HOT WORKING OF TITANIUM ALLOYS

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ABSTRACT

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Titanium, due to its high strength to weight ratio, especially at elevated temperatures and its other useful properties, has found increasing application in the aerospace industry. These same properties have however made the mechanical shaping of the metal difficult and, at least partial hot working a necessity. Four types of hot working are used to shape titanium alloys, forging, superplastic forming, extrusion, and rolling, each similar to conventional shaping of materials, but with its own problems and requirements due to the properties of titanium.

Forging can be carried out conventionally, or isothermally, with isothermal forging having a marked advantage in terms of capital costs and quality of finished product and fast becoming the favoured forging technology. Superplastic forming, taking advantage of the phenomenum of superplasticity in titanium, permits forming of thin sheets of alloy, in a manner analogus to the forming of thermoplastic plastic sheet. The result has been a technique making possible the production of titanium sheet metal products of complex shape, with substantially reduced costs and weight requirements. The extrusion and rolling of titanium, except for a greater degree of lubrication being necessary to reduce titanium adherence to dies and rolls,are identical to that of steel or copper.

Working temperature, cooling rate and post shaping heattreatment can also have a substantial effect upon the properties of the finished product. As a result the entire thermomechanical processing of the metal must be carefully developed and controlled to insure production of parts with the required microstructure and desired properties.

ACKNOWLEDGEMENTS

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PREFACE

One of the requirements for the Master of Engineering degree at Concordia University is the preparation of a technical review paper. This paper on the Hot Working of Titanium Alloys, prepared under the supervision of Dr. Hugh McQueen is being submitted to fulfill this requirement.

The purpose of this paper has been to provide a survey of the technology of hot working of titanium and its alloys. The material was gathered from many sources, most in the form of technical papers or review articles and these unfortunately were not always in agreement. In writing this paper I have attempted to as often as possible obtain confirmation of material or at least include only those statements that are not in conflict with several other opinions. Due to the fact that research into the properties and workability of titanium alloys has only been going on for some 30 years the material is for the most part empirical. As a result a section on the crystal and phase structure of titanium was also included to permit the reader to get a complete picture of the metal and its properties. In the section actually dealing with hot working, references to terms from this section on microstructure have been made, rather than redefining the microstructure in question each time it occurs.

Working on this paper has, for me, been an education and I hope that it proves as interesting for someone else to read as it was for me to work on.

CHAPTER 1

INTRODUCTION

In the last 30 years titanium has developed from a relatively unknown metal confined only to the laboratory, into a high technology product, fast becoming the mainstay of the aerospace industry. Along with this growth in its application has come a proportional growth in knowledge about the metal and its properties, both pure and when alloyed with a variety of other elements.[3]

The properties that make titanium so popular a material are its high strength to weight ratio at ambient and elevated temperatures, (density 60% that of steel), high toughness and ductility, and good corrosion resistance. It is the first three properties that make titanium a valuable metal in aerospace applications, and the last three that make it extremely useful in certain applications in the chemical and processing industries.[2]

Regardless of the final use of the metal, if it is to be shaped mechanically into an end product then at some stage in its processing it will have to undergo hot working.

Titanium, unfortunately, is extremely chemically reactive at elevated temperatures (over 420°C) and thus requires special techniques and handling to prevent deterioration of the properties of the metal. At the same time due to the variety of different microstructures that can be obtained for each titanium alloy a wide range of physical properties are possible. Through the control of the temperature and deformation during hot working and by later heat treating the metal it is possible to form the desired microstructure and its accompanying desired physical properties.[1] It is these techniques of hot working and their varied effects upon the structure and properties of the shaped titanium that is the subject of this technical report.

CHAPTER 2

ALLOYS, PHASES AND CRYSTAL STRUCTURE

The behaviour of metals is determined to a large extent by their crystal structure, with each different crystal structure that a metal or its alloys forms being referred to as a separate phase and having a different set of properties and characteristics from the other phases that may form.

Pure unalloyed titanium has only two solid phases, up to 882°C (1625°F) it exists in an alpha phase which has a closed-packed hexagonal (hcp) crystal structure, above this temperature its atoms rearrange themselves into the beta phase which has a body centered cubic (bcc) structure. This beta phase remains up to the melting point, at 1725°C (3135°F).(7)

Commercially pure titanium is however only 99.0% to 99.5% titanium the remainder being, iron, oxygen, nitrogen and occasionally hydrogen, all of these elements but particularly the last three, as can be seen from Figs, 1, 2 and 3, have an adverse effect on ductility and toughness while increasing the tensile strength of the metal. These three gases are therefore avoided as alloying elements if at all possible. A variety of other elements can however be added to titanium to increase its strength and still retain a high degree of ductility and toughness as well as providing a wide range of other useful properties.(1)(2) The addition of these alloying elements results in the formation of many additional phases as well as combination of phases, whose dependance on composition and temperature are recorded in phase diagrams.

Those elements that form useful alloys may be divided into four categories reflecting the manner in which they interact with titanium. Shown in Fig.4 a - d are the phase diagrams of these four modes of interaction. The first type is a continuous solution of alpha and beta titanium, Fig.4a, and occurs only with zirconium and hafnium, both of which belong to the fourth group of elements as does titanium and therefore have similar properties and structures. The second type of interaction is a continuous solid solution only with beta titanium, Fig. 4b, and occurs with vanadium, niobium, tantalum, and molybdenum. The third phase diagram, Fig.4c, is representative of peritectoid reactions where the alloying elements form

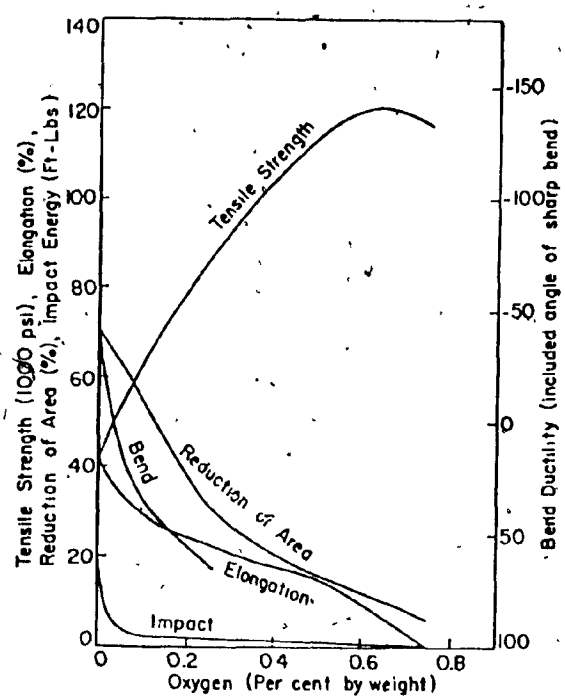


Figure 1. Effect of Oxygen on Mechanical Properties of Titanium. (After L. D. Jaffe, Watertown Arsenal)

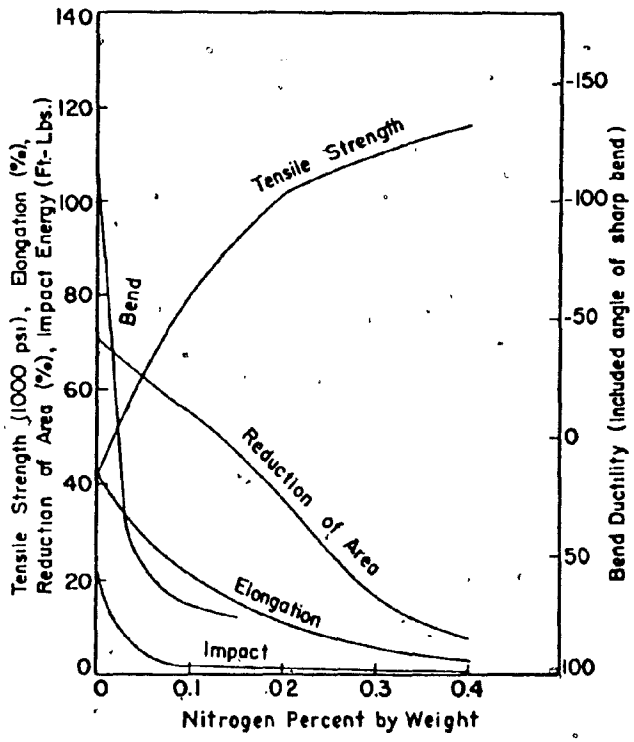


Figure 2. Effect of Nitrogen on Mechanical Properties of Titanium. (After L. D. Jaffe, Watertown Arsenal)

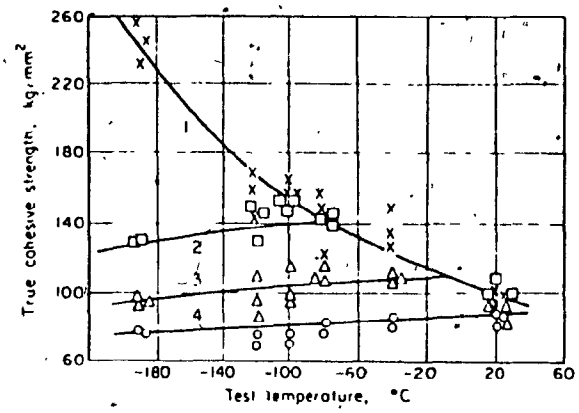


FIG. 3 Dependence of true fracture stress on test temperature for titanium containing various amounts of hydrogen. 1. Vacuum-annealed titanium. 2. Titanium containing 0·012 wt.% H_2. 3. Titanium containing 0·035 wt.% H_2. 4. Titanium containing 0·060 wt.% H_2.

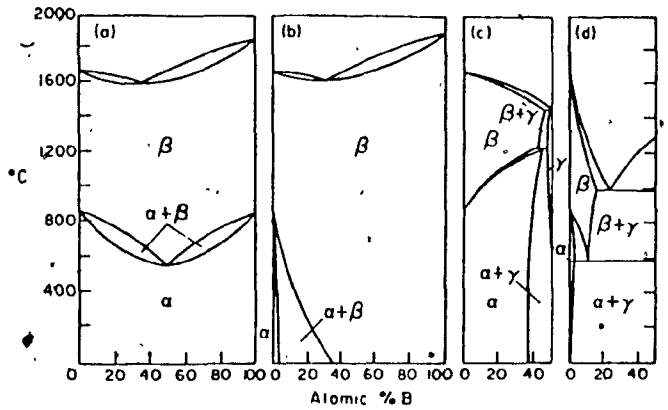


FIG. 4 The main types of phase diagram of the binary titanium systems.

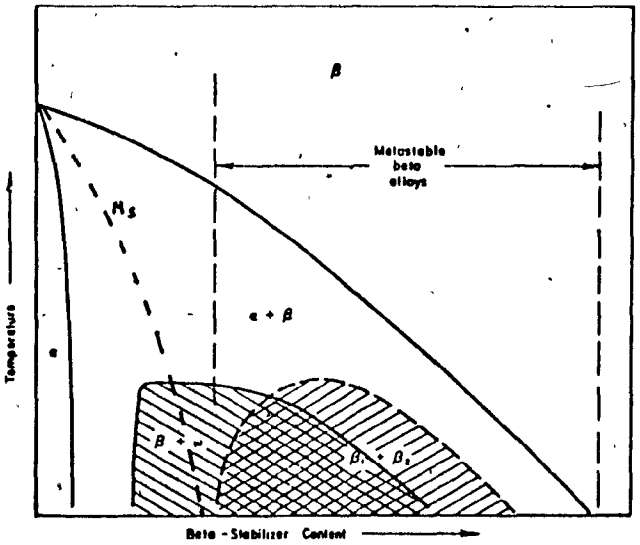


Fig 4e Schematic phase diagram showing the relative composition range for $\beta \rightarrow \alpha$, $\beta \rightarrow \omega + \beta$, $\beta \rightarrow \alpha_2 + \beta$ and $\beta \rightarrow \beta_1 + \beta_2$ reactions in the titanium base beta isomorphous system.[4,5]

one or more metallides with titanium. This reaction occurs with the elements aluminum, tin, boron, silicon and carbon. The fourth phase diagram, Fig.4d, represents those reactions that are eutectoid in the solid state as well as forming some stoichiometric compounds. The elements falling into this group include chromium, manganese, iron, cobalt nickel, copper, and beryllium.(2)

The metals of the first two phase diagrams, those forming continuous solid solutions with titanium do so due to the similarity of their atomic radii and their electrochemical properties. This permits these metals to replace atoms in the titanium crystal lattice, while the arrangement of the electrons in the d- and s- shells permits uniform interionic distribution of electrons. The resulting limited distortion of the crystal lattice along with the almost uniform electron distribution results in the alloy retaining a high degree of plasticity while increasing its tensile strength, even at high alloy contents.

The metals of the fourth phase diagram also form solid solutions with titanium, however due to differing numbers of external electrons as well as different atomic radii these are limited rather than continuous solutions. As the number of electrons in the d- shell increases (Fig.5) the solubility of the metals in beta titanium decreases and this is reflected in the eutectoid phase diagram. Beyond their solubility limit these alloying metals form intermetallic compounds (fig.5) The solubility of these compounds as well as their precipitation as dispersed phases within the titanium can materially affect the properties of the alloys in which they occur. Beryllium included in the above group of metals is an exception to the rule. Although it has a considerable smaller atomic radius than titanium and is electrochemically quite different it too undergoes a eutectoid reaction, with limited solubility in, and several intermetallic compounds with, titanium.

As the electrochemical properties of the alloying elements become even further removed from those of titanium or the difference in atomic radii becomes large, the solubility of the various titanium based phases decreases and the phase reactions themselves gradually shift from eutectoid to peritectoid, as shown in the third phase diagram.(2)

Element	Cr	Mn	Fe	Co	Ni	Cu	Be
Electron Shells	$3d^5$ $4s^1$	$3d^5$ $4s^2$	$3d^6$ $4s^2$	$3d^7$ $4s^2$	$3d^8$ $4s^2$	$3d^{10}$ $4s^1$	$2s^2$
Composition of Compound	$TiCr_2$	TiMn	TiFe Ti_2Fe	Ti_2Co	Ti_2Ni	Ti_2Cu	TiBe $TiBe_2$ $TiBe_4$ etc.
Melting Point (°C)	1350	1200	1500	1055	1015	990	—

Fig. 5. Electron shells and eutectoid compounds of the elements forming eutectoid alloys with titanium

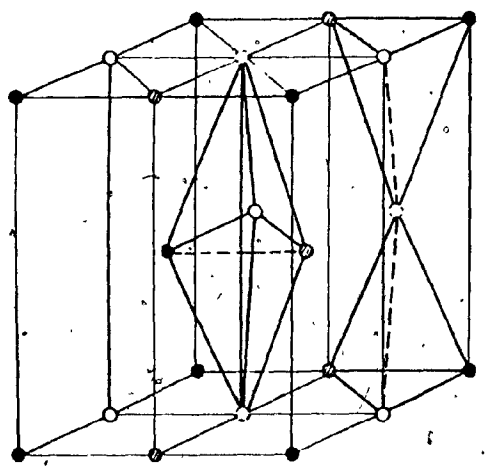


Fig. 6 The DO_{19} structure. The unit cell, composed of four regular hcp cells contains eight atoms. The four sublattices are indicated by different shadings of the circles. The heavy lines show some of the tetrahedra that are formed when the atomic displacements are taken into account.

TABLE 1 - PHYSICAL PROPERTIES

Nominal Composition, %	ASTM No.	Military	Aerospace Material Specifications: Bars & Forgings	Sheet & Plate	Tubing	Wire	Average Physical Properties: Coefficient of Linear Thermal Expansion, 10^{-4} in/in/F: 70 212 F	70 400 F	70-600 F	70-800 F	70 1,000 F	70 1,200 F	70-1,500 F	Modulus of Elasticity, 10^4 Psi*	Modulus of Rigidity, 10^4 Psi*
Commercially Pure															
99.5 Ti	B 265 (Gr. 1)	—	—	—	—	—	4.8	—	5.1	—	5.4	5.6	5.6	14.9	5.6
	B 348 (Gr. 1)	—	—	—	—	—									
	B 381 (Gr 1)	—	—	—	—	—									
99.2 Ti	B 265 (Gr. 2)	MIL T 9046	—	4902	4941	4951	4.8	—	5.1	—	5.4	5.6	5.6	14.9	5.6
	B 348 (Gr. 2)	—	—	—	4942	—									
	B 381 (Gr. 2)	—	—	—	—	—									
99.1 Ti	B 265 (Gr. 3)	MIL T 9046	—	4900	—	—	4.8	—	5.1	—	5.4	5.6	5.6	15.0	5.6
	B 348 (Gr. 3)	—	—	—	—	—									
	B 381 (Gr. 3)	—	—	—	—	—									
99.0 Ti	B 265 (Gr 4)	MIL T 9046	4921	4901	—	—	4.8	—	5.1	—	5.4	5.6	5.6	15.1	5.6
	B 348 (Gr. 4)	MIL T 9047	—	—	—	—									
	B 381 (Gr. 4)	—	—	—	—	—									
99.2 Ti†	B 265 (Gr. 7)	—	—	—	—	—	4.8	—	5.1	—	5.4	5.6	5.6	14.9	5.6
	B 348 (Gr. 7)	—	—	—	—	—									
	B 381 (Gr. 7)	—	—	—	—	—									
98.9 ‡	—	—	—	—	—	—	—	—	—	—	—	—	—	—	14.9
Alpha Alloys															
5 Al, 2.5 Sn	B-265 (Gr. 6)	MIL T 9046	4926	4910	—	4953	5.2	—	5.3	—	5.3	5.4	5.6	16.0	—
	B 348 (Gr 6)	MIL T 9047	4966	—	—	—									
	B 381 (Gr. 6)	—	—	—	—	—									
5 Al, 2.5 Sn (low O_2)	—	MIL-T 9046	4924	4909	—	—	5.2	—	5.3	—	5.4	5.5	5.6	16.0	—
	—	MIL T 9047	—	—	—	—									
Near Alpha															
8 Al, 1 Mo, 1 V	—	MIL T 9046	4972	4915	—	4955	4.7	—	5.0	—	5.6	5.7	—	18.0	6.8
	—	MIL T 9047	4973	4916	—	—									
11 Sn, 1 Mo, 2.25 Al, 5.0 Zr, 1 Mo, 0.2 Si	—	MIL T 9047	4974	—	—	—	4.7	—	5.1	—	5.2	—	—	16.5	—
6 Al, 2 Sn, 4 Zr, 2 Mo	—	MIL T 9046	4975	—	—	—	4.3	—	4.5	—	4.5	—	—	16.5	—
5 Al, 5 Sn, 2 Zr, 2 Mo, 0.25 Si	—	—	—	—	—	—	—	—	—	—	—	—	5.7	16.5	—
6 Al, 2 Cb, 1 Ta, 1 Mo	—	MIL T 9046	—	—	—	—	—	—	—	—	—	5.0	—	17.5	—
6 Al, 2 Sn, 1.5 Zr, 1 Mo, 0.35 Bi, 0.1 Si	—	—	—	—	—	—	—	—	—	—	—	—	—	—	—
Alpha Beta Alloys															
8 Mn	—	MIL-T 9046	—	4908	[illegible]	—	4.8	5.1	5.4	5.7	6.0	6.5	7.0	16.4	7.0
3 Al, 2.5 V	—	—	—	—	4943 4944	—	5.3	—	5.5	—	5.5	—	—	15.5	—
6 Al, 4 V	B 265 (Gr. 5)	MIL T-9046	4928	4911	—	4954	4.8	5.0	5.1	5.2	5.3	5.4	—	16.5	6.1
	B 348 (Gr 5)	MIL T 9047		4906	—	—									
	B 381 (Gr. 5)	—	4565	—	—	—									
6 Al, 4 V (low O_2)	—	MIL-T-9046	4930	4907	—	4956	4.8	5.0	5.1	5.2	5.3	5.4	—	16.5	6.1
	—	MIL T 9047	—	—	—	—									
6 Al, 6 V, 2 Sn	—	MIL T 9046	4971	4918	—	—	5.0	—	5.2	—	5.3	—	—	18.0	—
	—	MIL-T 9047	4978	—	—	—									
	—	—	4979	—	—	—									
7 Al, 4 Mo	—	MIL-T 9047	4970	—	—	—	5.0	5.1	5.2	5.4	5.6	5.8	6.2	16.5	6.5
6 Al, 2 Sn, 4 Zr, 6 Mo	—	—	4981	—	—	—	5.0	5.1	5.2	5.3	5.3	—	—	16.5	—
6 Al, 2 Sn, 2 Zr, 2 Mo, 2 Cr, 0.25 Si	—	—	—	—	—	—	—	—	5.1	—	—	—	—	17.7	6.7
10 V, 2 Fe, 3 Al	—	—	—	—	—	—	—	—	—	—	—	—	—	16.2	—
Beta Alloys															
13 V, 11 Cr, 3 Al	—	MIL T 9046	—	4917	—	—	5.2	—	5.6	—	5.9	—	—	14.7	6.2
	—	MIL T 9047	—	—	—	—									
8 Mo, 8 V, 2 Fe, 3 Al	—	—	—	—	—	—	—	—	—	—	—	—	—	15.5	—
3 Al, 8 V, 6 Cr, 4 Mo, 4 Zr	—	—	—	—	—	—	—	—	—	5.38 (to 900 F)	—	—	—	15.3	—
11.5 Mo, 6 Zr, 4.5 Sn	—	MIL T 9047	4977	—	—	4980	—	—	—	—	—	—	—	15.0	—

TABLE 2 - MECHANICAL PROPERTIES

Nominal Composition, %	Average Mechanical Properties: Room Temperature: Tensile Strength, Psi	Yield Strength, Psi	Elongation, %	Reduction in Area, %	Extreme Temperatures: Test Temp, F	Tensile Strength, Psi	Yield Strength, Psi	Elongation, %	Reduction in Area, %	Hardness	Applications and Characteristics
Commercially Pure											
99.5 Ti	48,000	35,000	30	55	600	22,000	14,000	32	80	Bhn 120	Airframes, chemical, desalination, and marine parts; plate type heat exchangers, cold spun or pressed parts; platinized anodes, high formability.
99.2 Ti	63,000	50,000	28	50	600	28,000	17,000	35	75	Bhn 200	Airframes, aircraft engines, marine and chemical parts; heat exchangers, condenser and evaporator tubing, high formability
99.1 Ti	75,000	65,000	25	45	600	34,000	20,000	34	75	Bhn 225	Chemical, marine, airframe, and aircraft engine parts which require formability, strength, weldability, and corrosion resistance
99.0 Ti	96,000	85,000	20	40	600	45,000	25,000	25	70	Bhn 265	Chemical, marine, airframe, and aircraft engine parts; surgical implants, high speed fans; gas compressors, good formability and corrosion resistance, high strength
99.2 Ti†	63,000	50,000	28	50	600	27,000	16,000	37	75	Bhn 200	Good corrosion resistance for chemical industry applications where media is mildly reducing or varies between oxidizing and reducing.
98.9 ‡	76,000	65,000	25	42	400 600	50,000 47,000	36,000 35,000	37 32	—	—	Same as 0.2 Pd alloy (above)
Alpha Alloys											
5 Al, 2.5 Sn	125,000	117,000	16	40	600	82,000	65,000	18	45	Rc 36	Weldable alloy for forgings and sheet metal parts such as aircraft engine compressor blades and ducting, steam turbine blades, good oxidation resistance and strength at 600 to 1,100 F; good stability at elevated temperatures.
5 Al, 2.5 Sn (low O_2)	117,000	108,000	16	—	—320 —423	180,000 229,000	168,000 206,000	16 15	— —	Rc 35	Special grade for high pressure cryogenic vessels operating down to —423 F.
Near Alpha											
8 Al, 1 Mo, 1 V	145,000	138,000	15	28	600 800 1,000	115,000 107,000 90,000	90,000 82,000 75,000	20 20 25	38 44 55	Rc 35	Airframe and jet engine parts requiring high strength to 850 F, good creep and toughness properties, good weldability.
11 Sn, 1 Mo, 2.25 Al, 5.0 Zr, 1 Mo, 0.2 Si	160,000	144,000	15	35	600 800 1,000	130,000 120,000 110,000	110,000 98,000 85,000	20 22 24	44 48 50	Rc 36	Airframes, blades, discs, wheels, spacers, and fasteners for turbine engines.
6 Al, 2 Sn, 4 Zr, 2 Mo	142,000	130,000	15	35	600 800 1,000	112,000 102,000 91,000	85,000 75,000 71,000	16 21 26	42 55 60	Rc 32	Parts and cases for jet engine-compressors, airframe skin components.
5 Al, 5 Sn, 2 Zr, 2 Mo, 0.25 Si	152,000	140,000	13	—	600 800 1,000	115,000 113,000 100,000	82,000 77,000 73,000	15 17 19	— — —	—	Jet engine parts, high creep strength to 1,000 F.
6 Al, 2 Cb, 1 Ta, 1 Mo	124,000	110,000	13	34	600 800 1,000	85,000 75,000 70,000	67,000 60,000 55,000	20 20 20	— — —	Rc 30	High toughness, moderate strength, good resistance to seawater and hot salt stress corrosion; good weldability.
6 Al, 2 Sn, 1.5 Zr, 1 Mo, 0.35 Bi, 0.1 Si	147,000	137,000	11	—	900	105,000	85,000	15	—	—	Jet engine discs and blades requiring extra creep resistance and stability
Alpha-Beta Alloys											
8 Mo	137,000	125,000	15	32	600	104,000	82,000	18	—	—	Aircraft sheet components, structural sections, and skins; good formability, moderate strength
3 Al, 2.5 V	100,000	85,000	20	—	600	70,000	50,000	25	—	—	Aircraft hydraulic tubing, foil, combines strength, weldability, and formability
6 Al, 4 V	144,000	134,000	14	30	600 800 1,000	105,000 97,000 77,000	95,000 83,000 62,000	14 18 35	35 40 50	Rc 36	Rocket motor cases, blades and discs for aircraft turbines and compressors, structural forgings and fasteners; pressure vessels, gas and chemical pumps, cryogenic parts, ordnance equipment; marine components, steam turbine blades.
	170,000	160,000	10	25	600 800 1,000	125,000 116,000 95,000	102,000 90,000 70,000	10 12 22	28 35 45	Rc 41	
6 Al, 4 V (low O_2)	130,000	120,000	15	35	—320	220,000	205,000	14	—	Rc 35	High pressure cryogenic vessels operating down to —320 F
6 Al, 6 V, 2 Sn	155,000 185,000	145,000 170,000	14 10	30 20	600 600	135,000 142,000	117,000 130,000	18 12	42 28	Rc 38 Rc 42	Rocket motor cases, ordnance components, structural aircraft parts and landing gears, responds well to heat treatments, good hardenability
7 Al, 4 Mo	160,000	150,000	16	22	600 800	127,000 123,000	108,000 104,000	18 20	50 55	Rc 38 Rc 42	Airframes and jet engine parts for operation at up to 800 F, missile forgings, ordnance equipment.
6 Al, 2 Sn, 4 Zr, 6 Mo	184,000	170,000	10	23	600 800 1,000	148,000 138,000 123,000	122,000 110,000 95,000	18 19 19	55 67 70	—	Components for advanced jet engines.
6 Al, 2 Sn, 2 Zr, 2 Mo, 2 Cr, 0.25 Si	185,000	165,000	11	33	600	142,000	117,000	14	27	—	Strength, fracture toughness in heavy sections, landing gear wheels.
10 V, 2 Fe, 3 Al	185,000	174,000	10	19	400 600	162,000 160,000	152,000 142,000	13 13	33 42	—	Heavy airframe structural components requiring toughness at high strengths.
Beta Alloys											
	177,000	170,000	8	—	600	128,000	115,000	19	—	—	High strength fasteners
13 V, 11 Cr, 3 Al	185,000	175,000	8	—	800	160,000	120,000	12	—	Rc 40	High strength fasteners, aerospace components, honeycomb panels, good formability, heat treatable
8 Mo, 8 V, 2 Fe, 3 Al	190,000	180,000	8	—	600	164,000	142,000	15	—	Rc 40	High-strength, tough airframe sheet, plate, fasteners, and forged components.
3 Al, 8 V, 6 Cr, 4 Mo, 4 Zr	210,000	200,000	7	—	600 800	150,000 136,000	130,000 110,000	20 17	— —	Rc 42	High strength fasteners, torsion bars, aerospace components.
	128,000	121,000	15	—	600	105,000	95,000	22	—	—	Parts requiring formability and corrosion resistance
11.5 Mo, 6 Zr, 4.5 Sn	201,000	191,000	11	—	600	131,000	123,000	16	—	—	High strength fasteners, high strength aircraft sheet parts.

All of the above alloying elements may be loosely classified into alpha and beta stabilizers depending on their solubility in the alpha or beta phase and their effect on the transformation temperature of beta to alpha (beta transus). The alpha stabilizers are the elements of phase diagrams 4a and c. Aluminum and carbon along with oxygen and nitrogen being strong stabilizing elements with tin, zirconium, hafnium and silicon having reduced effects and in some cases slightly depressing the beta transus.

The beta stabilizing elements are those of phase diagrams 4b and d and are capable of greatly reducing the beta transus. The elements in this group include vanadium, chromium, manganese, iron, cobalt, nickel, copper, niobium, tantalum and molybdenum. Although some of these elements undergo eutectoid reactions with titanium, due to the extreme slowness of these reactions, for the purposes of thermomechanical working these alloys may be treated as metastable beta isomorphous alloys and will react in a similar fashion.

Commercial titanium alloys may be classified according to the type of phase structure that predominates at room temperatures. The classifications currently in use are: alpha, alpha + beta, beta, near alpha or pseudo alpha, and near beta or pseudo beta, these last two categories refer to alloys having an alpha or beta structure but containing no more than 2% of the other phase. A list of commercial titanium alloys their compositions, properties and applications may be found in tables 1 and 2. The alloys listed are American and English; Russian and French developed alloys may vary slightly in composition but still fall into at least the three basic categories alpha, alpha + beta, and beta.

2.1 ALPHA ALLOYS

Alpha alloys have excellent low temperature ductility, good weld ductility, good high temperature creep strength and a flat strength - temperature characteristic curve. Their tensile strength is however low relative to other titanium alloys and despite the effectiveness of solid solution strengthening due to the addition of alpha stabilising alloying elements, (5 - 10 ksi per percent alloy addition) the total quantity of alloy elements that may be added is restricted by the formation of the $alpha_2$ phase. $Alpha_2$ is a coherent ordered phase based on Ti_3X compounds,

has a DO 19 (see Fig.6) structure, forms upon aging of supersaturated alloys with aluminum, gallium, indium, tin and lead, and is undesirable due to its marked reduction of ductility. The maximum alpha stabilizing alloy content to avoid excessive $alpha_2$ may be expressed in equivalent aluminum content (Al*) by the expression developed by Rosenberg[8].

$$Al^* = Al + \frac{Sn}{3} + \frac{Zr}{6} + 10(O + C + 2\ N) \leqslant 9\ wt\%$$

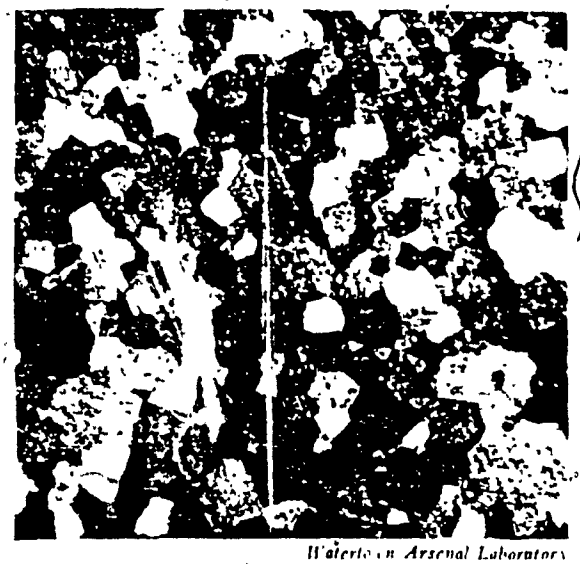
The mode in which alpha 2 is precipitated depends on the alloy system being considered as the greater the degree of lattice mismatch in the alloy, the smaller the particles and the volume fraction of alpha 2 that precipitates. Of the above elements, aluminum, having the lowest mismatch, tends to produce large quantities of alpha 2 however, the addition of gallium or to a lesser extent zirconium or silicon, all of which increase mismatch, may be used to control or minimize deleterious effects. At the same time the resulting fine alpha 2 structure which is produced may prove to be a useful source of precipitation strengthening if properly controlled.

Copper and silicon too, have a fair degree of solubility in the alpha phase and at high enough concentrations and if properly heat treated also undergo precipitation processes. These processes, involving Ti_3Si_4 and Ti_2Cu precipitates, result in a different microstructure but occur in a similar manner and produce properties similar to tempered martensitic structures.[3][4][5](See Fig. 7,8)

An acicular microstructure, solute lean but similar in morphology to martensite, and often called alpha alloy martensite, may also be obtained, by cooling from the beta field. The original beta grains are still visible and three structures are possible depending upon the cooling rate. Rapid cooling produces colonies of plates separated by low angle grain boundaries of alpha titanium. Slower cooling produces a "basketweave" structure of alpha platelets with beta stabilizing impurities segregated to the interlamellar regions. Slower cooling and a reduced quantity of beta stabilizing impurities results in a serrated alpha structure where the interlamellar regions are not distinguishable.(see fig.9, 10)

Fig.8 Fine lamellar α Ti_2Cu structure which is formed in a Ti-8%, Cu alloy during rapid cooling (~ 300 C/sec) from 950 C

Fig.7 3: Titanium silicides in a β-phase matrix in Ti-8.5 wt.%Mo-2.0 wt.%Si quenched from 1000°C.



Watertown Arsenal Laboratory

Fig. 9 Equiaxed Alpha Structure—High Purity Titanium Furnace-cooled from 1700°F. ×100

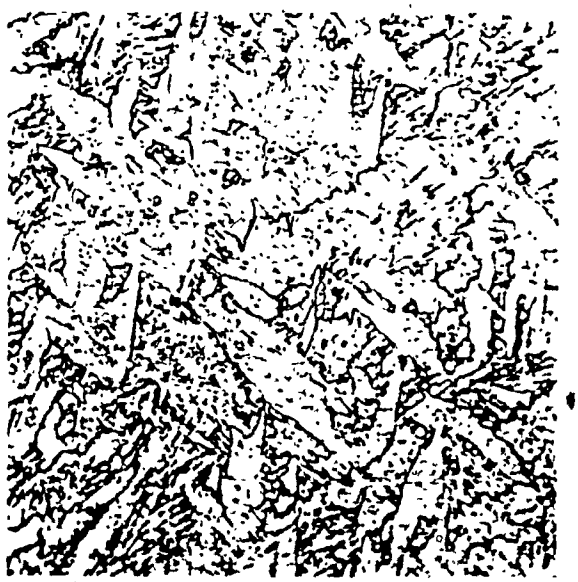


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Fig. 10 Serrated Alpha Structure—High Purity Titanium Water-quenched from 2000°F. ×100

Watertown Arsenal Laboratory

Fig. 11 Acicular Alpha Structure—Commercially Pure Titanium Water-quenched from 1650°F. ×100

Generally speaking, in alpha alloys the aging of these acicular microstructures results in a somewhat lower tensile strength, and ductility but an improved high temperature creep strength and fracture toughness.[4] With a view to later thermomechanical working, the optimum properties of low strength and high ductility, are obtained by slow cooling from the beta field, the result is an equiaxed polygonal grain structure of alpha or alpha + beta, depending upon the alloy composition.[1] (see Fig. 11)

2.2 BETA ALLOYS

When titanium has been alloyed with sufficient beta stabilizing elements, it is able to retain the body center cubic structure while being cooled from above the beta transus to room temperature. The alloys in this group being called beta alloys, are divided into two groups, stable-beta and metastable-beta.

Stable-beta alloys are possible only with beta-isomorphus alloying elements and at such high alloy content as to be more like intermediate alloys in the system of the beta stabilizers than a titanium alloy. Also, due to the high alloy content, stable beta alloys are too heavy and not strong enough to compete with titanium alloys and lack sufficient high temperature strength or oxidation resistance to compete with superalloys.

Meta-stable beta on the other hand has sufficient beta stabilizing elements to retain the beta phase only on rapid cooling from the beta transus but not sufficient to make beta the only stable phase at all temperatures. As a result the high ductility and excellent solution annealed formability of the body centered cubic crystal structure, with its surplus of active slip systems, is retained making possible cold working of the alloy. Afterwards, aging at the appropriate temperature will precipitate the hcp alpha phase resulting in maximum tensile strengths for titanium. This high tensile strength, combined with the inherent deep hardenability of beta alloys and their high fracture toughness makes them extremely useful in fabrication of heavy sections. Only in those applications where stiffness is critical is it impossible to use beta alloys, this being due to their low elastic modulus and cyclic fatigue strength.

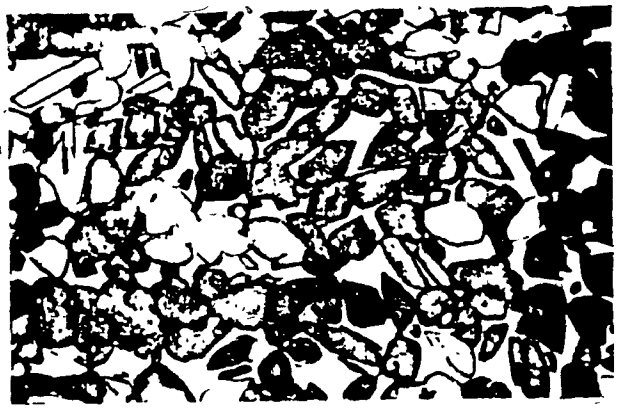
During the quenching of metastable beta alloys two other decomposition phases may be precipitated. The first is the omega phase, a homogeneously coherent precipitate limited to relatively lean beta alloys, while the second is beta prime, the result of a phase separation reaction beta $\longrightarrow$ beta + beta prime, where beta prime is also a homogeneously nucleated coherent precipitate but confined to richer alloys. Both of these phases but especially omega are undesirable due to their detrimental effects upon the properties of the beta alloys. The properties, occurence, structure and removal of these phases will be discussed later, in the section on alpha + beta alloys, but for now it is sufficient to say that proper selection of alloying elements and heat treatment can control this problem and even utilize it to provide improved properties.(5)(9)(11)(12)

Metastable beta may also undergo a stress induced transformation to martensite, or mechanical twinning. The type of transformation, its extent and the stresses at which it occurs are all dependent on alloying content. When it occurs at low stresses it results in low yield strength, but in more solute rich alloys it occurs above the beta phase yield strength and provides enhanced strength and ductility. Despite the many advantages they offer, due to their low high temperature creep strength, applications for beta alloys remain restricted to relatively low temperature applications.(1)(3)(4)(5)(6)(7)

2.3 ALPHA + BETA ALLOYS

The alpha + beta alloys are those, that fall between the two previously mentioned groups, and as a result, have large volume fractions of both alpha and beta titanium, as well as under the proper conditions, other phases and, or compounds all at equilibrium.

In cooling from the beta to the alpha + beta phase a certain portion of the metastable beta will be transformed into alpha by one of two processes. At slow cooling rates nucleation and growth processes occur, while for faster rates, martensite transformation occurs when the Ms temperature is crossed.

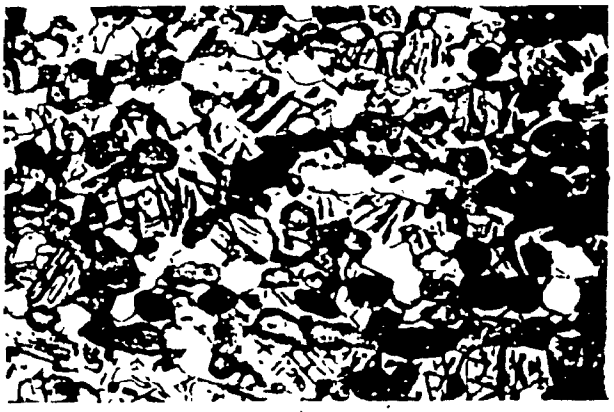


Medium globular and plate alpha —1 hr 900 C, furnace cool.

(A)

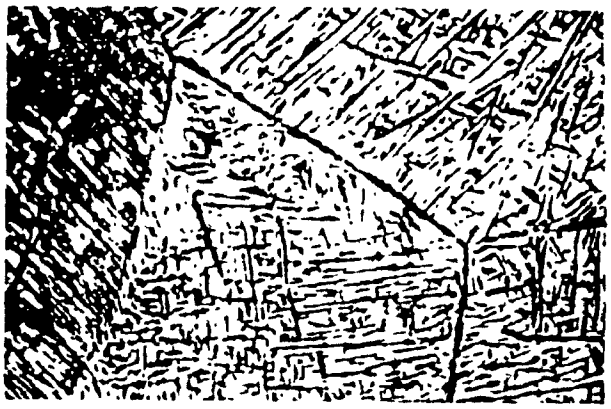
Coarse Widmanstatten type alpha —1 hr 1000 C, furnace cool.

(A)



Mixed original and Widmanstatten type alpha (proportion 2:1)—1 hr 900 C, still air cool.

(B)



Medium Widmanstatten type alpha — 1 hr 1000 C, still air cool

(B)

Mixed original and Widmanstatten type alpha (proportion 1:2)—1 hr 930 C, still air cool.

(C) Fig.12

Fine Widmanstatten type alpha —1 hr 1000 C, water quench. All 500.

(C) Fig.13

Fig.14 'Aligned α' Microstructure in Ti685. Heat Treated 1hr 1080°C FC and Aged.

Fig.15 'Basketweave α' Microstructure in Ti685 Heat Treated 1hr 1080°C AC and Aged.

(A)

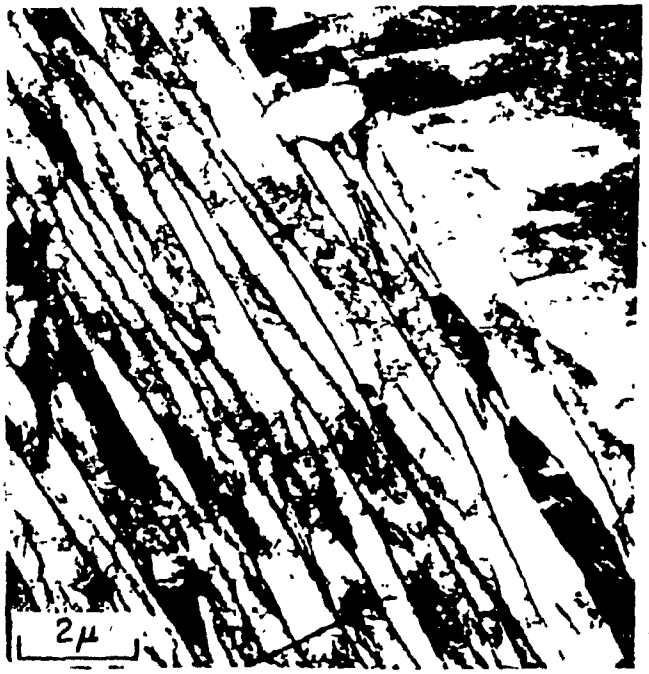
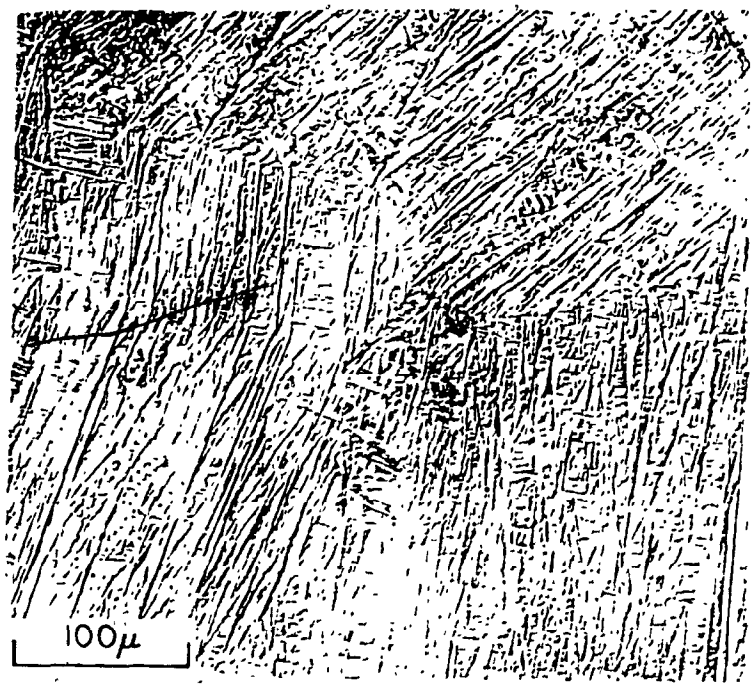


Fig.16 Massive martensite in Ti-1.78 wt.%Cu quenched from 900°C

A) Optical micrograph showing large colonies.

B) Thin foil micrograph showing individual plates within the colonies.



A

B

Fig.17: Acicular martensite in Ti-12 wt.%V quenched from 900°C.

A) Optical micrograph.

B) Thin foil micrograph showing lenticular-shaped plates, some of which are internally twinned.

The nucleation and growth process involves the development from metastable beta of equiaxed, grain boundary, and/or Widmanstatten alpha. For sufficiently slow cooling rates equiaxed alpha appears in a beta matrix, however as heat treatment and composition are changed it can also appear combined with Widmanstatten alpha or disappear altogether leaving only Widmanstatten alpha. The presence, grain size, and percentage of equiaxed alpha formed is dependant upon working temperature, cooling rate and alpha to beta transformation temperature, as is the structure of whatever Widmanstatten alpha is formed.(see fig. 12) At high transformation temperatures and slow cooling rates the Widmanstatten alpha occurs as clusters of parallel plates that appear to have nucleated at the grain boundary alpha.(see fig. 13a,14) As the transformation temperatures decrease due to increased beta stabilizers or increased cooling rate the clusters of plates become progressively smaller and less ordered.(see fig. 13b,c,15) Fewer and fewer plates are parallel and they appear to have nucleated throughout the original grain in many different directions forming the so-called "basketweave" structure.

Whichever Widmanstatten structure develops, the alpha platelets are surrounded by layers of the original beta phase enriched in beta stabilizers that diffuse out as the alpha forms, the thickness of these layers being dependant on the quantity of beta stabilizing elements in the alloy and the cooling rate.(3)(5)

The martensite transformation process, on the other hand, is a shear type transformation of the bcc beta phase to either hexagonal martensite called alpha prime or orthorombic martensite called alpha double prime. The more common of these two, the hexagonal alpha prime martensite, has two possible structres "massive martensites" and "acicular martensites". Massive martensite consists of colonies of parallel plates separated by low angle grain boundaries, each colony having the same Burgers vector orientation to the beta matrix. This type of martensite occurs for alloys of high Ms temperature, which from fig. 4e would indicate the near alpha alloys. For lower Ms temperatures, or greater amounts of beta stabilizers, the acicular martensite forms, with the colonies degenerating into individual plates each with a different varient of the Burgers orientation.(see fig. 16,17)

The second type of titanium martensite, the alpha double prime with an orthorhombic structure occurs with some binary alloys such as Ti-Mo, Ti-Nb, Ti-W and Ti-Re but not in others like Ti-V. (see fig. 18) In those alloys where alpha double prime does not occur, such as Ti-V, it has been found that ternary additions of aluminum tend to stabilize the orthorombic phase, over a fairly wide composition range.

Other martensites have also been reported with fcc and fco microstructures, there is however, some doubt as to whether these phases actually exist, in that they have been detected only in thin foils and have proved impossible to isolate in bulk quantities.

Both of the above martensites are metastable phases and will therefore with proper heat treatment decompose to alpha and beta phases. In hexagonal martensite the products of this decomposition depend on whether the alloy is beta-isomorphous or beta-eutectoid. In beta-ismorphous alloys the decomposition is directly to alpha + beta, through nucleation and growth processes of the beta phase occuring at martensite plate boundaries and on the internal martensite substructure. As the beta phase forms the remaining martensite approaches the equilibrium composition of the alpha phase at the tempering temperature. (fig. 19a) In beta eutectoid alloys the decomposition is to alpha phase plus intermetallic compound. In active eutectoid alloys this reaction is fairly fast but may be carried out in stages to permit forming before complete hardening occurs, in the slower eutectoids however the metastable beta phase forms first and the intermetallic compound very slowly after that, thus presenting no problem.

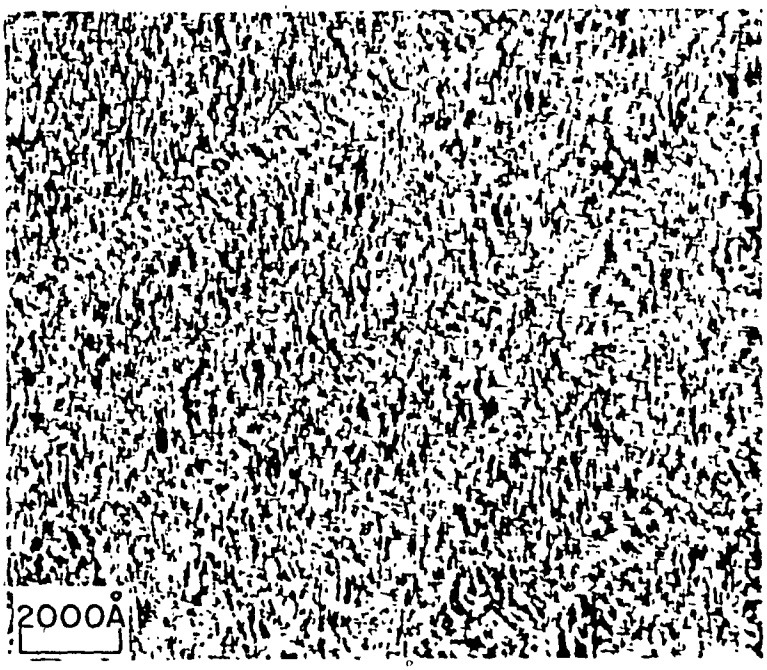
Orthorhombic martensite also decompose in two different ways. For those alloys of high Ms temperatures fine particles of alpha phase first precipitate out uniformly thus enriching the alpha double prime matrix with beta stabilizing elements. Eventually the alpha phase coarsens, alpha + beta lamellar regions nucleate at former beta grain boundaries and grow out to consume the martensite matrix. (see 19b) In alloys with fairly low Ms temperatures tempering occurs by a simple reversal or reshearing of the martensite back to beta phase which later decomposes.(3)(4)(5)(8)(9)(10)

Direct decomposition of the metastable beta phase to alpha phase, except for ageing for long periods of time, appears to be difficult to achieve and

Fig.18 Thin foil micrograph of α" in Ti-8.5 wt.%Mo-0.5 wt.%Si alloy quenched from 950°C, showing $\{111\}_o$ twins.

Fig.19 β-phase precipitates formed during tempering of α' martensite in Ti-6Al-4V

a) showing nucleation at α' plate boundaries and within the α' plates,



b) Tempered α" in a Ti-18 wt.%W alloy showing uniformly nucleated fine -phase precipitates.

as such is only infrequently seen. The usual method of decomposition is through intermediate products such as omega and beta prime phases, these have been examined superficially in the section on beta alloys but will be handled in greater detail below.

Omega phase may be formed as a metastable precipitate either on quenching metastable beta alloys whose Ms is just below room temperature or by aging, slightly more solute rich alloys, at 300 - 450°C. It is a homogeneously nucleated coherent precipitate the composition of which is not precisely known, however it occurs as extremely small particles, with a very high density and is solute poor compared to the beta phase in which it forms. (see fig. 20, 21, 22) As was stated earlier the amount of omega formed may be controllable by changes in alloy content, (see fig.23) such as addition of aluminum, tin or beta stabilizers as well as by altering heat treatment temperatures.(3)(4)(5)(9)(11)(12)

Beta prime forms between 200 and 500°C as a result of a phase separation of the beta phase into beta + beta prime. It is a uniformly distributed, coherent, bcc phase of indeterminate composition but is solute lean. The morphology of beta prime is variable depending on composition differences and misfit between the two bcc phases. (see fig.24) There is a close association between the formation of beta and omega, as beta prime forms, either during low temperature aging of alloys in which omega formation is sluggish, or in alloys which have sufficient beta stabilizers to suppress omega as well as make the alpha phase formation sluggish. As in the case of omega phase the addition of certain alloying elements such as aluminum appreciably reduce the stability of the beta prime precipitates (3)(4)(6)

Both omega and beta prime being metastable, will upon further aging, decompose and precipitate alpha phase. In omega the nucleation occurs at points of lattice mismatch or if mismatch is low at grain boundaries, and results in an elongated, plate or needle like morphology. (see fig. 25,26) In beta phase, nucleation may also occur on the beta particles and results in a morphology similar to omega. For aging of either phase at temperatures above their stability point alpha will precipitate out directly from the beta. For alloys lean in beta stabilizers or high in aluminum content, it appears as Widmanstatten alpha while for those rich in

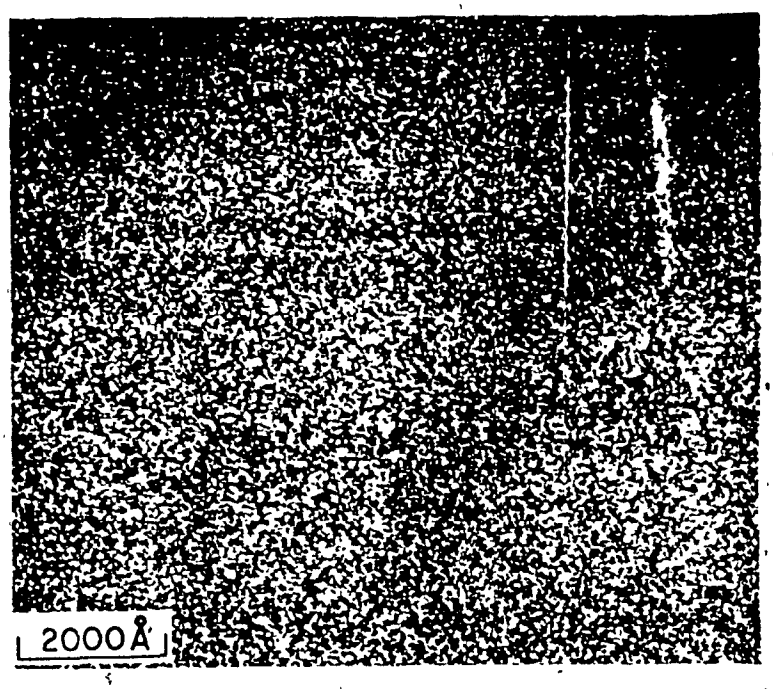


Fig. 20 Dark field electron micrograph showing athermal ω-phase particles in Ti-11.5Mo-4.5Sn-6Zr quenched from 900°C.

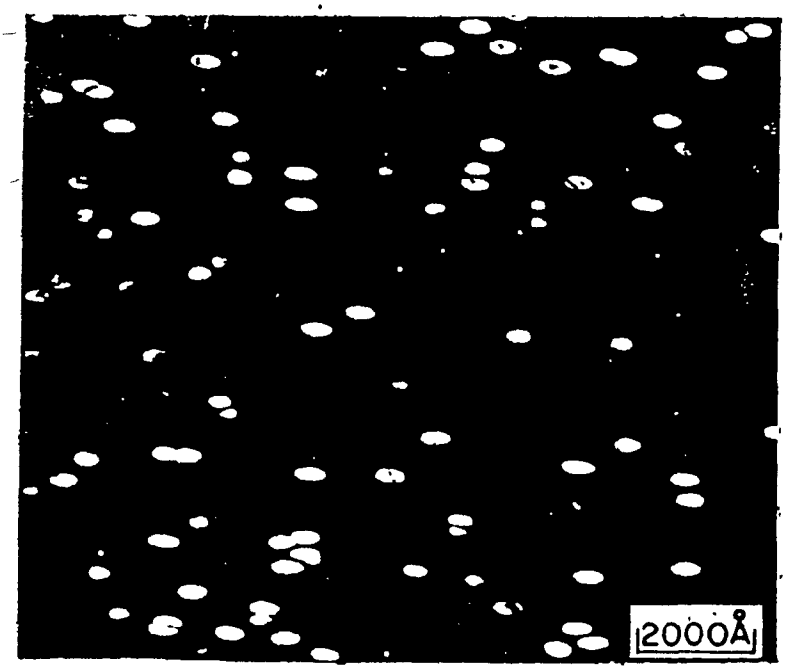


Fig. 21 Dark field electron micrograph of ellipsoidal omega phase in Ti-11.5Mo-4.5Sn-6Zr. Ellipsoidal omega forms in alloys with low lattice misfit between the omega precipitate and the bcc matrix. Alloys of low misfit include Ti-Mo, Ti-Ta, and Ti-Nb.

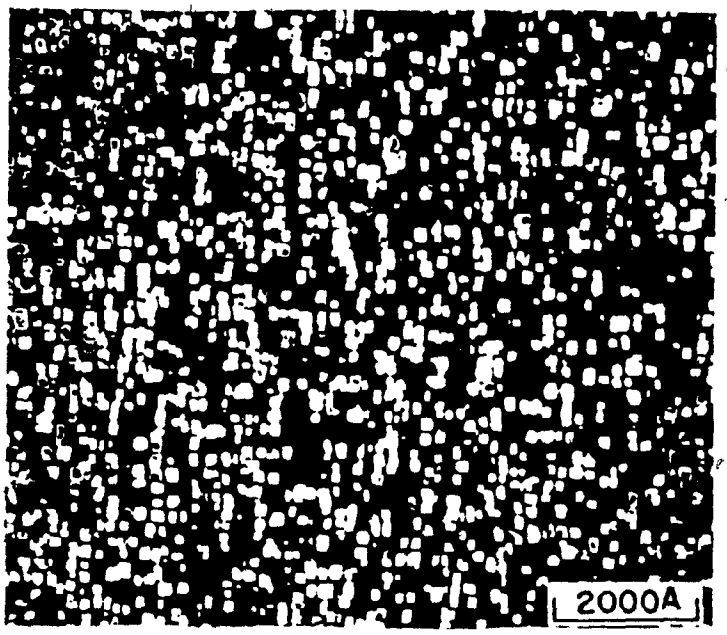


Fig.22 Dark field electron micrograph of cuboidal omega phase in Ti-10 wt.%Fe. Cuboidal omega forms in alloys with high lattice misfit between the omega precipitate and the bcc matrix. Alloys of high misfit include, Ti-Fe, Ti-V, Ti-Cr, Ti-Ni, and Ti-Mn.

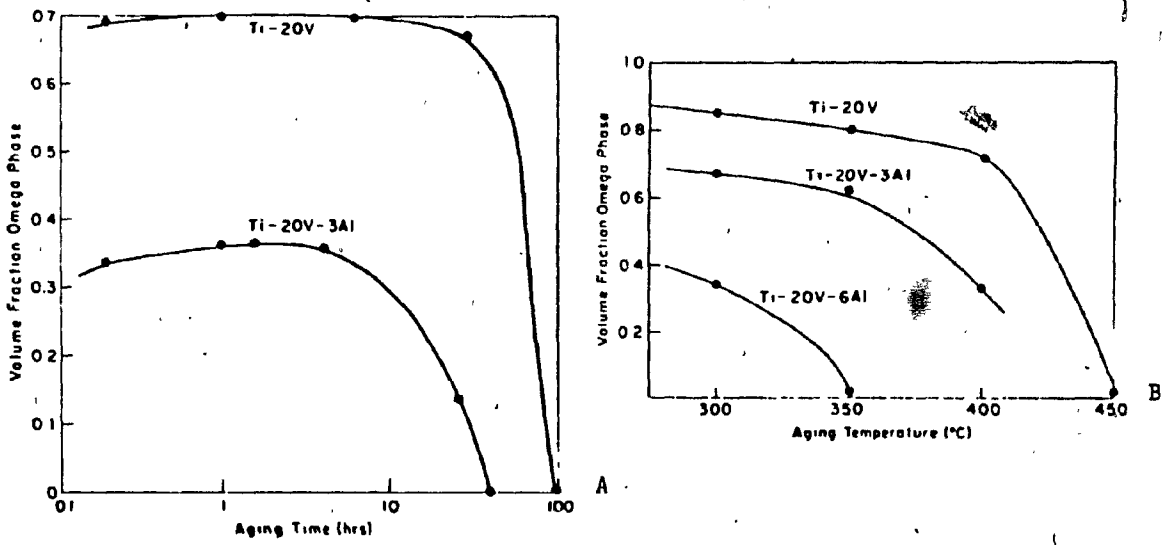


Fig.23 Showing the influence of ternary Al additions on the stability of ω-phase in Ti-V alloys.
a) Volume fraction vs. time at 400°C.
b) Metastable equilibrium volume fraction vs. aging temperature.

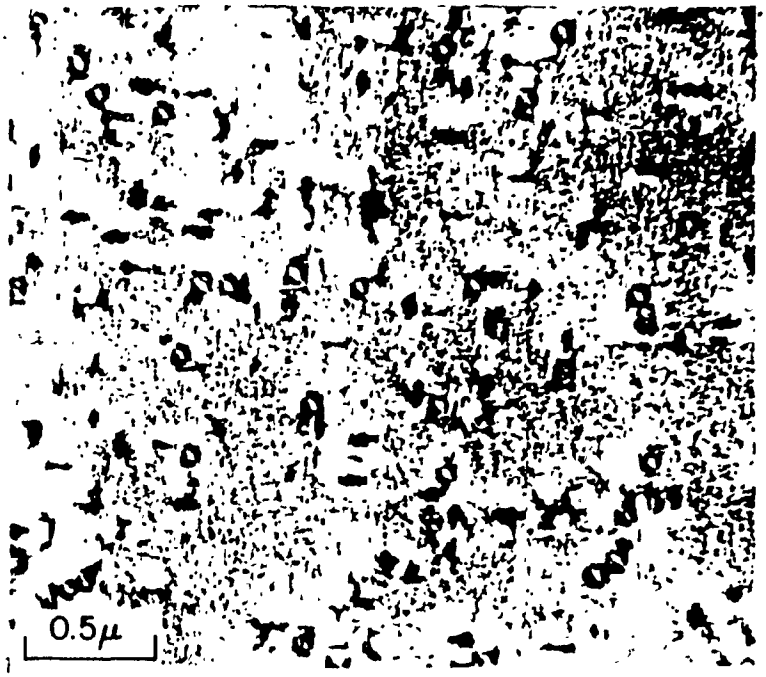
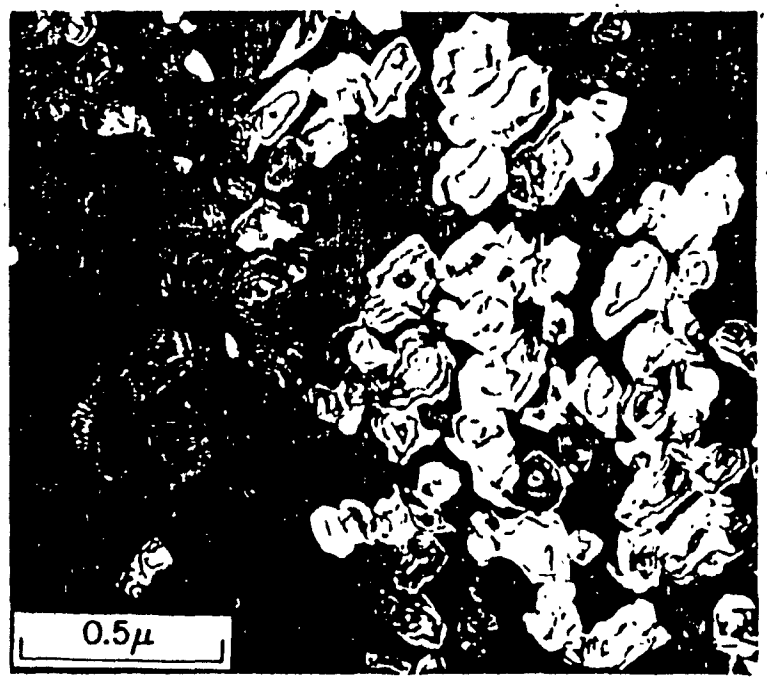


Fig. 24 : Thin foil micrograph showing β' precipitates in a β matrix in Ti-20 at.%V-8 at.%Zr alloy.

Fig. 25 : Heterogeneous nucleation of α-phase at dislocations and subboundaries in Ti-11.5Mo-4.5Sn-6Zr quenched from 900°C and aged at 480°C for 5 min. The white, mottled, background structure is ω-phase.



A

B

Fig.26 : (a) "Blocky" α which has formed in situ from ω-phase. Ti-8 wt.%Fe solution treated at 950°C and directly aged at 500°C.

(b) Dark field electron micrograph showing uniformly distributed α-plates which have nucleated at β:ω interfaces.

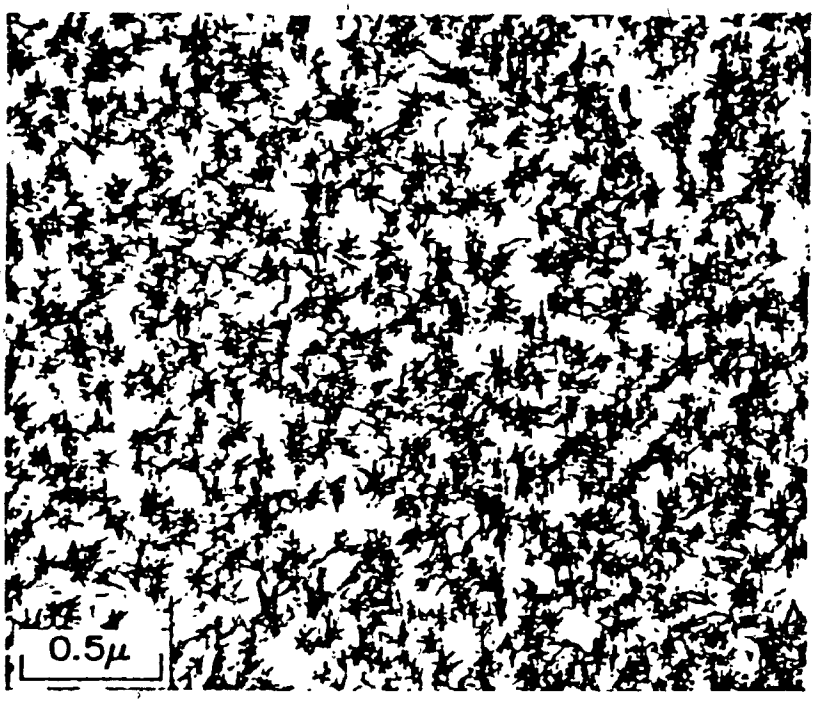


Fig.27

Fig.28

Thin foil micrographs showing the two morphologies which result from the β' → α reaction.

Fig.28: rafts of small α-particles within prior β' regions,
Fig.27: needles of α which have nucleated within the β' regions.

beta stabilizers and with little or no aluminum, alpha appears as lenticular plates or rafts without the normal Burgers orientation to the beta matrix.(see fig.27, 28)(3)(4)(5)(9)(11)(12)

One last form of decomposition of the high temperature beta phase is possible, this is the non martensitic decomposition of eutectoid alpha + beta alloys directly to alpha plus intermetallic compound. The relative activity of this reaction varies for different alloying elements and tends to decrease proportional to eutectoid temperature (see fig.29). In the sluggish eutectoids the reaction is extremely slow and retention of metastable beta is easy, in active eutectoids however the speed of the reaction makes it impossible to prevent precipitation. The decomposition of these alloys results in the formation a very fine lamellar constituent, analogous to pearlite (see fig. 30). In alloys isothermally reacted below the temperature at which this pearlitic transformation occurs, a non-lamellar eutectoid structure similar to bainite is formed. This structure, despite its reduced ductility appears to offer attractive mechanical properties if properly heat treated and research to this end is continuing.(see fig. 31) (3)(4)(5)(13)(14)(15)(16)

The various microstructures described in the previous sections may occur by themselves or in combination with various other phases, and quite often the optimum mechanical properties occur as a result of a mixture of phases. In the nucleation and growth microstructures the optimum combination of strength and toughness occurs when a mixed structure of 90% Widmanstatten and 10% equiaxed microstructures has been achieved. While in the case of martensite, good strength and ductility levels are achieved when tempering results in a fine alpha + beta structure. The other phases mentioned, the precipitates such as euctectoids, omega and beta prime show an ability to increase tensile strength through precipitation hardening but generally adversely effect ductility and have no elevated temperature stability making their range of application fairly limited.

All of the alpha + beta alloys have poor deep hardenability and for sections thicker than one inch that must be hardened it is necessary to use beta alloys.

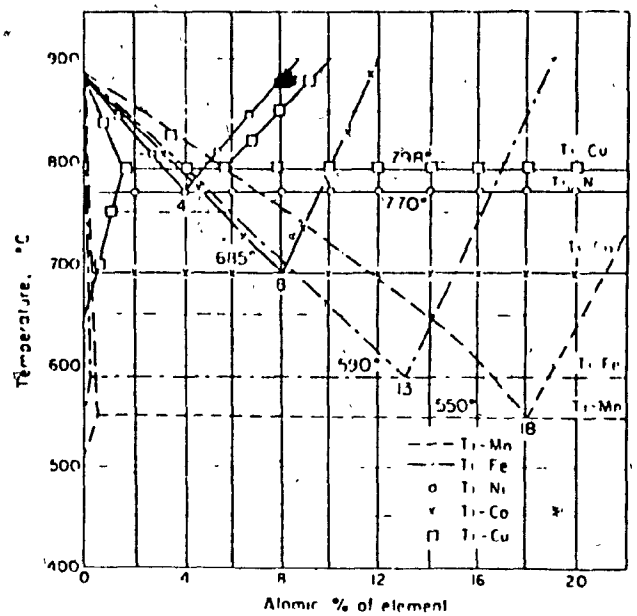


Fig. 29 Titanium systems with limited β- and α-solid solutions and eutectic and eutectoid reactions (Cu, Mn, Fe, Co, Ni).

Fig. 30 : Thin foil micrograph of lamellar eutectoid constituent consisting of α + Ti_2Cu in a Ti-8 wt.%Cu alloy.

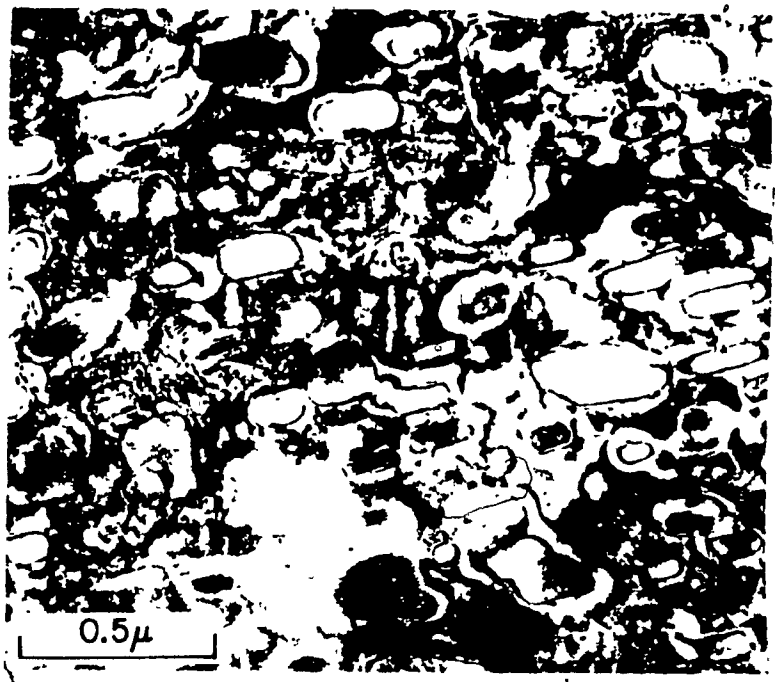


Fig. 31 : Thin foil micrograph of non-lamellar $\alpha + Ti_2Co$ eutectoid decomposition product.

CHAPTER 3

MECHANICAL SHAPING OF TITANIUM

Titanium and its alloys are not generally regarded as "well behaved" when compared to other structural metals and as a result the shaping of this metal through the various standards techniques of forging, extrusion and rolling has proved to be troublesome. The high ductility, high strength even at elevated temperatures, as well as the other useful properties of titanium are in large part dependent upon the texture, grain size and phase composition of the finished product, and these in turn are quite often dependent upon the entire thermomechanical treatment the metal has received. Even slight variations in this thermomechanical treatment can result in an altered microstructure and a wide scattering of properties between supposedly identical parts,low cycle fatigue strength and crack tolerance being particularly sensitive.(3)(24)

On top of this the earlier mentioned chemical reactivity of titanium at elevated temperatures and the detrimental effects of oxygen, nitrogen and hydrogen absorbed as a result, present problems of their own. The use of noble gas and vacuum processing as well as protective coatings and removal of contaminated metal by machining or pickling have proved quite successful but have made the handling of the metal more complex and expensive.(17)(18)(19) Still another problem affecting the hot working of titanium alloys is the low thermal conductivity and diffusivity of titanium. As a result the considerable quantities of internal heat generated during deformation are unable to diffuse throughout the workpieces and the resulting "hotspots" adversely affect crystal structure. Efforts must therefore be taken to insure even deformation of all titanium alloys undergoing hot working.(1)(20)

In general it has been found that through appropriate types and degrees of deformation combined with and followed by appropriate heat treatments it is possible to quite accurately control grain structure in titanium. It is however necessary to control all aspects of processing and their effects on crystal structure, as even the type, and size of grains within the ingot, prior to processing are important. In the sections to follow, the various types of hot working used on titanium, as well as the techniques of macro

and microstructural control for each hot working process will be described and discussed.

3.1 FORGING

During the initial development of forging techniques for titanium alloys concern was directed only to static properties, with limited attention being paid to control of microstructure in either the billet or the finished forging. However as a result of failures which occured under certain testing and service conditions, the critical relationship between internal structure and dynamic properties was driven home. The resulting changes in forging and heat treatment techniques provided a high degree of control and were substantially responsible for the increased quality and reproducibility of titanium alloy forgings. Now with the development of the new generation of aircraft and engines, have come still further demands for improved properties and their control through optimized microstructure. It has therefore become necessary to not only develop new improved alloys but also to examine and control microstructure, forging and heat treatment at every step of manufacture.(21)(24)

This control begins at the earliest stages of development, which due to the empirical nature of forging involves several trial forgings from varying preform shapes using different temperature ranges, and followed by laborious cutup testing. The optimized forging route, with all its variable factors is, once determined, carefully recorded as it must be accurately reproduced to duplicate desired properties.(21)(24)

The forging route is however also influenced by the properties of the material to be forged. From each and every billet received, slices are cut and examined, carefully, to insure uniformity of structure, freedom from porosity and any other defects that could persist through forging. At the same time the microstructure is examined and the beta transus found. From this data, the optimized forging route working temperature for each billet, in degrees above or below the beta transus, is calculated. The accuracy of the entire process is continually checked through the selection at regular intervals, of completed forgings for cut up, testing and examination.(21)

Of prime importance in the entire process is insuring that preheating and heat treatment temperature are very accurately controlled. As a result the heating rates and soaking times for a particular forging are carefully predetermined, and the temperature in the furnace controlled to +5°C Also important in terms of temperature control is the internal heat generated by deformation, this should at all times be kept at a minimum and if at all possible uniformly distributed throughout the forging. Failure to do so may result in hot spots and their accompanying changes in crystal structure (see Fig.42). Overall temperature control is particularly important in those forgings where it is desired to retain a specific percentage of the primary alpha phase. In this case the maximum temperature of the forging due to all forms of heating may not exceed the temperature at which the specified amount of alpha remains.(21)(27)(29)

In forging processes the addition of lubricants, makes deformation substantially easier even at elevated temperatures. In titanium alloys the lubricants used serve two purposes not only do they reduce friction but they also cut back on absorption of oxygen, nitrogen and hydrogen. As a result not only are the surface cracking problems due to brittleness reduced but die wear due to the hardened gaseous alloyed surface layer is substantially decreased.(19)(24)(27)(29)

Once the forging has been completed it is heat treated to bring out the required microstructure with its desired properties. Even here however care must be taken as titanium shows a tendency to sag under its own weight at elevated temperatures. For large and uniform sections this is not a problem but for thin variable ones such as compressor blade forgings, considerable twisting, bowing and sagging may occur. Correction of this problem is not possible in the cold state due to the limited workability of titanium alloys when cold but may be carried out at the solutioning or annealing temperature, or prevented through the use of special jigs. Forgings having contaminated surfaces are likely to crack during straightening so contamination must be removed beforehand.(27)

Two types of forging methods are used on titanium alloys, conventional forging using warm dies, and isothermal forging using dies heated to the same temperature as the work piece. Conventional forging is more extensively used as it may be carried out on the same type of equipment as

forgings for other metals, and because it is the older and more established process.

Conventional forging may itself be further subdivided into two categories. The first of these is alpha + beta forging which is carried out below the beta transus. It is used for those alloys, alpha and near alpha, that undergo excessive grain growth in the beta phase field, as well as for those forgings for which a very fine equiaxed microstructure has been specified. The temperature range available for alpha + beta forging is small, the upper temperature being high in the alpha + beta range and the lower limit being set by either the stiffness of the alloy or the onset of shear cracking. As a result frequent reheating of the workpiece is necessary with consequently increased risk of overheating and contamination.(3)(20)(27)

The microstructure produced by alpha + beta forging is a combination of equiaxied (primary) alpha and transformed beta (Widmanstatten alpha). The percentage of each being dependent upon the maximum forging temperature used, which controls the amount of primary alpha retained in the forging. As can be seen from fig.32 for the alloy Hylite 50 (Ti-4Al-40Mo -2Sn-.5Si) increasing amounts of Widmanstatten alpha can improve some properties and degrade others. In the case of this alloy the minimum permissible content of primary alpha is 20 percent for acceptable properties, more may however be retained if other combinations of properties are desired. In fig.23 are shown the properties of the alloy IMI679 (Ti-11Sn-5Zr-2.5Al-1Mo-2Si) that does not respond as favourably to increasing Widmanstatten alpha content and in this case 25 or 35 percent primary alpha would be needed for acceptable properties.(3)(20)(30)

The second type of conventional forging is beta forging. This term is however not completely accurate, although the workpiece is initially heated and worked above the beta transus, as a result of cooling, working is actually completed well below the transus. Even so, the effects of beta forging can be very noticeable in certain alloys, such as those with appreciable quantities of beta-stabilizers, where the resulting microstructures are very different from those of alpha + beta forgings. It has therefore become the practice to ensure that at least 75 percent of forging reduction is carried out below the beta transus. The single

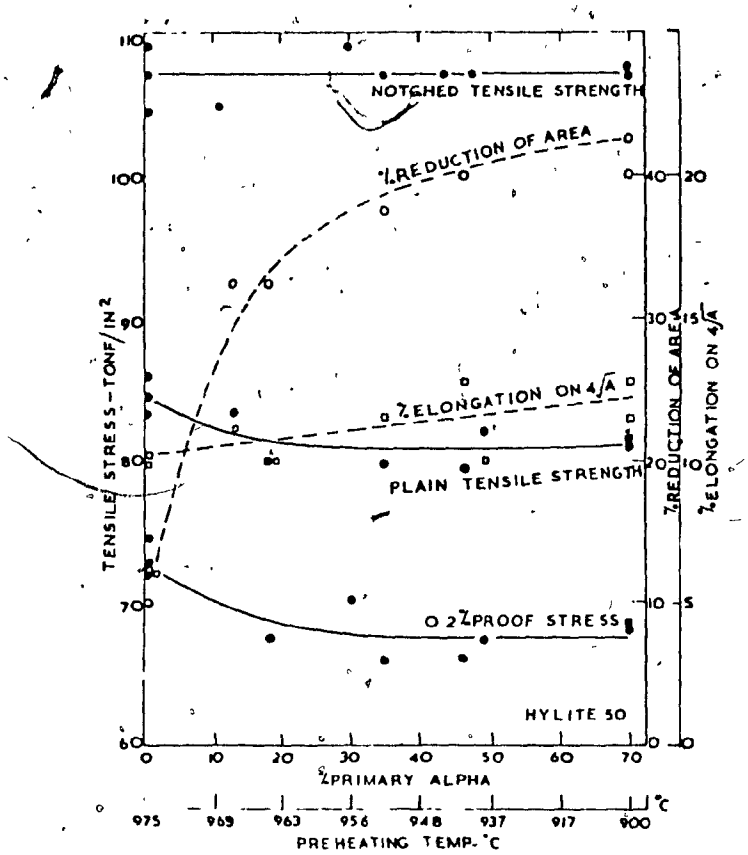


FIG. 32 Illustrating the variation in plain and notched tensile properties of fully heat-treated Hylite 50 bar with decreasing amounts of original alpha in the microstructure.

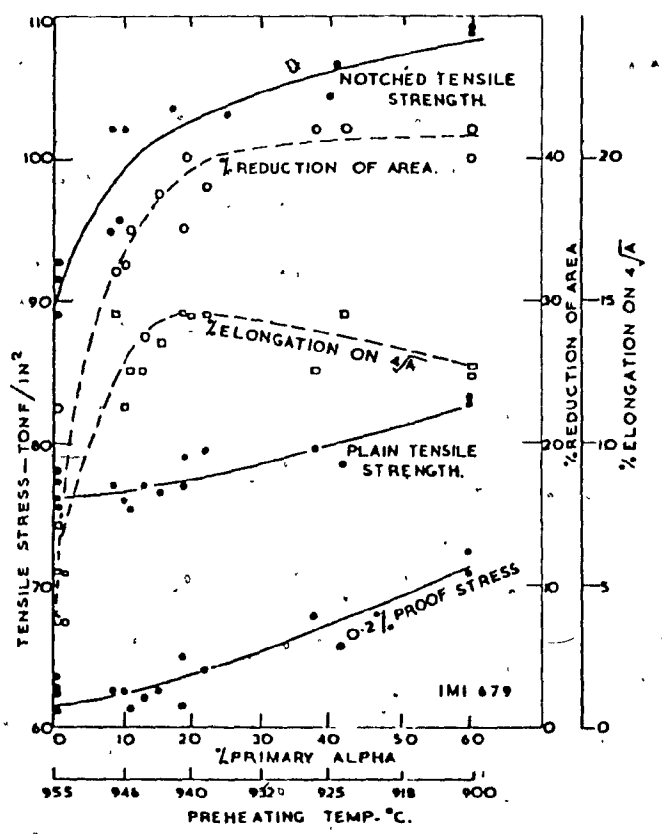
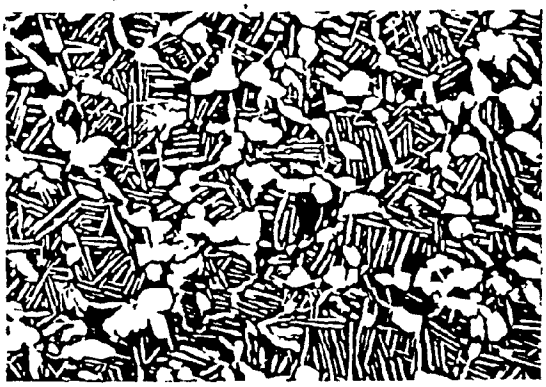


FIG. 33 Illustrating the variation in plain and notched tensile properties of fully heat-treated IMI.679 bar with decreasing amounts of original alpha in the microstructure.

(a)



(b)

Fig. 34 (a) and (b) Typical fully heat-treated microstructures (longitudinal sections) containing 20% original alpha. (a) IMI.679. (b) Hylite 50. × 500.

greatest advantage of beta forging is that it provides a much greater temperature range for working than is possible with alpha + beta forging. This makes necessary fewer forging and annealing steps for complete shaping and lower flow stresses, which result in better filling of the die and closer tolerances on the finished product. Although the beta forging of titanium alloys is viewed with a certain amount of suspicion by some users of titanium, it has been shown that for certain combinations of properties it can be as or more effective than alpha + beta forging.(3)(20)(24)(27)

From work done by Kuhlman and Billman(20) it is possible to compare the effects of different types of processing on two commonly used alpha + beta alloys. These alloys, their compositions and the type of processing used are to be found in table 3, while the resulting properties are shown in table 4 and microstructures in figs. 35 to 41. In general it can be seen that macrostructural refinement is most efficiently achieved through alpha + beta forging followed by a beta anneal to permit recrystallization. This is due to the fact that the size of newly formed grains after the beta anneal are inversely proportional to the amount of alpha + beta deformation. In the case of working in the beta field the critical strain limits need to achieve recrystallization, with accompanying grain refinement, are never reached because dynamic recovery occurs readily in the single phase structure.

Microstructure is similarly affected by amount and temperature of deformation as well as cooling rate and any subsequent thermal treatment. This can be clearly seen in the different microstructures developed due to the different processing options. Conventional processing results in equiaxed primary alpha in a transformed beta (Widmanstatten alpha) matrix with high tensile strength and ductility but only moderate fracture toughness. Option 1 substitutes a two step recrystallization anneal for the mill anneal, the resulting 80 to 90 percent equiaxed alpha microstructure is coarser than that for conventional processing with slightly reduced tensile strength but improved fracture toughness. In option 2 a beta solution treatment followed by an alpha + beta anneal is substituted for the mill anneal. The beta solutionning destroys any microstructural refinement resulting from the alpha + beta working by recrystallizing the beta grains, which on cooling form Widmanstatten alpha. The result is further decreased tensile strength and ductility but increase of fracture

Table 3 — Available Options for Obtaining High Fracture Toughness in Ti-6Al-4V and Ti-6Al-6V-2Sn Forgings

Composition Options*	Al	Sn	V	Fe	O_2	C	N	H	Cu	Other
Standard										
Ti 6Al-4V	5.5-6.75	—	3 5-4.5	0.30	0.20	0.10	0 05	0.015	—	0.40
Ti-6Al 6V-2Sn	5 0-6.0	1.5-2 6	5 0-6.0	0.35 1.0	0 20	0.05	0 04	0.015	0.35-1.0	0 30
Special										
Ti-6Al-4V	5 5-6 2	—	3 5 4.5	0.25	0.13	0.08	0 05	0 0125	—	0.30
Ti-6Al 6V-2Sn	5.0 6 0	1 5-2.5	5 0-6.0	0 35 1.0	0 13	0 05	0.04	0 015	0 35 1.0	0 30

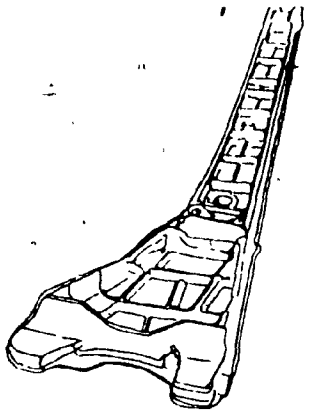
* Alloying element contents are maximum unless expressed as a range.

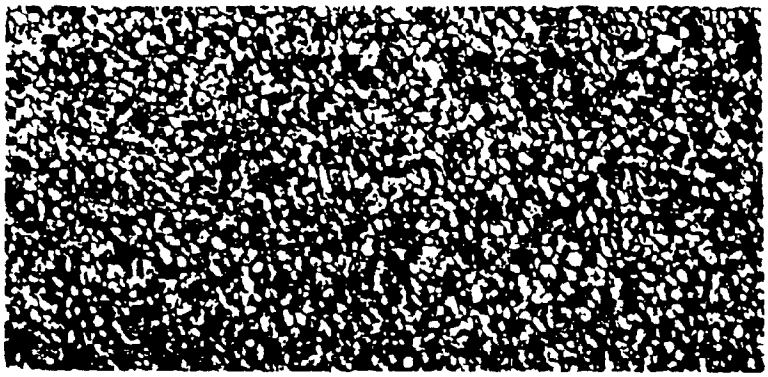
Processing Options	Composition	Forging Process	Heat Treatment
Conventional	Standard	$\alpha + \beta$	Mill anneal (MA)
1a	Standard	$\alpha + \beta$	Recrystallization anneal (RA)
1b	Special	$\alpha + \beta$	Recrystallization anneal (RA)
2a	Standard	$\alpha + \beta$	β Solution treatment + anneal (β A)
2b	Special	$\alpha + \beta$	β Solution treatment + anneal (β A)
3a	Standard	β preform, $\alpha + \beta$ block & finish	Mill anneal (MA)
3b	Standard	β preform β block, $\alpha + \beta$ finish	Mill anneal (MA)
4	Special	β preform β block, $\alpha + \beta$ finish	Duplex solution treatment + anneal (duplex STAN)

Table 4 — Typical Guaranteed Properties for Composition/Processing Options

Alloy and Option	Minimum Room-Temperature Tensile Properties: Yield Strength, MPa (10^3 psi)	Tensile Strength, MPa (10^3 psi)	Elongation, %	Reduction in Area, %	Fracture Toughness, MPa$\sqrt{m}$ (ksi$\sqrt{in.}$)
Ti-6Al-4V					
Conventional	827-862 (120-125)	896-931 (130-135)	10	20-25	44-55 (40-50)
1a	793-827 (115-120)	862-896 (125-130)	10	20-25	60-71 (55-65)
1b	758-793 (110-115)	827-862 (120-125)	10	20-25	77-88 (70-80)
2a	793-862 (115 125)	931-1000 (135-145)	6-8	10-15	82-93 (75-85)
2b	758-827 (110-120)	827-862 (120-125)	6-8	10-15	88 99 (80 90)
3a	813-848 (118-123)	896-931 (130-135)	10	20-25	66-77 (60-70)
3b	813-848 (118-123)	862 896 (125 130)	8-10	15 20	71-88 (65 80)
4	758-793 (110-115)	827-862 (120-125)	6-8	10-15	82-93 (75-85)
Ti-6Al-6V-2Sn					
Conventional	931-965 (135 140)	1000-1034 (145-150)	8 10	20-25	33-44 (30-40)
1a (from reference 4)	896-965 (130-140)	965-1034 (140-150)	8 10	20-25	44-55 (40-50)
1b	—	—	—	—	—
2a	862 931 (125-135)	965-1034 (140-150)	6-8	10-12	55-66 (50-60)
2b	—	—	—	—	—
3a	896 931 (130 135)	1000-1034 (145-150)	8-10	15-20	44-55 (40-50)
3b	896-931 (130 135)	965-1034 (140 150)	6-8	12-20	55-66 (50 60)
4	—	—	—	—	—

Fig.35 — Conventionally Processed Ti-6Al-4V Wingspar Forging



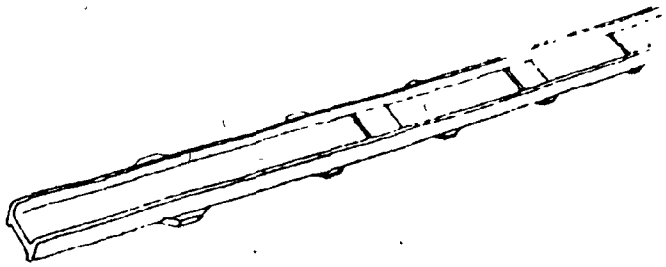


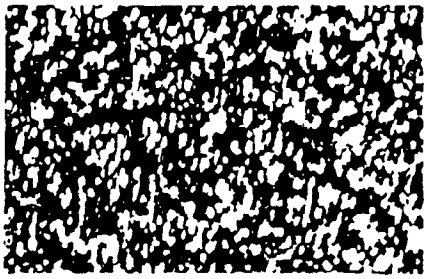
2660 mm (105 in.) long, 193 kg (426 lb) 100×

Orientation	Option	Room Temperature Tensile Properties: Yield Strength, MPa (10^3 psi)	Tensile Strength, MPa (10^3 psi)	Elongation, %	Reduction in Area, %	Fracture Toughness, MPa√m (ksi√in.)*
Longitudinal	Conventional	938 1000 (136 145)	979 1062 (142 154)	15 18	29 40	49 9 56 4 L T (45 4 51 3)
Long transverse	Conventional	938 985 (136 141)	958 1034 (139 150)	16 19	30 39	54 9 57 4 S T (49 9 52.2)
Short transverse	Conventional	903 958 (131 139)	958 1014 (139 147)	13 19	23 41	51 6 52 6 T L (46 9 47.5)

* K_{Ic} specimen orientation per ASTM E399

Fig.36 — Conventionally Processed Ti-6Al-6V-2Sn Wingspar Forging



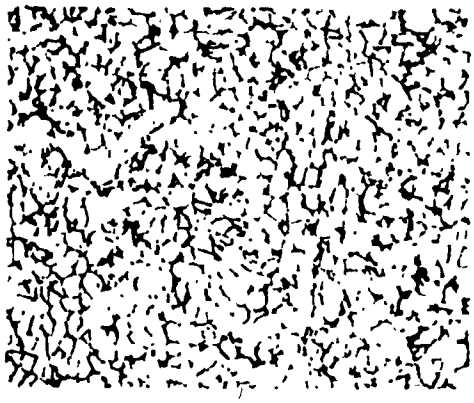


3640 mm (143 in.) long, 143 kg (316 lb) 100×

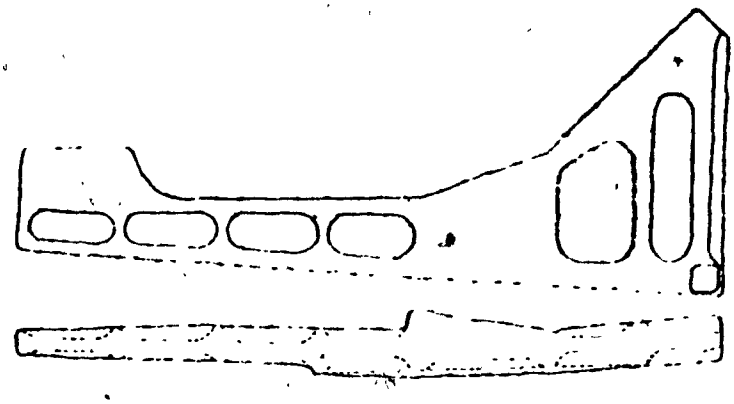
Orientation	Option	Room Temperature Tensile Properties: Yield Strength, MPa (10^3 psi)	Tensile Strength, MPa (10^3 psi)	Elongation, %	Reduction in Area, %	Fracture Toughness, MPa√m (ksi√in.)*
Longitudinal	Conventional	1063 1125 (154 163)	1139 1194 (165 173)	17 19	23 40	—
Long transverse	Conventional	1090 1139 (158 165)	1145 1201 (166 174)	14 18	20 31	—
Short transverse	Conventional	1035 1063 (150 154)	1111 1145 (161-166)	13-17	20 29	—

* K_{Ic} specimen orientation per ASTM E399 Fracture toughness not routinely measured, but is typically 33 44 MPa√m (30 40 ksi√in.)

Fig. 37 — Option 1b Ti-6Al-4V Support Beam Forging



100X



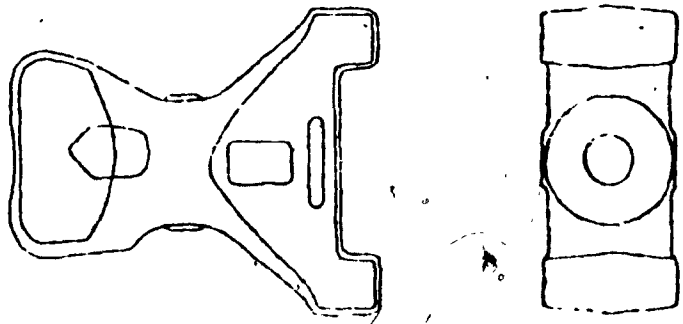
1955 mm (77 in.) long, 442 kg (974 lb)

Orientation	Option	Room Temperature Tensile Properties: Yield Strength, MPa (10^3 psi)	Tensile Strength, MPa (10^3 psi)	Elongation, %	Reduction in Area, %	Fracture Toughness, MPa$\sqrt{m}$ (ksi$\sqrt{in}$)*
Longitudinal	1b	814 834 (118 121)	896 910 (130 132)	16 18	42-48	83 2 93 9 L T (75 6 85 4)
Long transverse	1b	814 862 (118 125)	883 924 (128 134)	16 17	40 42	84 5 123 9 S T (76 8 112 6)
Short transverse	1b	777 807 (112 117)	869 910 (126 130)	10 13	30 34	

* K_{Ic} specimen orientation per ASTM E399.

Fig. 38 — Option 2a Ti-6Al-4V Helicopter Sleeve Forging

100X



508 mm (22 in.) long, 432 mm (17 in.) wide, 25 to 201 mm (1 to 8 in.) thick, 128 kg (282 lb)

Orientation	Option	Room-Temperature Tensile Properties: Yield Strength, MPa (10^3 psi)	Tensile Strength, MPa (10^3 psi)	Elongation, %	Reduction in Area, %	Fracture Toughness, MPa$\sqrt{m}$ (ksi$\sqrt{in}$)*
Longitudinal	2a	841-952 (122-138)	972 1041 (141-151)	7 10	10-13	—
Long transverse	2a	896 931 (130-135)	993 1083 (144-157)	8-10	10-14	—
Short transverse	2a	883 917 (128 133)	986-1055 (143-153)	7-11	13 17	—

* K_{Ic} specimen orientation per ASTM E399. Fracture toughness not guaranteed for this part, but 82-93 MPa$\sqrt{m}$ (75-85 ksi$\sqrt{in}$) is typically produced.

Fig. 39 — Ti-6Al-4V Wingspar Forgings

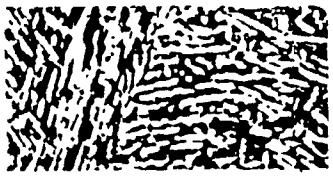
Orientation	Option	Room Temperature Tensile Properties: Yield Strength, MPa (10³ psi)	Tensile Strength, MPa (10³ psi)	Elongation, %	Reduction in Area, %	Fracture Toughness,* Mpa√m (ksi√in.)*
Longitudinal	3a	89[illegible]-925 (129-134)	959-987 (139-143)	12-17	25-42	63-4-81 L-T (58.5-73.7)
	3b	842-925 (122-134)	918-980 (133-142)	11-14	21-27	65-95.7 (59-87)
Long transverse	3a	842-932 (123-135)	918-987 (133-143)	14-17	30-41	68.1-69 S-T (61.9-62.7)
	3b	863-925 (125-134)	932-973 (135-141)	12-14	21-32	83-3-89 (75.7-80.8)
Short transverse	3a	814-876 (118-127)	904-959 (131-139)	11-16	27-30	69-2-79.3 T-L (62.9-72.1)
	3b	814-869 (118-126)	883-938 (128-136)	8-12	15-28	80.7-102 (73.4-92.7)

* K_{Ic} specimen orientation per ASTM E399

All 100X

Option 3a, α + β work — High level

Modest level



Option 3b, α + β work — Moderate level

Low level

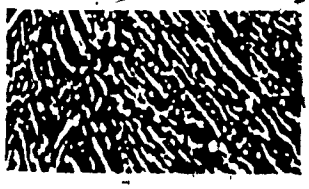
Fig. 40 — Ti-6Al-6V-2Sn Wingspar Forgings

Orientation	Option	Room Temperature Tensile Properties: Yield Strength, MPa (10³ psi)	Tensile Strength, MPa (10³ psi)	Elongation, %	Reduction in Area, %	Fracture Toughness,* Mpa√m (ksi√in.)*
Longitudinal	3a	1007-1056 (145-153)	1083-1104 (157-160)	10-11	19-25	54-8-57.6 L-T (49.8-55.4)
	3b	1007-1042 (146-151)	1104-1145 (160-166)	6-11	17-21	66-2-86.8 (60.1-78.9)
Long transverse	3a	1007-1041 (145-148)	1053-1090 (154-158)	14-17	21-34	— S-T
	3b	945-966 (137-140)	1070-1076 (155-156)	8-11	13-20	—
Short transverse	3a	1021 (148)	1090 (158)	14	36	56-7-59.4 T-L (51.5-54.0)
	3b	973 (141)	1076 (156)	13	26	67.4-70.8 (61.3-64.4)

* K_{Ic} specimen orientation per ASTM E399

All 100X

Option 3a, α + β work — High level



Modest level

Option 3b, α + β work — Moderate level

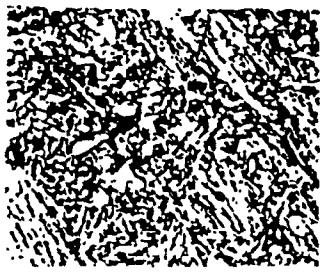
Low level

Fig. 41 – Ti-6Al-4V Forgings

Orientation	Option	Room-Temperature Tensile Properties: Yield Strength, MPa (10^3 psi)	Tensile Strength, MPa (10^3 psi)	Elongation, %	Reduction in Area, %	Fracture Toughness, MPa$\sqrt{m}$ (ksi$\sqrt{in.}$)*
Longitudinal	4	855-903 (130-131)	1000-1007 (145-146)	12	12-27	98.6 T-L (89.6)
Long transverse	4	862-910 (125-132)	972-1007 (141-146)	12	22-42	90.2 S-T (82.0)
Short transverse	4	848-862 (123-125)	952-979 (138-142)	12	28-36	82.1-99.2 T-S (74.6-90.2)

* K_{Ic} specimen orientation per ASTM E399.

Option 4 100X



1000X

toughness. Option 3 involving differing amounts of beta forging (a), (b) followed by alpha + beta forging and a mill anneal, showed the importance of extensive alpha + beta forging for optimum properties. In 3a only the preform was beta forged and as a result equiaxed primary alpha was formed, while in 3b both the preform and the blocking were beta forged and the result was the formation of high aspect ratio elongated alpha. Of these the equiaxed alpha because of grain refinement, had better tensile strength and ductility but slightly lower fracture toughness. In any case forging above the beta transus resulted in yield strengths and ductility below the maximum attainable and increased property variations, across the forging. Option 4 had exactly the same forging option as 3b but used a special composition alloy and a double or duplex solution treatment prior to annealing. The duplex solutioning helps to overcome some of the microstructural nonuniformities which result from extensive beta forging followed by limited alpha + beta working. The resulting structure is essentially a transformed beta (Widmanstatten alpha) matrix with occasional primary (equiaxed) alpha grains scattered throughout. The tensile strength and ductility of this structure are inferior to those of 3b while the fracture toughness is among the highest that may be achieved.

The above comparison of mechanical properties versus type of processing points out the wide range of properties available and how they may be selected and optimized through control of forging and heat treatment. At the same time, attention must be paid to the design of forgings, mechanical pretreatment of stock and rate of deformation. Failure to do so can result in defects or inferior microstructures in the finished product. Some examples of such defects are shown in Figures 42 to 53, along with explanations.

The second method of forging, isothermal forging, does not result in substantially different microstructures, it simply makes use of the certain strain rate phenomena to make easier the forging of titanium alloys. Isothermal forging is accomplished in practice by maintaining the forging dies and the workpiece at the same temperature throughout the forging process. In conventional forging the difference in temperature between die and workpiece and the low thermal conductivity of titanium combine to cause localized chilling of the workpieces. This chilling increases stress and can cause cracking unless forging is stopped and the workpiece reheated. In isothermal forging there is no chilling and so the forging may be

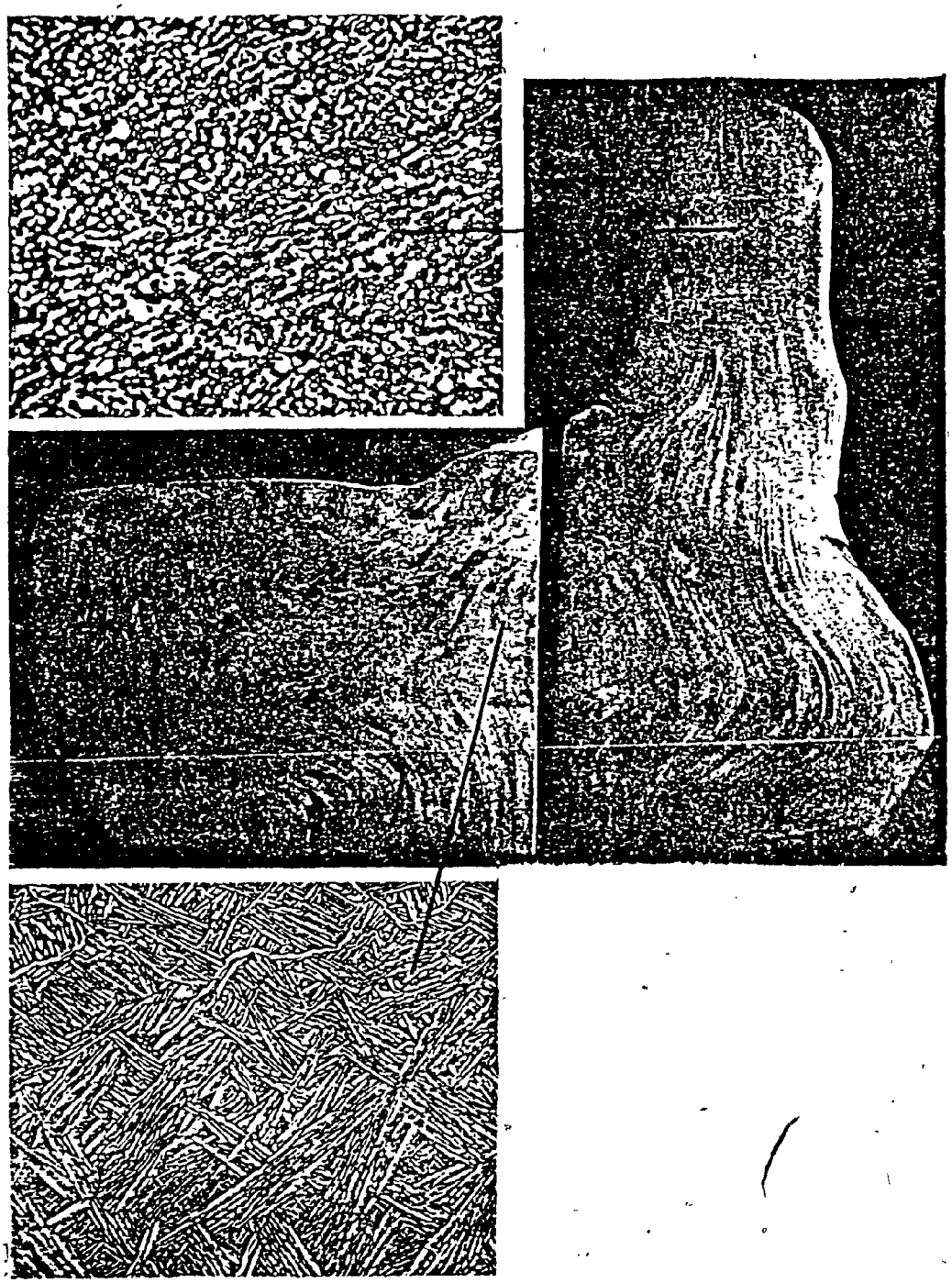


Fig.42 Effect of localized overheating due to temperature build up during forging. In this case the beta transus has been exceeded and there is no primary alpha remaining in the structure. Cooling has resulted in the precipitation of alpha in continuous form at the original beta grain boundaries and as basket-weave Widmanstatten alpha within the grains.

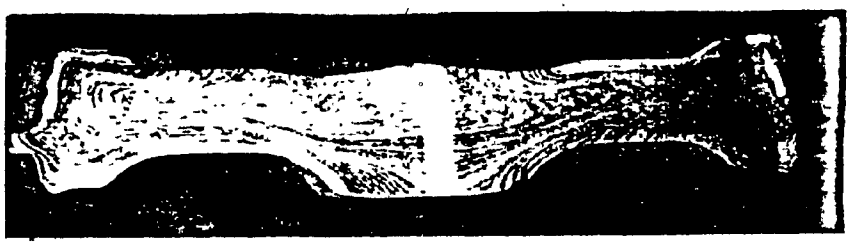


Fig.43 Non-symmetrical grain flow in a 20 inch diameter disc pressing. When the billet length to diameter ratio exceeds 8:1 the billet axis tends to move off center as the upsetting operation proceeds. This problem can be minimized by using forging stock with a length to diameter ratio between 1½ and 2 to 1, and by rotating the billet between blows.

Fig.44 Initial 10 inch upset blank showing "dead metal" zone

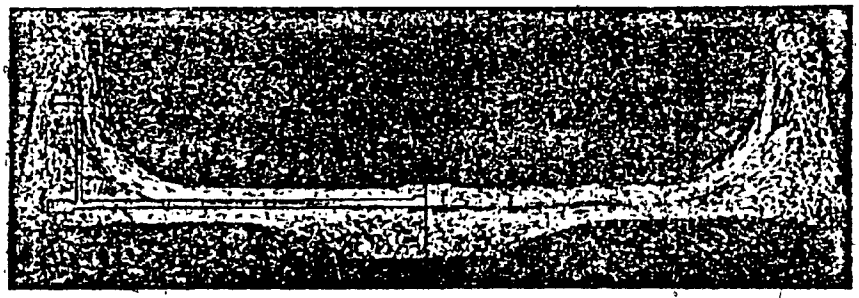


Fig.45 Effect of "dead metal" zones developed during original upsetting of billet, fig.44, on microstructure of a 20 inch diameter press forged disc in Ti-4Al-4Mn. This problem has been minimized as a result of the much finer grain structures now demanded in billets. It is however taken in to consideration, and forgings are designed in such a way that these zones are removed during final machining or deformed more uniformly to improve microstructure.

Fig.46 Surface "breakup" on an extruded stem originating from coarse saw cuts on the billet face.

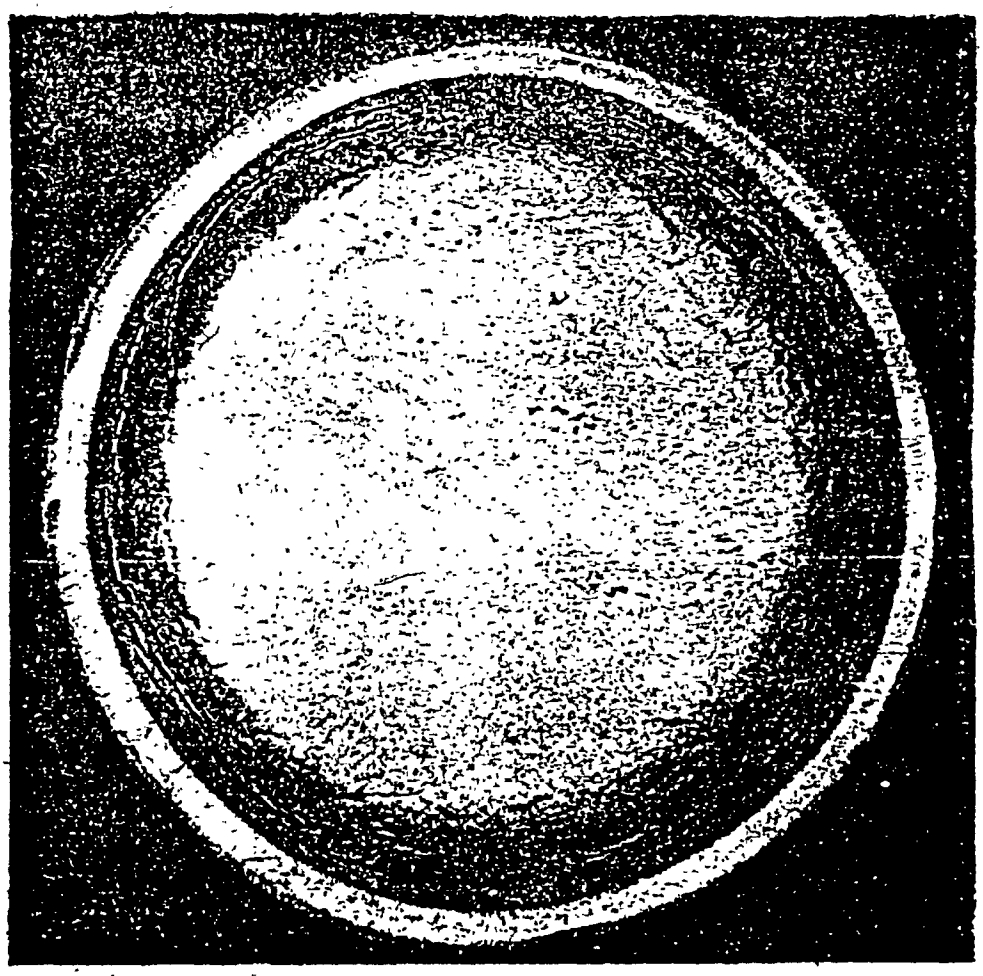


Fig.47 Chill cracked press forging in IMI 679. This problem occurs when forging at the lower end of the forging temperature range, with some alloys being more susceptible than others. It may be minimized through careful attention to handling and insulating techniques.

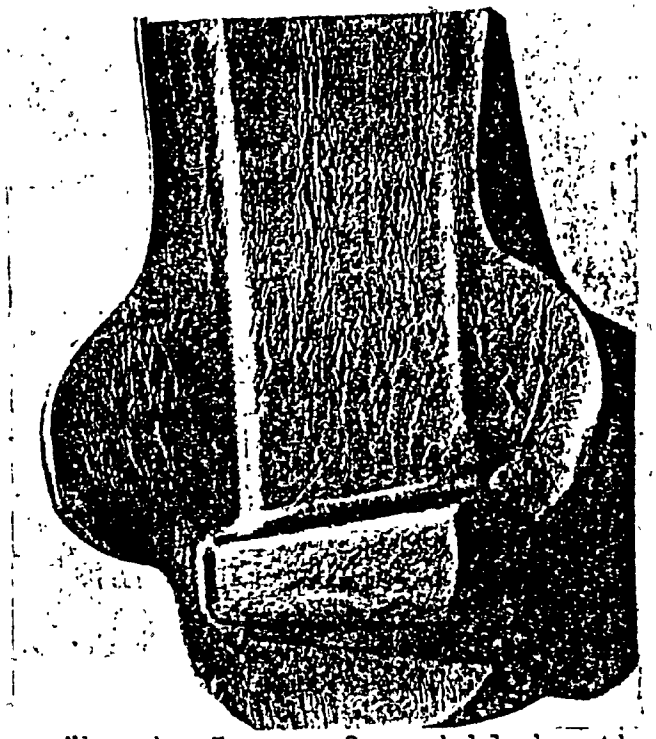


Fig.48 Surface "breakup" on a forged blade, the result of localized surface shear which occurs after lubricant breakdown during forging.

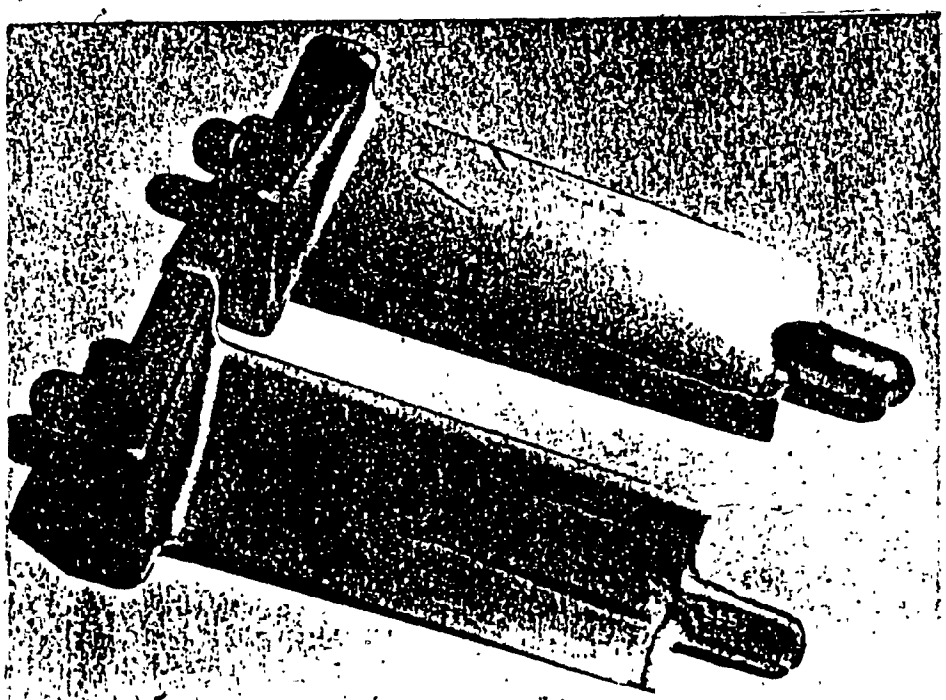


Fig.49 Longitudinal streaks produced by heavy deformation such as single blow finished forgings. Streaks occur at points where pronounced flow emerges at the surface.

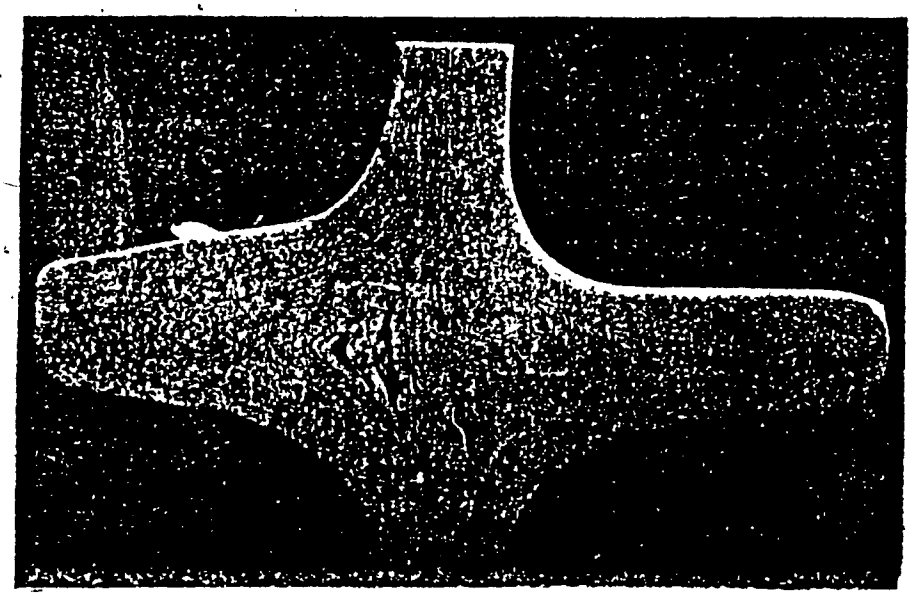


Fig.50 Segregated streaks in snubber section of Ti-6Al-4V blade forging.

Fig.51 High proportion of alpha phase in streak shown in Fig.50. (x315)

Fig.52 General microstructure in snubber section of blade shown in Fig.50. (x315)

In this case aluminium content in the defect region was 50% higher than normal

Fig.53 Example of type of streak defect which occurs in silicon containing titanium alloys. The streak shown consists of numerous titanium silicide particles, the concentration being more than three times that of the normal alloy.

carried out in one step and at slower ram speeds than conventional forging. In titanium which is strain rate sensitive this slower forging speed substantially reduces flow stress requirements as can be seen from Fig. 54. At the same time isothermal forging has been developed in such a way as to retain the good flow characteristics of superplastic forming while operating at strain rates above those usually associated with superplastic behaviour. This is done by insuring that all workpieces have the fine equiaxed microstructure necessary for superplasticity. As a result of the easier flow associated with superplastic behaviour thinner forgings, with improved property uniformity and higher strengths are possible, with accompanying savings in material, time, and energy.(3)(22)(23)(25)

Along with these advantages, however, come problems unique to operating at tempertures of 900 to 1000°C. At these temperatures the only die materials that have proved satisfactory are the superalloys such as IN 100 developed for use in the hottest portions of jet engines. Similarly the lubricants used in conventional forgings such as glasses, molten salts, and graphites have proved successful to a degree but where surface finish is important only boron nitride has proved suitable.(3)(22)(23)(25)

All alloys may be worked isothermally in the alpha + beta or the beta field, however little if any isothermal forging is carried out at beta phase temperatures. This is because alloys that are isothermally beta forged have coarse, completely transformed microstructures with inferior tensile strength and ductility which is unaffected by any subsequent heat treatment. In conventional forging it is possible to overcome this problem by forging partly at beta temperatures and partly at alpha + beta temperatures, in isothermal forging this is difficult and offers few advantages. On the other hand isothermal alpha + beta forging results in a fine, grained structure of equiaxed primary alpha in a matrix of transformed (Widmanstatten alpha) beta, the amount of primary alpha being dependent upon the forging temperature, as can be seen from Fig.55. At the same time if it should prove necessary to alter the ratio of primary alpha to transformed beta, solution heat treating is all that is necessary. The mechanical properties resulting from this fine grained structure are as good or, better than those for conventional alpha + beta forgings with generally high tensile and fatigue properties and improved ductility. Other advantages of isothermal forging are almost no anisotropy, no

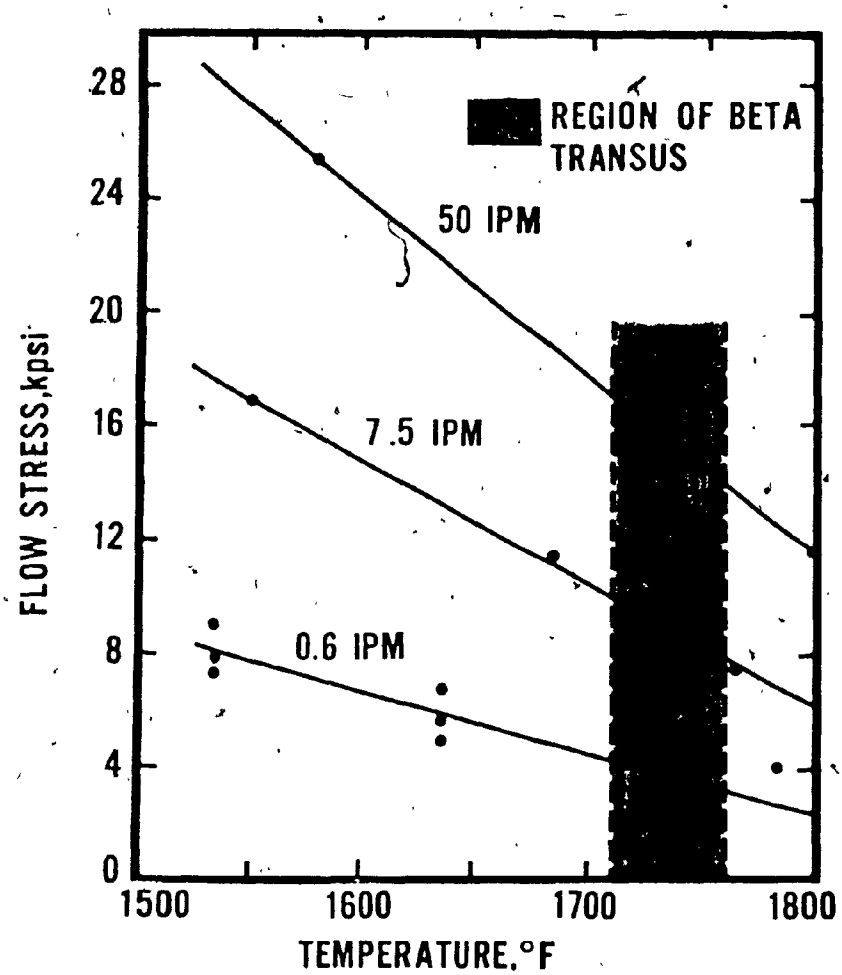


Fig. 54 Effect of strain rate and temperature on flow stress of Ti-6Al-6V-2Sn alloy under isothermal forging conditions.

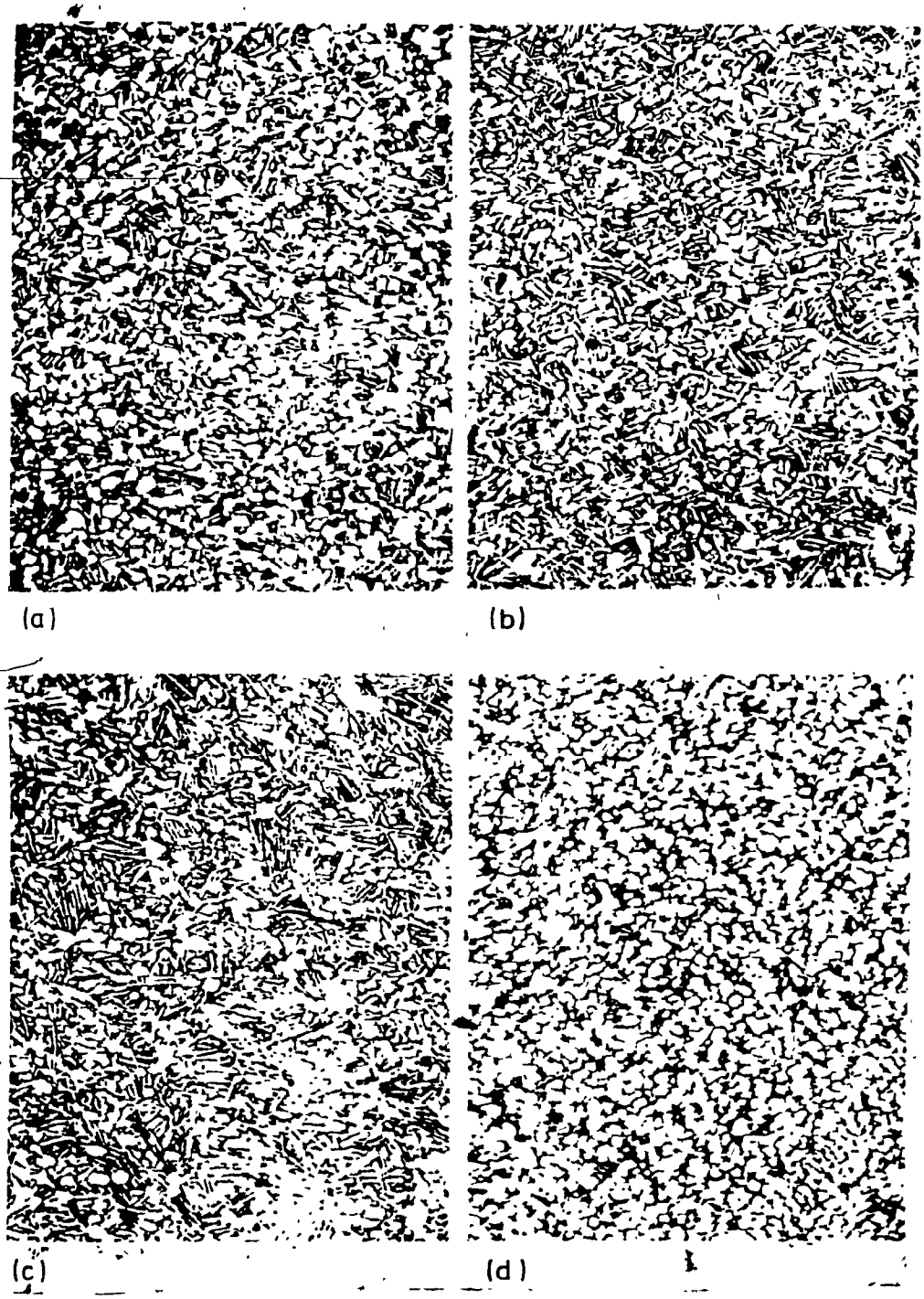


Fig.55 Microstructure of isothermally forged Ti-6Al-4V compressor wheel, a) to c) 950°C, d) 900°C. a) Wheel center, b) and d) half-way out, c) wheel periphery.

brittle alpha phase formation due to absorption of gases and no internal stress left after forging, which makes annealing unnecessary.[3][22][23][25]

A comparison of the two types of forging indicates that isothermal forging, even though it is still comparitively "under development", is the better of the two processes. Forgings produced isothermally are thinner and more accurate, with considerably reduced or non-existant post forging, machining requirements. They are inherently more flaw free and permit better and more accurate control of microstructure than conventional forgings, and the reduced stress levels and material requirements result in considerable savings on capital investment. This cost saving is particularly true of larger components, which as conventional forgings require very high stress levels and as a result very large and powerful presses. These advantages more than outweigh the more complex technology of isothermal forging, such as heated superalloy dies, special lubricants and inert gas forging envelopes. Given the increasing demands for high quality titanium forgings for the new generation of engines and aircraft it is only a matter of time until the majority of titanium forgings are being done isothermally.

3.2 SUPERPLASTIC FORMING

Superplasticity was mentioned in the earlier section as one of the factors or mechanisms involved in isothermal forging, it was not however clearly explained. Basically, superplasticity is the ability of a material to achieve extensive elongation, generally more than 200 percent and as much as 1000 percent, under tensile stress, combined with a high level of resistance to the plastic instability (necking) usually associated with tensile deformation.(31)(32)(33)

This behaviour has been put to use in the process known as superplastic forming to produce from thin sheet titanium and its alloys a wide range of shapes and products while still developing microstructures and properties as good as those of forged products. At the same time because the process results in finished parts requiring less material and little or no machining it has substantially reduced part costs and the material loss ratio's normal to titanium products.(31)(33)

Laboratory testing has revealed which of the titanium alloys are capable of superplastic deformation, and the conditions which are necessary for this to occur. A fine, equiaxed grain structure, preferably two phased and stable at the working temperature is normally required. A single phase microstructure may be used but has a tendency to undergo grain growth which produces undesirable microstructure in the finished product. At the same time the deformation must occur between 0.4 and 0.7 T_m (absolute melting point temperature) and at strain rates of 10^{-6} to 10^{-3} sec^{-1}, relatively low for most metal forming operations.(31)(33)

One measure of the degree to which a metal will become superplastic is the sensivity of the flow stress, to the strain rate imposed on the material. The classic equation relating flow stress (σ) to strain rate ($\dot{\varepsilon}$) is:

$$\sigma = K\dot{\varepsilon}^{m} \qquad (1)$$

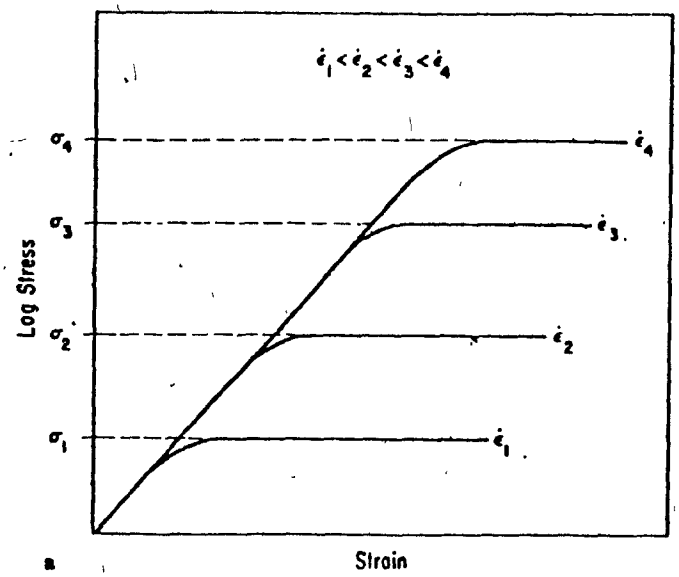
This may be determined experimentally for each alloy by testing tensile specimens at given strain rates and temperatures and determining the stress required to maintain that strain rate. By testing across a range of strain rates it is possible to prepare a series of stress-strain curves such as shown in figure 56a. This type of curve is however inconvenient to work with and the usual practice is to graph the data as a log-log plot, such as that shown in fig. 56b. By taking natural logarithms of equation 1:

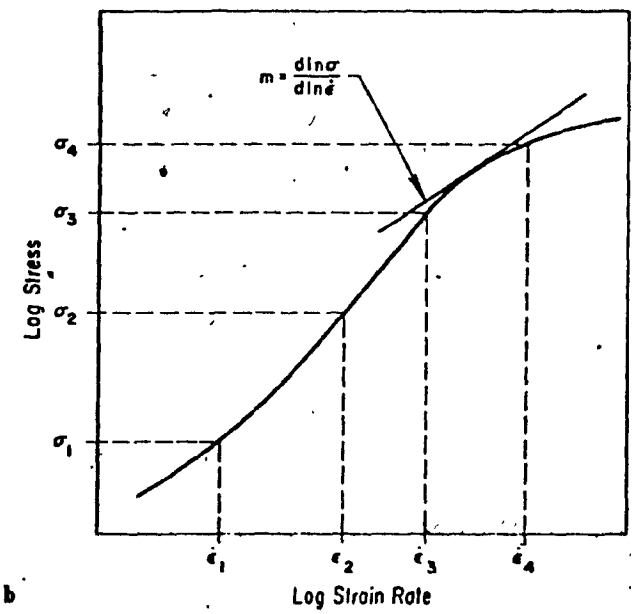
$$\ln \sigma = m \ln \dot{\varepsilon} + \ln K \qquad (2)$$

In this equation m is a measure of the strain rate sensitivity. Differentiating the result and finally solving for m the following expression is obtained:

$$m \quad \frac{d \ln \sigma}{d \ln \dot{\varepsilon}} \qquad (3)$$

From this equation, 3, it can be seen that the higher the value of m, the more strain rate sensitive and therefore, the more superplastic the material. This equation may then be related, to the log-log plot of fig. 56b, where, the slope of the stress-strain curve can be used to establish actual experimental values of m. In fig. 56c values of m have been plotted as a function of log strain rate. As can be seen from this curve, m reaches a maximum at intermediate strain rates, the low m value near $\dot{\varepsilon}_1$, indicating the region of creep, while that near $\dot{\varepsilon}_4$ indicates hot working. Values of m may vary from 0.0 to 1.0 with any value over 0.5 being considered an indication of superplasticity, naturally higher values of m, indicating still greater superplastic properties, are therefore desirable, especially in titanium.(31)(33)





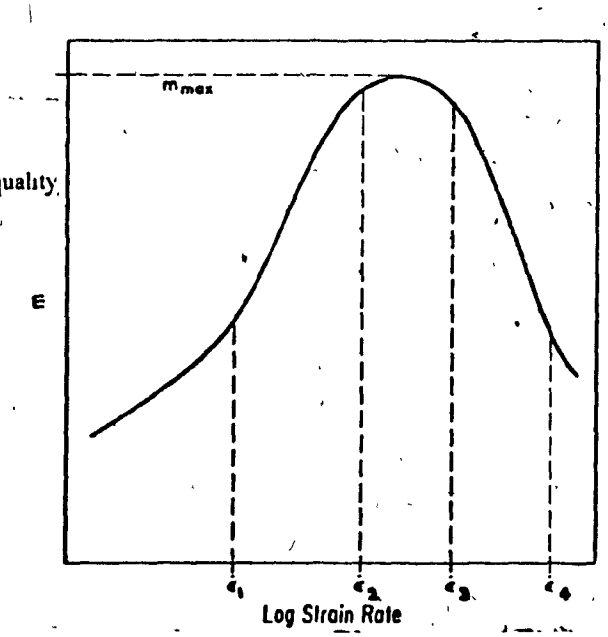


Fig.56 Tensile tests made in vacuum at high temperatures and varying strain rates enable engineers to determine an alloy's superplastic quality – it is indicated by the highest m value.

The technology of superplastic forming is really quite simple and is in many ways similar to that used to form thermoplastic sheet, the work piece being deformed in a manner analogous to that of a diaphragm. The sheet is inserted into the die cavity, the die closed, purged of air using an inert gas, such as argon, heated to the working temperature, and then formed by increasing the gas pressure in the upper cavity. Typical operating ranges for Ti-6Al-4V are 15 to 150 psig pressure at 900 to 950°C for from 15 minutes to 4 hours. The inert gas used during forming protects the titanium from gas absorption embrittlement, while permitting a slow pressure increase in the die cavity, to provide the slow strain rates necessary for superplastic forming. At the same time the use of gas to cause deformation means, that only a single female die is necessary, rather than the mated dies of conventional forging and this die may be much larger in area than conventional dies.(31)(33)

The slow forming rates and use of thin sheet make possible the forming of shapes far more complex than those possible by conventional forming techniques. Parts formed can include one or all of the following features, integral end flanges, beads in both webs and corners, short run-out joggles, and bend radii less than the thickness of the sheet. An example of the design freedom, and the advantages this provides may be seen in components developed for the B-1 bomber. One particular part the nacelle center beam frame, eight of which were required on each plane to separate the B-1's two engines, was manufactured conventionally and superplastically. (see fig. 57) The conventional part was formed as eight separate components joined together by 96 fasteners, while the superplastic part was only a single component which due to beading of the web was thinner, 30 percent lighter and 50 percent cheaper to produce. Difficulties encountered in superplastic forming are similar to those in isothermal forging, the need for high temperature die materials and the heating of the dies and workpieces. Fortunately the superalloys used in isothermal forging are also suitable here and due to the low strain rate levels used during the process it has been possible to carry out experimental trials of ceramic dies which have proved troublesome but adequate and considerably cheaper.(31)(33)

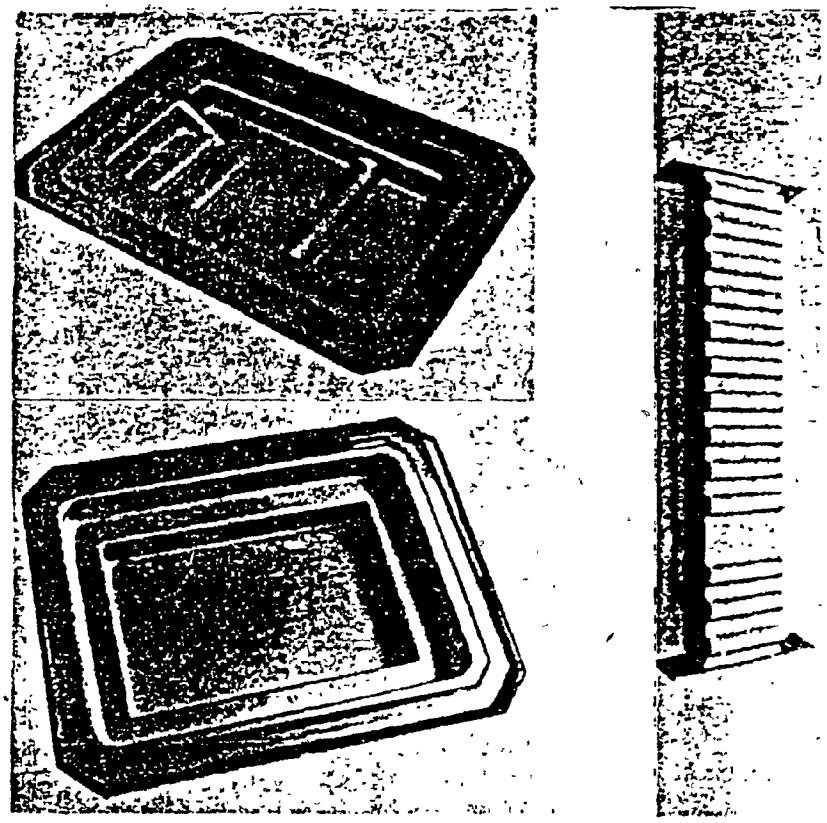


Fig.57 - Examples of parts superplastically formed from titanium alloy sheet.

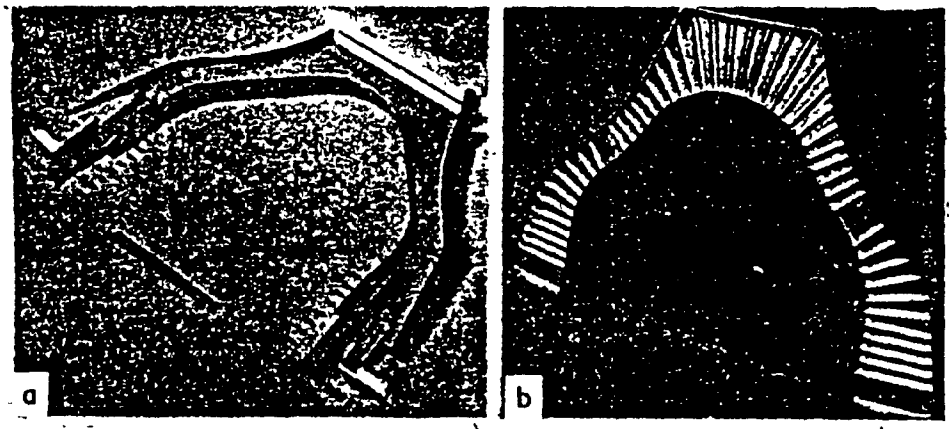


Fig.57 - B-1 nacelle center beam frame. (a) Conventionally fabricated part containing 104 separate components, (b) same part superplastically formed from a single sheet of Ti-6Al-4V.

TABLE 5

HEAT TREATMENT AND GRAIN SIZE OF Ti-6Al-4V

Specimen No.	Heat Treatment* Temperature (C)	Time (h)	Grain Size (μm)
HC-1	590	64	4.1
HC-3	704	8	4.2
HC-4	760	8	4.2
HC-5	870	2@	4.9

HEAT TREATMENT AND GRAIN SIZE OF BETA III

Specimen No.	Heat Treatment* Temperature (C)	Time(min)	Initial Grain Size (μm)
6#	-	-	see text
7	510	240	see text
8	775	2	8
9	775	30	28
10	775	180	33

* Air cool.

\# As-cold worked.

@ Cool 56C/h to 704C + air cool.

Grain size determined by linear intercept method, counting all interphase boundaries and corrected to true grain diameter. No detectable change between pre-test and post-test grain sizes.

TABLE 6

FLOW PROPERTIES OF Ti-6Al-4V AT 815C

Specimen No.	Grain Size* (μm)	σ_o (MPa)	m_{max}	$\dot{\varepsilon}$ at m_{max} (s^{-1})
	Present Investigation			
HC-1	4.1	3.77	.883	3.4×10^{-4}
HC-3	4.2	3.01	.876	2.0×10^{-4}
HC-4	4.2	3.11	.921	2.9×10^{-4}
HC-5	4.9	4.83	.838	1.6×10^{-4}
	Conventionally Processed Sheet			
LA-1-A	7.6	2.88	.624	3.0×10^{-5}
LA-1-A(R)	7.6	2.33	.786	4.2×10^{-5}

To date almost all superplastic forming has been done with the alloy Ti-6AL-4V at temperatures of 900 to 950°C, with no special processing to enhance its superplastic properties. Research to provide improved superplasticity has therefore concentrated on either, new alloys or special mill processing steps. Any improvements achieved would permit reduction of temperatures, and/or forming pressures and times, particular importance being attached to reduced temperatures with their accompanying cheaper die materials.

The microstructures developed during superplastic forming are similar to those that result from isothermal forging, fine grained equiaxed alpha + beta crystal structure at lower forming temperatures, and primary alpha + Widmanstatten alpha at higher temperatures, especially if the alloy has been cold worked prior to forming. Research carried out by Froes, et al[33], in an attempt to lower forming temperature by improving superplasticity was quite successful. By substantially cold reducing and then annealing workpieces it was possible to reduce grain size, in samples of Ti-6AL-4V and thereby increase m values. As a result it became as easy to form the metal at 815°C as it had previously been at 925°C. Compared to conventionally processed sheet material at the same temperature it was possible to use higher strain rates, and shorter forming times at slightly higher pressures. At the same time the fine grained alpha + beta crystal structure that was developed by cold working and annealing was retained after forming due to the almost complete absence of recrystallization or grain growth at the lower temperature, (see tables 5 and 6). During the same research the superplastic properties of the metastable beta alloy, Beta III, (Ti-11.5 Mo - 6Zr - 4.5 Sn), were also investigated. In this case the alloy was cold reduced and then given various annealling treatments to produce a variety of grain sizes which were subsequently superplastically deformed in the temperature range 650 to 815°C. However, due to the fact that Beta III has a transus temperature of only 745°C the deformation occured both above and below the beta transus and was complicated by recrystallization and grain growth. In general the best superplastic behaviour was observed below the beta transus, with optimum

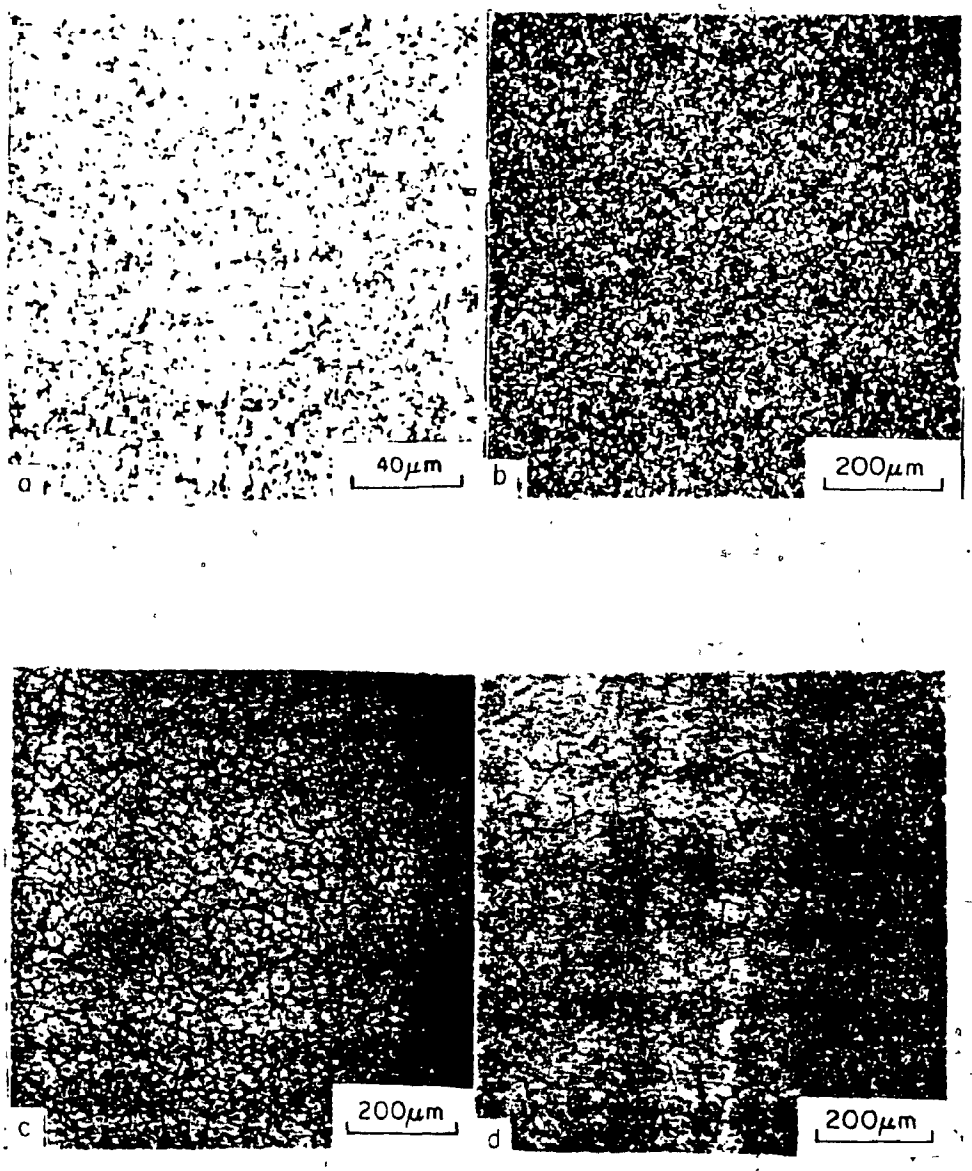


Fig.58 Micrographs of Beta III rod annealed 2 minutes at 775°C. (a) Optical micrograph, as annealed, (b) optical micrograph, post-test (704°C), (c) optical micrograph, post-test (760°C), (d) optical micrograph, post-test (815°).

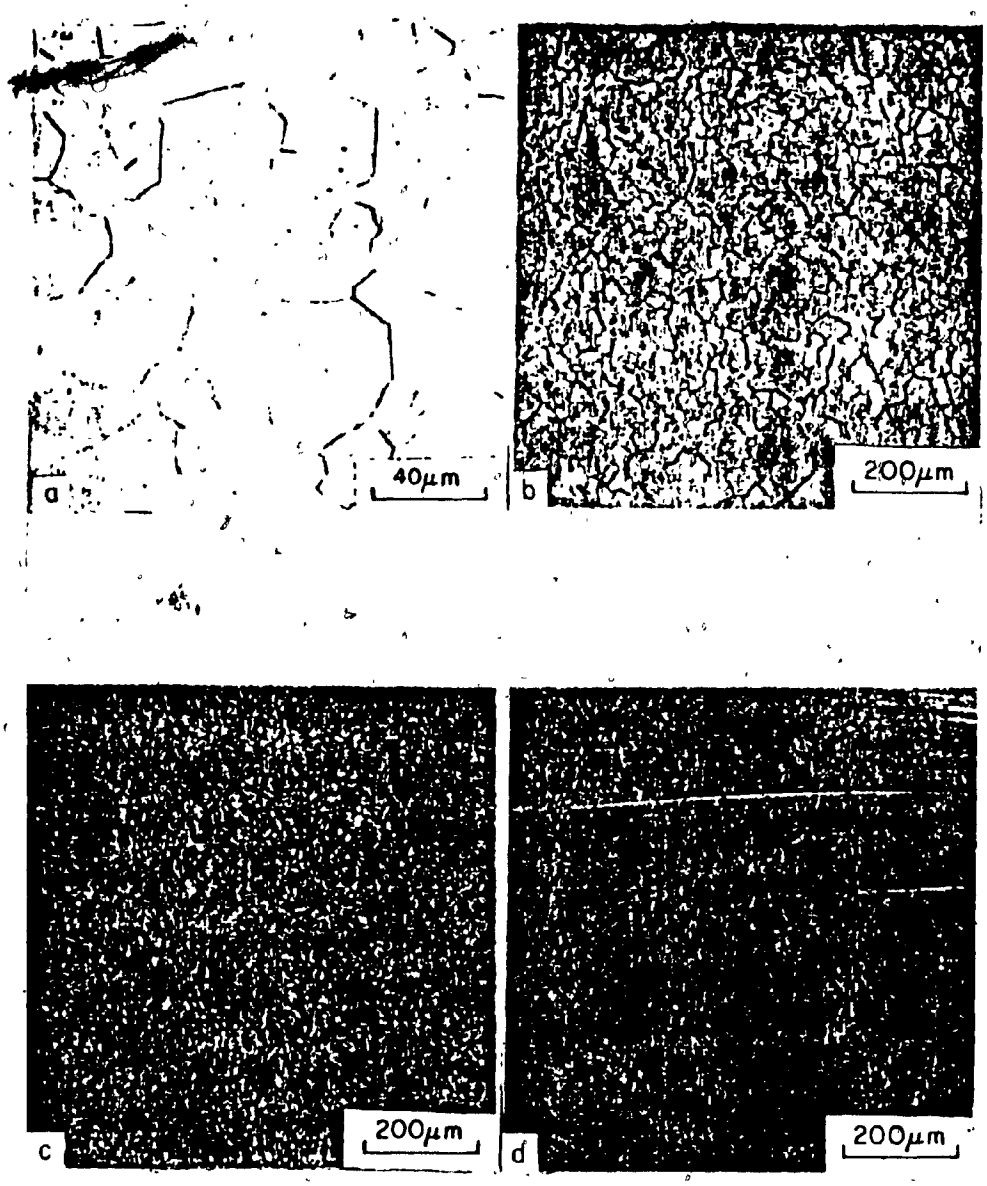


Fig. 59 Optical micrographs of Beta III rod, annealed for 180 minutes at 775°C. (a) As annealed, (b) post-test (704°C), (c) post-test (760°C), (d) post-test (815°C).

properties occuring at 730°C. Above the beta transus, rapid grain growth degraded superplasticity and resulted in coarse, transformed and generally undesirable microstructures. Compared to superplastic forming of Ti-6Al-4V, the forming of Beta III required more time and higher pressures at similar points below the beta transus. The greatest potential for superplastic forming in Beta III appears to exist in the ultrafine grain size resulting from cold working or cold working plus aging, when deformed at temperatures from 700 to 650°C. At these temperatures there is little grain growth and the retained fine grain structure encourages superplastic behaviour and results in optimum properties in finished products. (see fig. 58,59) These results indicate that, although not all titanium alloys may be as easy to form superplastically as Ti-6Al-4V, all or most of them may be worked in this fashion once the optimum forming temperature and grain size have been determined.

Superplastic forming appears to offer substantial advantages over conventional forming techniques, its greater design flexibility, lower fabrication costs, and reduced material requirements, alone are sufficient to make it a useful manufacturing technology. However when these are combined with excellent product microstructures and properties, and the growing demand for large titanium parts in new aircraft and jet engines, it is almost certain that superplastic forming will revolutionize the fabrication of titanium alloy sheet products.

3.3 EXTRUSION

The extrusion of titanium and its alloys is a process considerably more complicated than the extrusion of more conventional metals such as copper, aluminum, steel and nickel alloys. The reason for this being not only the gas absorption at high temperatures and the low thermal diffusivity that have been mentioned earlier but an unfortunate tendency on the part of titanium to adhere to the working surface of extrusion dies. As a result it is impossible to extrude titanium without a high degree of lubrication and even this may not be sufficient if the correct shape of die is not used. In a flat faced die the occurance of a dead metal zone results in titanium sliding over titanium and a breakdown in the lubricant film at the die.[34]

Several different techniques have been developed to overcome this problem most involving cone shaped dies and billets with tapered ends. In the Soviet Union however, another technique has been developed using flat dies, but with special cone shaped inserts made of lubricant material, usually glass, which are inserted between the die and the billet, thus supplying lubrication and so permitting extrusion. Care must however be taken to insure that the insert and the extrusion press chamber are above the melting point of the glass lubricant otherwise it will function as an abrasive and damage the walls of the extrusion chamber.[34] Another technique to correct this problem is to can the titanium in steel prior to heating and extrusion, this also eliminates gas absorption embrittlement during processing, and once the billet has been extruded the titanium may be declad through pickling or machining.[37]

The lubricants used in extrusion processes, while similar to those used in forging processes, must be formulated to provide a greater degree of lubrication due to the increased metal flow during extrusion. Glasses are the most popular lubricants although they have to be physically separated from the metal after extrusion which increases costs. Molten salts have also proved effective, boron nitrate being particularly good where high quality surface finishes are desirable. A considerable interest has also been shown in graphite — containing, hydrocarbon — based lubricants.

Although the hydrocarbon lubricants by themselves tend to burn off during extrusion, considerable improvement has been achieved through the addition of varying percentages of additives which alone provide no special degree of lubrication but in combination with the hydrocarbons display a synergistic effect, mutally lowering friction. Some of the additives that have proved effective are MoO_3, MoS_2 as well as suspension of lead and aluminum[35,36].

Other than the earlier mentioned difficulties the extrusion of titanium is carried out in a manner similar to that of other metals. The billet to be extruded is hot worked to improve microstructure, then pierced or drilled out to provide a location for the mandrel, heated to the extrusion temperature and then extruded. The earlier mentioned lubricant may be applied either before the final heating or just before extrusion, although it is frequently applied at both points in the process. Almost any press capable of extruding copper or steel is sufficiently powerful to extrude titanium and may be used for this purpose. Care must however be taken to lubricate the mandrel, as well as the die, for titanium sticking to the mandrel surface will damage the inside of the extruded pipe.[30][38][40]

Of the many factors involved in the process of extrusion only four appear to have a primary effect upon the pressures required to carry out deformation. These four factors are billet temperature, alloy composition, ram speed and the logarithm of the extrusion ratio, other factors such as die angle or container temperature have only a limited effect on the working pressure. From the graphs in figs. 60, 61, and 62 the relationship of temperature and extrusion ratio to extrusion pressure for various alloys of titanium can be clearly seen while from equation 4 the direct effect of velocity(V) on strain rate($\dot{\epsilon}$) for strain rate sensitive materials such as titanium can be seen[36]

$$\dot{\epsilon} = \frac{4V\, D_o^2 \tan\theta}{D_e^3} \qquad (4)$$

D= die diam., o= entrance, e= exit, θ= die semiangle.

These same factors along with cooling rate after extrusion, also control the grain size and microstructure of the product and therefore its properties.

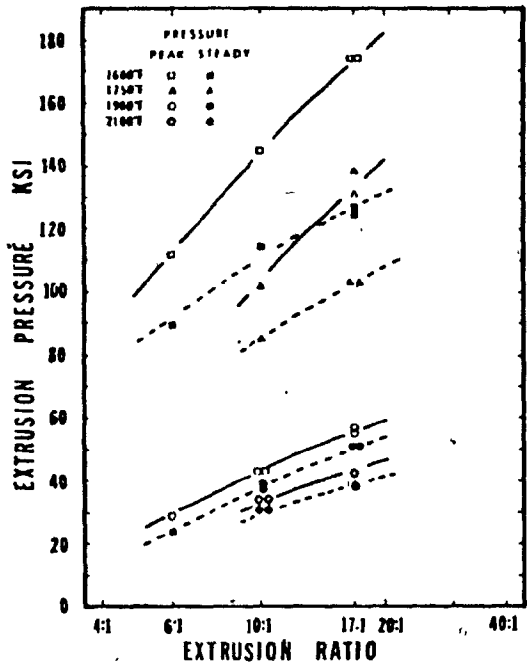


Fig. 60 Deformation pressure as a function of extrusion ratio for the extrusion of Ti-5Al-2.5Sn at various billet preheat temperatures.

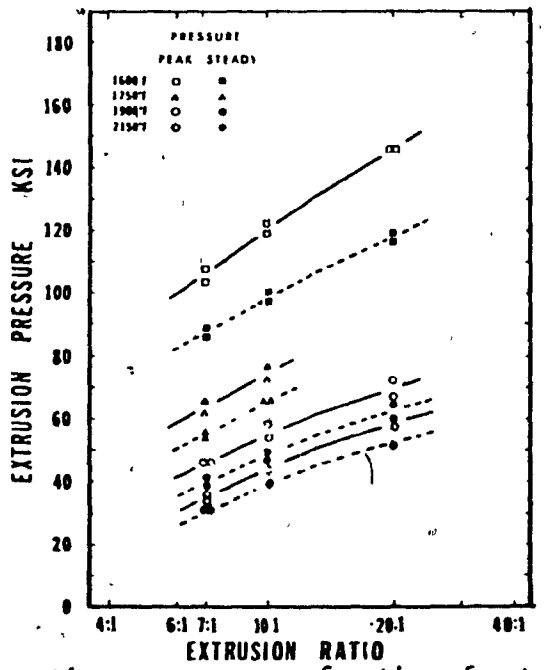


Fig. 61 Deformation pressure as a function of extrusion ratio for the extrusion of Ti-6Al-4V at various billet preheat temperatures.

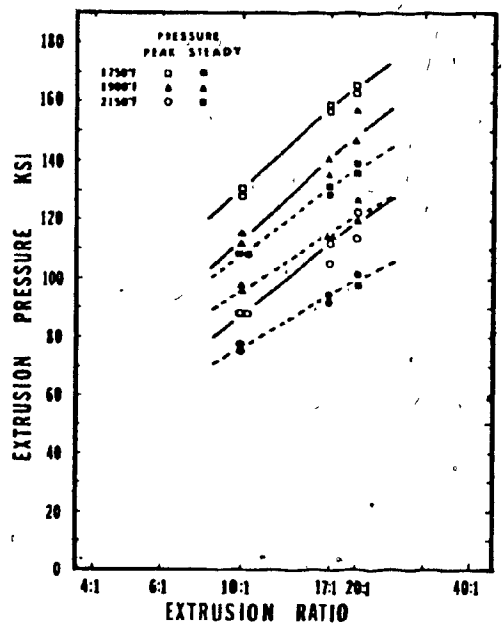


Fig. 62. Deformation pressure as a function of extrusion ratio for the extrusion of Ti-13V-11Cr-3Al at various billet preheat temperatures.

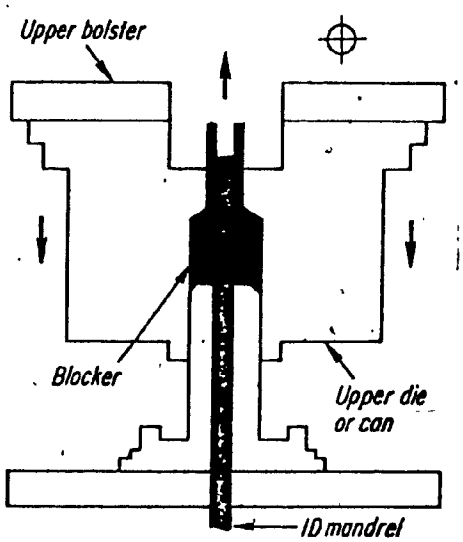


Fig. 63 Cross section of the extrusion die assembly; pipe being formed is in color.

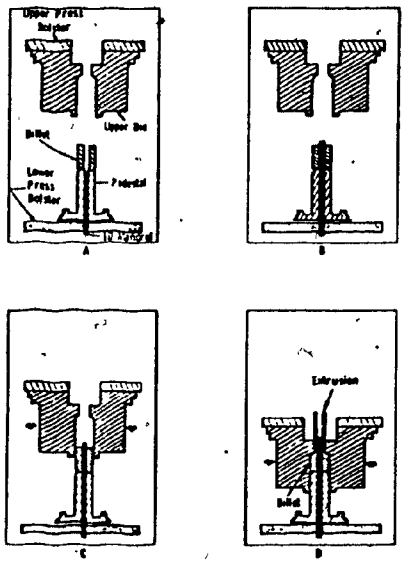


Fig. 64 Sequence utilized in conjunction with 20,000 ton press to fabricate extruded hollow.

To get a better understanding of the effects of these variables, a comparison of the results of extrusion at different temperatures, with different cooling rates, of three different alloys will be carried out from the work of Gurney and Male.[38] The three alloys reported on in the paper, were Ti-5Al - 2.5 Sn an alpha alloy, Ti-6Al -4V an alpha + beta alloy and Ti-13V -11 Cr-3Al a beta alloy. Each of these alloys was extruded at temperatures from 870°C (1600°F) to 1180°C (2150°F) and then either air cooled or water quenched.

In the case of Ti-5Al-2.5 Sn, bars extruded below the beta transus regardless of cooling rate had fibrous structures of fine beta particles in an alpha matrix. (see fig. 65 a,c) Extrusion above the beta transus, with air cooling producing coarse plate like (acicular) Widmanstatten alpha structure and water cooling producing alpha prime martensites. (See fig. 65 b,d). In those samples extruded above the beta transus prior beta grain boundaries were clearly visible, their size increasing with increasing extrusion temperature, indicating that recrystallization had occured. Annealing of these samples produced no visible optical changes in structure. The properties of these various microstructures may be seen in Fig.66. Generally below the beta transus there is almost no difference in properties between air cooled and water cooled alloy as reflected by their similar microstructures. However above the beta transus the water quenched sample due to its finer microstructure has higher strength and toughness than the air cooled sample but as both have similarly tangled microstructures, almost equal ductility.

For the alpha + beta alloy Ti-6Al-4V, bars extruded below the beta transus again had a fibrous structure in this case of elongated alpha interspersed with a fine grained structure of alpha + beta, regardless of cooling rate. (see fig. 67a, c) Those samples extruded above the beta transus however, upon air cooling, developed a structure of relatively slightly elongated primary alpha in a matrix of Widmanstatten alpha (basketweave), while when quenched, developed a structure of alpha prime martensite. (see Fig.67 b,d) The extruded samples of this alloy were heat treated in two different

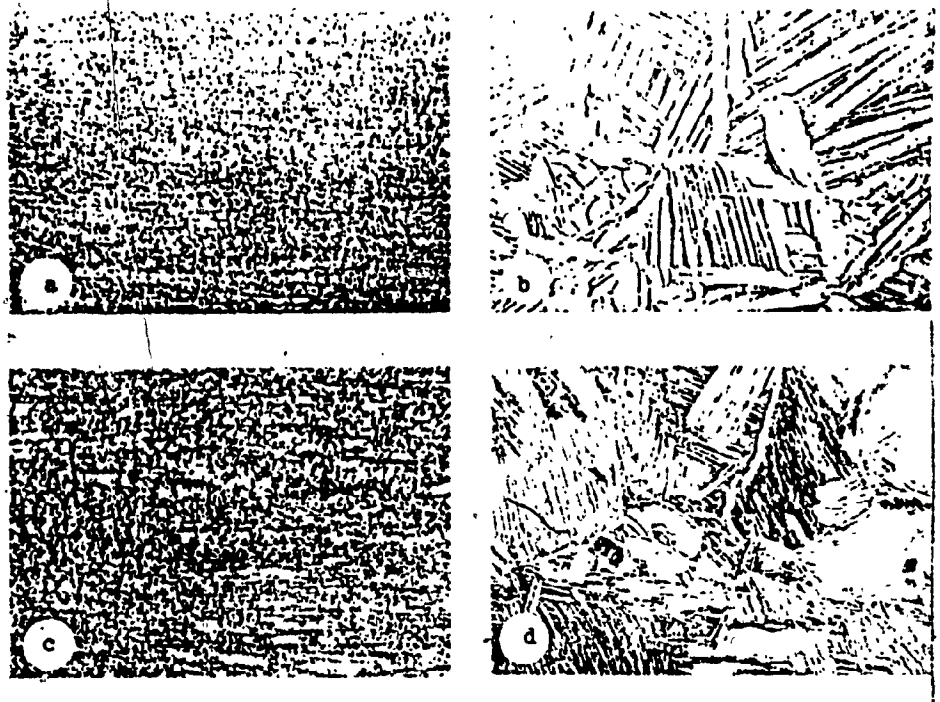


Fig.65 Structures of Ti-5Al-2.5Sn alloy bars extruded at various temperatures. Magnification 250X. (a) 1625°F, air cooled; (b) 1900°F air cooled; (c) 1625°F, water quenched; (d) 1900°F water quenched.

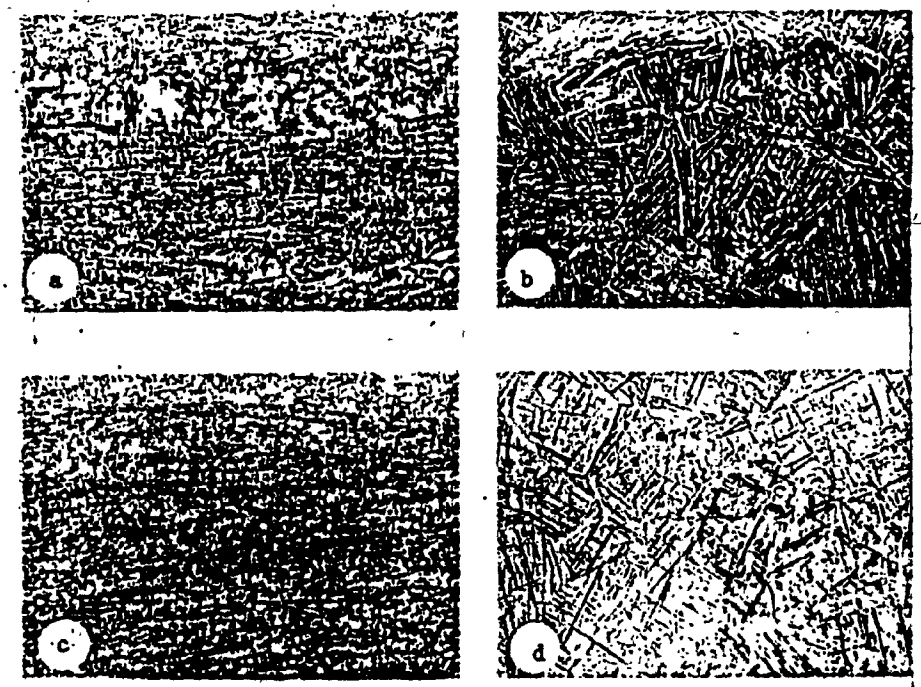


Fig.67 Structures of Ti-6Al-4V alloy bars extruded at various temperatures. Magnification 250X. (a) 1600°F air cooled; (b) 1900°F air cooled; (c) 1600°F water quenched; (d) 1900°F water quenched.

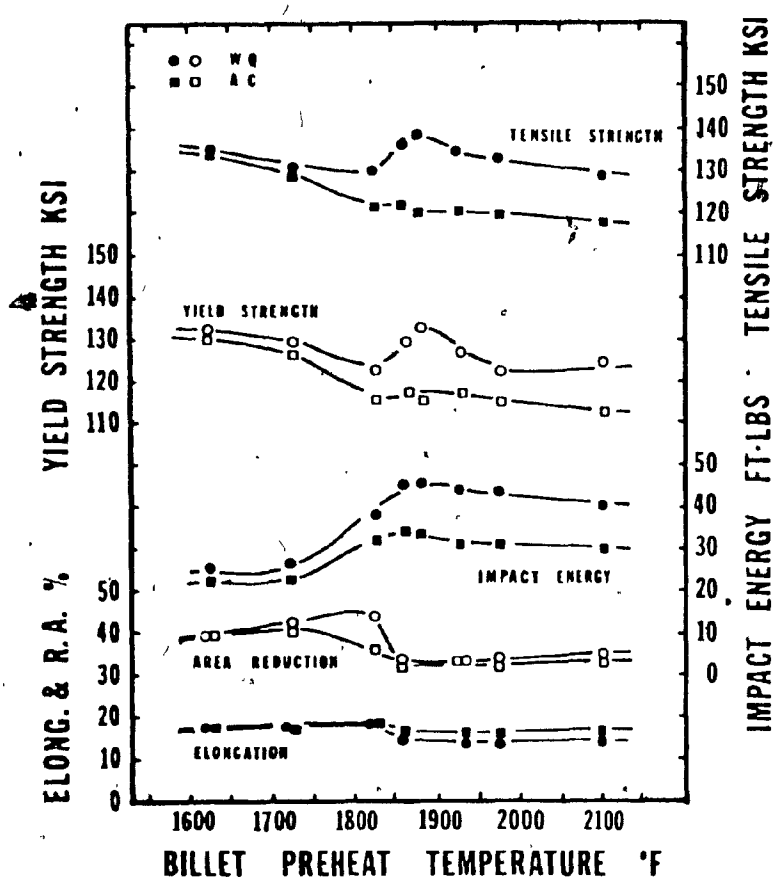


Fig.66 Variation of room temperature mechanical properties with billet preheat temperature for Ti-5Al-2.5 Sn bars extruded at 10:1 ratio and subsequently aged for one hr at 1000°F.

ways, while all samples were given the standard four hour aging, a separate set of air cooled samples were also solution treated and aged. As can be seen from Fig.68 ductility and toughness are effectively the same for each type of treatment, strengths on the other hand vary considerably. Aging of the air cooled material, produced a coarse alpha + beta structure in all samples and this appreciably decreased strength. Aging of water quenched material on the other hand had no effect below the beta transus but for those samples extruded above the beta transus aging tempered the alpha prime to fine grained alpha + beta which resulted in high strength values. Finally the effects of solutioning and aging varied with extrusion temperatures, below the beta transus it permitted recrystallization and produced a fine grained microstructure of primary alpha particles in a matrix of alpha prime and beta with properties equivalent to those of water quenched material. Above the beta transus however it permitted recrystallization, nucleation and growth of randomly oriented needles of primary alpha, in a matrix of alpha prime and beta the resulting microstructures being similar to Widmanstatten alpha produced strengths almost equal to those of alloy extruded below the beta transus (see fig. 69).

The third alloy, Ti-13V-11Cr-3Al, being a beta alloy, typically, showed no deformation structure of any kind. The samples worked below the beta transus, both air and water cooled, were almost identical, showing large beta grains with some slight evidence of a second phase (see fig. 70, a,c) In the material worked above the beta transus however, the quenched sample had large beta grains with no second phase showing, while the air cooled sample showed a coarse precipitate localized within the grains (see fig.70 b,d). Aging of the material resulted in almost identical, fine, uniformly distributed precipipate structures for all the samples except those worked above the beta transus and air cooled. In these samples the precipitate was not uniformly distributed but rather very coarse and unevenly scattered from grain to grain, the coarsest structure occuring at 1120°C (2050°F). (see fig. 71) Based on fig. 72 for the properties of this alloy it can be assumed that the precipitate is omega phase, as strength levels are high while ductility and toughness values are low, typical effects of this phase on titanium properties. The decrease in room temperature strength for billets extruded at 1120°C (2050°F) is probably therefore, a point where the coherency of the omega phase is

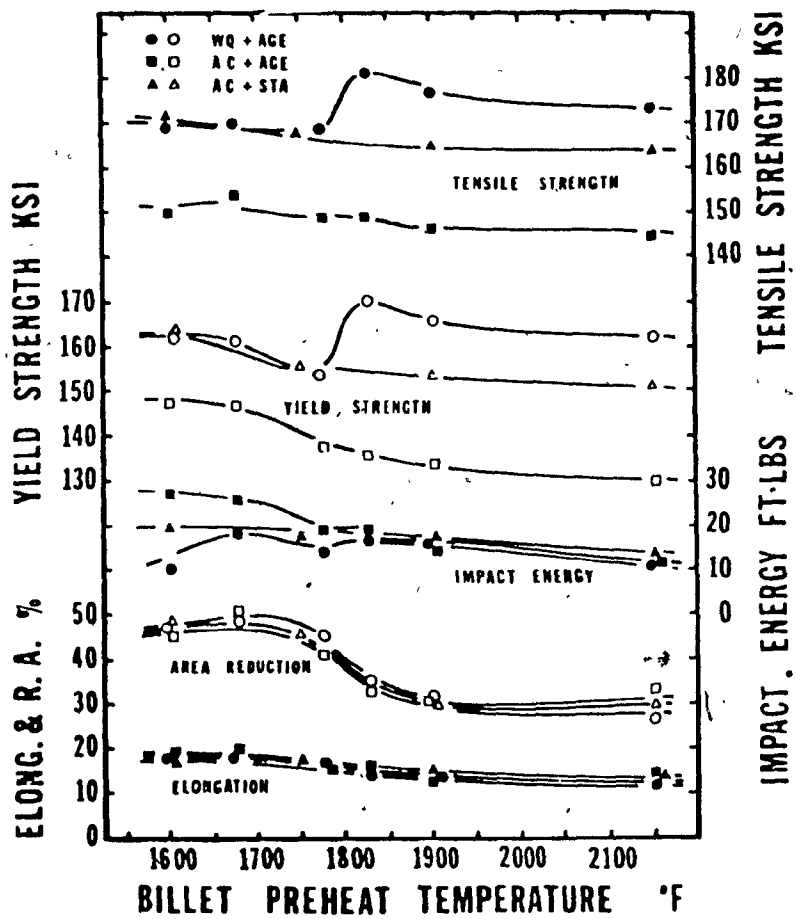


Fig.68 Variation of room temperature mechanical properties with billet preheat temperature for Ti-6Al-4V bars ex-10:1 ratio and subsequently aged for four hrs at 1000°F.

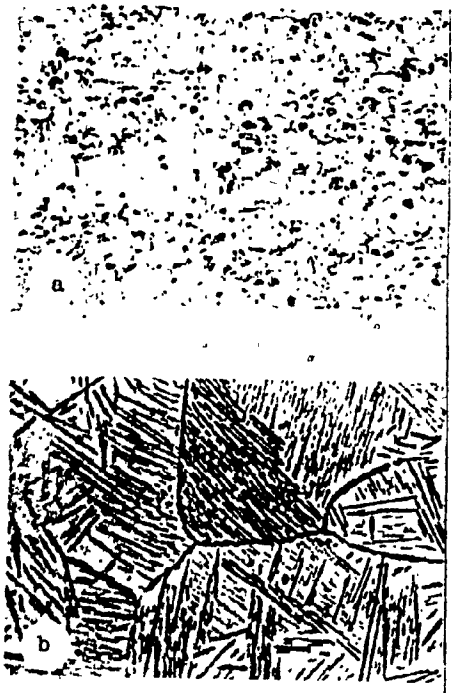


Fig. 69 Structures of Ti-6Al-4V alloy bars extruded at various billet preheat temperatures and allowed to air cool, and then solution treated at 1750°F for one hr, water quenched and aged at 1000°F for four hrs. a) 1750°F; b) 1900°F.

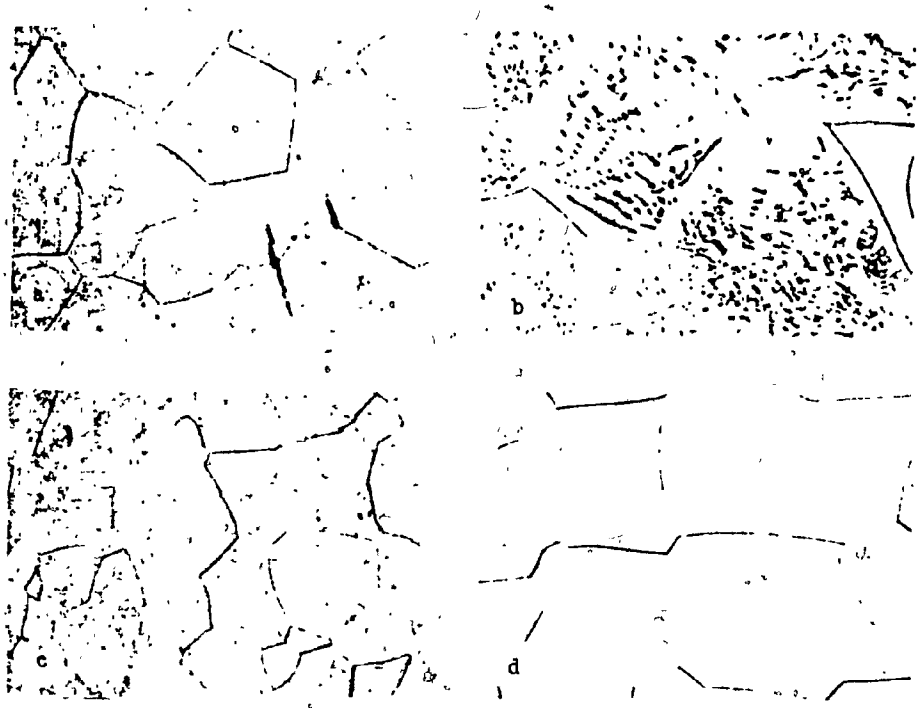


Fig.70 Structures of Ti-13V-11Cr-3Al alloy bars extruded at various temperatures. Magnification 250X. (a) 1750°F air cooled; (b) 2050°F air cooled; (c) 1750°F water quenched; (d) 2050°F water quenched.

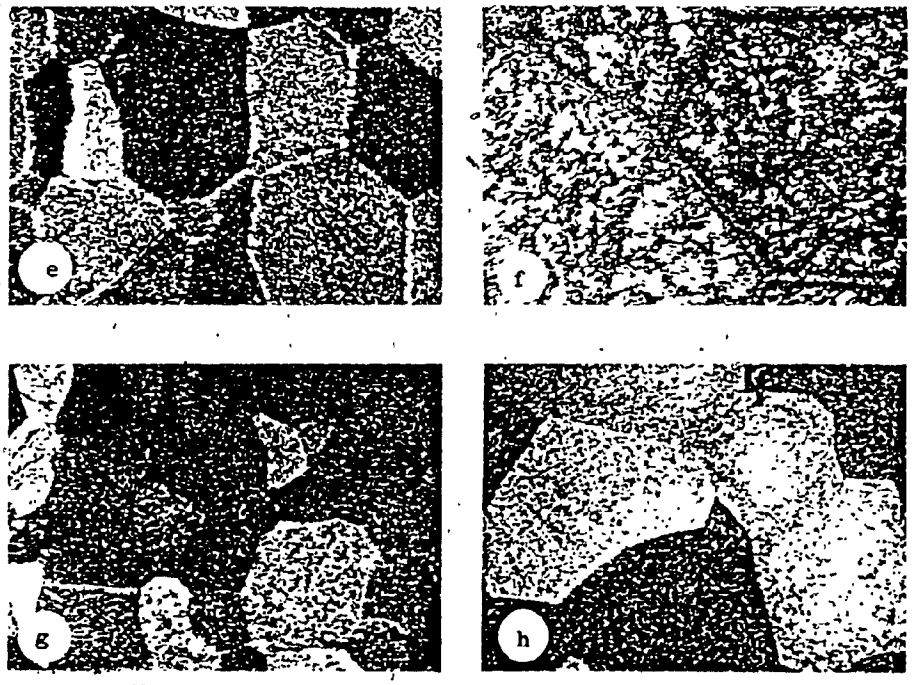


Fig.71 (e) 1750°F air cooled plus 24 hrs at 900°F;
(f) 2050°F air cooled plus 24 hrs at 900°F;
(g) 1750°F water quenched plus 24 hrs at 900°F;
(h) 2050°F water quenched plus 24 hrs at 900°F.

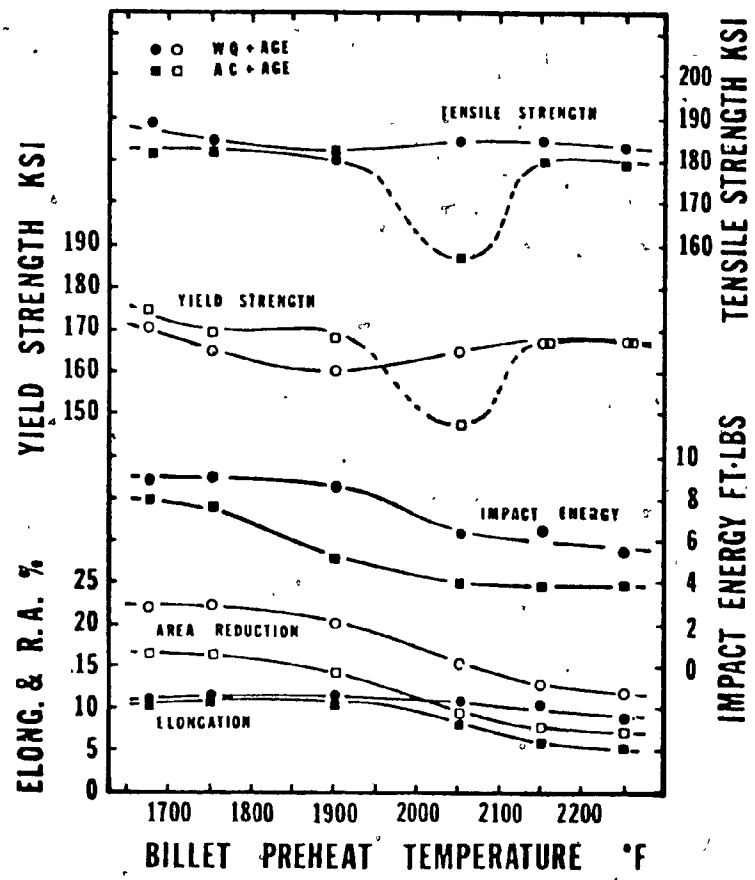


Fig. 72 Variation of room temperature mechanical properties with billet preheat temperatures for Ti-13V-11Cr-3Al bars extruded at 17:1 ratio and subsequently aged for 24 hrs at 900°F.

disturbed, as suggested by the microstructure at this point. (see fig. 71f) At the same time the difference in impact energy and area reduction between water and air cooled samples is due to the finer particle size of precipitate in the water quenched material, the finer particles making the material more ductile.

The above characteristics of extruded titanium alloys of various compositions has been partly confirmed in a paper by Chait and Stead[41]. In their paper however they used different alpha + beta and beta alloys and yet obtained similar microstructures and properties. This indicates that the behaviour attributed to each of the alloys mentioned earlier may be regarded as being representative of its titanium alloy group, alpha, alpha + beta and beta rather than just its alloying element composition.

The majority of tubing produced by hot working is later cold drawn to smaller diameters and thinner wall thicknesses. The properties desired in hot extruded titanium, high ductility and low strength are therefore not necessarily the same as those of other titanium products but are obtained and optimized in exactly the same way, through control of thermo-mechanical processes.

3.4 HOT ROLLING

The hot rolling of titanium and its alloys is carried out in a manner similar to that of other metals, subject only to the limits imposed by the properties of titanium itself. Hot rolling takes place between 730°C and 820°C, but may be carried out at temperatures as low as 590°C. The material to be rolled is heated to the working temperature, descaled if necessary, placed in the rolling mill and reduced in size. As the thickness of the workpiece decreases it becomes necessary to reduce its temperature to minimize gas absorption. When the sheets of titanium have been reduced to a thickness of about 0.1 inch they are hot pack rolled, several of the thin sheets being piled on top of each other with a top and bottom sheets of stainless steel or nickel being added. These are then composite rolled at one time, thereby reducing elastic deformation in the rolls, while the top and bottom sheet provide protection against gas absorption.(1)

The problems that occur in this process have for the most part been mentioned before in one of the earlier forming processes and may be solved in a similar manner. The tendency of titanium to adhere to the rolls as it is being deformed may be controlled through the application of appropriate lubricants and these will at the same time serve to reduce roll friction and gas absorption. Other problems encountered are chill cracking due to localized cooling and hot spots causing crystal structure changes. However through control of rate of deformation these two difficulties may be made to cancel out. As the metal is deformed between the rolls it is kinetically heated and if the amount of deformation is chosen correctly then the heat generated can be exactly that necessary to overcome the chilling effect of the rolls.(1)

Although hot rolling is generally carried out in the alpha + beta region it may also be done in the beta region. However as in the case of beta forging a substantial degree of the finished rolling must be carried out below the beta transus to obtain optimum properties. In fig.73 are shown examples of the microstructures resulting from alpha + beta rolling and beta rolling.(28)(43)

Two types of macrostructural defects occur in hot rolled titanium, the first of these is a localized problem called banding and the second is the phenomenum of texturing of rolled sheet resulting in anisotropy. Banding is the result of local kinetic overheating which exceeds the beta transus, followed by quenching due to adjacent cooler metal, the result, especially in alloys that may form martensites is a hardened band of transformed metal running across the sheet. If this is not corrected by deforming to greater than 50% or avoided through proper temperature control it can lead to brittle cracking of the sheet surface.(42) Anisotropy on the other hand is a bulk property, it is the possession of different properties in different directions, and occurs when, during rolling, a metal becomes textured. It is the result of a fabrication sequence which encourages a strong preferred orientation in a component and occurs most often in hcp crystal microstructures, such as titanium alloys rich in alpha phase. Although it does have drawbacks, in certain services it can be a useful source of added strength and otherwise may be corrected to a certain extent through beta heat treatment which tends to randomize a developed preferred orientation.(44)(26)

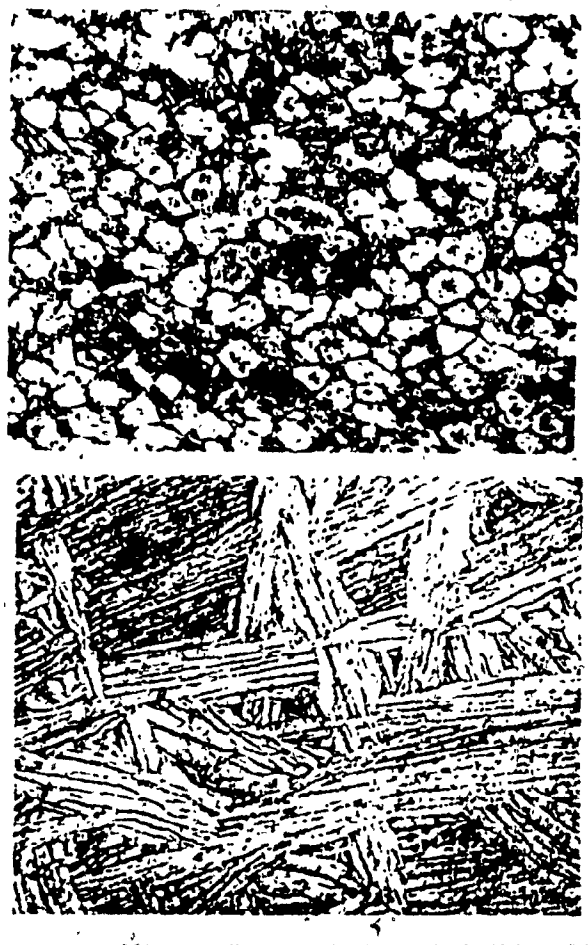


Fig. 73 Microstructure of IMI 679 billet (a) alpha–beta rolled; (b) beta rolled. × 500.

CHAPTER 4

CONCLUSION

Titanium and its alloys are among the most difficult of the structural metals to form and at the same time among the most promising in terms of strength, density and corrosion resistance. However, slowly but surely technology has been, and is being developed to handle this rather cantankerous metal, and as further research into its properties is done its shaping and bending to the tasks we find for it will become easier. Already in only a few short years techniques geared especially to the properties of titanium such as isothermal and superplastic forming have been developed. These combined with better control of microstructure have permitted a degree of control of finished part properties that was undreamed of 15 years ago.

Now, with the growing use of titanium in the latest generation of aircraft and the engines that power them, has come an increase in research that may be able to break the final barrier, cost, and if not make titanium as cheap as steel at least bring it into common usage with the other metals that serve mankind.

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