 1000 angstroms and the length is in the submicron

range. The alpha-needles become finer with decreasing aging temperature and greater in number. For example, in the microstructure of Beta-III aged at 800° F., many alpha-needles are very fine and have a diameter of 100–400 angstroms and a length of 1000–6000 angstroms. Some ellipsoidal particles also are found. As a result of the fineness and small inter particle spacing of these precipitants, their ability to act as barriers to slip becomes more effective. Therefore, the observed shape of the strength-aging temperature curves (FIGS. 1, 3 and 4) is expected. Equiaxed alpha-phase also is present when the solution treatment is carried out below the beta-transus.

As will be appreciated, the present studies have been directed primarily at methods of controlling the microstructure of the metastable beta-alloys. The primary goal has been to produce a fine homogeneous distribution of the alpha-phase in the beta-matrix. The treatments employed to produce such a distribution have included cold-rolling, pre-aging and a combination of these two separate treatments. Cold-rolling of solution treated Beta-III produces a uniformly-distributed high density of dislocation. Based upon the result of a microstructure studies, twinning rarely was observed and the shape of the grains was not affected by 10% cold-rolling in the fine grain materials used in this study. Obviously, other materials having different grain characteristics will require variations in the extent of the cold-working although cold-working. Unfortunately, the density of the dislocation accomplished by cold-rolling cannot be determined with certainty since, in the electron microscopy studies, the dislocations were in contrast and the deformed grains appear almost completely black. However, it can be shown that the dislocations lie within a range of 10^{-11} to 10^{-12} cm.²

are preferentially precipitated on dislocations. When the alpha-forming range, the result in structure in a finer dispersion of alpha-phase in a beta-matrix then that found in the unworked, single-aged Beta-III. Again considering the microstructure of Beta-III cold-rolled 10% and aged at 800° F., extremely fine (50–150 angstroms) ellipsoidal alpha-particles are produced. These alpha-particles can be compared with the alpha-needles produced by single aged solution treatment of Beta-III at 800° F. In the cold-rolled Beta-III aged at 800° F., no evidence of omega-phase can be detected by electron diffraction experiment. The alpha-particles are distributed throughout the beta-matrix but tend to align themselves along $\langle 112 \rangle$ beta. The studies further indicate that the alpha-particles are preferentially precipitated on dislocations. When the final aging temperature is above 900° F., alpha-needles rather than alpha-ellipsoids are observed and the degree of refinement becomes much less significant. For example, when the Beta-III is cold-rolled 10% and aged at 1000° C., the typical size of alpha-needles is 500–1000 angstroms in diameter and 0.1–0.6 micron in length.

The foregoing microstructural observations can account for the improved strength properties of Beta-III aged at the 750–900° F. range and for the acceleration of the $\beta \rightarrow \alpha + \beta$ transformation. Apparently, these properties arise from the ease of alpha-phase nucleation on the dislocation structure and produced by cold-work.

Warm-working at temperatures of 800° or 1000° F. prior to aging also altered the alpha-beta microstructure but rolling at 600° F. produced essentially the same effects as cold-working at room temperature. Apparently, little advantage can be gained in deforming the material at the higher temperatures. Warm-working at 600° F. resulted in a small improvement in tensile properties, notably in the yield strength, but it is doubtful whether the added complexity of equipment and control required for this treatment is warranted. Transformation times are accelerated by warm-working but to a decreasing extent with higher deformation temperatures. This is expected since the studies show a growth in the equiaxed primary alpha grains at the higher working temperatures.

As has been demonstrated, multi-step aging with the initial aging steps performed in the omega-forming range followed by aging in the alpha-forming range is quite effective in reducing the times needed to reach full strength. Depending on the pre-aging treatment, the strength levels of the alloys can be increased over those of single aged materials. The omega-phase precipitates in the beta-matrix as uniformly distributed, coherent particles which, for the aging treatments and materials used in this study, were small (50–100 angstroms) ellipsoids. Subsequent aging in the alpha-phase temperatures below the reversion temperatures of the omega-phase produces the precipitation of the alpha-phase at or near omega-particles. As a result, the final microstructures are a complex mixture of the fine alpha in the beta-matrix.

When the specimens are cold-worked and pre-aged before final aging in the alpha plus beta region, the microstructure becomes even more complex. For example, in the microstructure of Beta-III solution treated at 1350° F., cold-rolled 10%, pre-aged 60 minutes at 600° F. and finally aged at 1380 minutes at 840° F., numerous very fine alpha-needles (50–150 angstrom diameter, 300–500 angstroms long) were distributed in a more uniform manner than was found to be the case when the Beta-III specimen was thermomechanically treated in an identical manner except for the lack of cold-rolling. Excellent strength of the cold-rolled and pre-aged alloy undoubtedly arises from a high density of defects and precipitates phases. As is apparent, by comparing FIGS. 1 and 4, the strength levels are equivalent for cold-reduced-aged and cold-reduced multi-step aged materials. However the transformation times for the latter are reduced by a factor of 2–4 in comparison with the former treatments. Consequently, significant economies in processing are realized.

Obviously many modifications and variations of the present invention are possible in the light of the above teachings. It is therefore to be understood that within the scope of the appended claims the invention may be practiced otherwise than as specifically described.

What is claimed is:

1. A method of thermomechanically treating stock titanium alloys to obtain a desired strength and ductility, said stock alloys being selected from a group having a sufficiently large atom fraction of a beta-phase stabilizer to produce upon solution treatment at a fixed temperature slightly below its beta-phase transus an intermediate material having a beta-phase as a major constituent, said method comprising:

heating said stock alloy to said fixed temperature and rapidly cooling it to form said intermediate material, working said intermediate material at a sufficiently low temperature to provide as a result of said working at least a 5 percent plastic deformation, and finally isothermally aging said deformed material at a temperature within the $\beta \rightarrow \alpha + \beta$ phase reaction zone below the nose of the $\beta \rightarrow \alpha + \beta$ curve of the TTT diagram of the deformed material, said aging being conducted for a period of time sufficient to develop said desired strength and ductility.

2. The method of claim 1 wherein said deformation is performed within a temperature range extending from room temperature to about 600° F.

3. The method of claim 1 wherein said stock titanium alloy has a chemical composition including about 11% molybdenum, 6 zirconium and 4.5 tin (percents by weight), and said aging is performed within a temperature range 700°–900° F.

4. The method of claim 1 further including the step of pre-aging said work-deformed material in the omega-forming temperature zone of its TTT diagram.

5. The method of claim 4 wherein said pre-aging is multi-stepped to the extent that the deformed material first is pre-aged at one temperature and then pre-aged at a higher temperature within said omega-forming temperature zone.

6. The method of claim 4 wherein said stock alloy has a chemical composition including about 11% molybdenum, 6% zirconium and 4.5 percent tin (percents by weight), and wherein said pre-aging is performed at a temperature of about 600° F. for about 60 minutes.

7. The method of claim 6 wherein said pre-aging is multi-stepped and includes a first pre-aging at about 600° F. for about two minutes and a second pre-aging at about 700° F. for about two minutes.

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8. The method of claim 7 wherein said final aging is performed at about 840° F.

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