Feb. 9, 1965 J. R. NEWMAN 3,169,085

METHOD OF PRODUCING TITANIUM BASE STRIP

Original Filed March 19, 1958 2 Sheets-Sheet 1

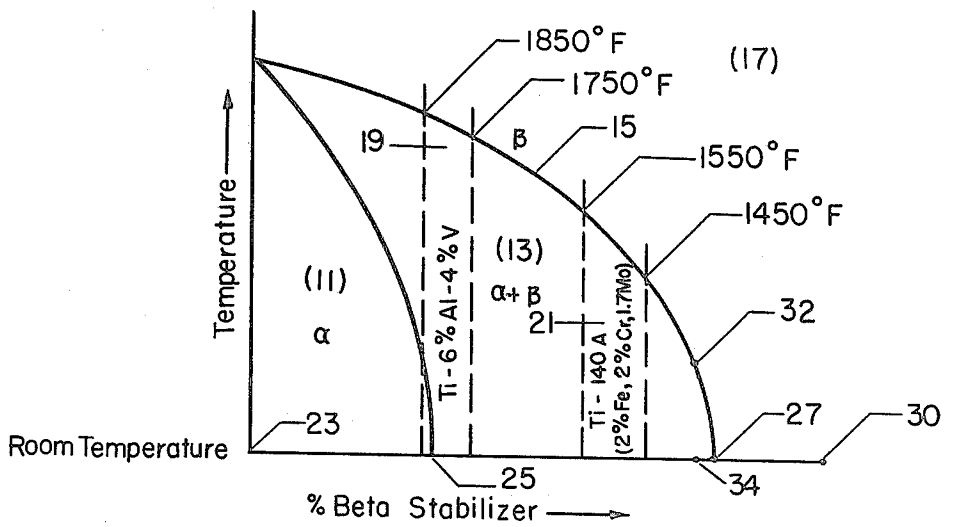

FIG. 1

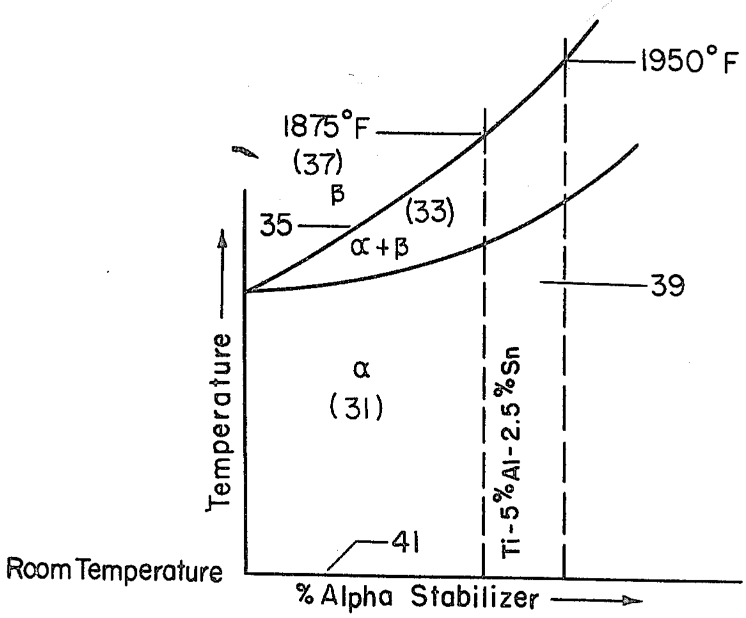

FIG. 2

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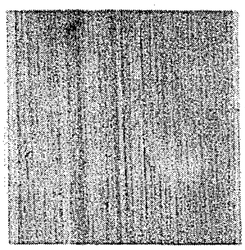

FIG. 3

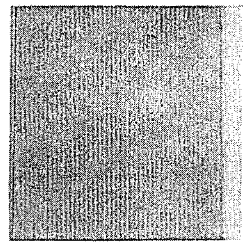

FIG. 4

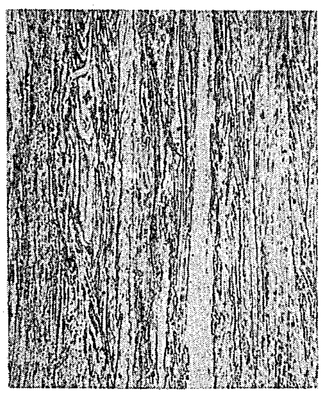

FIG. 5

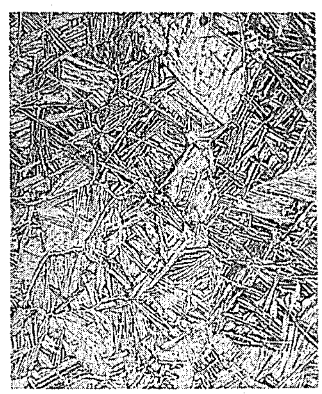

FIG. 6

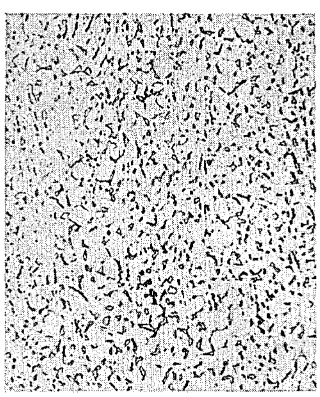

FIG. 7

FIG. 8

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3,169,085
METHOD OF PRODUCING TITANIUM BASE STRIP

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Continuation of application Ser. No. 722,543, Mar. 19, 1958. This application Feb. 20, 1963, Ser. No. 271,830

5 Claims. (Cl. 148—11.5)

This is a continuation of application Serial No. 722,543, filed March 19, 1958, now abandoned.

This invention relates to an improved method of producing flat rolled titanium and titanium alloy products and relates in particular to the production of cold rolled titanium and titanium alloy strip.

In the production of commercially available metals the most economical and advantageous manufacturing means of securing flat cold rolled products is that of producing such metal in strip form. Strip metal may be continuously annealed, pickled and cold rolled. It also may be coiled and thus relatively large pieces may be far more easily processed and handled than metal in sheet form. Most metals produced commercially, such as steel, lend themselves to the manufacture of strip. However, it has not been considered practical to produce titanium alloy in strip form.

Titanium and titanium base alloys are highly susceptible to the directional effects of continuous reduction in one direction. The manufacture of strip metal demands that all rolling or metal reduction be done in one direction. Although directionality may be substantially reduced in cold roll strip processing of commercially pure titanium, titanium alloy strip produced in a substantially conventional manner displays considerable physical anisotropy. This anisotropy manifests itself primarily as differences in mechanical properties and elastic moduli as measured parallel (longitudinal) and perpendicular (transverse) to the rolling direction. Such directional properties are highly undesirable in flat rolled metal products. Mechanical properties are inconsistent in that they vary with the direction in which they are measured. Also, a possible result of this anisotropy is visible striations usually referred to as "ribbing" which appear in the stretch-forming of commercially pure grades, and which sometimes appear in the course of mill processing. Even when directionality appears to have been minimized by conventional annealing and cold rolling practices, such as may be the case for commercially pure titanium, ribbing may occur. Strip materials exhibiting these defects are unattractive and affect the commercial value of the product.

The difficulties created by unidirectional rolling of titanium can be overcome by cross rolling sheets during hot rolling. Cross rolling titanium and titanium alloy sheets reduces the amount of crystallographic orientation which avoids directional properties. Cross rolling on hand sheet mills is the present conventional method of producing titanium alloy sheets. Conventional strip production methods require continuous rolling in one direction, and such material inherently possesses an oriented crystallographic structure in the direction of rolling that is responsible for the undesirable directional properties.

Cross rolling is more expensive and time consuming than strip processing since the sheets are rolled individually or in small packs, and generally this practice does not produce surface quality or gauge control as satisfactory as strip processing. The development of a process in accordance with this invention to produce isotropic titanium and its alloys by continuous strip methods therefore represents a significant improvement in the production of flat rolled titanium and titanium alloy products.

It has now been found that by employing conventional rolling practices in conjunction with the heat treatment of the present invention, titanium and titanium alloy in the

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form of continuous strip may be produced that exhibits little or no anisotropy.

In general, the present invention relates to the interposing of a step into the process of cold rolling titanium and titanium alloy strip, whereby the titanium strip, after hot rolling, is heated to a temperature near or above the beta transus temperature of the alloy being treated. The preferred embodiment of the present invention is to heat treat titanium, alpha structure titanium alloy and alpha-beta structure titanium alloy strip to a temperature near the beta transus temperature of the alloy involved. In carrying out the method of the present invention it is preferable to initiate such heat treatment immediately after hot rolling and prior to cold rolling.

It is, therefore, the object of the present invention to provide a method of cold rolling titanium and titanium alloy strip without producing the undesirable properties of directionality.

It is also an object of the present invention to provide a method whereby cold rolled titanium and titanium alloy may be produced in strip form with properties corresponding to cross rolled titanium and titanium alloy sheet.

A still further object of the present invention is to provide a process including the step of heating titanium and titanium alloy strip to a temperature near or above the beta transus temperature of said metal after hot rolling the strip to thereby produce strip substantially free from the effects of directionality.

Other objects and advantageous features will be obvious from the following description when taken in conjunction with the accompanying photomicrographs and drawings in which:

FIGURE 1 is an illustrative schematic phase diagram of a typical titanium alloy that contains alloying ingredients considered to be beta promoters showing the beta transus temperatures of two typical alloys that contain beta promoters;

FIG. 2 is a schematic diagram that illustrates a typical titanium alloy that contains alpha stabilizers only;

FIG. 3 is a photograph of the surface of Ti 140A cold rolled alloy strip processed in the conventional manner and showing a typical ribbing condition;

FIG. 4 is a photograph of the surface of Ti 140A cold rolled alloy strip processed in accordance with the present invention and showing a defect-free surface;

FIG. 5 is a reproduction of a photomicrograph taken at 500 diameters of Ti–6% Al–4% V alloy strip in the hot rolled condition;

FIG. 6 is a reproduction of a photomicrograph taken at 500 diameters of the alloy of FIG. 5 after being annealed above the beta transus of the metal;

FIG. 7 is a reproduction of a photomicrograph taken at 500 diameters of the material of FIGS. 5 and 6 after being annealed above the beta transus, cold rolled and annealed, and

FIG. 8 is a reproduction of a photomicrograph taken at 500 diameters of a Ti–6% Al–4% V alloy sheet that was cross rolled and annealed in the conventional manner.

It is generally accepted in the metals industry that one should not heat treat titanium and titanium base alloys at temperatures above or near the all-beta range because such a procedure results in a sharp loss of ductility, bendability and generally effects embrittlement of the metal. Modern publications on the subject of titanium and titanium alloys have warned against such heat treatment. However, I have found that such a heat treatment effects an equiaxed structure in the metal which wipes out the crystallographic orientation produced during hot rolling. If such a heat treatment is given under controlled conditions, ductility may be substantially retained or may be re-attained by proper subsequent cold rolling and annealing.

There are three basic stages in the conventional continuous strip practice:

(1) Hot rolling a thick slab to a coiled band.

(2) Annealing to produce mechanical properties and a microstructure amenable to cold rolling.

(3) Cold rolling with annealing cycles between each cold roll.

The hot rolling of titanium and its alloys has been described in general terms by most titanium producers. Hot strip rolling is usually conducted in the alpha or alpha plus beta field. This is followed by an anneal below the beta transus temperature, sometimes a time as short as three minutes being satisfactory but sometimes requiring up to eight hours at the annealing temperature. Cold reductions range from about 5% to as high as 50% per cycle, with intermediate anneals similar to those used in softening the hot rolled product. As previously revealed, this practice results in various degrees of directionality in titanium and its alloys.

In accordance with the present invention, heat treatment at, near or above the beta transus before cold rolling or between the cold rolling steps of any titanium or titanium base alloy that exhibits a room temperature alpha or alpha-beta microstructure will permit subsequent annealing and cold rolling of these materials to produce a product with the desired non-oriented structure similar to that obtained by cross rolling sheets of titanium. Such product does not exhibit anisotropy or directionality.

When alpha or alpha plus beta titanium alloy is subjected to beta heat treatment, on cooling the beta grains or crystals transform to alpha plus beta in a platelet-type array, the final proportions of alpha and beta depending on the specific alloy composition and cooling rate. Time at temperature and cooling rate from the all beta field are relatively immaterial with respect to anisotropy, affecting mainly the size of the beta grains and the size and spacing of the transformed platelets.

Broadly speaking, the beta stabilizers are Mn, Mo, Cr, Fe, V, W, Cb, Ta, Ni, Co, Si, Be, Zr and Cu. Within this broad category only certain of the elements mentioned are presently suitable for producing mixed phase alpha beta alloys. These are the elements which form with titanium, beta-isomorphous systems or beta-eutectoid systems in which the decomposition of the beta phase into eutectoid is so sluggish that the alloys behave like those in a beta-isomorphous system. The most practical beta stabilizing elements are V, Mo, Cb, Ta, Mn, Cr and Fe. Within this group Mo, V, Cb and Ta are true beta-isomorphous formers in titanium and Mn, Cr, and Fe are slow eutectoid formers in titanium.

The diagram shown by FIG. 1 is illustrative of a typical phase diagram of titanium alloyed with beta stabilizers. The temperatures and amounts of beta stabilizers made are indicated generally along the left hand side and bottom, respectively, of a diagram. Since the diagram is illustrative only, representing no actual alloy system, no actual temperatures or specific alloying additions are given along the ordinate and abscissa of FIG. 1, it being understood that the exact shape of the curve of any given diagram would depend on the actual alloying additions. However, accurate diagrams similar to FIG. 1 and showing actual temperatures and percent beta additions may be established for any given alloying system. Area 11 of the diagram of FIG. 1 represents those alloying concentrations and temperatures wherein the crystalline structure of the material is composed of an all-alpha phase. The area 13 represents the temperatures and concentrations wherein the alloys exhibit a mixed structure of alpha and beta. The line designated as 15 is the beta transus temperature where the crystalline structure of the alloy changes from a mixture of alpha and beta phases to a structure composed of all beta. The area above the line 15, designated as 17, is the all-beta field where the structure is composed entirely of the beta phase. Areas 19 and 21 represent the approximate ranges of beta stabilizers and critical temperatures of two specific alloys containing beta stabilizing materials such as Ti–6% Al–4% V and 2% Fe, 2% Cr, 1.7% Mo, Ti bal., respectively. As can be seen, Ti–6% Al–4% V has a beta transus temperature from about 1750° F. to 1850° F. and Ti 140A (2% Fe–2% Cr–1.7% Mo) shows a beta transus temperature of about 1450° F. to 1550° F. The temperature indicated for these alloys approximates the actual range of temperatures that encompass their beta transus temperatures but bear no actual relationship to the illustrative diagram of FIG. 1. It is necessary to estimate the beta transus as a range for a given type of alloy due to variations in chemical composition that are necessarily imposed by melting practices. A more accurate beta transus may be established for any given lot of a given alloy. Thus, the diagram of FIG. 1 represents the type of non-equilibrium phase structure obtained with the sluggish eutectoid formers such as Cr, Fe and Mn as well as the true beta-isomorphous elements.

The diagram shown by FIG. 2 represents a phase diagram showing the phases of a typical alloy that contains no beta stabilizers, but instead contains alpha stabilizing ingredients only. As in FIG. 1 the diagram is illustrative only and represents no given alloy system. Typical alpha stabilizing materials are aluminum, oxygen, nitrogen, carbon, tin, antimony and cerium. Tin and cerium are generally regarded to be substantially neutral elements. In this diagram, area 31 represents the all-alpha phase structure, area 33 the alpha plus beta phase structure, and area 37 the all-beta structure. Line 35 is the beta transus for this type of alloy. Area 39 illustrates a typical alpha stabilized titanium alloy showing the actual approximate beta transformation temperatures of such an alloy.

In a typical titanium alloy system there exists a series of analyses wherein the room temperature microstructure reveals an essentially all-alpha structure. This area is represented in the phase diagram of FIG. 1 as those compositions falling on the line between points 23 and 25 of area 11 and the entire room temperature line 41 of the diagram of FIG. 2. Such alloys are amenable to the process of the present invention in that they may be heat treated to near and above the beta transus temperature; thus directionality in the processing of strip may be eliminated by a beta heat treatment. Other alloys are of an analysis wherein they exhibit a mixed alpha plus beta structure at room temperature. These alloys are illustrated by the line between points 25 and 27 of FIG. 1. The alpha-beta room temperature alloys are also amenable to the process of the present invention because they too may be heat treated to and above the beta transus 15 during the strip processing. Alloys that contain additions of beta promoters and stabilizers in such amounts that they exhibit an all-beta structure at room temperature are represented on the diagram of FIG. 1 as those compositions theoretically existing between point 27 to and beyond point 30. The heat treatment of the present invention is not applicable to the all-beta structure alloys in that these alloys are already at or above the beta transus while at room temperature. Directionality may be a problem with the all-beta alloys as with the all-alpha or alpha-beta structure titanium alloy systems, but the solution of this problem in connection with these alloys may not be solved with a "beta anneal" such as has been found possible in the present invention. It may be possible to eliminate directionality in the all-beta alloys by means of a heat treatment during strip processing; however, such heat treatment may not be determined by means of the beta transus as in the process of the present invention.

Generally alloys of titanium and beta stabilizers are of the mixed-phase, alpha-beta or all-beta types, depending on the total amount of beta promoters present. Within the range of about 0.5% to 10% or 12% of beta promoters, the alloys are in general found to be of the mixed-phase, alpha-beta structure. With higher additions of beta promoters, the alloys are of the unstable or stable all-beta type, although the zone between the two is not definitely demarked.

The presence of alpha stabilizers alters the phase diagrams of such alloys in that these additions tend to raise the beta transus, thus requiring more beta stabilizing additions to effect an all-beta room temperature structure.

Titanium base alloys, during heat treatment in the alpha-beta field (area **13** of FIG. 1) exhibit greater or lesser amounts of one phase than the other depending on how close or far away they are from the beta transus. Such alloys at the more elevated temperatures near the beta transus temperature will exhibit a relatively high proportion of beta. The heat treatment of the present invention may effect the advantages sought when transformation to the beta is about 90% complete. Thus, in accordance with the present invention, near the beta transus is interpreted as encompassing the temperatures slightly below the beta transus wherein transformation to the beta is at least 90% complete.

Alpha-beta alloys that contain beta stabilizers in such amounts that their structure consists mainly of the beta phase at room temperature react slowly during heat treatment to the vicinity of their beta transus temperatures. The reason for such slow transformation is that heat treatment to and above the beta transus for these alloys involves low temperatures. For example, a binary alloy containing about 11% of a beta stabilizer, where additions of only 12% of such an alloying constituent would result in an all-beta room temperature structure, may require a heat treatment of only a few hundred degrees above room temperature before reaching the beta transus temperature. Such an alloy is illustrated in FIG. 1; point **34** represents the room temperature structure and point **32** the beta transus of such an alloy. As can be seen by the illustration, the beta transus temperature of such an alloy is much lower than the actual alloys depicted (Ti 140A and Ti–6%A1–4% V). To heat treat such an alloy to or near the beta transus may require an indefinite and unreasonable time to effect the destruction of directionality. Thus, it is preferred to apply the heat treatment of the present invention to alpha-beta alloys that exhibit a sufficiently high temperature beta transus to enable the destruction of directionality to take place near the beta transus within a reasonable time. It is preferred that the beta transus of such alloy be no lower than about 1000° F., because at this temperature beneficial effects will be realized within a reasonable time period. Also, in the process of the present invention it is preferred to confine the heat treatment to all-alpha alloys or alpha-beta alloys that contain about 80% or less of room temperature beta structure.

The preferred beta anneal of the present invention may be carried out on alpha or alpha-beta structure titanium of 80% or less room temperature beta at temperatures near or above the beta transus to the melting point of the metal involved, the exact temperature and time depending on the ultimate properties desired and handling techniques involved. Titanium held at temperatures far above the beta transus or that are held at near or above the beta transus for excessive amounts of time may be brittle when cooled to room temperature and exhibit large grains. Such brittleness may be alleviated to some extent by alpha or alpha-beta phase annealing or by furnace cooling from the beta anneal. However, the large grains may have an undesirable effect on subsequent mechanical properties of the strip. For example, for most of today's commercially available titanium, including the alpha stabilized alloys such as Ti 75A (commercially pure titanium that contains small amounts of C, O and N) and Ti–5% Al–2½% Sn, and the alpha-beta mixed structure alloys, such as Ti 140A (Ti–2% Fe–2% Cr–2% Mo), Ti–6% Al–4% V, Ti–4% Al–3%

Mo–1% V, Ti–3% Al–5% Cr, Ti–8% Mn and Ti–4% Al–4% Mn and including any titanium base alloy that contains as beta stabilizing additions up to a total of from about 10 to 15% of at least one beta stabilizing element selected from the group consisting of Mo, V, Cr, Fe. Mn, Cb and Ta and that contains 80% or less equilibrium beta structure at room temperature, it is preferred not to exceed about 50° F. above the beta transus of the metal or to conduct the beta anneal at temperatures below about 20° F. below the beta transus. It is generally undesirable to hold these metals at beta heat treatment temperatures for more than about 15 minutes. The exact upper limit of the above beta stabilizing additions depends upon the amount of alpha stabilizing elements present in the alloy; if the titanium base alloy is substantially free of alpha stabilizers and promoters the preferred upper limit is about 10%, but if there is present 1% or more of the alpha stabilizing elements then the preferred upper limit is about 15%. Usually merely heating these alloys to temperature and immediately cooling, such as quenching, will suffice to eliminate directional properties. It is to be understood, of course, that higher temperatures, i.e., up to the melting point of the metal involved, may be employed for this purpose and time is not a factor if subsequent treatment such as annealing, furnace cooling, cold rolling, etc., is so controlled and designed to effect an acceptable product.

Regardless of the analysis of the titanium base metal involved, for most applications of commercial grade strip, to avoid handling difficulties, unnecessary heat treatment, increased cold rolling steps and to obtain maximum physical and mechanical properties of the end product it is preferred that the beta anneal be controlled as to time and temperature to avoid grain growth of the beta structure to a grain size greater than about an ASTM grain size No. 1.

The adverse effects of anisotropy or directionality may be detected and observed in at least 3 different ways. First, observed ribbing may indicate the presence of this property during strip processing or when the final product is fabricated into desired articles. FIG. 3 shows ribbing on titanium alloy strip, Ti 140A, processed in the conventional manner of confining all anneals to the alpha-beta range, and FIG. 4 shows the surface of such material processed in accordance with the present invention. Second, the property may be observed by mechanically testing the rolled material during processing. The material that has excessive directionality exhibits variations in mechanical properties, particularly between the longitudinal (direction of rolling) tensile properties and the transverse (perpendicular to the direction of rolling) tensile properties. Thus, a difference in yield strength exceeding about 12,000 p.s.i. indicates the presence of directionality in most titanium alloys. A third method where directionality may be observed is through microexamination. A crystalline structure of the hot rolled strip taken longitudinally to the direction of rolling will always show a banded or directionally oriented grain structure such as is shown in FIG. 5. The alloy illustrated by FIG. 5 is a Ti–6% Al–4% V hot rolled strip material and shows a directional (banded) alpha-beta structure. FIG. 6 shows the alloy of FIG. 5 after heat treating at the beta transus. It is to be noted that the structure now consists of transformed beta (basket weave alpha-beta platelet structure) that is not banded. FIG. 7 shows the beta annealed structure obtained in FIG. 6 followed by cold rolling and conventional annealing and FIG. 8 shows the same alloy produced by hand mill sheet cross rolling practice. It should be noted that there is a definite similarity between the structure of FIGS. 7 and 8. The structure of FIGS. 7 and 8 consists of islands of beta in an alpha matrix. Thus it can be seen by microexamination that the method of the present invention permits the production of cold rolled strip while producing a structure substantially identical to that obtained by cross rolling sheet material.

The heat treatment above the beta transus to secure an equiaxed grained product is preferably conducted immediately after hot rolling and before cold rolling. A far better surface condition may be secured if all heavy scale and contamination created by a beta anneal is removed early in the processing cycle permitting more complete surface treatment between cold rolling and annealing cycles. However, advantageous features of the heat treatment may be secured by beta annealing between cold rolling operations.

Although, as described above, times at temperature is not a critical factor in the heat treatment of the present invention in reducing directionality, time at temperature does affect to some extent the ultimate mechanical properties of the finished strip. Directionality is reduced if the metal is heated to temperature and immediately cooled. In many instances such a treatment is preferred in that it avoids grain growth. Continuous annealing practices usually involve heating the strip to temperature for about 1 to 10 minutes. On the other hand it may be desirable to heat treat some of the alloys for much longer periods of time. In some instances it may be desirable to box anneal for periods as long as 100 hours.

Although the "beta anneal" of the present invention is preferably a separate anneal or heat treatment it may be incorporated as part of the hot rolling. If the final hot rolling is concluded at a temperature near or above the beta transus, the alloy will obviously cool from the beta field or near the beta field, and thus a reduction in directionality will be effected.

When employing the heat treatment of the present invention by finishing hot rolling near or above the beta transus temperature it is necessary to employ a subsequent anneal to effect the significant reduction in directionality desired. Such second heat treatment may be carried out at any temperature that is sufficiently high to soften the metal but may be as high as the temperature at which hot rolling was completed. It is unlikely heating such a metal to a temperature below about 800° F. would have any desired effect. Therefore, the preferred temperature for such second heat treatment is from about 800° F. to the finish hot rolling temperature. In repeating the hot rolling temperature or in exceeding near or above the beta transus one would, of course, be "beta annealing" in accordance with the present invention. The time of such heat treatment is not critical but practical considerations would dictate a preferred time similar to that of the "beta anneal" or for a time of from to temperature to about 100 hours.

A subsequent anneal in the alpha-beta field is frequently desirable to soften the as-beta heat treated metal unless the material has been cooled in the furnace or otherwise slowly cooled from the beta field.

In carrying out the process of the present invention a coil of alpha-beta structure titanium alloy was produced substantialy free of adverse directional properties. An ingot of titanium alloy, having the following approximate analysis, was put into production as strip:

	Front End of Coil, Percent	Back End of Coil, Percent
Fe	1.70	1.36
Cr	1.37	1.10
Mo	1.31	1.29
Ti	Bal.	Bal.

The major processing steps were as follows:

(*a*) *Hot strip roll.*—The ingot was hot rolled at a temperature slightly below 1500° F. The as-hot-rolled material was very brittle so that an accurate determination of directionality was unobtainable; however, the longitudinal properties are given in Table I below. Samples of the as-hot-rolled strip were given a short-time anneal at 1525° F. (approximately 50° F. below the betal transus) for 15 minutes and furnace cooled to soften the material and

enable transverse tests to be conducted to show directionality. These results are also shown in Table I below:

Table I

Direction* of Test	Heat Treatment	Yield Strength (p.s.i.)	Tensile Strength (p.s.i.)	Elong. in 2" (percent)
L	As-hot-rolled	139,420	160,820	5.0
L	1525° F	85,000	103,000	19.0
T	1525° F	114,000	120,000	11.0

*L stands for longitudianl testing or in the direction of rolling. T stands for transverse testing or in a direction perpendicular to the direction of rolling.

(*b*) *Beta anneal.*—The hot rolled coil was box annealed at about 1575° F. for from 3½ to 7 hours in an argon atmosphere and furnace cooled. The beta transus for this alloy had been previously determined to be about 1575° F. Strength properties were measured in three places across each end of the coil as shown in Table II below:

Table II

FRONT END

	Yield Strength (p.s.i.)	Tensile Strength (p.s.i.)	Elong. in 2" (percent)
1L	108,620	118,750	15.0
T	113,000	119,850	5.0
2L	109,280	117,500	14.0
T	103,260	120,660	11.0
3L	110,420	122,250	6.0
T	105,400	118,190	16.0
BACK END			
1L	111,880	117,440	16.0
T	108,980	123,280	6.0
2L	108,080	118,050	15.0
T	105,300	125,680	7.0
3L	104,020	117,380	13.0
T	108,020	117,600	7.0

(*c*) *Cold rolling.*—The strip was now side trimmed to 26 inches, preheated at from 200° F. to 300° F. and cold rolled on a United 4-high 56-inch reversing mill from a hot rolled gauge of about .117 inch to about .090 inch in about 5 passes.

It was intended that the strip should be heat treated at from about 1250° F. to 1275° F. for 5 or 10 minutes; however, temperature readings indicated that the annealing temperature exceeded 1300° F. Air cooling from this temperature probably resulted in an omega decomposition of beta which is very hard and brittle. Consequently the material was box annealed at 1150° F. to 1200° F. and furnace cooled as soon as it reached temperature to soften the material (to Rockwell C–27).

The second cold reduction was to about .068 inch gauge in about 6 passes. The strip was then continuously annealed at about 1180° F. to 1220° F. Hardness of the annealed material was Rockwell C26–27.

The third cold reduction was to about .053 inch in 8 passes. The strip was then continuously annealed at 1180° F. to 1220° F. for about 10 minutes.

The strip was, of course, pickled free of scale after each heat treatment.

After the third cold rolling the surface of the strip was ground to remove surface defects.

A fourth cold rolling reduced the gauge of the strip to .043 inch in 5 passes.

The material was then finally box annealed at 1200° F. to 1250° F. in an argon atmosphere for 4 to 7 hours and furnace cooled.

Twenty-three 100-inch sheets were cut from the treated strip. Table III shows the strength properties of these sheets. The sheets are numbered 1 to 23 from the front to the back of the coil.

Table III

Sheet Number	Gauge, inch	Yield Strength .2% off-set (p.s.i.)	Tensile Strength, p.s.i.	Elong. in 2″, percent	Bend Radius	Yield Strength Directionality, p.s.i.
1L	.070	96,200	119,240	13.0	3XT	—
T	.071	102,940	122,170	10.0	3XT	00
5L	.046	95,610	117,340	16.0	3XT	7,0
T	.047	103,200	121,960	13.0	3XT	00
10L	.046	96,890	117,930	16.0	3XT	7,0
T	.046	104,260	124,240	11.5	3XT	00
15L	.045	95,450	116,860	16.0	3XT	7,0
T	.046	105,540	122,840	15.0	3XT	00
20L	.043	94,040	116,850	16.0	3XT	10,0
T	.042	105,820	123,930	11.0	3XT	00
23L	.055	95,310	118,040	14.0	3XT	11,0
T	.055	103,940	121,640	10.0	3XT	00
						9,0

As may be observed by the actual processing of the titanium alloy strip as given hereinbefore and the test results obtained by the intermittent testing, the results of which are shown in Tables I to III, titanium and titanium alloy strip may be produced in accordance with the process of the present invention without adverse directional properties. Table I shows the directionality of the material after hot rolling. It may be observed that a 29,000 p.s.i. differential exists between the longitudinal and transverse yield strength of the softened material. Table II shows clearly that the "beta anneal" in this instance has reduced directionality to an acceptable figure. In 4 of the 6 sets of tests, the longitudinal yield strength was actually somewhat greater than the transverse. Table III shows the completed cold rolled annealed strip material possessing little directionality while exhibiting good ductility and bendability.

As further examples of the process of the present invention sheets of titanium alloy were produced by hot rolling in one direction only, thus simulating or reproducing the same effect as one would experience in attempting to roll strip material. These sheets were composed of an alloy commonly known as Ti 140A (Ti–2% Cr–2% Fe–1.7% Mo). This alloy had a beta transus of about 1515° F. The hot rolled sheets were produced by rolling sheet bar at temperatures of 1700° F. and 1300° F. and air cooling from the rolling temperatures. The sheet bars were reduced from approximately .250 inch gauge to approximately .127 inch gauge in 5 passes.

After descaling, physical properties were obtained as shown in Table IV:

Table IV

	.2% Offset Yield Strength, p.s.i.	Tensile Strength, p.s.i.	Elong. in 2″, percent	Directionality Difference	
				.2% Offset Yield Strength, p.s.i.	Tensile Strength, p.s.i.
1300° F. L	125,948	156,923	12.3	28,745	15,067
Rolling T	154,693	171,990	8.8		
1700° F. L	134,106	157,583	10.0	20,931	11,765
Rolling T	155,037	169,348	7.5		

Sheets rolled at 1300° F. were then heat treated at 1500° F. and 1600° F. which encompassed the beta transus temperature of the alloy being treated. The results of this heat treatment are shown in Table V.

Table V

Heat Treatment *	.2% Offset Yield Strength, p.s.i.	Tensile Strength, p.s.i.	El. in 2″, percent
1500° F., 2 hrs., W. Q. L	100,800	119,100	20.0
plus 1200° F., 24 hrs. T	101,800	121,500	15.0
1600° F., 2 hrs., F. C. L	93,140	113,700	12.0
1600° F., 2 hrs., F. C. T	97,520	115,200	11.0
1600° F., 2 hrs., W. Q. plus 1200° F., 2 hrs. L	117,600	133,000	6.0
1600° F., 2 hrs., W. Q. plus 1200° F., 2 hrs. T	125,600	129,100	7.0

* F. C. stands for furnace cooled and W. Q. for water quenched.

For the purpose of comparison, sheets rolled at 1300° F. and 1700° F. were also heat treated at 1200° F., 1300° F. and 1400° F. (1200° F., 1300° F. and 1400° F. being more than 20° below the transus temperature of this alloy with the 1700° F. hot rolling being above the transus temperature). The results of such comparative treatments are given in Table VI:

Table VI

Rolling Temp.	Heat Treatment	.2% Off-set Yield Strength, p.s.i.	Tensile Strength, p.s.i.	El. in 2″, Percent	Yield Strength Directionality p.s.i.
1300° F.	1300° F., ½ hr., W.Q. L	117,870	126,680	19.0	21,000
	1000° F., 24 hrs. T	139,270	140,510	18.0	21,000
1700° F.	1300° F., ½ hr., W.Q. L	117,350	124,150	17.0	2,000
	1000° F., 24 hrs. T	119,010	132,480	17.0	2,000
1300° F.	1400° F., ½ hr., W.Q. L	126,800	131,860	18.0	15,000
	1000° F., 24 hrs. T	142,530	143,320	17.0	15,000
1700° F.	1400° F., ½ hr., W.Q. L	122,400	125,980	15.0	1,000
	1000° F., 25 hrs. T	121,620	134,220	16.0	1,000
1700° F.	1200° F., 8 hrs., F.C. L	105,390	123,060	16.5	8,000
	Do. T	113,740	131,280	15.5	8,000
1700° F.	1400° F., 8 hrs., F.C. L	100,810	109,840	19.0	5,000
	Do. T	105,330	116,610	18.0	5,000

Again referring to Table IV given hereinbefore, directionality is shown to exceed 28,000 p.s.i. in yield strength for material as hot rolled and cooled from 1300° F., while material as hot rolled and cooled from 1700° F., well over the beta transus, exhibited a lower yield strength directionality of about 21,000 p.s.i. In Table V it is seen that the high directionality of the material rolled and cooled from 1300° F. is all but eliminated by heat treatment near the beta transus. Table VI shows the results of heat treating and solution treating and aging both the 1300° F. and the 1700° F. hot rolled material at temperatures well below the beta transus of the metal. It can be seen that very little benefit was obtained in the material rolled at 1300° F. but that directionality is substantially reduced for the material that was final hot rolled and cooled from 1700° F. or above the beta transus, even though the annealing temperature was not at about the transus temperature. Thus it is seen that the heat treatment of the present invention may be incorporated into the final hot rolling step.

Eight-inch wide strips of Ti–6% Al–4% V and Ti 140A (2% Fe–2% Cr–1.7% Mo) and commercially pure titanium (Ti 75A) of hot rolled bands were heat treated for about 10 minutes both above and below their beta transus temperatures. They were then annealed and cold rolled (in one direction only) in three cycles from a hot rolled gauge of about .140 inch to a cold rolled gauge of about .060 inch. Intermediate softening anneals were conducted at temperatures of about 1550° F. for the Ti–6%–4% V, 1250° F. for the Ti–2% Fe–2% Cr–1.7% Mo alloy and about 1400° F. for the Ti–75A. Yield strength directionality (longitudinal properties minus the transverse properties) was determined for both the hot rolled and annealed material and the cold rolled products. Results are shown in Table VII below:

Table VII

(a) Ti-6% Al-4% V, M 7948; BETA TRANSUS 1825/1850° F.

Anneal, ° F.	As Annealed		After 3 Cold Roll-Anneals	
	Bend Radius	Yield Strength Directionality Difference, p.s.i.	Ribbing	Directionality, p.s.i.
1350°, 15 hrs., FC	4.5×T	----------	Heavy	29,000
1400°, 21 hrs., FC	4.5×T	26,900	do	20,700
1550°, 8 hrs., FC	4.5×T	23,600	do	21,800
1775°, 6 hrs., FC	4.5×T	9,900	do	----------
1550°, 10 min., AC	4.5×T	38,300	do	30,500
1750°, 10 min., AC	4.5×T	29,900	do	27,700
1875°, 10 min., AC	6.0×T	8,100	None	6,700
1900°, 10 min., AC	6.7×T	8,800	do	5,800
1900°, 15 min., AC	6.7×T	2,100	do	5,200
1925°, 10 min., AC	6.7×T	1,250	do	1,050
(b) Ti-140A, M 7947; BETA TRANSUS 1525/1550° F.				
1250°, 6 hrs., FC	3.0×T	14,600	Heavy	17,500
1350°, 6 hrs., FC	3.0×T	16,800	do	17,500
1450°, 10 min., AC	3.0×T	4,400	do	17,500
1550°, 15 min., AC	3.0×T	4,400	None	2,700
1575°, 10 min., AC	4.0×T	----------	do	----------
1625°, 10 min., AC	7.5×T	7,500	do	10,000
(c) Ti-75A, M 8060; BETA TRANSUS 1770/1775° F.				
1375°, 10 min., AC	2.5×T	14,000	None	14,500
1750°, 10 min., AC	2.5×T	2,200	do	4,500

It is shown in Table VII that titanium alloy heated in the alpha or alpha-beta field (1350° F.–1750° F. for Ti–6% Al–4% V, 1250° F.–1450° F. for Ti 140A and 1375° F. for Ti 75A) exhibited high directionality both as heat treated and after three cold roll-anneal cycles. It is also shown that titanium alloy annealed in the beta field, i.e., 1875° F. for Ti–6% Al–4% V, 1550° F. for the Ti 140A and 1750° F. for Ti–75A, exhibits acceptable properties before and after three cold-roll plus convention anneal cycles.

Small panels of Ti–6% Al–4% V hot rolled strip were wired to stringers (coils of steel strip used to thread the processing strip mills) and run through the preheater and furnace of the anneal-pickle line. The preheater temperatures were varied from about 1850° F. to 2000° F. and the furnace temperature was maintained at about 1500° F. to 1525° F. The treated samples were in the preheater for from about 2 to 4 minutes and in the furnace for about 1 to 3 minutes. The beta transus of the test material was estimated to be about 1850° F. Results are shown in Table VIII below:

Table VIII

MECHANICAL PROPERTY RESULTS—PREHEATER TRANSFORMATION OF Ti-6% Al-4% V

Identification	Preheater Temp.		.2% Offset Yield Strength, p.s.i.	Tensile Strength, p.s.i.	Elongation, Percent	Transformed Grain Size	Contaminations, inch	Bend
M-7948	1,850° F	L	126,140	142,260	12.0	3-4	.001/.0015	OK.
	1,850° F	T	130,390	146,590	10.5			
	1,900° F	L	129,750	146,460	10.0	3	.001/.002	OK.
.150″	1,900° F	T	129,690	147,810	9.0			
	1,950° F	L	131,520	146,990	11.0	2-4	.0015/.0025	OK.
	1,950° F	T	130,110	146,410	8.5			
	2,000° F	L	134,280	147,370	11.0	1-3	.002/.0035	OK.
	2,000° F	T	130,630	146,600	7.0			
M-8210	1,850° F	L	127,270	141,060	12.0	2-4	.001/.002	OK.
	1,850° F	T	126,160	142,580	10.0			
	1,900° F	L	126,280	140,250	8.0	2-4	.001/.002	Cracked.
.160″	1,900° F	T	122,100	139,480	9.0			

As may be observed upon examination of Table VIII, short time continuous annealing of titanium alloy strip within the beta field results in low directional properties.

The eight-inch wide strips of commercially pure titanium (Ti-75A) and Ti-6% Al-4% V hot rolled bands mentioned previously were also processed to cold rolled strip (.140 inch gauge to .060 inch gauge) in 3 cold rolling cycles both in the conventional manner and by employing the improved method of the present invention. Table IX below gives the various heat treatments employed. Each cold rolling involved about a 25% reduction.

Table IX

ANNEALING CYCLES

Type	Conventional Hot Band Anneal	Improved Hot Band Anneal	Intermediary Anneal
Ti-75A	1375° to 1400° F. (C)*		1375° F.-1400° F. (C).
Ti-75A		1750° F., 10 min., A.C.	1375° F.-1400° F. (C).
Ti-6%Al-4%V	1550° F. (C)		1550° F. (C).
Ti-6%Al-4%V†		1850° F., 10 min, A.C.	1550° F. (C).

*(C) stands for a continuous anneal.
†This material was given a 1350° F. box anneal for 8 hours before cold rolling.

Table X below shows the lower directionality obtained by employing the method of the present invention as compared to the process now commercially employed for commercially pure titanium (Ti-75A) and alloys of titanium that contain beta stabilizing elements (Ti-6% Al-4% V):

Table X

COMPARISON OF CONVENTIONAL AND IMPROVED TITANIUM STRIP PROCESSES—TYPICAL MECHANICAL PROPERTIES.

Alloy System	Single Phase, All-Alpha		Alpha-Beta, Alpha Matrix	
Typical Grade / Beta Transus	Ti-75A 1725±25° F.		Ti-6%Al-4%V 1800+25° F.	
Process	Conventional	Improved	Conventional	Improved
T.S., Trans., p.s.i.	101,350	89,560	171,370	157,430
T.S., Long., p.s.i.	103,010	90,100	145,710	151,360
Y.S., Trans., p.s.i.	90,260	77,930	164,800	130,480
Y.S., Long., p.s.i.	79,430	74,000	130,490	126,650
Elong., Trans., %	23.0	22.5	7.0	12.5
Elong., Long., %	22.0	22.5	9.0	8.5
Elastic Modulus, Transverse, p.s.i.	18.0×10^6	16.8×10^6	15.5×10^6	15.9×10^6
Elastic Modulus, Longitudinal, p.s.i.	16.0×10^6	16.0×10^6	15.0×10^6	15.1×10^6

A coil of hot rolled band of Ti-6% Al-4% V that had been hot rolled at temperatures within the alpha-beta region (1700° F.-1400° F.) was beta annealed on an anneal-pickle line. The strip of the coil was continuously passed through the annealing furnace entering the preheater first and then passing through the annealing furnace itself. Conditions were as follows:

Heat	Preheater (beta anneal) ° F.	Furnace, ° F.	Line Speed, 6 f.p.m.	Gauge, inch
M-8210	1825	1400/1425	6	.150

Preheater temperature was actually oscillating from 1810-1870° F. for M-8210.

Microstructural samples from each end of the coil indicated complete transformation. Mechanical properties after the beta anneal were:

Table XI

Heat No.	Gauge, "	Location	Y.S. (p.s.i.)	T.S. (p.s.i.)	Elong., Percent	Bend Radius
M-8210	.150	Leading end-L.	131,180	136,060	8.0	1.000"-OK.
M-8210	.150	Leading end-T.	130,610	135,220	8.0	1.000"-OK.
M-8210	.137	Trailing end-L.	134,470	140,180	9.5	1.000"-OK.
M-8210	.137	Trailing end-T.	132,610	139,870	8.0	1.000"-OK.

Typical properties of the coil before the beta anneal were:

L	135,000	151,000	8.5%	4.5XT
T	163,000	171,000	8.0%	4.5XT

The above examples are given to illustrate the process of the present invention and in no way limit the invention to the exact processing steps set forth. For example, continuous anneals may be substituted for box anneals and box anneals may be substituted for continuous anneals. Conventional heat treatments such as are employed in softening the material and those used for hot rolling and for annealing between cold rolling cycles will vary in accordance to the exact alloy being processed and the ultimate properties desired, it being understood that these steps will be carried out by the conventional and commercially known practices to effect the desired results. It is also to be understood that to obtain both strong and ductile strip such as is presently in commercial demand, the all-beta heat treatment is preferably conducted at or as near the beta transus temperature as possible in that heat treatment at temperatures exceeding the beta transus to too great a degree will create a large grain structure which may be brittle and unmanageable and, treatment too far below the beta transus where less than 90% of the beta phase is present will not effect significant improvement in directional properties. It is therefore preferable that the heat treatment of the present invention be conducted at a temperature wherein the structure is composed of at least 90% beta phase and at a temperature of at least about 1000° F. It is also preferred to maintain the heat treatment at a temperature and a time wherein beta grain growth does not exceed a grain size of approximately ASTM No. 1.

I claim:

1. A process for manufacturing continuously rolled cold rolled strip of a metal selected from the group consisting of titanium, titanium-base alloys that have an all alpha room temperature structure, and titanium-base alloys that have a mixed alpha and beta room temperature structure, which comprises:

(*a*) hot rolling the said metal substantially in one direction so as to form a metal strip, subsequently

(*b*) heat treating said metal strip after hot rolling to a temperature of from near the beta transus of said metal to the melting point of said metal, and finally

(*c*) cold roll-softening anneal cycling said strip said softening anneal being at a temperature substantially below the heat treating temperature of step (*b*).

2. A process for manufacturing continuously rolled

cold rolled strip of a metal selected from the group consisting of titanium, titanium-base alloys that have an all alpha room temperature structure, and titanium-base alloys that have a mixed alpha and beta structure, the beta phase not exceeding 80% of the structure which comprises:

(a) hot rolling the said metal substantially in one direction so as to form a metal strip, subsequently

(b) heat treating said metal after hot rolling to a temperature of from near the beta transus of said metal to the melting point of said metal, and finally

(c) cold roll-softening anneal cycling said strip softening anneal being at a temperature substantially below the heat treating temperature step (b).

3. A process for manufacturing continuously rolled cold rolled strip of a titanium-base alloy that contains as beta stabilizing additions from 0% to about 15% of at least one titanium beta stabilizing addition selected from the group consisting of Mo, Cb, Ta, V, Fe, Cr and Mn, the beta phase not exceeding 80% of the room temperature structure which comprises:

(a) hot rolling the said metal substantially in one direction so as to form an elongated metal strip, subsequently

(b) heat treating said metal strip to a temperature of from about 20° F. below the beta transus of said alloy to about 50° F. above the betal transus of said alloy, and finally

(c) cold-roll-anneal cycling said metal strip so as to provide a finished product.

4. A process for manufacturing continuously rolled cold rolled strip of a metal selected from the group consisting of titanium, titanium-base alloys that have an all alpha room temperature structure and titanium-base alloys that have a mixed alpha and beta room temperature structure, the beta phase not exceeding 80% of the structure, which comprises:

(a) hot rolling said metal strip substantially in one direction at a temperature of from near the beta transus of said metal to the melting point of said metal, subsequently

(b) annealing said strip at a temperature between 800° F. and the beta transus of said metal, and finally

(c) cold-roll-anneal cycling said strip so as to form a finished product.

5. A process for manufacturing continuously rolled cold rolled strip of a titanium-base alloy that contains as beta stabilizing additions from 0% to about 10% of at least one titanium beta stabilizing addition selected from the group consisting of Mo, Cb, Ta, V, Fe, Cr, and Mn, the beta phase not exceeding 80% of the room temperature structure, which comprises:

(a) hot rolling said metal substantially in one direction so as to form an elongated strip and completing said hot rolling at a temperature of from about 20° F. below the beta transus of said alloy to about 50° F. above the beta transus of said alloy, subsequently

(b) annealing said hot rolled strip at a temperature between about 800° F. and the beta transus of said metal, and finally

(c) cold-roll-anneal cycling said metal strip so as to provide a finished product.

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DAVID L. RECK, *Primary Examiner*.

UNITED STATES PATENT OFFICE

CERTIFICATE OF CORRECTION

Patent No. 3,169,085

February 9, 1965

Jeremy R. Newman

It is hereby certified that error appears in the above numbered patent requiring correction and that the said Letters Patent should read as corrected below.

In the grant, lines 1 to 3, for "Jeremy R. Newman, of Wintersville, Ohio," read -- Jeremy R. Newman, of Wintersville, Ohio, assignor to Allegheny Ludlum Steel Corporation, of Brackenridge, Pennsylvania, a corporation of Pennsylvania, --; line 12, for "Jeremy R. Newman, his heirs" read -- Allegheny Ludlum Steel Corporation, its successors --; in the heading to the printed specification, line 4, for "Jeremy R. Newman, 232 Vires Drive, Wintersville, Ohio" read -- Jeremy R. Newman, Wintersville, Ohio, assignor to Allegheny Ludlum Steel Corporation, Brackenridge, Pa., a corporation of Pennsylvania --.

Signed and sealed this 13th day of July 1965.

(SEAL)
Attest:

ERNEST W. SWIDER
Attesting Officer

EDWARD J. BRENNER
Commissioner of Patents