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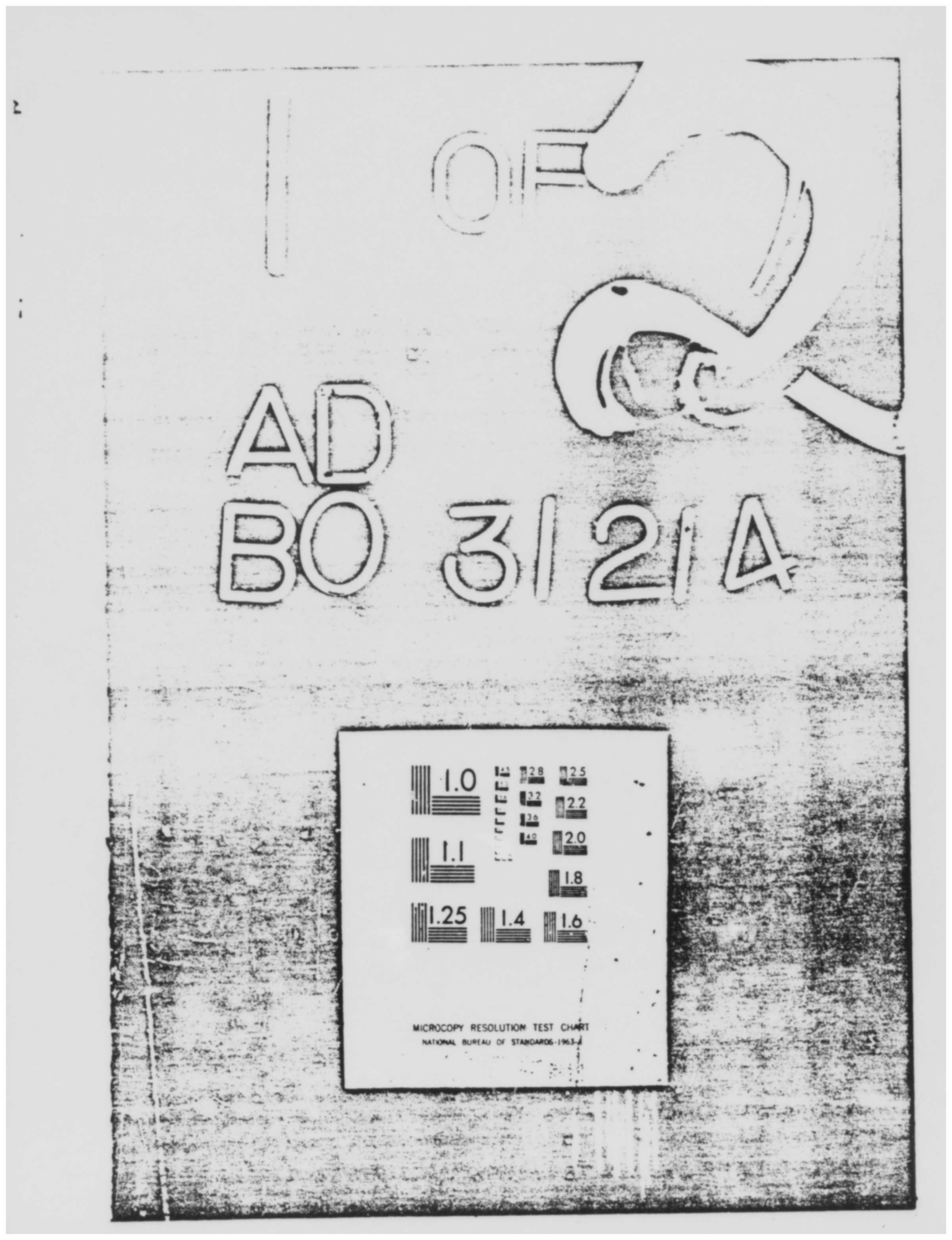
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ROLLING, FORMING AND JOINING TITANIUM-ALUMINIDE SHEET

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MAY 1978

TECHNICAL REPORT AFML-TR-78-59
Final Report February 1975 to September 1977

AIR FORCE MATERIALS LABORATORY
AIR FORCE WRIGHT AERONAUTICAL LABORATORIES
AIR FORCE SYSTEMS COMMAND
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This technical report has been reviewed and is approved.

HARRY A. LIEBITT
Project Engineer
High Temperature Materials Group

FOR THE COMMANDER

HENRY C. GRAHAM
Acting Chief, Processing and
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The preparation, deformation processing and mechanical testing of several titanium aluminide alloys based on Ti_3Al and TiAl were evaluated starting with preforms prepared from rotating electrode (REP) powders and cast ingot. The following alloys, expressed in weight percent, were evaluated: Ti_3Al-base alloys, Ti-16Al-10Nb, Ti-12Al-19Nb, Ti-13Al-20Nb-0.25Si; and TiAl-base alloys, Ti-36Al-5Nb and Ti-32Al-5Nb-5W.

Special HIP canning procedures were evaluated and developed to minimize

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inclusion of both soluble and insoluble gases and to avoid can-powder interaction during the HIP cycle which could produce cracking of the consolidated powders. Grain size after various thermal treatments to near the melting temperatures was the most significant indicator of workability. The one alloy, Ti-36Al-5Nb, which could not be made to achieve grain sizes beyond the powder size range, even after annealing just below the melting temperature, could not be successfully rolled. Although significant grain growth could be achieved with the other alloys, excessive grain growth in both the powder and cast preforms was not conducive to successful rolling.

Heated-roll rolling with roll temperatures to 875 C was used for the majority of trials. Breakdown and finishing procedures were established for all of the alloys. The finishing practices for the Ti_3Al-base alloys were established to achieve a fine grain microstructure for superplastic forming to facilitate conventional roll forming. Rolling of the TiAl-base alloys was consistently successful only with the Ti-32Al-5Nb-5W, which could be produced with an average grain size below 5 microns with the rolling conditions evaluated. Evaluations of conventional rolling of Ti-25Al-5Nb indicated that conventional rolling of this alloy was feasible.

The 0.05 to 0.06 inch thick Ti_3Al-alloy sheets produced during this program were evaluated for roll forming at Battelle with rolls heated to 425 C and for superplastic forming at Rockwell International. Roll forming in multiple passes was successfully achieved when the sheet temperature did not cool to below 540 C. Superplastic forming of the Ti_3Al-base alloys was successful; however, only the two-phase (α_2 + β) alloys exhibited m values in excess of 0.7 and up to 1.0.

Mechanical testing was performed on the alloys after HIP and after various thermomechanical treatments. No significant improvement in low temperature ductility was experienced with the processed sheet.

Data for this project were recorded in Battelle's Laboratory Record Books No. 32112, 32671, and 32356.

FOREWORD

This final technical report covers work performed on contract F33615-75-C-1168, Project No. 7021-01-67, during the period February, 1975 to September, 1977.

The investigation was conducted by the Battelle Memorial Institute, Columbus Laboratories, Columbus, Ohio, under the technical direction of Dr. H. A. Lipsitt, AFML/LLM, Wright-Patterson Air Force Base, Ohio.

Dr. A. L. Hoffmanner was program manager and Mr. D. D. Bhatt and Mr. G. E. Meyer were the responsible engineers.

TABLE OF CONTENTS

LIST OF TABLES

LIST OF ILLUSTRATIONS

LIST OF ILLUSTRATIONS (Cont'd)

LIST OF ILLUSTRATIONS (Cont'd)

LIST OF ILLUSTRATIONS (Cont'd)

SUMMARY

This final technical report describes the results of a program on "The Rolling, Forming and Joining of Titanium-Aluminide Sheet" performed during the period from February 17, 1975, through September 1977. The objectives of this program for the forming and joining of the titanium aluminides were to (1) establish consolidation and sheet-billet preparation practices to achieve desirable workability during rolling of predominantly α_2 and γ titanium aluminide-base alloys; (2) relate preparation practices, working conditions and post-rolling treatments to mechanical properties and microstructure of the product; (3) establish and demonstrate a methodology for heated-roll rolling of the alloys based on metallurgical and mechanical principles; (4) evaluate the mechanical properties of the product; (5) establish rolling-heat treatment practices to provide maximum room temperature ductility; and (6) utilize the sheet product to produce formed products of typical aerospace shapes.

The two alloys originally selected for this evaluation were:

- The two-phase, predominately γ alloy, Ti-36Al-5Nb (36 weight percent Al and 5 weight percent Nb) based on TiAl
- The essentially single phase α_2 alloy, Ti-16Al-10Nb (16 weight percent Al and 10 weight percent Nb) based on Ti_3Al

Sheet billet alloys were prepared by HIP of inert gas atomized, spherical powders produced by the rotating electrode (REP) technique from the cast alloys.

Companion programs performed concurrently and coordinated with this program provided newly developed alloys and/or practices which were also evaluated for their amenability to fabrication into sheet using heated rolls. These additional alloys and conditions included the following:

- Ti-16Al-10Nb rolled from as-cast ingot
- Ti-12Al-19Nb and Ti-13Al-20Nb-0.25Si rolled from cast and isothermally forged ingot
- Ti-32Al-5Nb-5W rolled from HIP-REP powder.

Considerable problems were initially encountered in producing the sheet billets by HIP of the powders. Although the first HIP series with thin (0.020 inch thick) CP-Ti cans were successful, the weight of the powder caused barrelling of the compact. The resulting crown is objectional because such features promote tensile stresses and subsequent fracture during the early stages of rolling. Attempts to avert this barrelling with heavier (0.093 inch thick) Ti-6Al-4V cans first resulted in severe thermally induced porosity (TIP) because the stiffness of the can required special handling procedures which affected outgassing of the argon cover used in can welding prior to sealing. A subsequent evaluation with vacuum can filling and outgassing resulted in cracking of all the Ti-16Al-10Nb and Ti-36Al-5Nb compacts due to reaction of the aluminide and the thick cladding during HIP. Successful HIP consolidation subsequently was achieved with vacuum can filling and outgassing using the thin CP-Ti cladding with external supporting frames and the thick cladding plasma-sprayed with alumina to provide a diffusion barrier.

Successful rolling of all Ti_3Al-base alloys (produced from powder or as-cast) to 0.05 to 0.06 inch thick sheet was achieved with billet preheat temperatures in the range of 870 C to 1200 C with the rolls controlled to temperatures between 700 C to 870 C. Trials on sheet specimens with ambient temperature rolls (conventional rolling) were not as successful as with roll heating, but appeared feasible at lower yield and higher roll forces.

Rolling of the TiAl-base alloys was significantly more difficult than the Ti_3Al-base alloys. The γ-alloy, Ti-36Al-5Nb, could only be rolled with partial success with a 870 C roll temperature and a 1425 C preheat temperature. The other γ-alloy, Ti-32Al-5Nb-5W, was successfully rolled at 1360 C to nominally 0.06 inch and thinner sheet. Light initial reductions appeared critical to achieve successful sheet from this alloy.

Rolling schedules and thermal treatments were developed for the Ti_3Al-base alloys to achieve microstructures for enhanced superplastic forming. These treatments achieved average grain diameters less than 15 microns for the Ti-16Al-10Nb program alloy and less than 10 microns for the two-phase Ti-13Al-20Nb-0.25Si alloy, with m values as large as 1.0 at 980 C. As expected, the lower rolling temperatures (below 1066 C) produced the finer grains and improved the superplastic performance.

Although the superplastic response of the Ti-32-5Nb-5W was not evaluated this α_2 + β alloy exhibited the finest grain size, less than 5 microns, in this evaluation.

The Ti-16Al-10Nb alloy was successfully roll formed to 2.5 to 3.0-inch diameter cylinders in multiple passes with preheat temperatures between 870 and 1010 C and a nominal forming-roll temperature of 425 C. Cracking during roll forming occurred when the sheet had cooled to below about 540 C.

Tensile, creep and fatigue tests on the program alloys were performed over the temperature range from -70 to 900 C for the as-HIPed material and the as-rolled and heat-treated Ti-16Al-10Nb sheet. These results showed that the 900 C ductility of the γ alloy was enhanced by vacuum annealing the HIP material at 1400 C. The test results on sheet of the Ti-16Al-10Nb alloy showed that the rolling conditions did not significantly affect the low room temperature ductility found in other investigations.

The details of this program are described in the following.

PREFACE

Titanium alloys containing higher than conventional aluminum concentrations (greater than about 8 percent) have been demonstrated to possess combinations of mechanical properties intermediate between the current Ti-base and the Ni-base superalloys, and have low densities. These properties are of interest for aircraft turbine engines, but poor ambient temperature ductility had limited further development. Additional evaluations of this alloy range were warranted by increased knowledge of this alloy class and mechanisms affecting ductility; greater experience and understanding of the use of materials of low ductility in aircraft engines; and improved processing procedures.

Previous work had indicated two potential composition ranges corresponding to alloys based on Ti_3Al (α_2) and on TiAl (γ). Although considerable information on Ti_3Al-based alloys was developed during the early, but recent, development of titanium alloys, no evaluation has taken place using current techniques and understanding. Alloy development and processing were pursued under U.S. Air Force sponsorship to improve room temperature ductility and elevated temperature properties, and to produce products of typical aircraft shapes from modern processing techniques utilizing some of the special properties of the titanium-aluminide-base alloys. Alloy development, casting and deformation processing programs were sponsored and coordinated by the Air Force over a period of approximately three years. This program on sheet fabrication and evaluation is part of this overall effort.

The results in this report are on processing and forming of Ti-aluminide sheet using rolls heated to temperatures as high as 875 C. Special procedures were thought to be necessary because of the difficulties experienced in preliminary work on extrusion and early tension data showed almost no ductility below about 700 C and a rapidly decreasing ductility with increasing strain rate.

Sheet bar was prepared from rotating electrode powder (REP) of both base alloys and from ingot of the Ti_3Al-base alloys. Special procedures were followed to insure the quality of the powder and ingot stocks. HIP processing of the REP powders was evaluated and developed

during this study. Procedures conducive to high purity were employed during HIP canning and resulted in can-compact interactions and failure during the HIP cycle. These problems were subsequently avoided by using ductile, low strength canning materials or diffusion barriers.

Heated-roll rolling of 0.050 inch sheet of the Ti_3Al-base alloys was consistently successful or the behavior was understood. However, failure was consistently experienced with the TiAl alloy containing the highest aluminum content. This failure, which was attributed to the high aluminum content and the effects of adsorbed films, was intergranular and appeared to result from incomplete particle bonding in fully densified compacts. Although the microstructures of the consolidated powders of this alloy appeared sound, grain growth beyond the powder-particle size range could not be achieved. These problems were averted with a recent generation alloy to which tungsten was added to replace approximately four percent of the aluminum.

Thermal-mechanical treatments based on heated-roll rolling were evaluated to obtain Ti_3Al-base alloy sheet with a fine-grain microstructure for superplastic forming; to improve room temperature ductility; and to provide a range of conditions for evaluation of conventional roll forming of cylindrical shapes. Superplastic forming of simulated aircraft was performed by Rockwell International using this sheet and conventional roll forming of cylindrical shapes successfully performed using rolls heated to 475 C. Tension, bend, creep and fatigue testing of the sheet product was performed over a wide range of temperatures to evaluate the sheet product for potential use primarily for static structures in gas turbine engines.

This program was conducted at Battelle-columbus Laboratories. Dr. A. L. Hoffmanner, Research Leader, Metalworking Section, was the Program Manager. The Project Engineer was D. D. Bhatt, Research Metallurgist. Other contributors to this program were G. S. Serio, Master Technician, C. T. Olofson (retired), Research Metallurgist, and G. E. Meyer, Principal Research Metallurgist, who has since left Battelle.

INTRODUCTION

The titanium aluminides represent an important class of materials to extend the strength, density and oxidation resistance of the titanium alloys from 475 C into the normal operating range of the Ni-base superalloys. Recognition of this potentional has existed for several years[1,2]*, but their development has been limited due to the low room temperature ductility. Among the factors which have warranted additional evaluations of this alloy class are (1) greater experience and knowledge of the use of materials of low ductility (e.g., cast superalloys) in gas turbine engines; (2) improved processing procedures (powder preparation, HIP, casting techniques, heated-roll rolling and superplastic forming); (3) increased knowledge of this alloy class; and (4) the need for an alloy system effective in the 900 to 1500 F range.

Two approaches were pursued to improve the room temperature ductility: alloy development and thermal-mechanical processing. An alloy development program[3,4] was sponsored and coordinated by the U.S. Air Force simultaneously with this rolling program.

One objective of this program was to evaluate thermal-mechanical treatments based on rolling to achieve improved room temperature ductility as well as improved properties at elevated temperatures. The overall objective of this program was to establish fabrication procedures and resulting properties for titanium aluminide-base alloys rolled to sheet of varying thicknesses, formed into complex shapes and joined together to scaled component assemblies typical of those used in aircraft turbine engines. The following three phases were intended to meet these objectives:

Phase A: Rolling

Starting from HIP consolidated rolling preforms, the heated rolling technique developed at Battelle was to be used to produce nominally 1.27 mm thick sheet in widths in excess of approximately 3.80 mm.

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* References are listed on page 113.

Phase B: Forming

Utilizing the sheet produced in Phase A, conventional forming methods were to be evaluated to fabricate scaled shapes typical of those used in aircraft turbine engines. In addition to the conventional forming techniques, superplastic forming was evaluated at Rockwell International[1] using sheet produced on this program.

Phase C: Joining

Utilizing the formed structures produced in Phase B, exploratory evaluations of joining by welding and diffusion bonding were planned.

Table 1 provides a list of the alloys investigated in this program; the alloy designations used throughout this report; and the relative success which was achieved. These alloys represent first generation alloys (e.g., binary alloys, Ti_3Al and TiAl); second generation alloys (e.g., the ternary alloys, Ti-25-5Nb and TiAl+Nb); and third generation alloys (e.g., the alloys Ti-24Al-11Nb and TiAl+Nb+W) in the development of the titanium-aluminide system. With the exception of the base alloys, the latter generations have resulted from an alloy development effort[3,4] of the overall Air Force-sponsored aluminide program (hereafter referred to as the Aluminide Committee).

At the start of the Committee effort, recent work at the Al-rich end of the Ti-Al and Ti-Al-X systems[5,6] refined the positions of phase boundaries. These results showed that the α_2-phase field exists in the range of 21 to about 37 atomic percent aluminum at 700 C in Ti-Al alloys. The α_2 phase is an ordered (DO_{19}) structure based on Ti_3Al (or Ti_3X) and exhibits a ductility dependence which decreases with increasing Al. The α_2 phase is stable to temperatures of about 1000 C, depending on the exact composition, above which it undergoes disordering to α. Transformation of α ($\alpha/\alpha + \beta$) initiates at temperatures above about 900 C and the initiation temperature increases with Al-concentration. For the Ti_3Al alloys in this program, the ordering kinetics are very rapid and appear nearly complete upon still-air cooling of nominally 0.10 inch thick sheet from temperatures above about 1000 C[7]. Ordering in these alloys was insured by aging for 8 hours at 900 C and still air cooling to room temperature. For the alloys in Table 1, the range of stabilities of the various phase fields are approximately: α_2 - below about 1100 C; $\alpha_2 + \alpha$ - 1090 to 1100 C; and $\alpha + \beta$ - 1140 to 1200 C.

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TABLE 1. TITANIUM-ALUMINIDE ALLOYS

Composition (Nominal Percent) Weight				Atomic				Powder Lot No.	Alloy Designation Used in this Report	Source	Current Status	Remarks
Al	Nb	W	Si	Al	Nb	W	Si					
35.5				49.4				395	TiAl	REP/NMI		
35.5	5			47.5	2.1			397	TiAl+Nb	REP/NMI	Partial rolling success at 1440 C	Initial TiAl-base program alloy
15.8	10			25	5			398	Ti-25Al-5Nb	REP/NMI	Successfully rolled	Ti_3Al-base program alloy
13.6	21.4		0.15	24	11		0.25	As-cast	Ti-24Al-11Nb-.25Si	P&W	Successfully rolled	
13.6	21.4			24	11			As-cast	Ti-24Al-11Nb	P&W	Successfully rolled	
32	5	5		47.9	2.1	1.1		606	TiAl+Nb+W	REP/NMI	Successfully rolled	Final TiAl-base program alloy

For aluminum concentrations above 37 atomic percent, two phases, α_2 and γ (based on TiAl with an $L1_0$ structure), exist at temperatures below about 900 C and up to Al concentrations of about 48 atomic percent.[8] The γ phase exists up to the solidus temperature (peritectic temperature of about 1450 C at 50 atomic percent Al).

Approximate definition of the phase boundaries in ternary alloys has been obtained from the binary phase diagram by considering the atomic percent of Al. This binary[9] is shown in Figure 1, and appears to be the best available published diagram based on the findings of the Aluminide Committee. This diagram shows that the Ti_3Al-base alloys in this program should be single phase below about 1000 C, although they were found to contain between about 1 to 15 volume percent β.[7] The position of the $\alpha_2 + \gamma/\gamma$ phase boundary indicates that the TiAl-base alloys may be two phase. This condition was observed for the TiAl + Nb alloy[8,10] which was found to exhibit less than about 5 volume percent of α_2 in a γ matrix.

Consolidation and rolling procedures were based on the phase relations in Figure 1 and conform to the following:

(1) Ti_3Al-Base Alloys

β Treatment - above 1175 C

- Thermal treatment to achieve an acicular, transformed α_2 structure and rapid recrystallization
- Hot working to break down the initial structure, either as-cast or after HIP
- HIP processing at 1220 C (maximum capability of the available HIP unit) to achieve complete consolidation

$\alpha + \beta$ Treatment - 1125 to 1175

- A high finishing temperature to obtain improved workability without the formation of large grains
- Typical temperature range for breakdown of powder compacts and ingot

α (α_2) Treatment - 980 to 1125 C

- Warm working for finishing reduction to achieve a fine-grain wrought structure
- Annealing treatments to obtain a fine, equiaxed, recrystallized structure after working in the same range

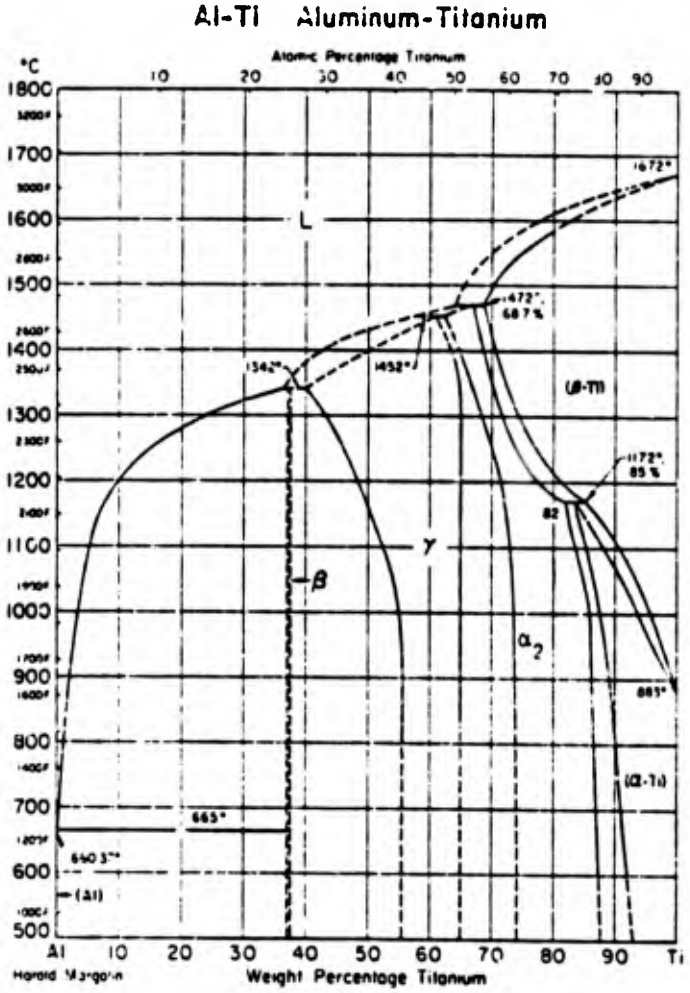


Figure 1. Ti-Al Binary Phase Diagram

(2) TiAl-Base Alloys (essentially all γ treatments)

- 1400 C annealing treatment to improve ductility of HIP-powder comparts
- 1200 to 1300 C for isothermal forging of HIP compacts to insure satisfactory consolidation.
- 1300 to 1415 C preheat temperature for rolling.

The rolling conditions for the Ti_3Al-base alloys were established mainly to evaluate thermal-mechanical treatments to improve room temperature ductility and to achieve a fine grain size (less than 10 microns) for subsequent superplastic forming. The major justification for selecting the rolling conditions for the TiAl-base alloys was based on observation of the rolling results, which were consistently successful only for TiAl + Nb + W alloy.

During this program, supporting results were provided by the efforts of the Aluminide Committee. This work included studies of material behavior,[3,8,10,11] alloy development,[3,8] processing of TiAl-base alloys.[10] and superplastic forming of TiAl-base alloys.[12]. With the exception of the latter, most of the results dealt with structural performance, which provided important insight, but not specific direction, to the production of sheet. The results from the companion studies are referenced where relevant in the following discussion of the fabrication of Ti-aluminide sheet.

EXPERIMENTAL PROCEDURE

Materials

The materials, Ti-25Al-5Nb, TiAl+Nb, and TiAl+Nb+W, upon which this program were based, were produced as rotating electrode powder (REP) with the compositions listed in Table 2. The powders were spherical, in the size range of -35 to 240 mesh, and had a loose density of nominally 65 percent of theoretical. The Ti-24Al-11Nb alloys were provided by Pratt Whitney Aircraft as either machined sections (nominally 1 inch x 1 inch x 3 inch) of a casting, TMCA V-5301, or as isothermally forged slabs (nominally 7 inch x 5 inch x 0.5 inch) from castings TMCA V5414. The compositions of the castings are listed in Table 2. A detailed description of the casting practices are provided in Reference 3. In addition to these materials, REP-TiAl powder (Lot No. 395) was used to evaluate HIP (hot isostatic pressing) parameters.

HIP (Hot Isostatic Pressing) Consolidation of Rolling Preforms

HIP processing of the REP powders was performed in five HIP processing runs, contrary to the original program plan, because of difficulties encountered during HIP processing. The first four HIP runs were largely exploratory because of the unanticipated results. These processing conditions are summarized in Table 3. The following materials and practices were standardized in the billet preparation procedure as summarized in Table 3:

(1) Cladding

Ti55A sheet, either 0.02 inch or 0.078 inch thick, and Ti-6Al-4V sheet, either 0.050 or 0.075 inch thick, were used for cladding.

(2) Can Design

Two can designs were used: (1) the swastika or lap joint design shown in Figure 1 was the preferred can structure because it was felt that HIP would improve the integrity of the joint, and (2) the butt-welded

TABLE 2. POWDER AND INGOT ANALYSES
(Weight Percent)

Element	Alloy Description				
	Ti-25Al-5Nb (Lot No. 398)	TiAl+Nb (Lot No. 397)	TiAl+Nb+W (Lot No. 606)	Ti-24Al-11Nb (TMCA V-5301)	Ti-24Al-11Nb-.25Si (TMCA V-5414)
Al	16.46	36.40	33.2	14.2	14.2
Nb	10.25	4.60	6.2	20.4	22.4
O	839 ppm	618 ppm	1200 ppm	1620 ppm	900 ppm
C	0.007	0.006	0.02		
H	41 ppm	13 ppm	20 ppm		
N	12 ppm	11 ppm	50 ppm	155 ppm	100 ppm
Fe	0.06	0.05	0.13		0.16
Si	0.015	0.033	0.04		0.14
W	<0.01	<0.01	6.8		
Cu	0.01	0.03	0.07		
Mo	<0.01	<0.01	0.012		
Sn	0.04	0.04	<0.01		
Ni	<0.01	<0.01	<0.01		
Zn	<0.01	<0.01	<0.01		
Mn	<0.01	<0.01	<0.01		
Mg	<0.01	<0.01	<0.01		

TABLE 3. SUMMARY OF HIP BILLET PROCESSING PARAMETERS AND CODE SYSTEM FOR BILLET IDENTIFICATION

HIP Run No.[a]	1	2	3			4			5
HIP Parameters									
Temperature, C.	1225	1225	1225			1225			1225
Pressure, ksi	20-15	25-30	26			26			26
Time, min	[illegible]	180	180			180			180
Other						Slow Cool			Slow Cool
Identification No.[b]	10	20	30	40	50	60	70	80	90
Cladding	Ti55A	Ti-6Al-4V	Ti-6Al-4V	Ti-6Al-4V	Ti-6Al-4V	Ti55A	Ti55A	Ti55A	Ti55A
Cladding thickness	0.020	0.050	0.075	0.075	0.075	0.020	0.078	0.078	0.078
Can Design	Swastika	Butt	Butt	Butt	Butt	Swastika	Butt	Butt Al_2O_3 Coat.	Butt
Outgassing procedure	Cold EB; no fill tube	Cold EB; no fill tube	Cold; vac. out gas	Warm; vac. out gas	Vac. fill and warm vac. out gas	Cold EB; no fill tube	Warm vac. out gas	Warm vac. out gas	Warm vac. out gas
Material Identification									
Ti-15.8Al-10Nb Lot No. 398 [c]	1810 to 1819	2820 to 2825	3830	3840	- -	4860	- -	4840	5890 to 5899
Ti-35.5Al Lot No. 395[c]			3530	3540 3541	3550	4560	4570 4571	4580	
Ti-35.5Al-5Nb Lot No. 397 [c]	1710 to 1719	2720 to 2725	3730	3740 3741	3750	4760	4770 4771	4780	
Ti-35Al-5Nb-1W Lot No. 606 [c]									5690 to 5699-5

(a) HIP run number was used for the first digit of the billet identification number.

(b) Can identification number was used for the third and fourth digit of the billet identification number for the first billet. Subsequent billets in the same series varied the fourth digit.

(c) Last number in the material lot number was used as the second digit for the identification numbers.

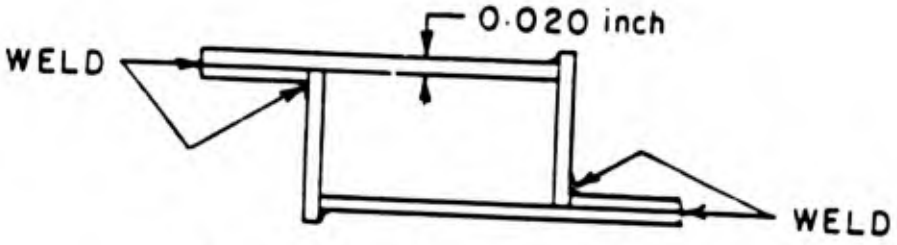
design shown in Figure 2 was used to accommodate heavier and higher strength can thicknesses which were not easily formed. The swastika-design can was prepared from cold formed cover plates and was restricted to moderate ductility and low strength material of relatively small thickness to facilitate forming. The following two nominal compact sizes were used for this evaluation: 6 x 3 x 0.5 inch and 9 x 6 x 0.5 inch. The actual can dimensions were calculated based on a 60 percent density of the loose powders.

In addition to the two can-joint designs, the butt-welded construction was used with both an open-joint and an attached fill tube for loading with powder. The fill tube was Ti55A of 0.375 inch outside diameter and 0.065 inch wall thickness. The tube length from the can was 18 inches long to facilitate sealing.

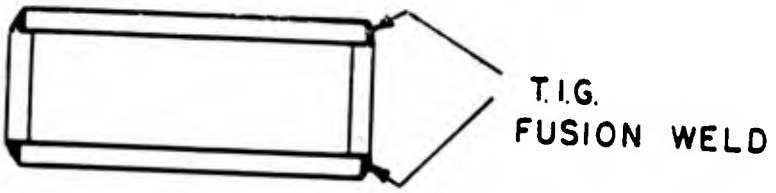
After forming or shearing the can segments to the appropriate sizes, the cans were partially welded, with either a side or fill tube left open, pickled with a HNO_3-HF solution, water and alcohol rinsed and dried. After this cleaning, the cans were filled using various filling procedures. One exception to the above procedures was used in HIP Run No. 4 with the Al_2O_3 plasma-sprayed cans. In this assembly procedure, the can segments were first cleaned, then plasma sprayed with the fill tube welded in place with no further chemical cleaning. All joining for can assembly was performed by TIG welding.

(3) Outgassing Procedure

With the exception of the vacuum filled cans, all cans were filled in air and densified on a vibratory table. In HIP Run Numbers 1 and 2, the cover for the open side of the filled cans was placed in position and the filled-assembled cans were placed in a glove box. The glove box was evacuated to 10^{-4} torr and then backfilled with argon. The open end was then sealed by TIG (GTA) welding except for one seam which was forced open to permit subsequent outgassing. After TIG welding, the cans were transferred to an EB (electron beam) welding chamber and outgassed at room temperature over night at a chamber pressure less than 10^{-4} torr. This procedure was exactly followed for the cans in HIP Run No. 1. For Run No. 2, difficulty was encountered in sealing the open seam of



a. Swastika or Lap-Joint Can Design



b. Butt-Weld Can Design

Figure 2. HIP Can-Joint Designs Evaluated for Fabricating Rolling Preforms

the thicker cans by EB welding. These cans were then returned to the glove box where they were placed in a fixture and a partial TIG weld pass made to reduce the seam opening. The cans were then successfully EB welded using the previously described procedures.

A modification of these procedures was used for a series of cans (No. 60 in Table 3) for HIP Run No. 4. This modification involved air filling, vacuum outgassing and EB welding, but eliminated the exposure to argon. This approach was feasible because the thin, soft cladding could be manipulated sufficiently in the EB chamber to permit reducing the joint gap sufficiently to permit sealing by EB welding.

Fill-tube filling in air and vacuum was evaluated in HIP Runs 3 and 4 (Can Identification Numbers 30,40,50,70 and 80) and used in Run 5. The following procedures were evaluated:

(1) Air fill and vacuum outgas to 10^{-5} torr

  (a) Room temperature outgas for in excess of 16 hours

  (b) 175 C vacuum outgas for in excess of 16 hours

(2) Vacuum fill and warm vacuum outgas to 10^{-5} torr as in (1)(b).

Prior to filling, all cans were evacuated and exposed to air twice to insure elimination of argon possibly contained from the TIG welding used for can fabrication.

The can-tube assemblies were filled nominally 0.75 inch into the fill tube and then outgassed. When outgassing was complete (no measurable static leak rate), the fill tube was locally heated, collapsed and then forged shut. The can-fill tube assemblies were then transferred to the EB welding chamber where EB welding was used to melt, seal and sever the tube one inch from the can. After sealing, all cans were helium leak tested to insure the absence of cracks, porosity and open seams in the can.

The preforms fabricated for HIP Runs 1 through 4 were designed to provide 6 x 3 x 0.5 inch compacts. The HIP preforms for HIP Run 5 were fabricated to provide the following compact sizes:

(1) Ti-25Al-5Nb

  (a) 6 x 3 x 0.5 inch compacts - Compact Numbers 5890 to 5898

  (b) 9 x 6 x 0.5 inch compact - Compact Number 5899

(2) TiAl+Nb+W

(a) 6 x 3 x 0.5 inch compacts - Compact Numbers 5690 to 5699-1, and 5699-2 and 5699-3

(b) 9 x 6 x 0.5 inch compacts - Compact Numbers 5699-4 and 5699-5.

(4) HIP Processing

The HIP cycle consisted of the following sequence:

(1) A 4-hour heat to 1225 C (+ 0/-15 C) to a pressure of 26,000 to 30,000 psi. Because of occasional system leaks, the nominal pressure was not always attained as shown in Table 3.

(2) A 3-hour hold at 1225 C and at nominal pressure. The only exception to this practice was HIP Run No. 1 which was held at temperature a total of 2.5 hours because of some minor problems with the HIP system.

(3) An under-pressure (argon) cool to approximately 80 C.

Fast and slow cooling rates were evaluated. The slow rate was approximately one-half the fast rate, which exhibited the following time-temperature behavior:

Hours from Start of Cooling	Temperature (C)
0	1225
0.5	900
1.0	700
1.5	550
2.0	440
2.5	350
3.0	290
.	
.	
.	
8.0	∿ 80

HIP compacts from Run No. 2 are shown in Figure 3. The HIP compacts were evaluated metallographically and occasionally by radiography and ultrasonics to establish if they were crack free. Tests for TIP (thermally induced porosity) were performed by heating in vacuum the Ti_3Al-base alloys

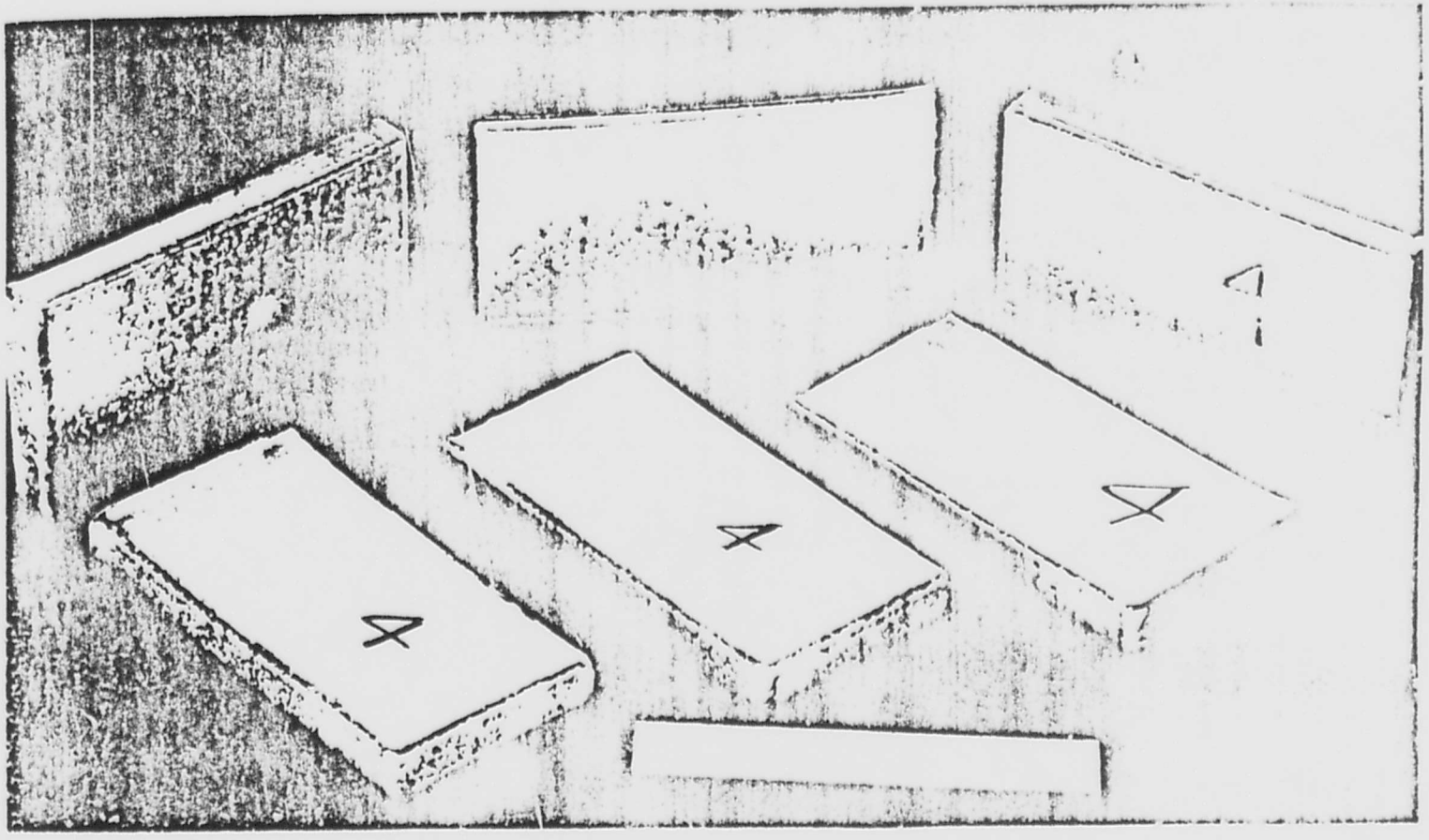


Figure 3. Typical HIP Compacts from the Second HIP Series

(2) TiAl+Nb+W

(a) 6 x 3 x 0.5 inch compacts - Compact Numbers 5690 to 5699-1, and 5699-2 and 5699-3

(b) 9 x 6 x 0.5 inch compacts - Compact Numbers 5699-4 and 5699-5.

(4) HIP Processing

The HIP cycle consisted of the following sequence:

(1) A 4-hour heat to 1225 C (+ 0/-15 C) to a pressure of 26,000 to 30,000 psi. Because of occasional system leaks, the nominal pressure was not always attained as shown in Table 3.

(2) A 3-hour hold at 1225 C and at nominal pressure. The only exception to this practice was HIP Run No. 1 which was held at temperature a total of 2.5 hours because of some minor problems with the HIP system.

(3) An under-pressure (argon) cool to approximately 80 C.

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1.5	550
2.0	440
2.5	350
3.0	290
.	
.	
.	
8.0	~ 80

HIP compacts from Run No. 2 are shown in Figure 3. The HIP compacts were evaluated metallographically and occasionally by radiography and ultrasonics to establish if they were crack free. Tests for TIP (thermally induced porosity) were performed by heating in vacuum the Ti_3Al-base alloys

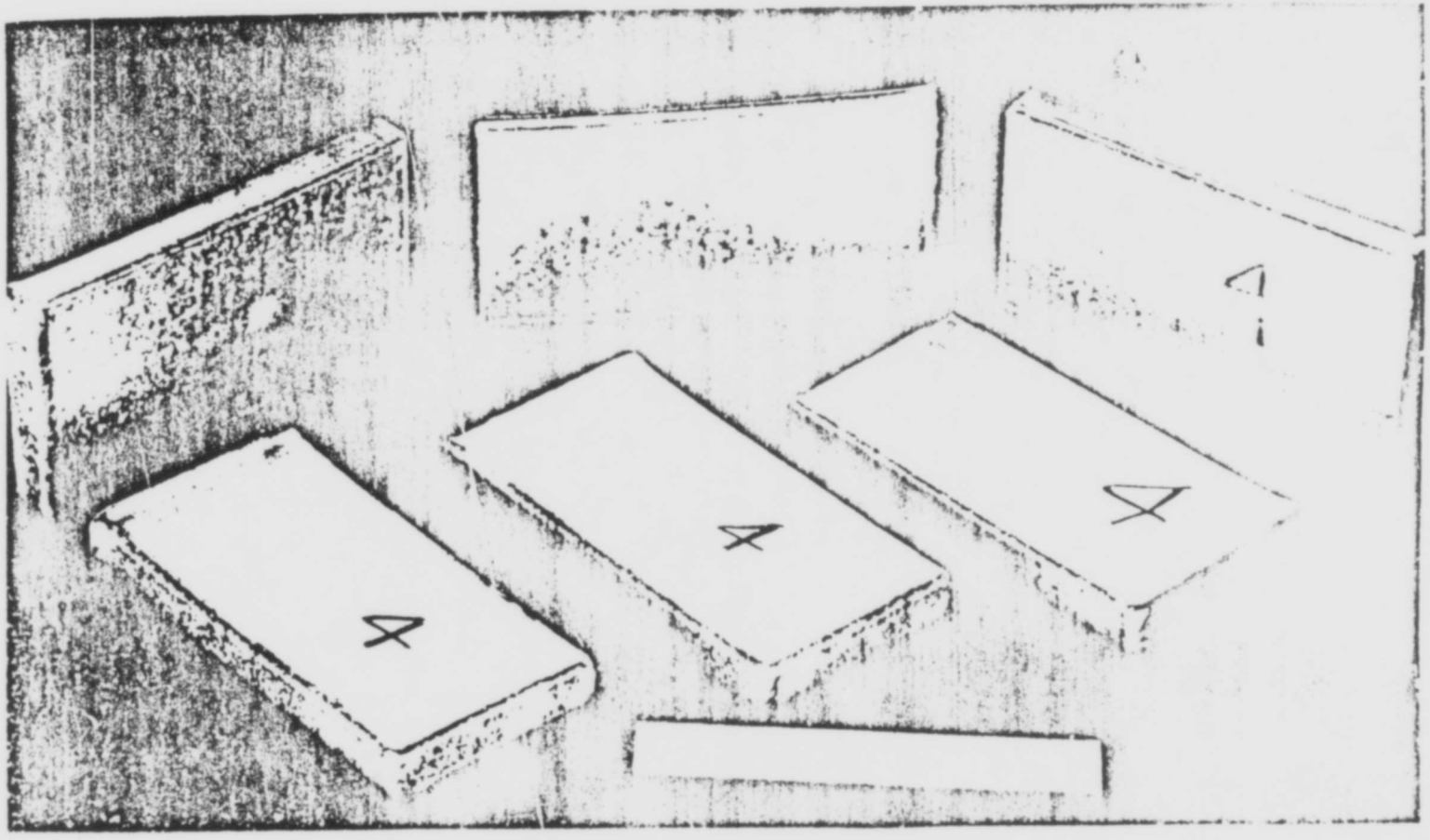


Figure 3. Typical HIP Compacts from the Second HIP Series

to 1275 C and the TiAl-base alloys to 1400 C in vacuum and, after cooling, visually examining for blisters in the cladding.

Post HIP Processing

After consolidation, the HIP compacts of the TiAl-base alloys were either rolled directly or given the following post-HIP treatments:

(1) Heated to 1400 C for three hours in vacuum and furnace cooled to room temperature to improve ductility and to test for gas absorption;

(2) Isothermally forged in vacuum at temperatures between 1280 to 1380 C to reductions in the range of 25 to 50 percent;

(3) Declad after either or both heating to 1400 C or isothermal forging and reclad with either Ti-6Al-4V or clad with 0.050 inch thick tantalum sheet and reclad with Ti-6Al-4V.

Rolling

Rolling was performed without coatings or lubrication at speeds in the range of 30 to 70 feet/min at reductions up to 50 percent per pass. Rolling was performed at Battelle usually with heated rolls, but occasionally conventionally, in the following mills:

(1) Two-high 8 x 12 inch Experimental Heated Rolling Mill[13]

This mill, originally manufactured by United Engineering and Foundry Company, was modified at Battelle for heated-roll rolling with roll temperatures up to 900 C. In addition to the basic mill design described in Reference 13, the water-cooled roll arbors were equipped with IN100 sleeves to permit achieving working-roll temperatures up to 900 C. The rolls in this mill were induction heated and monitored with a contact pyrometer.

This mill was equipped with calibrated, water-cooled load cells under each screw to provide recordings of the separating force.

(2) Four-high 16 x 24 inch Heated Rolling Mill[14]

This mill was originally manufactured by United Engineering and Foundry Company as a two-high 16 x 24 inch conventional hot mill and modified under U.S.A.F. contract[14] at Battelle to provide a heated-roll sheet mill for aerospace materials. The roll dimensions are as follows:

(a) Work Rolls: AF2-1DA-4 inch diameter x 20.5 inch length

(b) Back-up Rolls: H 11 - 13 inch diameter x 21 inch length.

The maximum operating temperature is 840 C which is attained with radiant-quartz lamps and monitored with an Ircon optical pyrometer. The size range of operation of this mill is 1.13 to 0.02 inches.

The four-high heated roll subassembly is installed into the housing of the 16 x 24 inch mill. The existing drive and controls are used to provide a continuous mill speed range of 0 to 300 feet per minute. The mill is equipped with strain bars on the mill stand columns and water-cooled load cells under each screw to provide recordings of separating froce. Figure 4 is a picture of this system.

The rolling preforms were preheated for 15 minutes after the surface temperature reached the preheat temperature and all intermediate preheating was for 5 minutes in a flowing argon atmosphere. Feeding and removal of the sheet during rolling was performed manually to eliminate or minimize contact with the feed table. The feed table on the two-high mill was heated to 500 C and no feed table was used on the four-high mill.

The early rolling trials showed that handling of the in-process sheet was a major source of chilling, cracking and high loads. To avert this problem, tongs wrapped with asbestos and widely spaced tungsten wire were heated and used for handling the sheet.

Rolling of Ti_3Al-base alloys was performed at preheat temperatures from 900 to 1200 C with roll temperatures of 125 C (conventional rolling) and at controlled temperatures in the range of 700 to 900 C. The TiAl-base alloys were rolled preheated in the range of 1150 to 1415 C with roll temperatures in the range of 800 to 900 C.

Post Rolling Treatments

Cooling after rolling was either in air or vermiculite, although furnace cooling was initially evaluated for the TiAl-base alloys. After cooling, the sheet was cut into the appropriate sizes for subsequent processing. The cutoff procedures were not extensively evaluated, although cracking was essentially eliminated with a water-cooled diamond wheel (4000 SFM) at a rate of less than 0.5 in/min. Conventional cutoff wheels usually produced burning of both alloy types and cracking of the TiAl-base alloys.

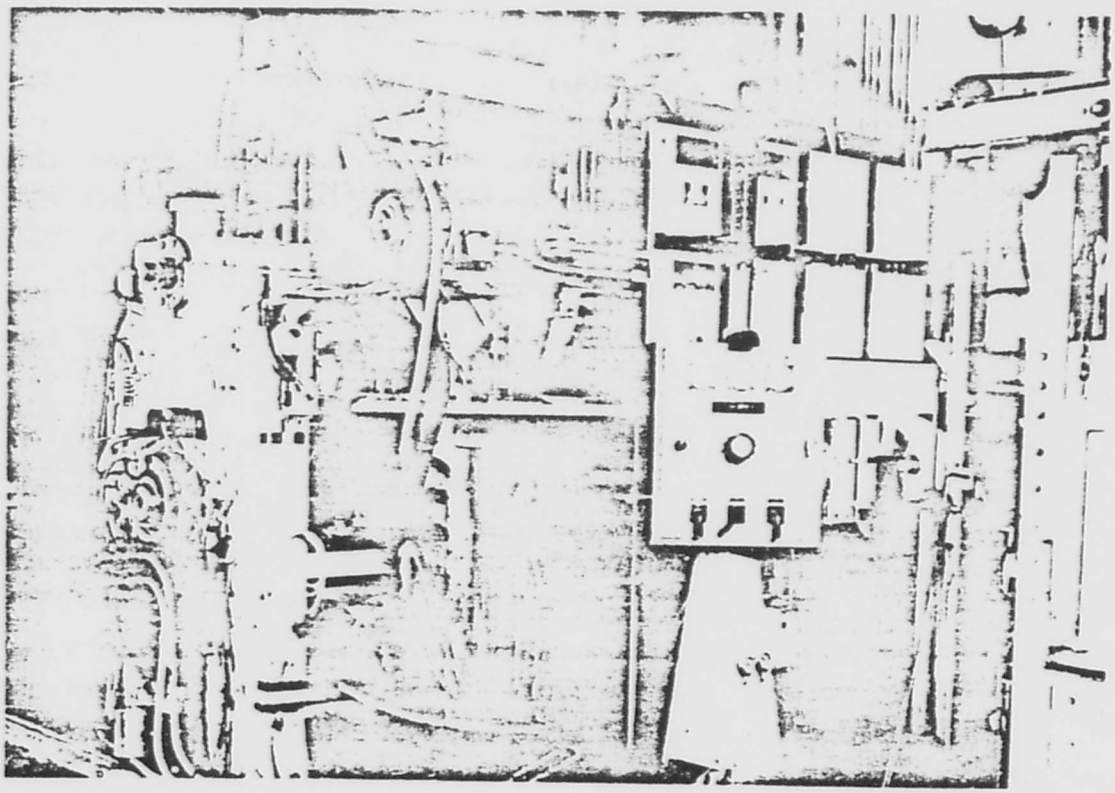


Figure 4. Four-High 16 x 24 Inch Heated Rolling Mill

Flattening of sheet in lengths less than 0.762 m was performed by placing the sheet between two 304 stainless steel plates and heating the assembly to 1000 C in an air furnace. After reaching temperature, the assembly was removed from the furnace, pressed at a pressure of 13.8 MPa for about 1 minute, removed from the press, and allowed to cool to below 500 C. The flatness of 1.52 mm sheet was typically less than 4.17 mm/m after this treatment. Both bare and clad sheet were treated in this manner. In general, flattening was an optional operation depending on the flatness of the as-rolled sheet, which was not entirely controllable. Sheet flatness was impaired by handling and became worse as the length and preheating temperature increased and the thickness decreased. Straightening with light rolling passes was not consistently successful.

Forming

Forming of Ti-25Al-5Nb was performed on as-rolled and ground sheet after various rolling schedules established to promote formability. Preliminary trials with vacuum forming were performed in an argon-atmosphere furnace with the ceramic tooling shown schematically in Figure 5. This work was discontinued at Battelle after limited success and subsequently continued at Rockwell International[12] with sheet provided from this program. Conventional rolling of cylindrical shapes was performed on the heated roll former shown in Figure 6.

Roll forming was performed on 1.27 to 1.52 mm thick ground sheet samples up to 100 mm wide, preheated to temperatures in the range of 980 to 1100 C with the temperature of the rolls maintained at 400 C. The roll speed was maintained manually at approximately 3.1 m per minute. Because of the stiffness of the sheet, multiple-pass reductions were required to produce cylindrical structures with diameters less than 76 mm. During rolling, the exit temperature of the in-process sheet was measured with a contact pyrometer after each pass. These results showed consistent success could be attained in the multiple pass sequence if the sheet temperature did not cool below 600 C.

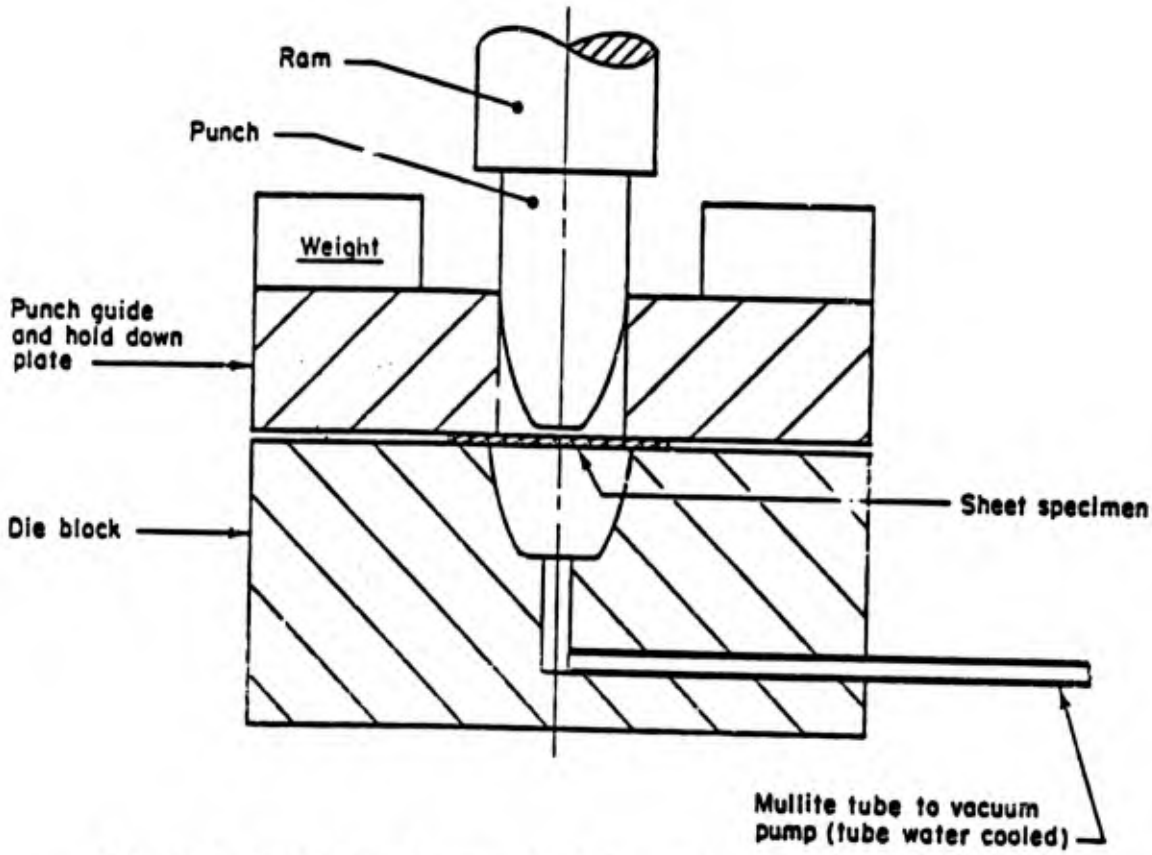


Figure 5. Schematic Diagram of Tooling Evaluated for Vacuum Forming of Ti-25Al-5Nb

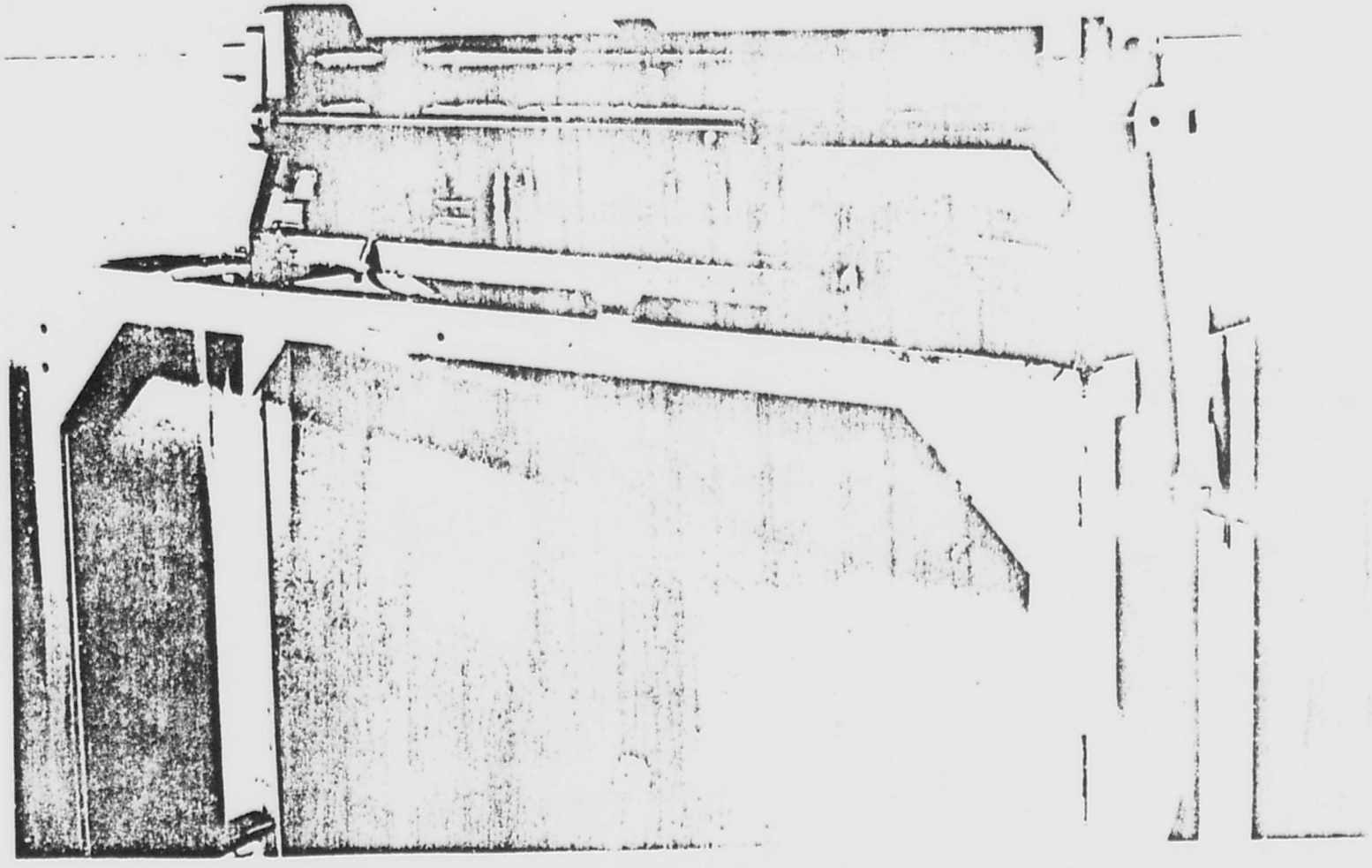


Figure 6. Photograph of Heated Three Roll Forming Equipment

Heat Treatment

All sheet samples were heat treated as rolled, either bare or clad, and after grit blasting. No special procedures were used for heat treatment. The actual heat treatments used are described in the discussion of results. More detail on the justification for these treatments is provided in Reference 3.

Sample Preparation

Sample preparation usually involved the following sequence of operation:

(1) Diamond wheel cutoff to the appropriate sample or coupon size followed by visual and dye penetrant inspection of the cut surfaces
(2) Heat treatment
(3) Grinding to remove the cladding (if sheet were rolled clad) and to provide the finished specimen shape.

Grinding was qualitatively evaluated to establish preferred grinding conditions to avoid cracking and/or crazing frequently encountered with both alloy types. Several wheels from Battelle's grinding wheel inventory and fluids available for grinding were evaluated after decladding with both alloys. Although no attempt was made to optimize these practices, preferred conditions were established which visually appeared to avert surface cracks and, upon subsequent grinding of samples with surface cracks, to remove surface produced by previous more abusive grinding conditions. The preferred grinding conditions which were employed are as follows:

(1) Grinding wheels (silicon carbide)
    39C100H8VK or 39C100I8VK
(2) Grinding fluid - Houghton 50 Grind (diluted 50:1)
(3) Wheel speed - 1370 m per minute
(4) Part speed - 2 m per minute.

Testing

The following testing and specimen designs were used with material after HIP and on the rolled and heat treated sheet product.

(1) Tensile testing

(a) Button head tensile specimens were used for testing of the material after HIP, to establish post-HIP processing conditions, and perform preliminary workability evaluations. This specimen design is shown in Figure 7.

(b) Modified subsize ASTM sheet tensile specimens, shown in Figure 8, were used for evaluation of the sheet.

All tensile testing was performed in an Irstron Universal Testing Machine at the indicated strain rates and temperatures achieved by heating in an electric-resistance furnace. Temperature control was typically ± 4 C.

(2) Bend testing from -73 to 550 C was performed by three-point loading over a 19 mm span at a cross head speed of 0.025 mm/minute. The specimens had the following nominal dimensions: 1.3 mm thickness, 19 mm width, and 38 mm length. Testing below room temperature was performed with the specimen and test fixture immersed in a constant temperature bath inside an insulated container. Elevated temperature testing was performed with the specimen and test fixture contained inside a resistance-heated furnace controlled to within ± 6 C of the preset temperature.

(3) Creep or stress-rupture testing were performed in Battelle-designed and constructed creep racks with the sheet specimen design shown in Figure 8. Prior to initiation of the test, the specimen was allowed to

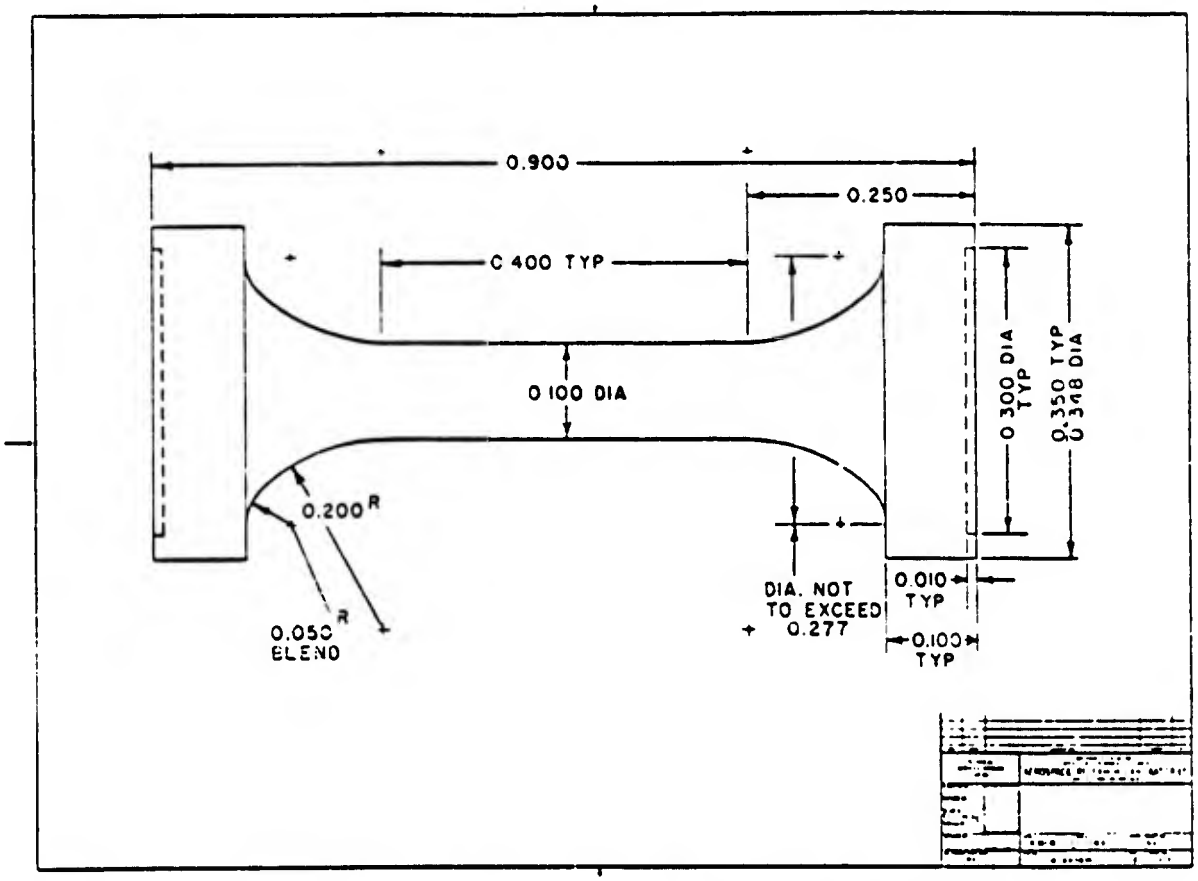


Figure 7. Button-Head Tensile Specimen

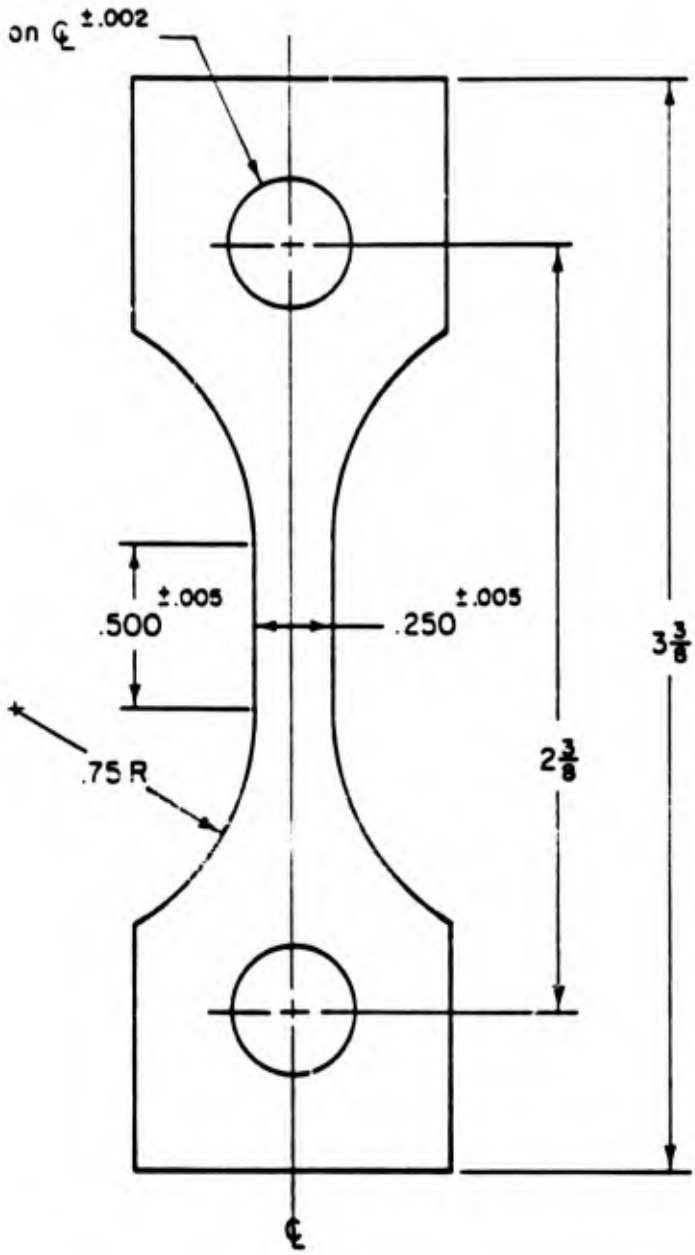


Figure 8. Modified ASTM Subsize Sheet Tensile Specimen (Dimensions in Inches)

equilibrate to the furnace temperature before load was applied. Temperature control during creep testing was typically $\pm$ 3 C and displacement was measured optically with a Pt-type strip gage to a sensitivity of 50 x 10^{-6} m/m.

(4) Fatigue testing was performed on a 2.5 KIP MTS machine with the specimen design in Figure 8. Testing was performed with a stress ratio (R) of 0.1 at a frequency of 30 Hz. Temperature was controlled to within $\pm$ 3 C with a calibrated thermocouple spot welded just outside the gage section.

In addition to the described procedures, metallography, usually with bright field or polarized light illumination, and other conventional testing and inspection procedures were used in this program. The conventional techniques and the particular reason they were used are discussed in the following section on the program results. Metallographic specimens for microstructural evaluation were etched with Kroll's reagent for bright-field illumination and were anodized with a 1:1 phosphoric acid: ethyl alcohol solution at 40 volts D.C. for 1 second for polarized light illumination. The latter was used to provide improved resolution of grains for grain size determinations.

RESULTS

The experimental results are presented in the following in sequence of the HIP trials for each of the steps in the processing sequence. The conditions used in these trials were summarized in Table 3. This sequence was used because HIP processing was found to be a critical factor affecting the quality of the starting material and, consequently, the quality of the rolled sheet.

HIP Processing Results

The HIP results for HIP Run No. 1 were considered successful with the exception that severe barreling of the cans, as a result of the small Ti-55A can thickness, resulted in a crown on the HIP compacts. This crown resulted in a compact thickness variation between 13.7 and 18.5 mm. Such a crown is known to cause tensile stresses which will promote failure in materials of marginal ductility. For this reason, sheet of greater thickness was planned for the next HIP Run No. 2.

Metallography and mechanical testing were performed to evaluate the quality of the HIP-processed material to establish the necessity for post-HIP annealing treatments and to determine preliminary rolling conditions. These evaluations were performed on three equal sections of HIP Compact numbers 1711 (TiAl+Nb) and 1811 (Ti-25Al-5Nb). These sections were given the following treatments:

(1) TiAl+Nb (No. 1711)
- 1711-1 As-HIP processed
- 1711-2 Annealed for 3 hours in vacuum at 1300 C and furnace cooled
- 1711-3 Annealed for 3 hours in vacuum at 1400 C and furnace cooled

(2) Ti-25Al-5Nb (No. 1811)
- 1811-1 As-HIP processed
- 1811-2 Annealed for 3 hours in vacuum at 1300 C and furnace cooled
- 1811-3 Annealed for 3 hours in vacuum at 1400 C and furnace cooled.

Visual observations of the specimens showed profuse cracking on the abrasive-cut surfaces of the TiAl+Nb samples. This cracking occurred to depths as great as 0.5 mm, but it was later found to be partially or completely averted by diamond-wheel cutoff. There were no indications of

thermally induced porosity from exposure to the elevated temperatures during annealing. Subsequent metallographic evaluations (Figures 9 through 14) of the material after HIP and after the thermal treatments showed some small, very sparce porosity. Examples of this porosity are shown in Figures 10, 11, and 13.

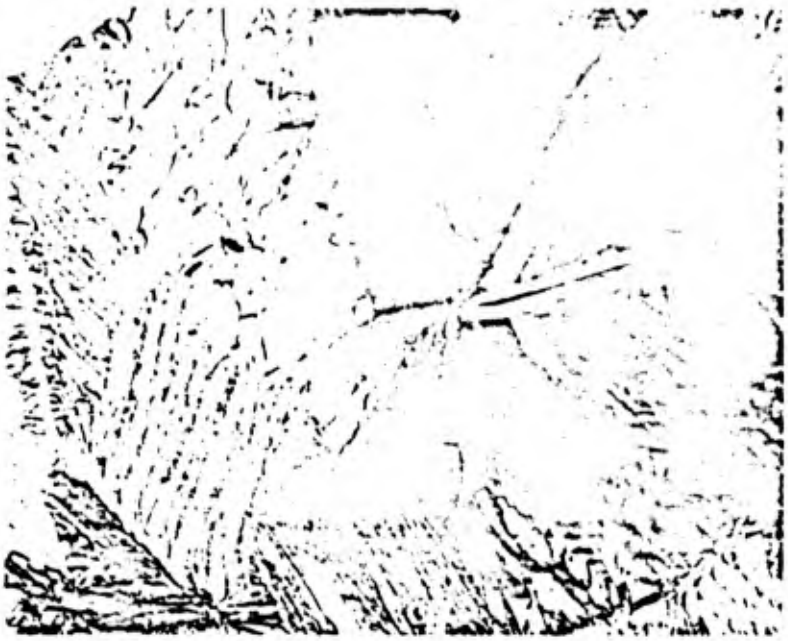
The micrographs showed that all Ti-25Al-5Nb structures exhibited a large prior beta grain size, on the order of 1000μ, and the coarseness of the α_2 plates appeared to increase with increasing annealing temperature. These observations indicate that the 1225 C HIP temperature may be too high and that higher annealing temperatures would not facilitate rolling because of the usual adverse effects from large grain size. Furthermore (as will be shown more clearly in subsequent micrographs), primary alpha usually forms as equiaxed grains along prior beta grains. The combined effects of the large beta grain size and the discontinuity in the microstructure caused by the primary alpha grains about the large prior grains indicated that special consideration must be given to break down (e.g., grain refinement) of the HIP structure before specific thermal-mechanical treatments were evaluated for the Ti-25Al-5Nb compacts. As anticipated, the microstructures in Figures 9 through 11 are essentially single-phase α_2 structures of differing degrees of coarseness.

The micrographs of TiAl-Nb show that the increase of the average grain size from the HIP structure, Figure 12, to the structure after annealing at 1400 C, Figure 14, was from approximately 90 to 130 microns. This increase was considered desirable because it showed that some grain boundary motion had occurred which indicated some integrity of the overall microstructures. However, this change in grain size was not as large as anticipated for most structures annealed near their melting temperatures.

The microstructures in Figures 12 through 14 appear to be single-phase γ structures. At higher magnifications, α_2 can usually be observed at low volume fractions.

Figure 9. Ti-25Al-5Nb As HIPed

Figure 10. Ti-25Al-5Nb After 3 Hours at 1300 C



100X

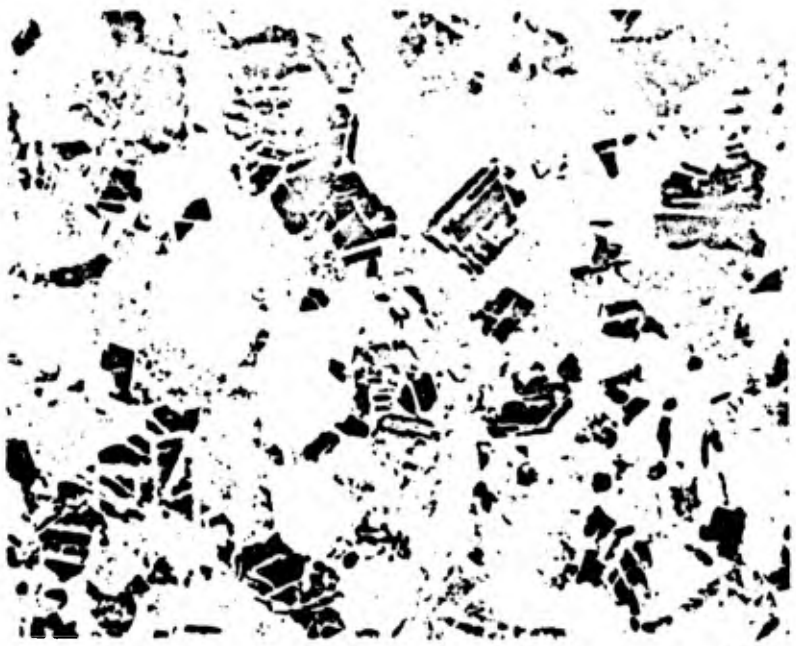
Figure 11. Ti-25Al-5Nb After 3 Hours at 1400C

100X

Figure 12. TiAl + Nb As HIPed

100X

Figure 13. TiAl + Nb After 3 Hours at 1300C



100X

Figure 14. TiAl + Nb After 3 Hours at 1400 C

Tensile tests were performed on the samples from which the micrographs in Figures 9 through 14 were obtained. These tests were performed with the button-head specimen shown in Figure 7 at a nominal strain rate of $10^{-3}sec^{-1}$ and cycled to $10^{-2}sec^{-1}$ to determine the strain rate sensitivity, m. The results of these tests presented in Table 4 indicate the following:

(1) Thermal treatments on Ti-25Al-5Nb HIP samples increased the R.T. tensile stress, but produced a negligible-to-degrading effect on 900 C ductility.

(2) The thermal treatment at 1400 C on TiAl+Nb HIP samples was definitely beneficial to 900 C ductility, which was improved nearly five fold.

(3) None of the samples were superplastic, as would be expected from the large grain size.

The 900 C test temperature to assess workability was calculated as a baseline temperature which would be the lowest temperature approached as a result of rolling under conditions most conducive to chilling. The basis for this selection is discussed in the discussion of the rolling results.

The rolling trials with the HIP Run No. 1 were successful for the Ti-25Al-5Nb alloy, but not for TiAl+Nb with and without the thermal treatments. The hot rolling failure analysis (as discussed in the section on Rolling) showed that failure was intergranular with no indication of recrystallization. Loose powder was also apparent about the fracture areas. In an attempt to avert these problems, heavier cladding of higher strength material (Ti-6Al-4V) was planned for HIP Run No. 2. The use of 1.27 mm thick Ti-6Al-4V cladding was planned to avert the crowning and to provide increased thermal protection on the initial passes. This cladding was used for both alloys.

TABLE 4. TENSILE TEST RESULTS ON Ti-25Al-5Nb AND TiAl+Nb AFTER HIP AND BEFORE THERMAL TREATMENTS

Specimen Number	Condition	Temperature, degrees F	Yield Stress, psi x 10^{-3}	Tensile Stress, psi x 10^{-3}	Elongation, percent	R.A., percent	Strain Rate, sec.$^{-1}$	m*	Remarks
• HIP SAMPLES									
(1) Ti-25Al-5Nb (No. 1811)									
1811-1A	As-HIPped	R.T.		41.6	--	--	0.001	--	
1811-2A	1300 C/3 hours	R.T.		46.3	--	--	0.001	--	
1811-3A	1400 C/3 hours	R.T.		54.2	--	--	0.001	--	
(7) Ti-25Al-5Nb (No. 1811)									
1811-1B	As-HIPped	900 C	24.0	70.3	7.6	17.2	0.001	0.00	
1811-2B	1300 C/3 hours	900 C	27.9	29.3	6.0	9.9	0.001	0.09	
1811-3B	1400 C/3 hours	900 C	27.3	28.9	9.1	11.6	0.001	0.00	
(1) TiAl+Nb (No. 1711)									
1711-1A	As-HIPped	900 C	26.6	28.3	10.8	14.1	0.001	0.1	
1711-1B	As-HIPped	900 C	29.0	31.7	10.7	20.1	0.001	--	Constant strain rate test
1711-2A	1300 C/3 hours	900 C	31.4	33.9	3.8	5.9	0.001	0.06	
1711-2B	1300 C/3 hours	900 C	28.7	33.7	3.4	7.1	0.001	--	Constant strain rate test
1711-3A	1400 C/3 hours	900 C	21.6	23.6	31.1	39.8	0.001	0.13	
1711-3B	1400 C/3 hours	900 C	19.9	21.9	34.6	67.8	0.001	--	Constant strain rate test

* m was determined by strain-rate cycling between 0.001 and 0.01 sec.$^{-1}$.

The thicker cladding of Ti-6Al-4V for HIP Run No. 2 could not be consistently cold formed so cans of the butt-welded design were used. After can filling, partial welding in an argon-atmosphere glove box and outgassing in an EB chamber, it was discovered that the open seam was distorted sufficiently to impair the quality of the EB weld by either or both a large joint gap or distortion during final welding. To correct this problem, the compacts were returned to the blove box for additional TIG welding to insure the success of the final EB welding. This practice restricted the joint opening for outgassing and may have produced some sintering of the powder to further reduce outgassing. Upon subsequent annealing of the TiAl+5Nb compacts from HIP Run No. 2, severe indications of TIP (thermally induced porosity) were readily observed. Some blisters were as large as 2 inches in diameter and 1 inch in height. Subsequently, the Ti-25Al-5Nb compacts from HIP Run No. 2 were given a TIP test in vacuum at 1200 C. The compacts of both alloys which did not exhibit TIP were rolled; however, occasional TIP was observed in the rolled sheet. For these reasons, air and vacuum outgassing with no exposure to inert gases were evaluated in HIP Run No. 3.

Three can-filling techniques with fill tubes were evaluated with 0.075-inch-thick Ti-6Al-4V cans in HIP Run No. 3. These techniques were as follows:

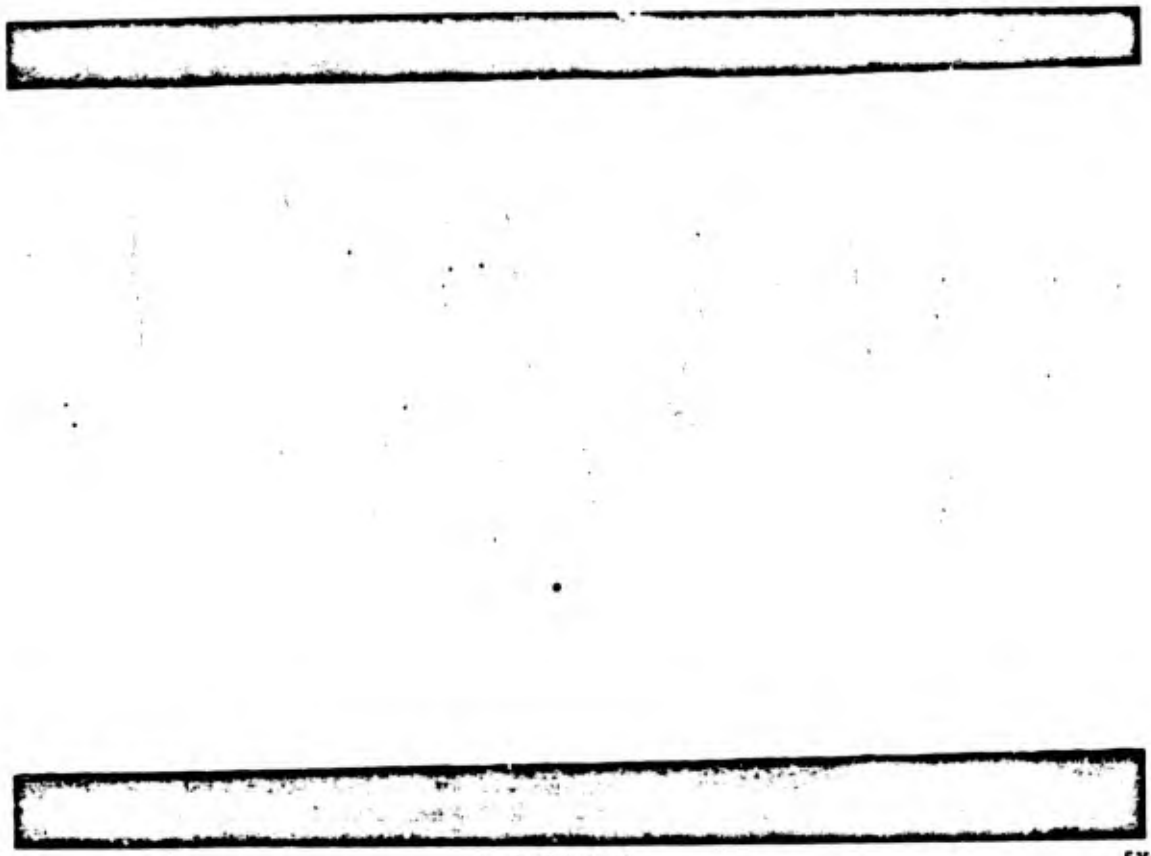
(1) Can Series 30: Can filling in air with cold vacuum outgassing
(2) Can Series 40: Can filling in air with warm (175 C) vacuum outgassing
(3) Can Series 50: Can filling in vacuum with warm (175 C) vacuum outgassing.

After outgassing was complete, the fill tubes were heated externally and forged closed with a press, without disturbing the vacuum. After closure, the cans were transferred to an EB chamber and the fill tube was severed by melting.

Evaluations of the HIP compacts from the 3 can series showed that all of the compacts for all of the alloys were severely cracked on the cut and polished surface. An example of this failure for TiAl compact No. 3530 is shown in Figure 15. The cracks on the Ti-25Al-5Nb compacts were more closely spaced (e.g., higher crack density) than for either of the two TiAl-base alloys.

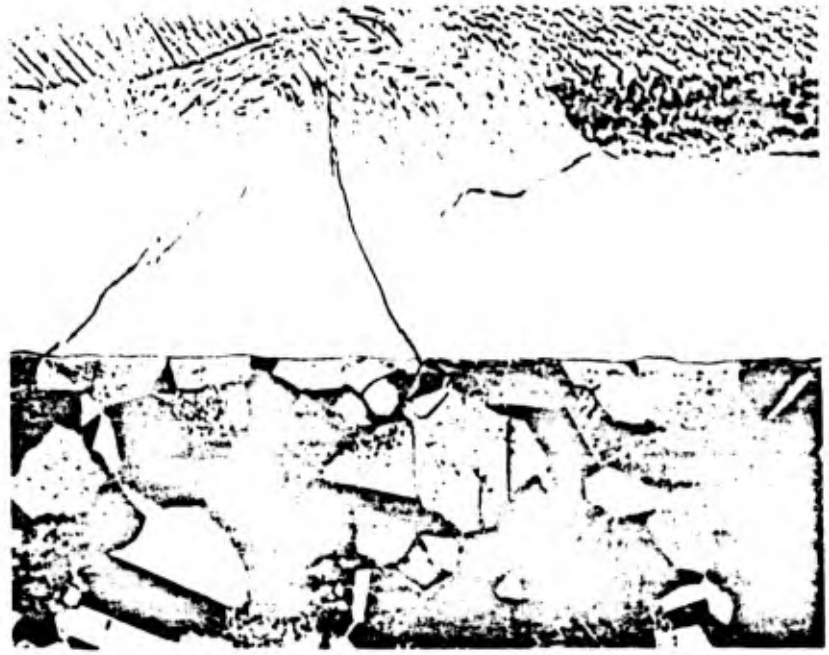
Metallographic evaluations of the compacts in HIP Run No. 3 showed that all exhibited cracking on the cut and polished surface, similar to Figure 15 and 16; and the cracks completely penetrated the aluminide on the observed cross sections and always penetrated into the Ti-6Al-4V cladding beyond the original interface. Additional evaluations were then performed to determine if abrasive or diamond wheel cutting was a cause for the cracking or the crack propagation.

It was hypothesized that the difference in coefficients of expansion of the cladding and compact provided the driving force for the crack initiation and propagation. The net effect of these conditions would be a tensile residual stress in the aluminide, which could be removed by careful grinding of the cladding. Therefore, grinding trials were performed to remove sequentially equal amounts of the cladding and the diffusion zone until only the consolidated aluminide remained. At this point, the aluminide was stress relieved by heating to 730 C for 1 hour, followed by a slow furnace cool (approximately 50 C/hr). This procedure was followed by an additional grinding and polishing of the cut surface, which, in most cases, no longer contained through cracks. This condition is shown in Figure 17 for the TiAl compact shown in Figures 15 and 16 before cladding removal. Surface grinding was then continued with intermediate dye-penetrant inspection to determine if all cracks had been removed. Following this procedure, some compacts from HIP Run No. 3 were salvaged for rolling trials without cladding. However, this salvaging technique was time consuming, expensive, and not always successful. Occasionally, cracks were found either to continue, or to propagate into the TiAl-base alloys with continued grinding. The greater success in salvaging was obtained with a 39C 100H 8VK grinding wheel rather than with a 32A 46I 8VBE wheel.



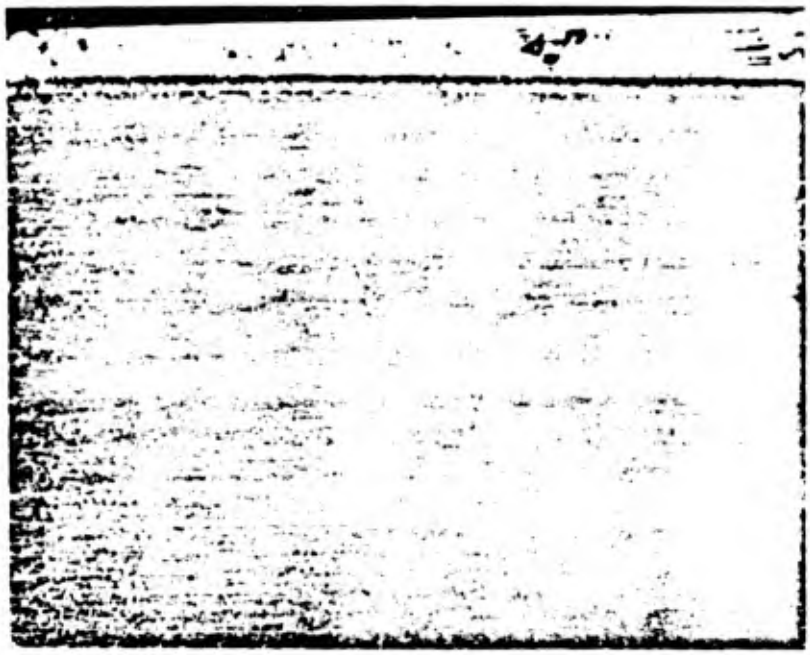
8H842 As Polished 5X

Figure 15. TiAl Compact No. 3530 Showing Profuse Cracking After HIP



8H766 Etchant 30 a/o Lactic, 5v/oHNO_3, 1v/oHF, Bal H_3O 100X

Figure 16. TiAl Compact (No. 3530) with Ti-6Al-4V Cover Plates (Top) Separated by Crack-Sensitive Diffusion Zone



9H840 As-Polished 5X

Figure 17. TiAl Compact (No. 3530) Surface Ground Below Diffusion Zone. Shows Propagation of Cracks from Near Original Cladding-Compact Interface

Based on the preceeding findings, it appeared that cracking originated in the diffusion zone between the aluminide and the Ti-6Al-4V cladding and propagated in both directions. The probable causes were the difference between the thermal expansion of the cladding and aluminide which would cause the aluminide to be stressed in tension on cooling; the greater thickness and strength of the 0.075-inch-thick Ti-6Al-4V cladding at temperatures below about 400 C; and the poor toughness of the aluminides below 500 C. The efficiency of the outgassing procedures was also briefly considered as a potential source of cracking by promoting adhesion between the cladding and the aluminide. However, this possibility did not appear significant because efficient diffusion bonding of the cladding and aluminide was observed in HIP Run Nos. 1 and 2. However, the blisters observed in Run No. 2, when sectioned, always resulted in separation of the can from the aluminide. This separation usually extended over less than 30 percent of the compact surface and the other portion showed a well bonded interface. Based on these deductions, it was concluded that the following conditions might be successfully employed for HIP consolidation in Run No. 4:

(1) Ti-55A 0.51 mm and 2.0 mm thick sheet to provide a cladding of lower strength than the aluminide and the absence of the aluminum in the cladding could promote the ductility of the diffusion zone

(2) Ti-55A cans, 2.0 mm thick, plasma-sprayed on the inside with alumina to provide a diffusion barrier.

These options were selected to determine if the larger thickness cladding of Ti-55A could be used, either with or without a diffusion barrier, and to re-evaluate the feasibility of using the 0.51 mm thick Ti-55A cladding which was apparently successful in HIP Run No. 1. This range of conditions appeared to be minimal to evaluate the deductions reached from the failure analysis of Run No. 3 and to provide some confidence that successful compacts would be produced.

HIP Run No. 4 was performed, as described in Table 3, with 0.51 and 2.0 mm thick Ti-55A cladding with and without plasma-sprayed alumina on the inside. Can Series No. 60 was essentially a duplication of Run No. 1 without exposure of the powder to argon. Can Series Nos. 70 and 80 were used to test the necessity for the aluminum oxide coating. In addition to these special preparations, the HIP cycle was a duplicate of Run Nos. 2 and 3 with the exception that slower under-pressure cool of approximately one-half the previous rate was used in Runs 1 through 3. The slower cool was used to promote thermal equilibration throughout the compacts, which could also minimize cracking.

The HIP compacts from Run No. 4 were evaluated by radiography, ultrasonic inspection (full scan pulse-echo with water coupling), TIP testing, and metallography. These compacts were determined to be completely sound and of high quality. Based on these findings, HIP Run No. 5 was established as the final run with scaled-up compacts, as previously described. A total of 24 compacts were prepared from Ti-25Al-5Nb and TiAl+Nb+W powders for consolidation to 12.7 x 76 x 152 mm and 12.7 x 152 x 230 mm compacts using the canning procedures previously evaluated in Can Series 70 in Run No. 4. The compacts from HIP Run No. 5 were inspected similar to Run No. 4 and were found to be completely successful.

No tensile testing was performed on the TiAl+Nb+W compacts because of the large grains found after the elevated temperature annealing treatments. An example of the large grain size and the compact-cladding interaction is shown in Figure 18 for TiAl+Nb+W Compact No. 5690, after the 1400 C, 3-hour thermal treatment. This observation is in marked contrast to the comparatively small grain size shown in Figure 14 for TiAl+Nb given the same processing history.

The HIP compacts from Run Nos. 1 through 5 were subsequently prepared for rolling. This preparation involved a 1400 C anneal for 3 hours for the TiAl-base alloys and, as noted, isothermal forging, decladding, and special cladding. These special procedures are described with the results on rolling in the following section.

Figure 18. Microstructure of TiAl+Nb+W Compact No. 5690 After Annealing at 1400 C for 3 Hours

Rolling

Heated-roll rolling, with preset roll temperatures in the range of 700 C to 870 C was used to roll the Ti_3Al- and TiAl-base alloys. In addition, some conventional rolling (125 C roll temperature) was performed with the Ti-25Al-5Nb. Most of the rolling was performed on Battelle's experimental two-high, heated rolling mill to produce nominally 4-inch-wide strip. Two campaigns were performed on the large heated-roll, four-high mill. Unless noted in the following discussion, rolling was performed on the small mill.

The initial approach to rolling these new alloys produced from powder or by casting was essentially no different than what would have been employed for more conventional alloys produced by the same techniques. Primary working either at low temperatures in the β range, to breakdown the initial structure and achieve a finer-grained recrystallized structure, was initially evaluated with the Ti_3Al-base alloys at approximately 1200 C. In an attempt to achieve an initial finer grain size, the preheat temperature was subsequently lowered to 1175 C, thought to be high in the α+β range. Pending the success of these trials, lower rolling temperatures and thermal treatments would be employed to achieve a fine-grain, hot-worked, wrought or recrystallized structure which might possess improved room temperature ductility and/or desirable superplastic behavior at temperatures below 1000 C. Similar considerations were the bases for establishing the initial rolling conditions for the TiAl-base alloys; however, the range of stability and effects of the α_2 transformation in these predominantly γ alloys were unknown.

The low strain-rate tensile-testing results presented in Table 4 indicated that the HIP structure was preferred for rolling the Ti-25Al-5Nb alloy, although the 900 C ductility appeared marginal, and the 1400 C for 3 hour anneal was definitely beneficial for TiAl+Nb. Based on these results, it appeared that TiAl+Nb should be more successful in rolling. The large β grain size observed with Ti-25Al-5Nb after HIP and the thermal treatments indicated that grain

boundary mobility was rapid at temperatures above at least 1200 C and that the large grain size may have been a significant factor resulting in the marginal ductility. These preliminary observations indicated that rolling below 1200 C to achieve a finer grain structure probably was feasible and desirable.

The most notable and encouraging feature of the TiAl+Nb+W and Ti-25Al-5Nb microstructures (see, for example, Figures 9 through 11) was the average prior β grain size which was in the range of 2 to 20 times the average powder particle size. This observation showed that the prior particle boundaries had been eliminated. Contrary to this observation, the range of average grain diameters for TiAl+Nb after HIP and the annealing treatments was well within the original powder particle size range (nominally 60 to 500μ) even after thermal treatments very near the melting temperature. These observations may be significant in understanding the rolling behavior of the two alloys and the special procedures used in preparation for rolling.

Because few problems were encountered in rolling the Ti_3Al-base alloys, when compared with TiAl-base alloys, the alloys will be discussed separately in the following.

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Rolling of Ti_3Al-Base Alloys

No special post-HIP processing was employed with the Ti_3Al-base alloys, although some of the cast alloys were isothermally forged to provide a suitable geometry for the sheet billet. Using the identification established in Tables 2 and 3, Table 5 summarizes the rolling results for the Ti_3Al-base alloys. Unless specified otherwise, the sheet billet was heated for 15 minutes at the preheat temperature at the initiation of rolling and subsequently for 5 minutes before each pass.

The data in Table 5 are presented in the sequence of HIP Billet (Compact) Number for Ti-25Al-5Nb and in the chronological sequence of receipt from Pratt and Whitney of cast and forged sheet billets of Ti-24Al-11Nb. After initial breakdown, the partially rolled sheet was cut into approximately equal lengths to facilitate handling and to fit the furnace for subsequent processing. These cut lengths are indicated by the number or letter code following the original Compact or Billet Number, thereby permitting traceability to its source. Figure 19 is a photograph of the sheet before cladding removal. This figure shows the typical crazing of the cladding after rolling. This crazing was caused by the combined effects of oxidation and rolling.

A summary review of the results demonstrates the following:

(1) Rolling of HIP Ti-25Al-5Nb

   (a) Initial passes on the clad and bare HIP compacts were successfully performed in the temperature range of 1096 to 1200 C, at reductions per pass of 15 to 25 percent. Although lower initial pass temperatures were evaluated, these trials (see Compact Nos. 5894 through 5896 through 5899) were performed on compacts inadvertently

TABLE 5. ROLLING PARAMETERS AND RESULTS FOR Ti_3Al-BASE ALLOY, Ti-25Al-5Nb

Billet/Sheet No.	Preheat Temp. (C)	Roll Temp. (C)	Thickness (inch) Initial	Thickness (inch) Final	No. of Passes	Total Reduction (%)	Roll Speed (ft/min)	Surface Dimensions (inch)	Remarks
1812(4A)	1200	815	.54 - .73C	0.22C	3	63.3	68	3.5 x 14.5	
-1	1200	815	0.22C	0.07C	3	68.2	68	3.7 x 18.6	
-2	1100	815	0.22C	0.125C	1	43.2	68	3.65 x 10.6	
-3	980	815	0.22C	0.130C	1	41.0	68	3.65 x 10.2	
1813(4B)	1200	815	.54 - .73C	0.212C	4	64.7	68	3.5 x 15.0	
-1	1100	815	0.212C	0.165C	1	22.2	68	3.6 x 8.1	
	1200	815	0.165C	0.105C	1	36.4	68	3.65 x 12.6	
-2	1100	815	0.212C	0.140C	1	34.0	68	3.6 x 9.5	
	1200	815	0.140C	0.106C	1	24.3	68	3.65 x 12.5	
-3	1100	815	0.212C	0.105C	2	50.5	68	3.65 x 12.6	
2820(4C)	1175	815	0.62C	0.21C	4	58.8	63	3.5 x 15.0	
-1	1175	815	0.21C	0.113C	3	46.2	63	--	
	1120	815	0.113C	0.103C	2	8.8	63	--	
-2	1175	815	0.21C	0.113C	3	46.2	63	--	
	1120	815	0.113C	0.093C	1	17.7	63	--	
2821(4D)	1175	815	0.62C	0.21C	4	58.8	63	3.5 x 15.0	
-1	1175	815	0.21C	0.148C	2	34.3	63	--	
	1120	815	0.138C	0.104C	2	24.6	63	3.65 x 14.7	
-2	1175	815	0.21C	0.120C	2	42.9	63	3.65 x 12.7	

TABLE 5. (Continued)

Billet/Sheet No.	Preheat Temp. (C)	Roll Temp. (C)	Thickness (inch) Initial	Thickness (inch) Final	No. of Passes	Total Reduction (%)	Roll Speed (ft/min)	Surface Dimensions (inch)	Remarks
2822(4E)	1175	815	0.62C	0.21C	4	58.8	63	3.5 x 15.0	
-1	1175	815	0.21C	0.150C	1	28.6	63	3.6 x 10.3	
-2	1175	815	0.21C	0.145C	1	31.0	63	3.6 x 10.5	
2823(4F)	1175	815	0.62C	0.21C	4	58.8	63	3.5 x 15.0	
-1	1175	815	0.21C	0.120C	2	42.9	63		
	1120	815	0.120C	0.100C	2	16.7	63	3.65 x 15.3	
-2	1175	815	0.21C	0.150C	1	28.6	63	3.6 x 10.3	
2824(4G)	1175	815	0.62C	0.21C	4	58.8	63	3.5 x 15.0	
-1	1120	815	0.21C	0.140C	2	33.3	63	3.6 x 11.1	
-2	1120	815	0.21C	0.153C	2	27.1	63	3.6 x 10.1	
2825(4H)	1175	815	0.62C	0.21C	4	58.8	63	3.5 x 15.0	
-1	1120	815	0.21C	0.144C	2	31.4	63	3.6 x 10.6	
-2	1120	815	0.21C	0.146C	2	30.5	63	3.6 x 10.2	
-21	980	870	0.146C	0.120C	2	17.8	63	3.5 x 4.3	
-22	925	870	0.146C	0.117C	2	19.9	63	3.5 x 4.3	
-23	870	870	0.146C	0.122C	3	16.4	63	3.5 x 4.25	

TABLE 5. (Continued)

Billet/Sheet No.	Preheat Temp. (C)	Roll Temp. (C)	Thickness (inch) Initial	Thickness (inch) Final	No. of Passes	Total Reduction (%)	Roll Speed (ft/min)	Surface Dimensions (inch)	Remarks
As-cast 1[a,b]	1135	815	0.313	0.055	8	82.4	67	4.0 x 14.0	Minor edge cracks
As-cast 2[a,b]	1135	815	0.345	0.062	8	82.0	67	3.7 x 6.7	Minor edge cracks
As-cast 3[b]	1135	815	0.309	0.085	6	72.5	67	3.1 x 14.0	Minor edge cracks at ends attributed to handling
As-cast 4[b]	1135	815	0.360	0.065	8	81.9	67	3.0 x 20.0	Minor edge cracks
3830	1135	815	0.450	0.150		66.7	67	3.6 x 28.5	Ground on all surfaces to remove cracks & successfully rolled bare
3840A	1135	815	0.398	0.090		77.4	67	2.2 x 9.5	
3840B	1135	815	0.398	0.070		82.4	67	2.5 x 10.5	
4860	1135	815	0.460	0.150		67.4	67	3.5 x 24.5	Cladding removed by grinding before rolling
4880	1175	815	0.680C	0.210C	5	69.1	67	3.0 x 19.0	
-1	870	815	0.210C	0.145C	3	31.0	67	3.0 x 14.0	Solution treated at 1200C for 1 hr surface cracks present
-2	870	815	0.210C	0.145C	3	31.0	67	3.0 x 14.0	

(a) Irregularly shaped sections of castings.

(b) Surface cracks observed on all sections of castings from TMCA HV-5123 before rolling. Provided by Pratt & Whitney to obtain sheet for welding trials.

TABLE 5. (Continued)

Billet/Sheet No.	Preheat Temp. (C)	Roll Temp. (C)	Thickness (inch) Initial	Thickness (inch) Final	No. of Passes	Total Reduction (%)	Roll Speed (ft/min)	Surface Dimensions (inch)	Remarks
Air Force Sheet									
-1	1000	n.r.	0.060-0.080	n.r.	n.r.	n.r.	n.r.	n.r.	
-2	1000	n.r.	0.060-0.080	n.r.	n.r.	n.r.	n.r.	n.r.	
-3	1000	n.r.	0.060-0.080	n.r.	n.r.	n.r.	n.r.	n.r.	Details recorded by Dr. H. Mendiratta, AFML
-4	1000	n.r.	0.060-0.080	n.r.	n.r.	n.r.	n.r.	n.r.	
-5	1000	n.r.	0.060-0.080	n.r.	n.r.	n.r.	n.r.	n.r.	
-6	1000	n.r.	0.060-0.080	n.r.	n.r.	n.r.	n.r.	n.r.	
5890	1125	815	0.720C	0.110C	6	84.7	67	3.5 x 21.0	
-1	1125	815	0.110C	0.095C	3	13.6	67	3.5 x 8.2	
-2	1095	815	0.110C	0.095C	3	13.6	67	3.5 x 16.1	
5891	1175	815	0.702C	0.260C	5	63.0	50	3.6 x 15.6	
-1	1175	815	0.260C	0.160C	3	38.5	50	3.8 x 14.3	
-11	1175	120	0.160C	0.070C	3	56.0	100	4.3 x 8.5	Cross rolled conventionally; one edge crack
-2	1175	815	0.260C	0.155C	3	40.4	50	3.8 x 14.2	
-21	1125	815	0.155C	0.090C	4	42.0	73	3.8 x 14.4	
-22	1125	815	0.155C	0.087C	4	43.9	105	3.8 x 14.8	

TABLE 5. (Continued)

Billet/Sheet No.	Preheat Temp. (C)	Roll Temp. (C)	Thickness (inch) Initial	Thickness (inch) Final	No. of Passes	Total Reduction (%)	Roll Speed (ft/min)	Surface Dimensions (inch)	Remarks
5892	1175	815	0.680C	0.300C	3	55.9	50	3.5 x 14.0	
-1	1093	815	0.300C	0.096C	8	68.0	50	3.8 x 15.5	
-2	1093	815	0.300C	0.098C	8	67.3	50	3.8 x 22.5	Minor surface cracks
5893	1175	815	0.680C	0.290C	3	57.4	50	3.5 x 14.3	
-1	1066	815	0.290C	0.090C	6	69.0	63;105	3.8 x 23.5	Roll speed was increased after the 3rd pass
-2	1066	815	0.290C	0.090C	6	69.0	73	3.8 x 27.0	
5894(c)	1125	720	0.750C	0.630C	1	16.0	67	6.8 x 11.0	One large crack
-1	1125	720	0.630C	0.266C	5	57.8	67	6.8 x 6.3	Section of 5894 cross rolled
-2	1125	720	0.630C	0.269C	4	57.3	67	6.8 x 7.0	Section of 5894 cross rolled
5895	1125	720	0.730C	0.132C	5	81.9	67	3.5 x 34.4	
-1	1070	720	0.132C	0.091C	1	31.1	67	3.5 x 19.2	
-2	1070	720	0.132C	0.099C	2	25.0	67	4.6 x 6.6	Cracked about tong-hold area
-3	1070	720	0.132C	0.103C	1	22.0	67	4.6 x 8.0	
-4	1070	720	0.132C	0.094C	1	28.8	67	3.5 x 9.0	

(c) Indicated compacts were heated to 1357 C inadvertently and furnace cooled to room temperature.

TABLE 5. (Continued)

Billet/Sheet No.	Preheat Temp. (C)	Roll Temp. (C)	Thickness (inch) Initial	Thickness (inch) Final	No. of Passes	Total Reduction (%)	Roll Speed (ft/min)	Surface Dimensions (inch)	Remarks
5896[c]	1095	720	0.713C	0.600C	1	15.8	67	3.5 x 7.6	
	1040	720	0.600C	0.400C	1	33.3	67	3.6 x 11.2	Severely cracked
5897[c]	1125	720	0.720C	0.506C	2	29.7	67	3.5 x 9.1	Severely cracked
5898[c]	1125	720	0.720C	0.430C	2	40.3	67	3.5 x 7.3	Severe local cracking removed
	1125	720	0.430C	0.230C	2	46.5	67	3.6 x 9.4	Severe local cracking removed
	1100	720	0.230C	0.070C	2	69.6	67	3.6 x 11.0	Severely cracked
5899[c]	1125	720	0.720C	0.521C	3	27.6	67	3.5 x 6.2	Severe local cracking removed
	1125	720	0.521C	0.186C	4	64.3	67	3.6 x 16.8	

TABLE 5. ROLLING PARAMETERS AND RESULTS FOR Ti_3Al-BASE ALLOY, Ti-24Al-11Nb

Billet/Sheet No.	Preheat Temp. (C)	Roll Temp. (C)	Thickness (inch) Initial	Thickness (inch) Final	No. of Passes	Total Reduction (%)	Roll Speed (ft/min)	Surface Dimensions (inch)	Remarks
24-11-A	1175	815	0.500	0.090	9	82	50	16.0 x 2.3	Isothermally upset forged 25% at 1205 C
24-11-B	1175	815	0.500	0.084	11	83	50	11.0 x 3.6	Isotheramlly upset forged at 1205 C
24-11-3[d]	1125	720	0.345	0.098	4	72	67	31.5 x 4.6	Cut into 5 pieces 3A, 3B, 3C, 3D, and 3E
24-11-3A	1092/1025[e]	720	0.098	0.062	2	37	67	9.2 x 4.7	
24-11-3B	1038/1000[e]	720	0.098	0.062	2	37	67	8.4 x 4.6	
24-11-3C	1055/1080[e]	720	0.098	0.062	2	37	67	6.2 x 4.6	Large longitudinal crack on the second pass
24-11-3D	1052/1075[e]	720	0.098	0.062	2	37	67	8.2 x 4.6	
24-11-3E	1075	720	0.098	0.054	3	45	67	8.5 x 4.7	
24-11-4A[d]	1150	815	0.516	0.257	4	50	59/80	8.5 x 4.5	Roll speed was increased after second pass
	1125	815	0.257	0.124	4	52	80	16.3 x 4.5	
	1066	815	0.124	0.055	13	56	80	40 x 4.5	Cut into halves 4A1 and 4A2

(d) Received from Pratt & Whitney Florida.

(e) Preheat temperature of first pass/second pass.

TABLE 5. (Continued)

Billet/Sheet No.	Preheat Temp. (C)	Roll Temp. (C)	Thickness (inch) Initial	Thickness (inch) Final	No. of Passes	Total Reduction (%)	Roll Speed (ft/min)	Surface Dimensions (inch)	Remarks
24-11-4B[d]	1093	815	0.522	0.285	4	45	59/80	7.0 x 4.5	Roll speed was increased after second pass
	1066	815	0.285	0.063	13	78	80	27.0 x 4.5	
24-11-5A[d]	1125	815	0.528	0.125	6	76	80	18.5 x 4.5	
	1066	815	0.125	0.055	7	56	80	40.0 x 4.5	Cut into two halves, 5A1 and 5A2
24-11-5B[d]	1066	815	0.542	0.280	4	48	59/80	8.5 x 4.5	Roll speed was increased after second pass
	1010	815	0.280	0.045	16	79	80	34.0 x 4.5	Cut into two halves, 5B1 and 5B2

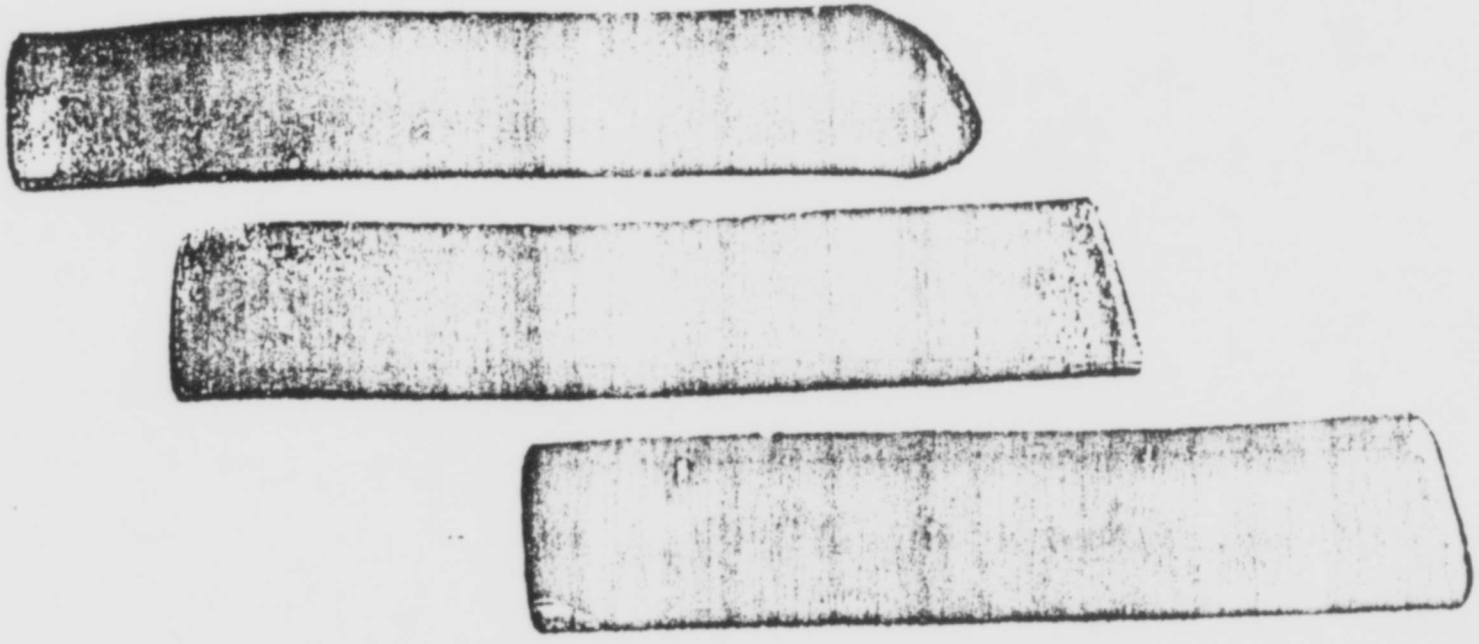

Figure 19. As-Rolled Ti-25Al-5Nb Sheet from Compact No. 1813 (-1,-2,-3) Before Removal of Ti Cladding

heated to 1360 C during the initiation of the annealing cycle for a lot of TiAl compacts.

(b) Finishing was successfully performed at temperatures in the range of 870 to 1200 C at reductions of 15 to 43 percent on both bare and clad sheet. However, failure (similar to that experienced with TiAl+Nb) occurred at the lower finishing temperatures for samples previously annealed at 1200 C.

(c) Roll speeds in the range of 50 to 105 SFM were employed with the same apparent success.

(d) Roll temperatures in the range of 800 to 870 C did not appear to affect rolling performance, although limited rolling with rolls at 125 C did promote edge cracking and high forces.

(e) The surface temperature of the sheet immediately after rolling dropped to between 150 and 290 C for a nominal roll temperature of 815 C and between 360 and 400 C for a nominal roll temperature of 125 C.

(2) Rolling of as-cast and as-cast and isothermally forged Ti-24Al-11Nb

(a) The results for the Ti-24Al-11Nb alloy essentially duplicated the results with Ti-25Al-5Nb, although the Ti-24Al-5Nb alloy appeared to undergo larger reductions at lower temperatures more successfully.

(b) The single crack appearing in Sheet No. 24-11-3B appears random and may have been associated with a material defect.

The force measurements obtained from the later rolling trials for all alloys are presented in the Appendix. The load recults from earlier trials are not included because their accuracy became questionalbe after finding that the adhesive had failed on some of the strain gages

within the load cells on the experimental two-high mill. After this finding, the load cells were repaired with new gages and water cooling was adapted to both mills to maintain the load cell temperatures in the range of 20 to 30 C.

These measurements are summarized in Figures 20 through 24 as the calculated flow stress[16,17] as a function of the process parameters: temperature, reduction, and the average effective strain rate[16,17] derived from the reduction and the roll speed. The calculated quantities, flow stress ($\bar{\varepsilon}$) and strain rate ($\dot{\bar{\varepsilon}}$), were used in place of separating force and roll speed to eliminate the complicating effects of different sheet dimensions, roll speeds, and reductions from the discussions. The majority of these calculated results are for the Ti-24Al-11Nb alloy which was always rolled without cladding. Accurate rolling measurements were obtained with the Ti-25Al-5Nb alloy compacts from HIP Run No. 5, but the cladding caused some uncertainty in determining the sheet width and interpreting the separating force due to the combined effects on this force of the aluminide and cladding.

Figure 20 shows the flow stress dependence on temperature at three rolling speeds and strain rates for compacts from Run No. 5, rolled with a roll temperature of 815 C. The flow stress at the lowest strain rate (4 sec^{-1}) decreased with increasing temperature, as anticipated. However, the results at the higher strain rates show a positive dependence on temperature, particularly at a roll speed of 77 feet per minute. These results are similar to the findings for Ti-24Al-11Nb shown in Figure 21. Factors which contributed to the spread of the data, in addition to experimental error, were as follows:

(1) Roll speed (see Figure 20)
(2) Prior reduction-temperature history (see Figures 22 and 23)
(3) Reduction per pass (see Figure 24)
(4) Sheet thickness (see Figure 25).

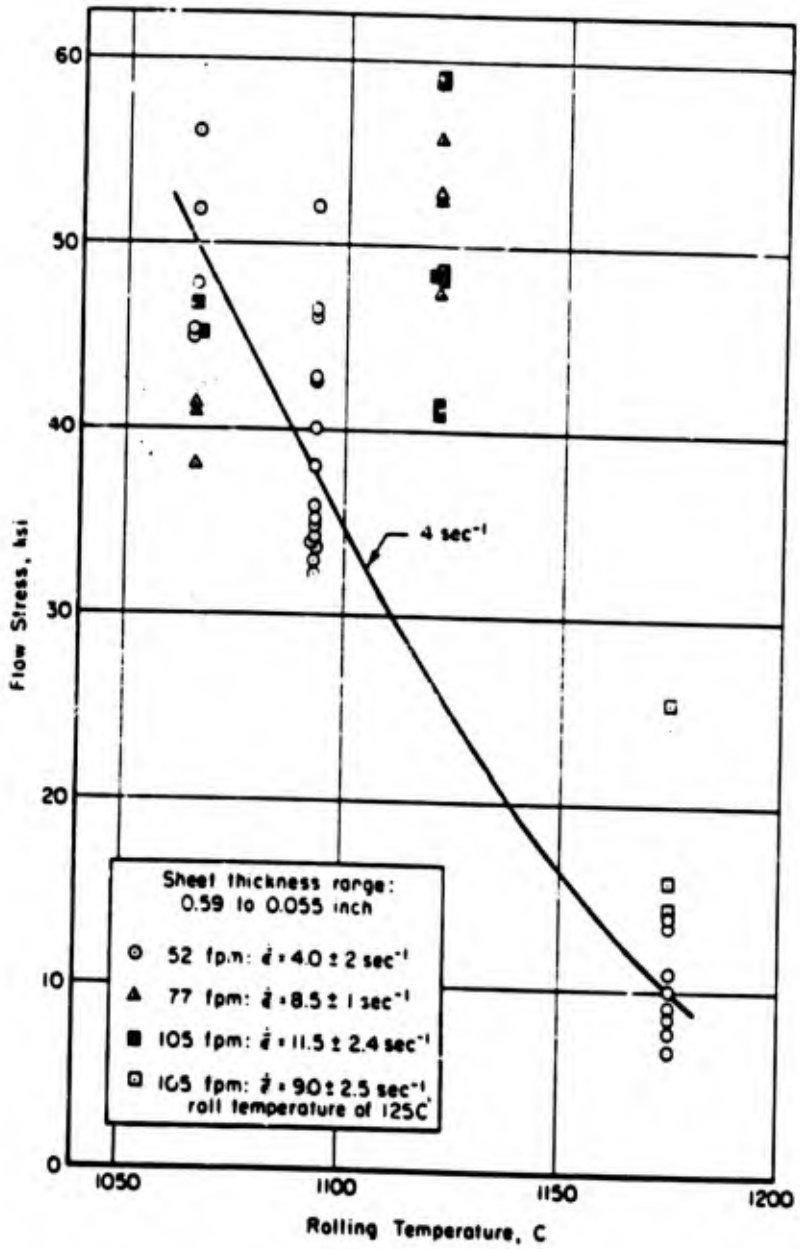

Figure 20. Flow Stress Dependence on Temperature for Ti-25Al-5Nb

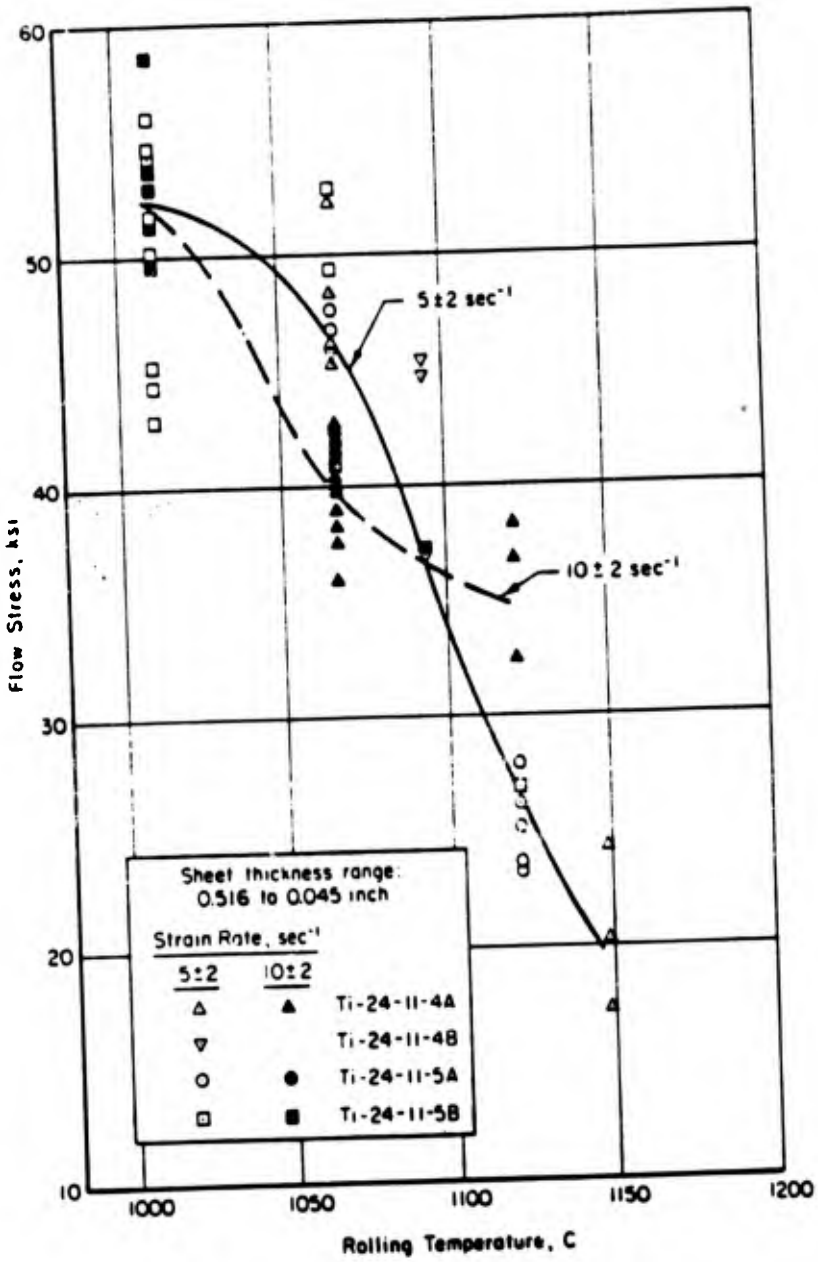

Figure 21. Flow Stress Dependence on Temperature for Ti-24Al-11Nb

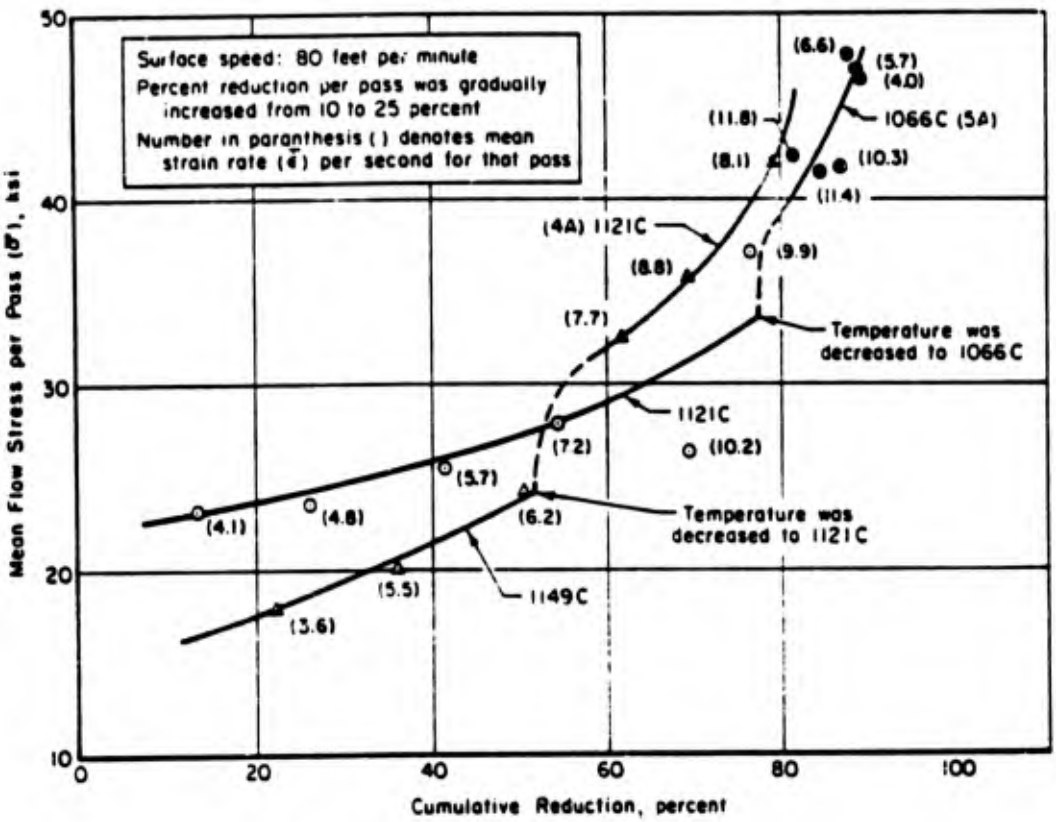

Figure 22. Effect of Cumulative Reduction and Temperature on Flow Stress During Initial Passes

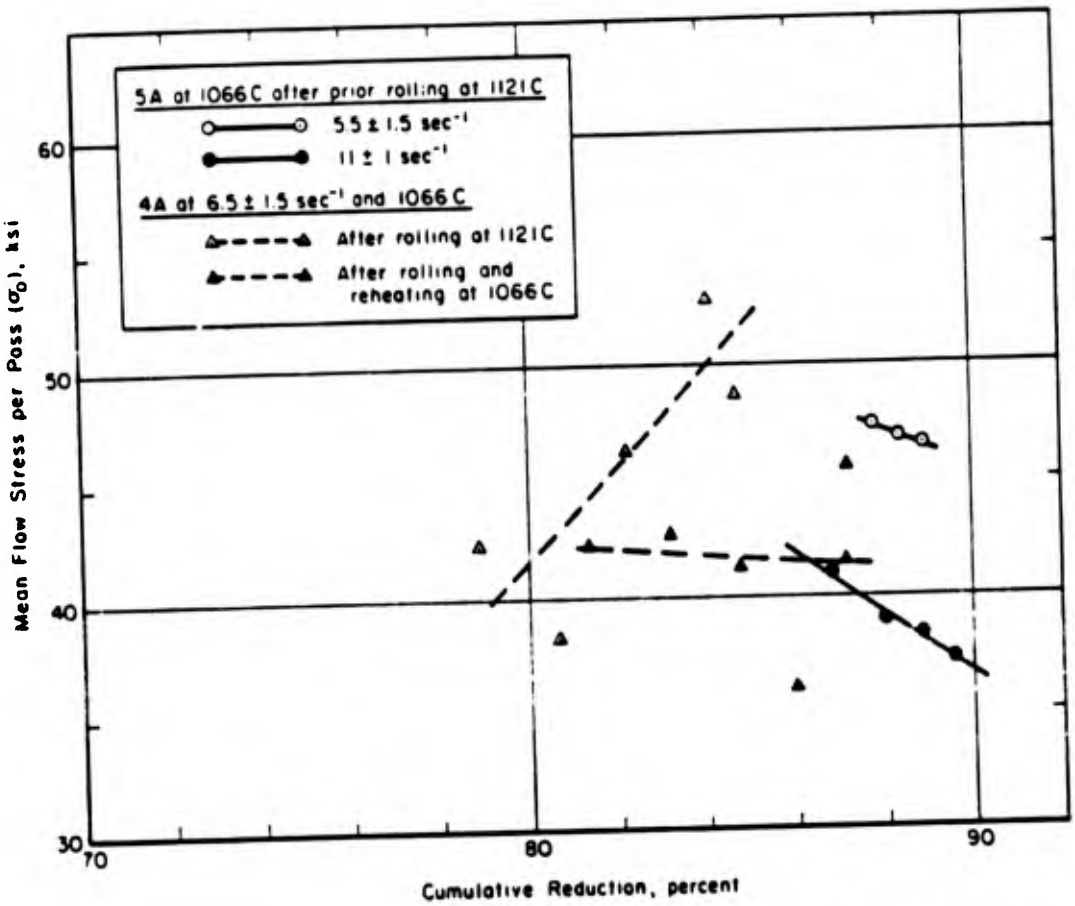

Figure 23. Effect of Cumulative Reduction and Strain Rate on Flow Stress During Finishing Passes

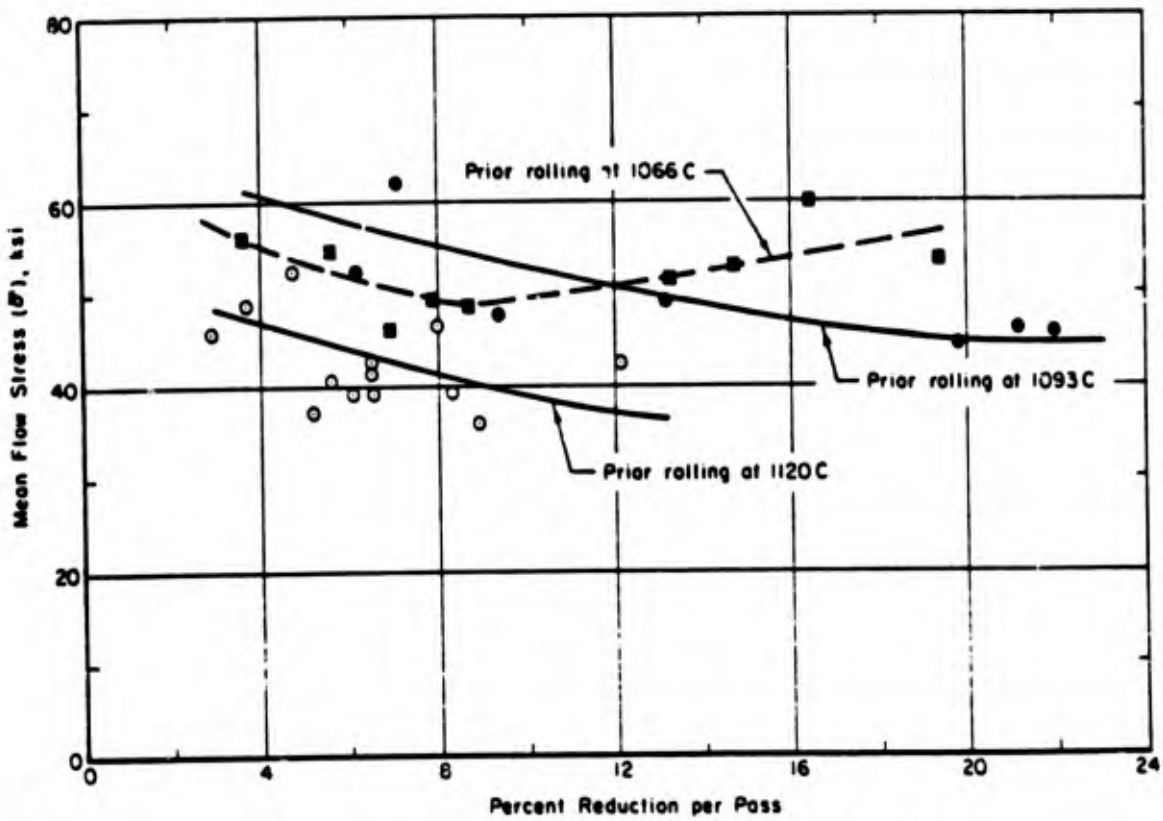

Figure 24. Flow Stress Dependence on Reduction per Pass

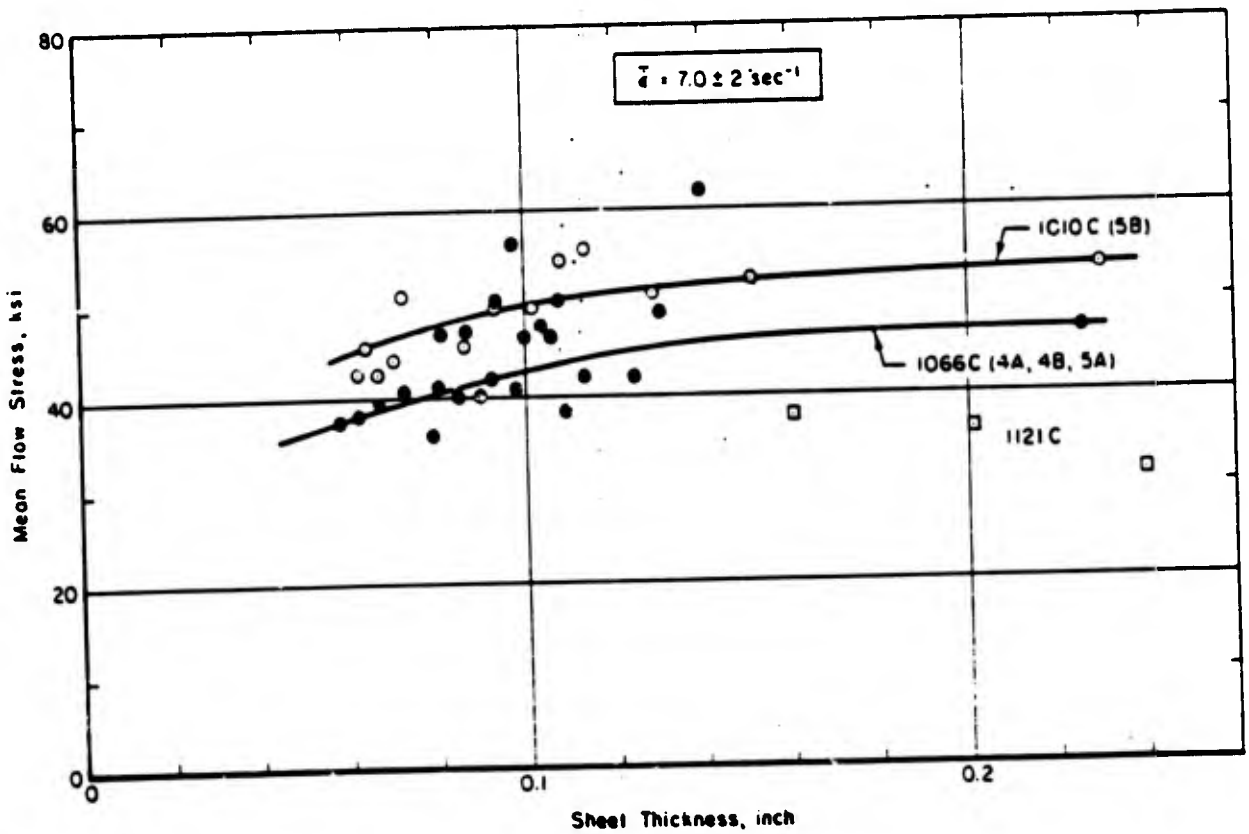

Figure 25. Effect of Sheet Thickness on Flow Stress

Both alloys exhibit lower flow stresses at higher strain rates in the temperature range 1000 to 1150 C. Figures 22 and 23 demonstrate the effects of a change in the rolling temperature on flow stress at different strain rates during the deformation history of a compact. The exact cause for the observed behavior can only be speculated at this time, but may be associated with the predominantly dynamic recrystallization at temperatures below about 1100 C, significant recrystallization during heating at higher temperatures and the formation of β in the Ti-25Al-5Nb alloy above about 1125 C and in the Ti-24Al-11Nb alloy above about 1070 C.

The dependence of flow stress on reduction per pass in Figure 24 shows a relatively small effect of reduction, but a change in deformation behavior, over the range evaluated, below 1166 C.

The effect of heat transfer on the rolling of aluminide sheet was evaluated by rolling at various roll speeds (Figure 20); evaluating the flow stress dependence on sheet thickness at one nominal strain rate (Figure 25), and by observing the separation forces at the initiation, middle, and final positions of several rolling passes as documented in the Appendix. The effect of roll speed in Figure 20 does not appear sufficient to reach any obvious conclusions. The flow stress dependence on thickness in Figure 25 shows a decreased flow stress with decreasing thickness. This observation probably results from an approach to a fine steady-state grain size with continued rolling, but does not indicate a significant effect of chilling. Furthermore, the results in the Appendix show a near-constant separating force, which shows that chilling was relatively insignificant in affecting the flow stress. However, the majority of the results in the Appendix were obtained after special handling precautions were instituted to avoid chilling from handling of the sheet with tongs.

Microstructures of the Ti-25Al-5Nb alloy after rolling in the range of 1066 to 1204 C are shown in Figures 26 through 29. Figures 26, 27, 28, and 29, obtained from sheet after rolling at 1066, 1093, and 1121 C, respectively, exhibit similar microstructures with the exception that the grains coarsen as the rolling temperature is increased. Figure 29, a bright field micrograph,

Figure 26. Longitudinal Section of Sheet 5895B After Rolling at 1066 C. Polarized Light Illumination at 250 X

Figure 27. Longitudinal Section of Sheet 5890B1R After Rolling at 1093 C. Polarized Illumination at 250X. Average Grain Size = 11 Microns

Figure 28. Longitudinal Section of Sheet 5895 After Rolling at 1121 C. Polarized Light Illumination at 250 X

Figure 29. Longitudinal Section of Sheet 5895 After Rolling at 1121 C. Bright Field Illumination at 250 X

is shown to demonstrate the structure from this type of illumination, which did not provide as high grain boundary constrast as polarized light. All of these structures show a hot-worked α_2 microstructure , with little or no discernable β . Recrystalilization had been observed in both Ti_3Al-base alloys after rolling at temperatures as low as 985 C. However, thermal treatments of 2 hours at 985 C and 8 hours at 870 C produced a more equiaxed structure with little or no recrystallization in sheet previously rolled at temperatures 200 C higher than the annealing temperatures. Typical microstructures obtained from the material after rolling at 1093 C (see Figure 27) and annealed at 870 C for 8 hours and 1093 C for 1 hour are shown in Figures 30 and 31. These observations confirm the findings for both Ti_3Al-base alloys that from room temperature up to the α/β transus, the rolling conditions determine grain size and that thermal treatments at temperatures equal to or below the rolling temperature promote a relatively small amount of grain growth to achieve a more equiaxed grain shape. The grain size of the as-rolled Ti-25Al-5Nb material is irregular with average values of 8 to 14 microns for rolling temperatures in the range of 1060 to 1120 C. At the higher temperatures in this range a duplex microstructure appears typical with the larger grains being elongated and 3 to 5 times the average.

A wide range in the general scale of the microstructures can be observed in sheet rolled near the α/β transus. This range is demonstrated in Figures 32 and 33 with sheet rolled at 1175 C. Sheet 5891A shown in Figure 32 was rolled at 1175 C and air cooled. This sheet was sectioned to provide 5891A1 which was finished conventionally (roll temperature of 125 C) in three passes to provide the structure in Figure 33. The microstructures are similar except for the scale, since both structures were produced by working in the high $\alpha+\beta$ range. However, cooling, obviously, was more rapid to achieve the finer α_2 structure in Figure 33. The practice used to achieve the fine grain size observed in Figure 33 might be difficult to reproduce with the same results (e.g., without observing Figure 32). Therefore, the feasibility of achieving the same result more consistently was evaluated with the following practice:

(1) Heat previously rolled sheet into the low temperature end of the β range (1225 C) for a short period of time (15 minutes);

Figure 30. Longitudinal Section of Sheet 5890B2 After Rolling at 1093 C and Annealing for 8 hours at 870 C. Polarized Light Illumination at 250X. Average Grain Size = 12 Microns

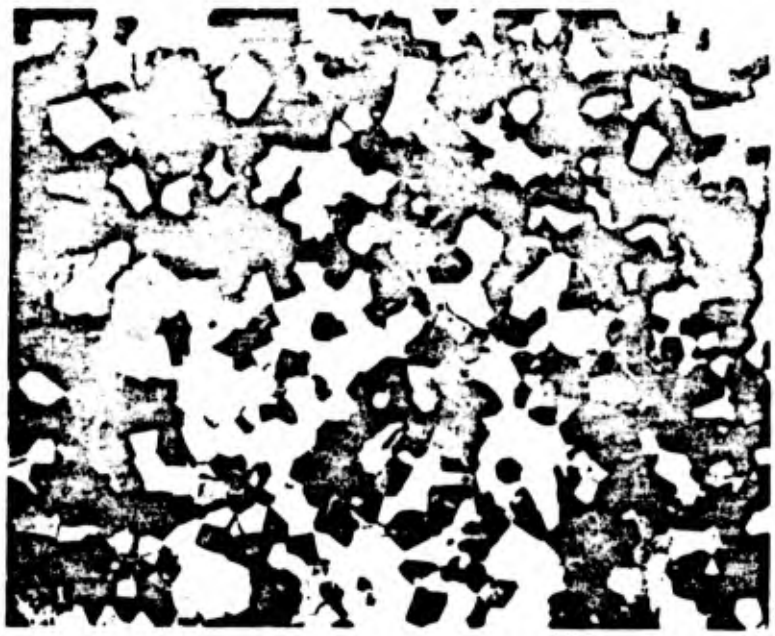

Figure 31. Longitudinal Section of Sheet 5892A After Rolling at 1093 C and Annealing for 1 Hour at 1093 C Plus 8 Hours at 870 C. Polarized Light Illumination at 250X. Average Grain Size = 13.5 Microns

Figure 32. Longitudinal Section of Sheet 5891A As Rolled at 1175 C. Polarized Light Illumination at 250X

Figure 33. Longitudinal Section of Sheet 5891A1 As Rolled at 1175 C Conventionally. Polarized Light Illumination at 250X

(2) Oil or water quench to achieve a very fine α_2 microstructure;

(3) Roll at a low temperature (980 C) to achieve more equiaxed grains either by rolling or by subsequent thermal treatment.

The result of this trial with sheet 589|AHIR is shown in Figure 34. The matrix in this microstructure is very fine-grain α_2, but the 20 percent reduction at 980 C caused recrystallization about the prior β grain boundaries. This approach was not pursued further to achieve a fine grain size because the solution treatment appeared to promote cracking during subsequent rolling.

Rolling at higher temperatures resulted in a fully α_2 structure transformed from β of large grain size. Microstructures from sheet after rolling at 1200 C are shown in Figure 35 at two magnifications. These microstructures show a slow-cooled α_2 microstructure exhibiting no deformation.

The microstructures of the Ti-24Al-11Nb alloy were similar to the preceeding; however, some differences were noted between the various alloys of this general designation. The metallographic evaluations of the Ti-24Al-11Nb alloy were performed with sheet billets 24-11-A and 24-11-3 (0.25% Si). As shown in Table 2, both alloys depart from the Ti-24Al-11Nb composition. The results with Ti-24Al-11Nb were nearly a duplicate of the general metallography of the Ti-25Al-5Nb alloy with the exception that Ti-24Al-11Nb generally showed more primary α_2 and, in general, a finer α_2 transformation product: and fine β grains about the primary α_2.

Figure 36 is the microstructure of 24-11-A after rolling at 1175 C. This microstructure apparently was produced by rolling at a low temperature in the α+β range because it consistsof a low volume fraction of α_2 needles and worked α_2 grains. Rolling this sheet at 870 C did not appear to change significantly the irregular microstructure as shown in Figure 37. The average grain size in both photomicrographs is less than 11 microns. However, it was found that a very fine grain matrix could be obtained after rolling by annealing the structure in Figure 36 at either 870 C for 2 hours or 1190 C for 15

Figure 34. Longitudinal Section of Sheet 5891AH1R After Rolling at 1175 C Plus Solution Treatment for 15 Minutes at 1225 C and Oil Quenched and Finish Rolling at 870 C to a 20% Reduction Followed by Oil Quenching. Polarized Light Illumination at 250X

(2) Oil or water quench to achieve a very fine α_2 microstructure;
(3) Roll at a low temperature (980 C) to achieve more equiaxed grains either by rolling or by subsequent thermal treatment.

The result of this trial with sheet 589|AHIR is shown in Figure 34. The matrix in this microstructure is very fine-grain α_2, but the 20 percent reduction at 980 C caused recrystallization about the prior β grain boundaries. This approach was not pursued further to achieve a fine grain size because the solution treatment appeared to promote cracking during subsequent rolling.

Rolling at higher temperatures resulted in a fully α_2 structure transformed from β of large grain size. Microstructures from sheet after rolling at 1200 C are shown in Figure 35 at two magnifications. These microstructures show a slow-cooled α_2 microstructure exhibiting no deformation.

The microstructures of the Ti-24Al-11Nb alloy were similar to the preceeding; however, some differences were noted between the various alloys of this general designation. The metallographic evaluations of the Ti-24Al-11Nb alloy were performed with sheet billets 24-11-A and 24-11-3 (0.25% Si). As shown in Table 2, both alloys depart from the Ti-24Al-11Nb composition. The results with Ti-24Al-11Nb were nearly a duplicate of the general metallography of the Ti-25Al-5Nb alloy with the exception that Ti-24Al-11Nb generally showed more primary α_2 and, in general, a finer α_2 transformation product: and fine β grains about the primary α_2.

Figure 36 is the microstructure of 24-11-A after rolling at 1175 C. This microstructure apparently was produced by rolling at a low temperature in the α+β range because it consists of a low volume fraction of α_2 needles and worked α_2 grains. Rolling this sheet at 870 C did not appear to change significantly the irregular microstructure as shown in Figure 37. The average grain size in both photomicrographs is less than 11 microns. However, it was found that a very fine grain matrix could be obtained after rolling by annealing the structure in Figure 36 at either 870 C for 2 hours or 1190 C for 15

Figure 34. Longitudinal Section of Sheet 5891AH1R After Rolling at 1175 C Plus Solution Treatment for 15 Minutes at 1225 C and Oil Quenched and Finish Rolling at 870 C to a 20% Reduction Followed by Oil Quenching. Polarized Light Illumination at 250X

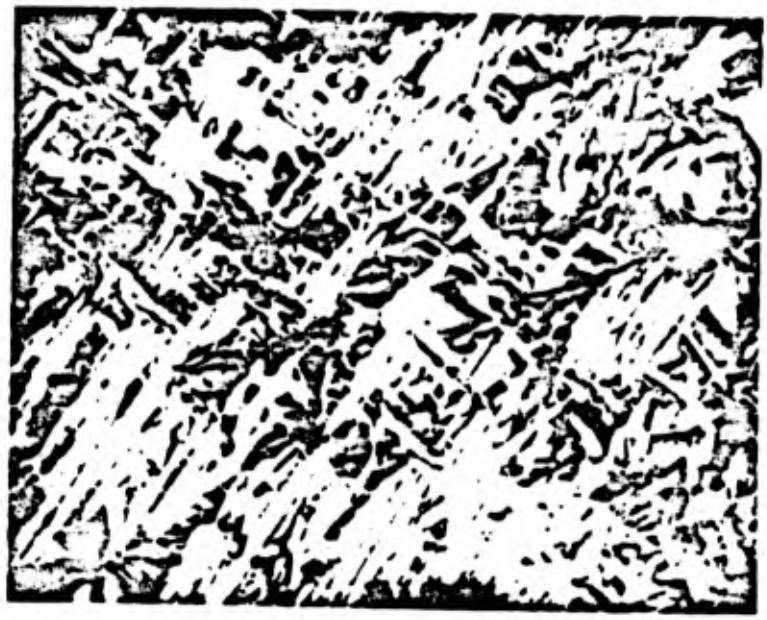

Figure 35. Transverse Section of Sheet Rolled From 5.3 to 2.5 mm at 1200 C and Vermiculite Cooled. Bright Field Illumination at 250X

Figure 36. Ti-24Al-11Nb (24-11-A) Sheet Microstructure After Rolling at 1175 C Followed by Air Cooling. Polarized Light Illumination at 250X

Figure 37. Ti-24Al-11Nb Rolled at 1175 C and Air Cooled, Then Rolled (20%) at 870 C and Oil Quenched. Polarized Light Illumination at 250X

minutes, and probably other conditions, as shown in Figure 38. This microstructure, obtained after annealing and rolling at 870 C, is also similar to the structures obtained after a 2 hour, 870 C anneal. A major feature of the 1190 C anneal was the development of an equiaxed dispersion of primary α_2 as shown in Figure 39. This structure exhibits a fine prior β grain size decorated by the primary α_2 which appears at random elsewhere in the matrix. This microstructure was used to achieve a fine-grain structure after rolling with results similar to Figure 38.

The sheet produced from 24-11-3 (e.g., ingot with 0.25 percent Si) contained a large volume fraction of primary α_2 after rolling first at 1125 C and finishing at 1080 C. The resulting microstructures are shown in Figures 40 and 41. Figure 40 shows a duplex structure with the larger grains (primary α_2) averaging less than 5 microns and the smaller grains appearing barely resolvable at 250X. Further evaluations at higher magnifications (see Figure 41) were not conclusive, but indicated that the grain size of the matrix is larger than anticipated from Figure 40; some of the primary α_2 had recrystallized; and that the very fine grains in the matrix are probably β.

The sheet product, always in the as-rolled condition, was evaluated by formability testing.[12] These results showed that 24-11-3 exhibited m values of 1 and elongations in excess of 450 percent at strain rates of 10^{-3} sec^{-1}. All other material provided inferior performance[12]; however, Ti-25Al-5Nb from HIP Runs 4 and 5, which had finer and more equiaxed grains than sheet from previous runs, has not yet been tested. The successful results with 24-11-3 appear attributable to its unique microstructure, consisting of fine grains which may be associated with the silicon or rolling on the four-high mill. Metallography on 24-11-4 and -5 was not performed; however, this sheet was given heavy reductions at the lower rolling temperatures and, based on the previous processing, this processing appeared to be an improvement over previous practices.

Rolling of TiAl-Base Alloys

The preparations and rolling histories for the TiAl-base alloys are summarized in Table 6 and in Appendix A. The rolling results with this base

Figure 38. Longitudinal Section of Ti-24Al-11Nb After Rolling at 1175 C and Air Cooling; Annealing for 2 Hours at 870 C; and Rolling at 870 C to a Reduction of 20 Percent. Polarized Light Illumination at 250X

Figure 39. Longitudinal Section of Ti-24Al-11Nb Sheet (24-11-AH2) After Annealing for 15 Minutes at 1190 C Followed by Oil Quenching. Bright Field Illumination at 250X

Figure 40. Longitudinal Section of 24-11-3 After Initial Rolling at 1125 C and Finishing at 1080 C. Polarized Light Illumination at 250X

Figure 41. Transverse Section of 24-11-3 (Identical Process as in Figure 39). Bright Field Illumination at 750X

TABLE 6. ROLLING PARAMETERS AND RESULTS FOR TiAl-BASE ALLOY, TiAl+Nb

Billet/Sheet No.	Preheat Temp. (C)	Roll Temp. (C)	Thickness (inch) Initial	Thickness (inch) Final	No. of Passes	Total Reduction (%)	Roll Speed (ft/min)	Surface Dimensions (inch)	Remarks
1712(3A)	1200	815	0.54-0.73	0.21	4	65.0	68		Severely cracked
1713(3B)	1200	815	0.54-0.73	0.52	1	~10	68		Cracked
2720(3A)	1230	815	0.66	0.59	1	~12	68		Badly cracked
2721(3B)	1200	815	0.65	0.59	1	~10	68		Cracked on sides and trailing end. Double cladding of 0.02 inch. Ti 6Al4V was used.
2722(3C)	1370	870		0.45	44		22		TiAl internally oxidized and cracked
2723(3D)	1370	870		0.2	71		22-29	3.6 x 27.5	
2724(3E)	1370	870		0.49	51		33	3.9 x 8	Badly oxidized and cracked compact was sectioned out-gas and reclad
2725(3F)									Isothermally forged to small thick-ness ~0.02 inch at 1350 C. Was not rolled.

TABLE 6. (Continued) ROLLING PARAMETERS AND RESULTS FOR TiAl-BASE ALLOY, TiAl+Nb+W
(Preheat temperature of 1343 C and roll temperature of 870 C)

Billet/Sheet Number	Prior History	Cladding Design	Thickness (inch) Initial (inch)	Thickness (inch) Final (inch)	No. of Passes	Total Reduction With Clad	Roll Speed (fpm)	Nominal Surface Dimensions With Clad (inch x inch)	Average Reduction Per Pass	Remarks
3691A		0.37 inch thick Ti-6Al-4V cladding	1.036 C	0.152 C	11	85.4	30	29 x 3.8	10/15[a]	Essentially crack free
3691B		0.37 inch thick Ti-6Al-4V + 0.050 inch thick Ta	1.060 C	0.104 C	12	90.2	30	36 x 3.8	10/15	Crack free
3693A	Isothermally forged 26% at 1300 C		0.748 C	0.201 C	8	73.1	30	9.8 x 2.3	(1) (2) (3) (4) 10 / 20 / 25 / 30	Cracks which propagated rapidly observed after 4th pass
3693B	Isothermally forged 26% at 1300 C	0.37 inch Ti-6Al-4V clad picture frame + 0.10 inch thick Ta	0.922 C	0.108 C	13	88.3	30	39 x 3.8	(1-4) (5-8) (9-13) 10 / 15 / 20	Crack free
3693 C	Isothermally forged 26% at 1300 C	0.32 inch Ti-6Al-4V clad picture frame	0.813 C	0.112 C	9	86.2	30	30 x 2.6	15/20	Crack free
3693D	Isothermally forged 26% at 1300 C	0.32 inch Ti-6Al-4V clad picture frame with .05 inch Ta barrier	0.823 C	0.108 C	12	86.8	30	36 x 2.8	(1-4) (5-8) (9-13) 10 / 15 / 20	Crack free
3695A	Isothermally forged 26% at 1300 C	0.032 inch Ti-Al-4V clad picture frame	0.812 C	0.128 C	12	84.2	30	28 x 2.8	(1) (2) (3) (4) (5) 10 / 12 / 15 / 20 / 22	Cracks which propagated after 5th pass
3695B	Isothermally forged 26% at 1300 C	0.32 inch Ti-6Al-4V clad picture frame with 0.050 inch Ta barrier	0.835 C	0.166 C	12	87.3	30	39 x 3.8	(1-4) (5-8) (9-13) 10 / 15 / 20	Crack free

(a) Initial breakdown reductions/finishing reductions. Numbers in ( ) correspond to either the particular pass or the passes in a sequence used at the indicated reduction/pass.

alloy are less extensive than with Ti_3Al because a large part of the effort was involved in special preparation procedures to achieve crack-free sheet. These procedures were based on the following considerations:

(1) Special consolidation practices:
- Vacuum anneal at 1400 C for 3 hours
- Isothermal forging.

(2) Special cladding practices:
- Heavy Ti-6Al-4V cladding to minimize heat transfer
- Tantalum cladding to provide an inert, thermal barrier of nearly equivalent or higher strength than the TiAl.

(3) Reduction - strain rate schedules.

Early rolling trials with material from HIP Runs 1 and 2 showed that failures of the TiAl compacts during rolling were nearly catastrophic during either the first or second pass at preheat temperatures in the range of 1200 to 1400 C. Analyses of these failures showed the following:

(1) High temperature thermal treatments, within 60 C of the melting temperature, did not change the grain size significantly.

(2) Fracture was intergranular; the fracture surfaces did not appear metallic (e.g., the surfaces were dull and granular); tapping the sectioned rolled compacts resulted in considerable loss of powder, indicating the lack of complete consolidation.

(3) Deformation of the compact during rolling occurred by rapid thinning of the cladding, and rigid rotation of blocks of material between fracture surfaces and subsequent sliding and/or shear along these surfaces (see Figures 42 and 43).

(4) The relative amount of cladding thinning appeared to be reduced by rolling at temperatures above 1400 C.

(5) The rolling forces were not unusually high over the temperature range evaluated.

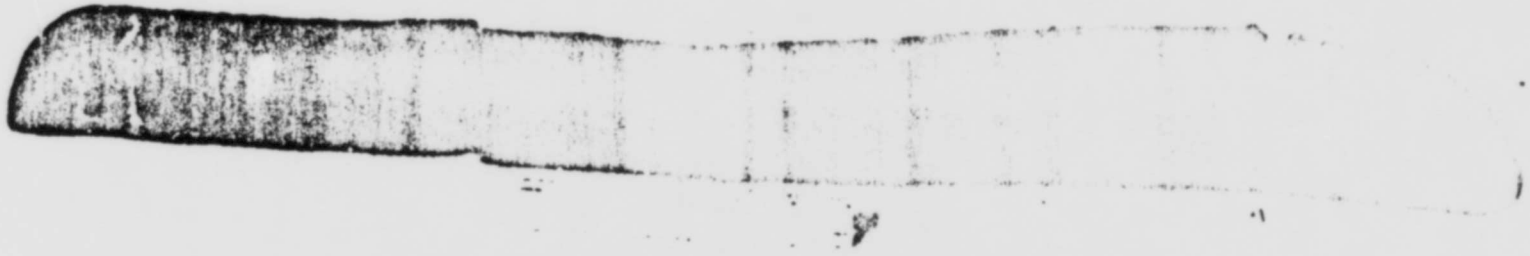

Figure 42. TiAl+Nb Sheet Failure After Rolling

Figure 43. TiAl+Nb Sheet Failure After Rolling
(Longitudinal Section at 5X)

These observations indicated the following potential sources and solutions for the poor rolling performance:

(1) The lack of significant grain growth after the annealing treatments, the non-metallic appearance of the fracture surfaces, and the observation of loose powder in the vicinity of the fractures indicated that the consolidation practices were inadequate for this alloy and that high temperature thermal treatments (to provide grain growth or sintering) and/or isothermal forging might avert these problems apparently associated with poor consolidation.

(2) The rapid thinning of the cladding indicated that
   (a) The cladding would not provide a significant thermal barrier
   (b) The rapidly deforming cladding could stress the compact surface in tension. These problems could be averted with a cladding more closely approximating the compact flow stress, but available materials were limited if reaction with the aluminide was to be avoided.

These deductions were also reinforced with

(1) The tensile test results in Table 4, which showed that the 1400 C annealing treatment improved 900 C ductility, which appeared sufficient,

(2) The difficulites in consolidation encountered by other investigators,[10] and

(3) The successful superplastic forging accomplished in the temperature range of 1066 to 1200 C[10].

The data in Table 6 and the Appendix document the results and consistent failures experienced in attempts to roll the TiAl+Nb alloy at preheat temperatures from 1090 to 1400 C using conventional practices for heated rolling of these HIP compacts. Roll temperatures from 800 to 900 C were used at roll speeds of 28 to 67 feet per minute, which corresponded to average strain rates in the range of 1 to 10 sec^{-1}, depending on reduction. Figures 42 and 43 show the typical failures experienced with the TiAl-base alloys.

A test plan was defined to determine the conditions necessary to achieve success in rolling the TiAl+Nb alloy while using advanced practices without the use of special cladding materials. This plan consisted of the following starting with the HIP compact:

(1) Vacuum anneal for 3 hours at 1400 C and vacuum furnace cool;
(2) Isothermal forge at 1300 C to in excess of 25 percent;
(3) Declad, grind, and inspect the forged sheet billet to insure the absence of surface cracks;
(4) Reclad the ground billets with 5.1 mm thick cover plates in a Ti-6Al-4V picture frame which was coated on the interior with Al_2O_3 + CrO_3 powder mixture to act as a parting compound and diffusion barrier;
(5) Roll the assemblies at a series of temperatures and inspect the product to determine the causes of failure and preferred rolling conditions.

This inspection was performed with Compact 4770, which is shown in Figure 44 after isothermal-creep forging in vacuum at 1300 C. This forging was performed with graphite dies at a maximum nominal pressure of 7.6 MPa (1100 psi) to achieve a 25 percent reduction in 4 minutes. This billet was declad, ground and sectioned into four sheet billets nominally 7.9 x 31.8 x 76.2 mm. Metallographic observation of the as-forged billet showed it to be crack-free, but porosity was occasionally observed at grain boundary triple points. This porosity was probably caused by the forging since there was no exposure of the powder to insoluble gases and warm vacuum outgassing was used to minimize gas inclusion. This porosity was minimal, but was of concern. However, preparations were continued because the material was the product of preferred practices (e.g., our options had been eliminated).

The coated cladding and billets were assembled and the assembly was EB welded approximately 12.7 mm away from the aluminide. These compacts were subsequently rolled at 20.4 m per minute at reductions of nominally 10 percent per pass at a roll temperature of 875 C. The total reduction was nominally

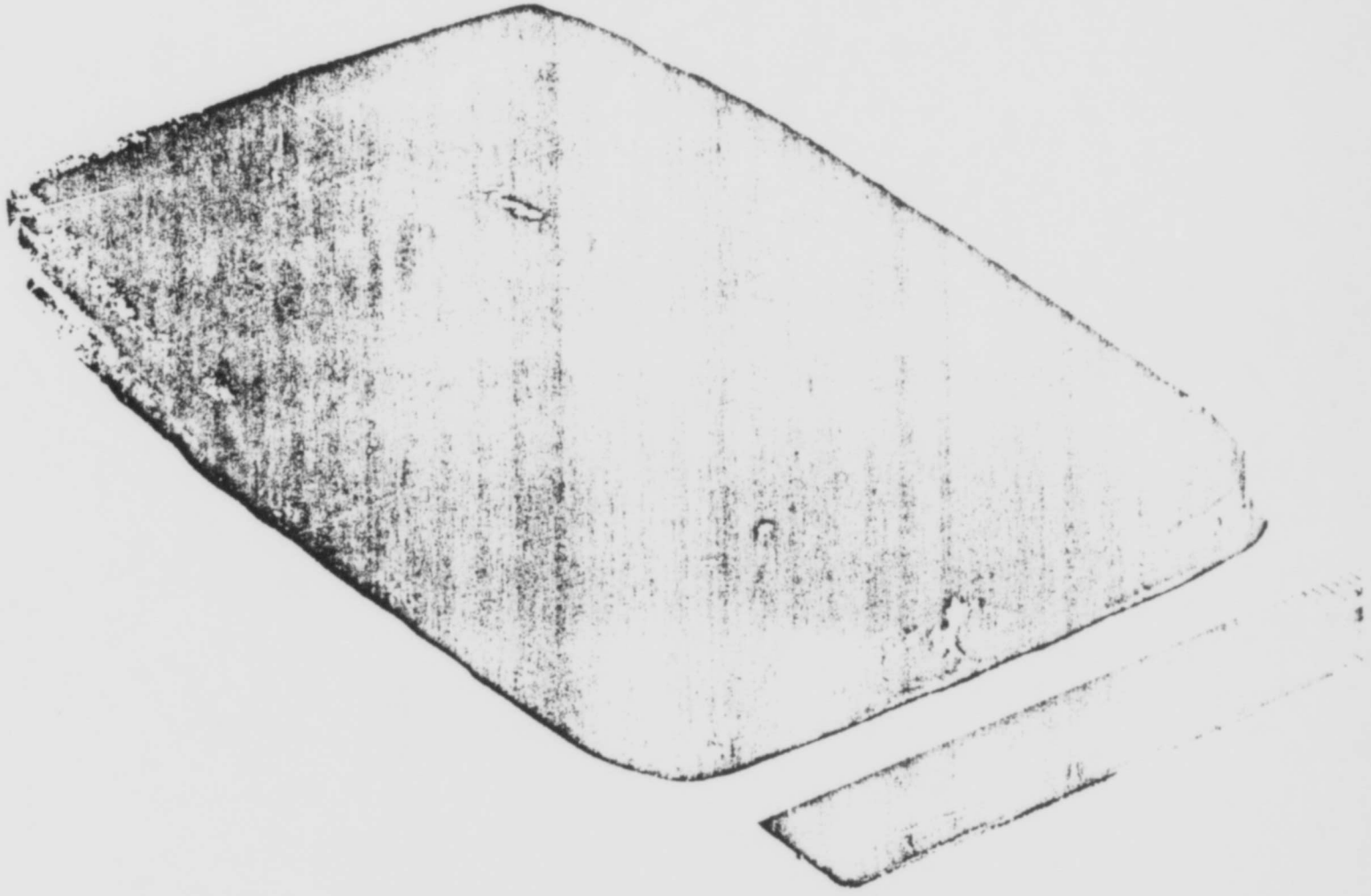

Figure 44. TiAl+Nb Billet Number 4770 After Isothermal Forging at 1300 C (Reduction in Thickness was 26%)

40 percent achieved in five passes. After each pass, the assemblies were reheated for 5 minutes after reaching the preheat temperature. The following billet-preheat temperatures were used:

4770A - 1440 C
4770B - 1330 C
4770C - 1400 C
4770D - 1370 C.

Radiographs of the resulting sheet are shown in Figure 45 and polished transverse sections are shown in Figure 46. Clearly, the incidence of cracking decreased with increasing temperature up to 1440 C, which is within 20 degrees of the melting temperature. Figure 45 shows that rolling at 1440 C almost produced crack-free sheet. The cracks occurring at either end may have been caused by handling, e.g., chilling due to contact with the tongs.

These results demonstrated the following:

(1) TiAl+Nb cannot be rolled with consistent success using rolls heated to temperatures below 875 C and at roll speeds between 9.1 to 21.3 m per minute.
(2) The metallic appearance of the sheet correlated with the observed ductility during rolling.
(3) The fracture surfaces (normal to the rolling direction) cannot originate from states of stress typically encountered in rolling conventional alloys isothermally.[16] The fractures appear to have originated from flow stress gradients induced either thermally, by heat transfer, or by the flow stress of the cladding being lower than the aluminide.

Observations of the HIP compacts of TiAl+Nb+W after the 1400 C annealing (see Figure 18) showed that this processing resulted in a very large grain size, contrary to observations with TiAl+Nb. Although the large grain size was objectional for workability, the fact that significant growth had occurred indicated that consolidation was complete. However, it should be recognized that all TiAl-base compacts were given the 1400 C anneal, unless specified otherwise, and the resulting large grain size of TiAl+Nb+W probably affected the rolling results until the compacts were recrystallized.

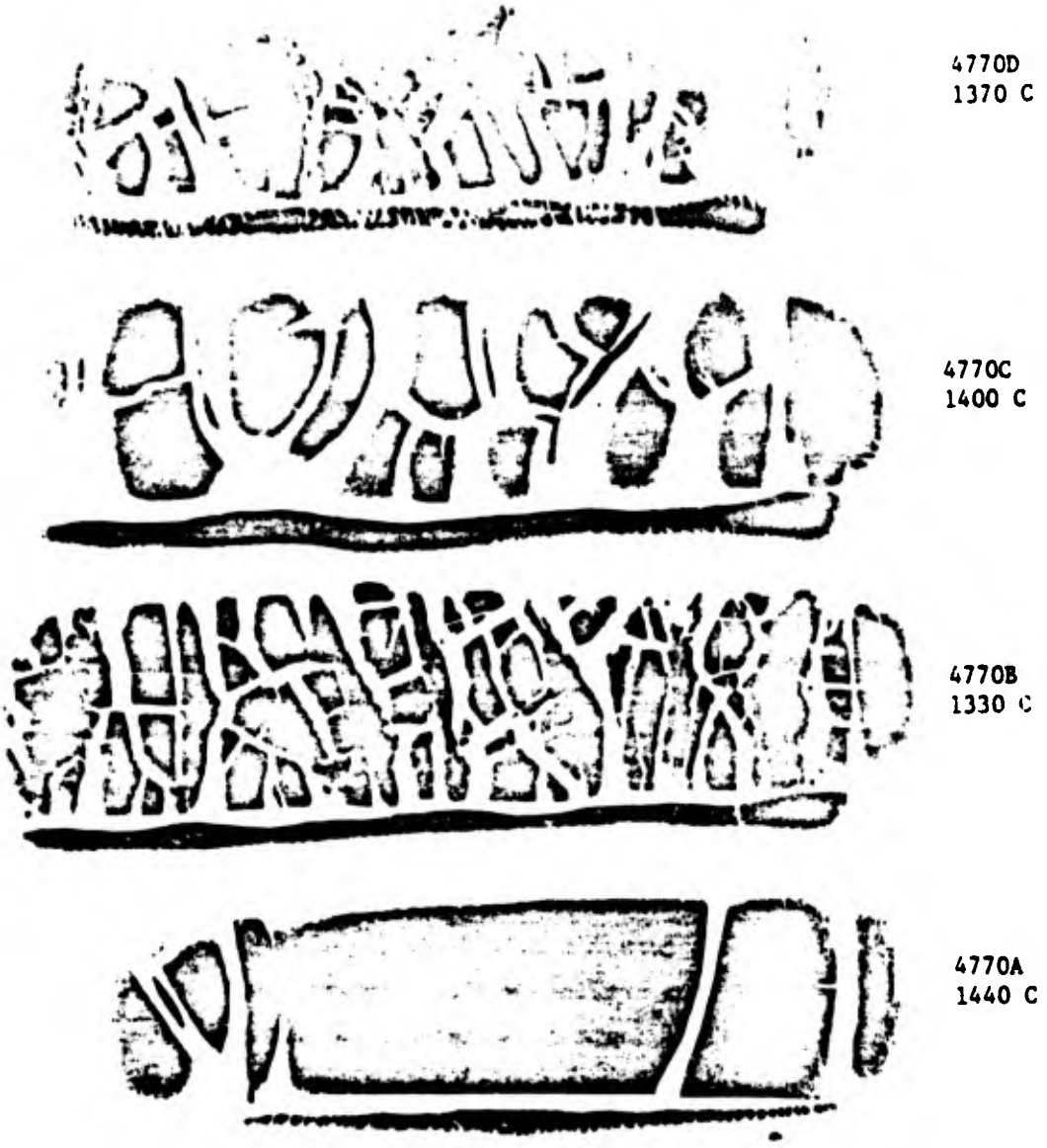

Figure 45. Radiographs of TiAl+Nb Billets 4770A Through D After Rolling at the Various Indicated Temperatures. Magnification: 1X

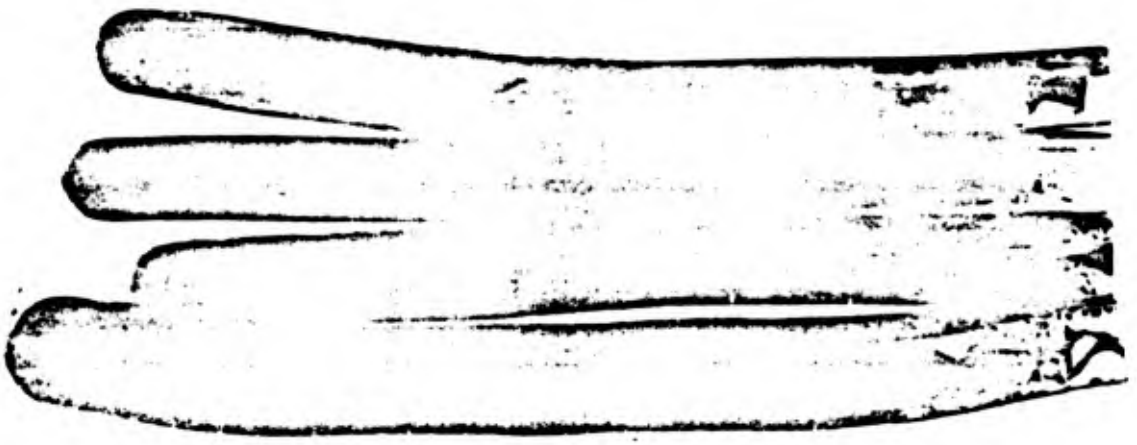

Figure 46. Longitudinal Sections of TiAl+Nb Packs Numbered 4770A-D. Billet Preheat Temperatures Top to Bottom were 1400 C, 1400 C, 1370 C, and 1330 C, Respectively

The results for TiAl+Nb+W are summarized in Table 6 in greater detail than for the other alloys because of the special procedures necessary to obtain successful rolling. Because initial trials were successful with a preheat temperature of 1350 C, this temperature was fixed for the entire rolling campaign. However, it was anticipated that lower preheat temperatures could be used with TiAl+Nb+W because of the extensive grain growth experienced at 1400 C. Because of the massive cladding, preheat times of 30 minutes were used with intermediate times of 15 minutes. All rolling was performed at a roll speed of 18 m per minute and a roll temperature of 870 C.

The results in Table 6 show that two conditions were associated with successful rolling:

(1) The use of 1.27 or 2.54-mm thick Ta cover plate on the surface of the billet

(2) Using light reductions (less than 15 percent) during the initial or breakdown passes.

No failures were experienced when the Ta sheet was used. However, with this exception, light reductions during the initial passes were necessary for successful rolling independent of the processing prior to rolling (e.g., annealed at 1400 C or annealed at 1400 C and isothermally forged at 1300 C) and the cladding design. The reduction-pass sequence which was consistently successful was as follows:

(1) First four passes - nominal 10 percent reduction per pass

(2) Second four passes - nominal 15 percent reduction per pass

(3) Subsequent passes - nominal 20 percent reduction per pass.

Rolling measurements for the TiAl-base alloys in the Appendix show that for the conditions evaluated the flow stresses were in the range of 34.5 to 138 MPa and the range for TiAl+Nb+W was typically 62.1 to 110.3 MPa (9,000 to 16,000 psi) at 1350 C.

Figures 47 and 48 are micrographs of the TiAl-base alloys after rolling. Figure 47 shows the profuse grain boundary fracture for sheet 2724 rolled at 1370 C typically observed after rolling the TiAl+Nb alloy. This sheet was produced from a Series 2 compact which satisfactorily passed the TIP test. Figure 48 is the microstructure of sheet 5691B after rolling to a reduction of 90 percent at 1340 C. This structure has a very fine grain size, less than 5 microns,

7H917 100X

a. Longitudinal Section of Sheet 2724

7H922 250X

b. Longitudinal Section of Sheet 2724

Figure 47. Longitudinal Section of TiAl+Nb Compact 2724 After Rolling to a Total Reduction of 51 Percent at 1370 C

Figure 48. TiAl+Nb+W Sheet 5691B After Rolling to A 90% Reduction at 1340 C. Polarized Light Illumination at 250X

and the near-equiaxed shape indicates that recrystallization was complete during the rolling cycle. The elongated structures in this figure were probably caused by incomplete solubility of the tungsten.

Forming

Vacuum and rigid-punch forming and roll forming were performed with Ti-25Al-5Nb sheet produced during this program. Vacuum and rigid-punch forming with Ti-25-5Nb Series 1 sheet was performed at displacement rates of 12.7 mm/hour and 25.4 mm/hour (average strain rates of 0.003/sec. and 0.006/sec., respectively) at 1370 and 1200 C with the die set described in Figure 5. This strain rate range was selected based on communication of data by Rockwell International[12]. All tests were performed in argon after the blank, at room temperature, had been in the dies for 10 minutes. At both temperatures, vacuum forming produced cup depths of approximately 12.7 mm. However, at 1370 C, for both 0.89 and 1.27 mm thick blanks at 25.4 mm/hour, and a 0.89 mm thick blank at 12.7 mm/hour displacement rate, a continuous load increase without decay was experienced beyond a cup depth of approximately 12.7 mm. This observation was originally attributed to either or both grain growth or internal oxidation. Metallography of a failed cup showed very coarse α_2 grains, severe surface oxidation, and indications of internal oxidation. The observed fracture was predominantly transgranular. However, because little oxidation had occurred on the fracture surfaces and cracks occurred through the oxidized surface, it appeared that fracture occurred near the completion of forming. Based on these observations at 1370 C, an attempt was made to form a cup from 1.27 mm thick strip at 1200 C. This cup was successfully formed to a 16.8 mm depth as shown in Figure 49.

These preliminary trials showed that forming at 1200 C is preferred over 1370 C to avoid severe oxidation and grain growth. However, the subsequent results of this program showed that temperatures below 1100 C would be preferred with sheet processed to achieve a finer grain size.

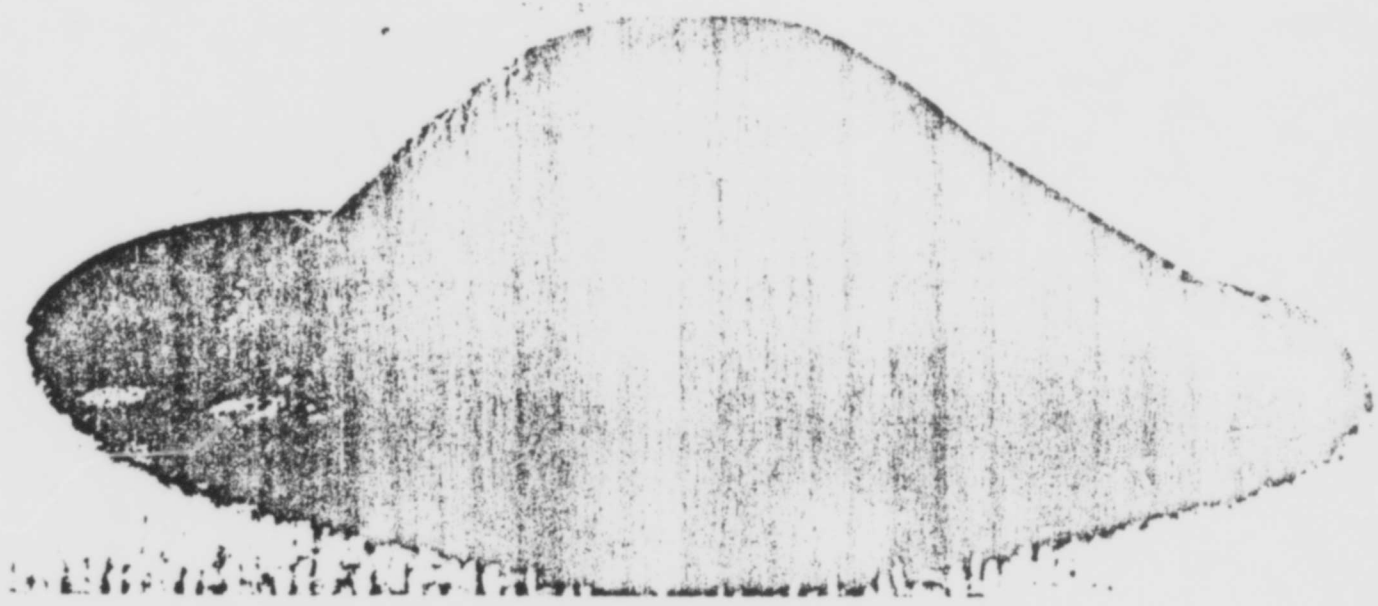

Figure 49. Ti-25Al-5Nb Cup Formed at 1200 C Before Chemical Polishing (Mag. 2.3X)

Roll forming of cylindrical shapes was performed with the conventional roll forming equipment shown in Figure 6 with rolls gas-flame heated to 400 $\pm$30 C. The fractional strain, $\bar{\varepsilon}$, during rolling of sheet of thickness t to a diameter D is approximated by the relation

$$\bar{\varepsilon} = t/D \quad ,$$

and the corresponding strain rate

$$\dot{\bar{\varepsilon}} = \frac{D_R R}{D} \quad , \quad \text{where}$$

D_R is the roll diameter and R is the rotational speed of the rolls. For sheet of thickness t = 1.27 mm formed to a diameter D = 76.2 mm with rolls of 51 mm diameter rotating at a speed of 3.05 m/min. or R = 19 rpm, the strain and strain rate are 0.017 (or 1.7%) and 13 min^{-1} (or 21% sec^{-1}), respectively. These findings, when considered with the tensile test results, indicated that Ti-25Al-5Nb could be successfully roll formed at temperatures above about 800 C.

Preliminary forming trials showed that multiple passes would be required to feed the sheet through the rolls at the relatively small part radii being attempted. After these trials, forming runs were made to provide the results in Table 7 and parts shown in Figure 50. These trials were essentially successful except for sheet 5895-4-2, which exhibited edge cracks, and sheet 5890-1-2, which had minor surface cracks. These surface cracks occurred after four passes without reheating. The temperature, measured with a contact pyrometer, after the final pass, was 315 C, less than the nominal roll temperature of 370 C. This low reading probably resulted from extraction of heat by the contact pyrometer.

These results clearly show that the Ti-25Al-5Nb alloy can be successfully roll formed using near conventional procedures. A preheat temperature of 870 C appears to be near the minimum temperature to obtain successful forming and chilling of the product to temperatures below about 400 C should be avoided.

TABLE 7. CONVENTIONAL ROLL FORMING OF BARE Ti-25Al-5Nb SHEET

Sheet Number	Nominal Dimensions Length x Width x Thickness, (inch)	Preheat Temperature (C)	Roll Temperature (C)	Number of Passes	Final Diameter (inch)	Remarks
5890-1						
-1	3.5 x 3.5 x 0.060	870	430	8	2.88	
-2	4.1 x 3.5 x 0.060	982	430	1st&2nd		
		927	430	3rd		
		900	430	4th		
		885	430	5th		
		877	430	6th		
		871	430	7th		
		860	371	8th		
		816	371	9th[(a)]		
			371	10th[(a)]		
			371	11th[(a)]	2.31	Minor surface cracks
			371	12th[(a)]		
5891-1-1	3.8 x 1.8 x 0.059	982	430	6	2.81	
5895-4-2	5.4 x 2.7 x 0.059	871	430	12	2.25	Minor edge cracks

(a) The specimen was not reheated between the passes.

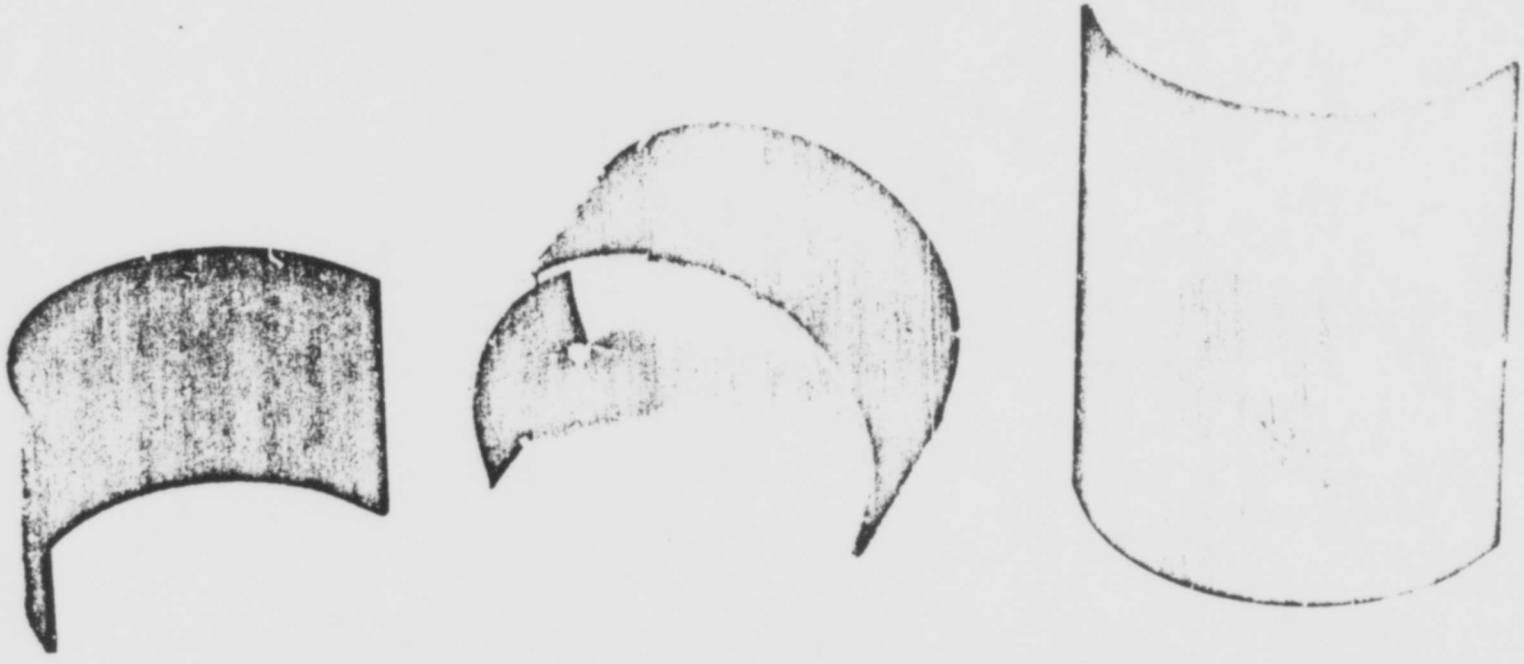

Figure 50. Roll-Formed Sections of Ti-25Al-5Nb

Mechanical Testing

Testing of sheet was performed using the specimens previously described in Figure 8 and in the discussion in the section on Testing. Both tensile and bend testing were performed in two campaigns corresponding to material obtained before and after HIP Run 3 (e.g., before and after the preferred HIP practices were developed). The results in Table 8 were obtained from sheet for which the compact was unaffected by the TIP tests and there was no indication of TIP during heat treatment. However, the microstructures of the material from this HIP run all showed some indications of porosity. The results in Table 8 show that there was no significant difference between the bend test results at room temperature and -73 C; the alloy begins to exhibit some ductility at 500 C; and rolling of solution treated material at temperatures in the range of 870 to 980 C appears to degrade the bend test properties.

The tests in Table 9 on transverse samples from as-rolled material were performed to obtain data to establish minimum roll-forming temperatures. Although these results showed some indications of ductile behavior at 538 C, this minimum temperature would not be satisfactory for roll forming.

The tensile test results in Table 10 and plotted in Figures 51 and 52 show that the Ti-25Al-5Nb alloy sheet exhibits increasing ductility at temperatures above approximately 500 C for both the solution treated and aged and the as-rolled conditions. Although the transverse ductility is consistently lower than the longitudinal, this same behavior was not consistently reflected in the yield and tensile stresses.

Indications of strain aging were observed during testing of solution treated and aged specimen 5892BFL6 at 427 C. Stress decrements as large as 6,000 psi were observed up to about maximum load. Unfortunately, this specimen was the only one tensile tested at 427 C.

An apparently anamolous response was experienced with specimens 5890B1RL3 and 5892BRT2, longitudinal and transverse specimens, respectively,

TABLE 8. BEND TEST RESULTS ON HEAT-TREATED Ti-25Al-5Nb SHEET FROM HIP RUN 2 (Displacement rate of 0.001 inch/minute)

Specimen Number	Specimen Orientation[a]	Specimen Temperature, C	Maximum Fiber Stress at Fracture, ksi	Maximum Total Fiber Strain at Fracture, Percent
		Sheet 2823-1 (Solution Treated and Aged)		
3a	L	RT	91.8	1.5
3b	L	-73	66.6	1.4
7a	T	RT	75.1	0.95
7b	T	-73	81.7	1.0
8a	T	500	96.9	3.3
9a	L	500	103.2	4.2
9b	L	RT	69.3	1.5
		Sheet 2825-2 (Solution Treated and Aged)		
2A (980 C)[b]	L	RT	51.4	1.5
2B (925 C)[b]	L	RT	45.9	1.5
2C (870 C)[b]	L	RT	92.6	1.8

(a) Specimen orientations L and T were sectioned parallel and perpendicular to the rolling directions.

(b) Rolled at indicated temperature after solution treatment.

TABLE 9. TRANSVERSE BEND TEST RESULTS ON AS-ROLLED Ti-25Al-5Nb FROM HIP RUN 5
(Displacement rate of 0.025 mm/minute)

Specimen Number	Test Temperature, F (C)	Maximum Fiber Stress at Fracture, ksi	Maximum Total Fiber Strain at Fracture, Percent	Hardness, HV-10
(1) Rolled at 2150 F (Finishing - 3 passes with rolls at 250 F):				460 ± 15
5891A-1-1	RT	127.3	0.68	
5891A-1-2	800 (427 C)	92.0	1.20	
5891A-1-3	1000 (538 C)	231.9	1.24	
(2) Rolled at 2050 F:				311 ± 7
5891-B-1-1	RT	79.6	0.70	
5891-B-1-3	1000 (538 C)	152.8	1.37(a)	
(3) Rolled at 2000 F:				349 ± 13
5892BR-1	RT	72.6	0.71	
5892BR-2	800 (427 C)	37.52	0.86	
(4) Rolled at 1900 F:				318 ± 10
5895B-1	RT	84.4	1.22	
5895B-2	500 (260 C)	85.2	1.27	
5895B-3	1000 (538 C)	132.6	1.61	

(a) Includes small plastic strain (≈0.55 percent).

TABLE 10. TENSILE TEST RESULTS FOR Ti-25Al-5Nb SHEET

Specimen Number[a]	Test Temperature, F (C)	Yield[b] Stress, ksi	Tensile[b] Strength, ksi	Plastic Elongation, Percent	Hardness, HV-10	Remarks
Sheet rolled at 2000 F (1093 C) and air cooled						
1812-2T1	RT	--	42.3	--		
1812-2L1	RT	--	57.6	--		
1812-2L2	(900 C)	18.2	29.9	41.7		n = 0.15
1812-2L3	(1120 C)	0.44	0.88	102		n = 0.4
Sheet rolled at 2100 F (1120 C) or 2150 F (1175 C)						
(1) Heat treatment: 2200 F for 1 hour and air cooled + 1600 F for 8 hours/A.C.:						
2823-1-1	RT	--	45.1	nil		
2823-1-2	RT	--	30.9	nil		
2823-1-4	932 (500 C)	64.1	72.9	5.0		
2823-1-5	932 (500 C)	61.8	64.1	3.0		
2823-1-6	1472 (800 C)	45.3	55.4	7.5		
2823-2-1	1472 (800 C)	38.3	51.4	20		
2823-2-2	1292 (700 C)	45.1	62.4	12		
2823-2-3	RT	--	22.0	nil		
2823-2-4	1472 (800 C)	--	39.8	8		
2823-2-5	1472 (800 C)	38.9	61.5	14		

TABLE 10. (Continued)

Specimen Number[a]	Test Temperature, F (C)	Yield[b] Stress, ksi	Tensile[b] Strength, ksi	Plastic Elongation, Percent	Hardness, HV-10	Remarks
(2) Heat treatment: 2200 F for 1 hour and air cooled and rolled as indicated:						
2825-21 (980 C)[c]	RT	--	22.3	nil		
2825-22 (925 C)[c]	RT	--	57.8	nil		
2825-23 (870 C)[c]	RT	--	49.9	nil		
Sheet rolled at 2000 F (1093 C)						
(1) Heat treatment: 2200 F for 1 hour and air cooled + 1600 F for 8 hours/A.C.:					276 ± 10	
5892BFL1[a]	RT	--	45.1	0.12		
5892BFT1[a]	500 (260 C)	--	--	--		Broke in the grips
5892BFL6	800 (427 C)	42.1	79.4	9.7		Serrated stress-strain curve
5892BFL2	1200 (649 C)	40.2	66.4	7.0		
(2) Heat treatment: 2000 F for 1 hour and air cooled + 1600 F for 8 hours/A.C.					267 ± 9	
5892AL1	RT	--	50.7	--		
5892AT1	500 (260 C)	44.4	66.0	6.0		
(3) Heat treatment: 1600 F for 8 hours and air cooled:					282 ± 11	
5890B2L1	RT	--	49.8	--		

TABLE 10. (Continued)

Specimen Number[a]	Test Temperature, F (C)	Yield[b] Stress, ksi	Tensile[b] Strength, ksi	Plastic Elongation, Percent	Hardness, HV-10	Remarks
5890B2T1	500 (260 C)	--	44.3	--		
5890B2L2	1200 (649 C)	57.8	66.6	1.84		
(4) Heat treatment: NONE (as-rolled at 2000 F):						
5890B1RL1	RT	--	40.9	0.14	321 ± 16	
5892BRT1	500 (260 C)	--	51.3	--		
5890B1RL2	1200 (649 C)	--	--	--		
5890B1RL3	1200 (649 C)	14.5	22.2	5.4		Broke in the grips
5892BRT2	1200 (649 C)	18.9	21.7	2.5		

(a) Orientation relative to rolling direction: L = Longitudinal and T = Transverse.

(b) All specimens tested at a nominal strain rate of 0.001 $sec.^{-1}$ except Specimen Numbers 2823-2-2 and -5, which were run at 0.01 $sec.^{-1}$ and Specimen Numbers 1812-2L2 and -2L3 were cycled between 0.001 and 0.01 sec^{-1}.

(c) Rolling-aging temperature.

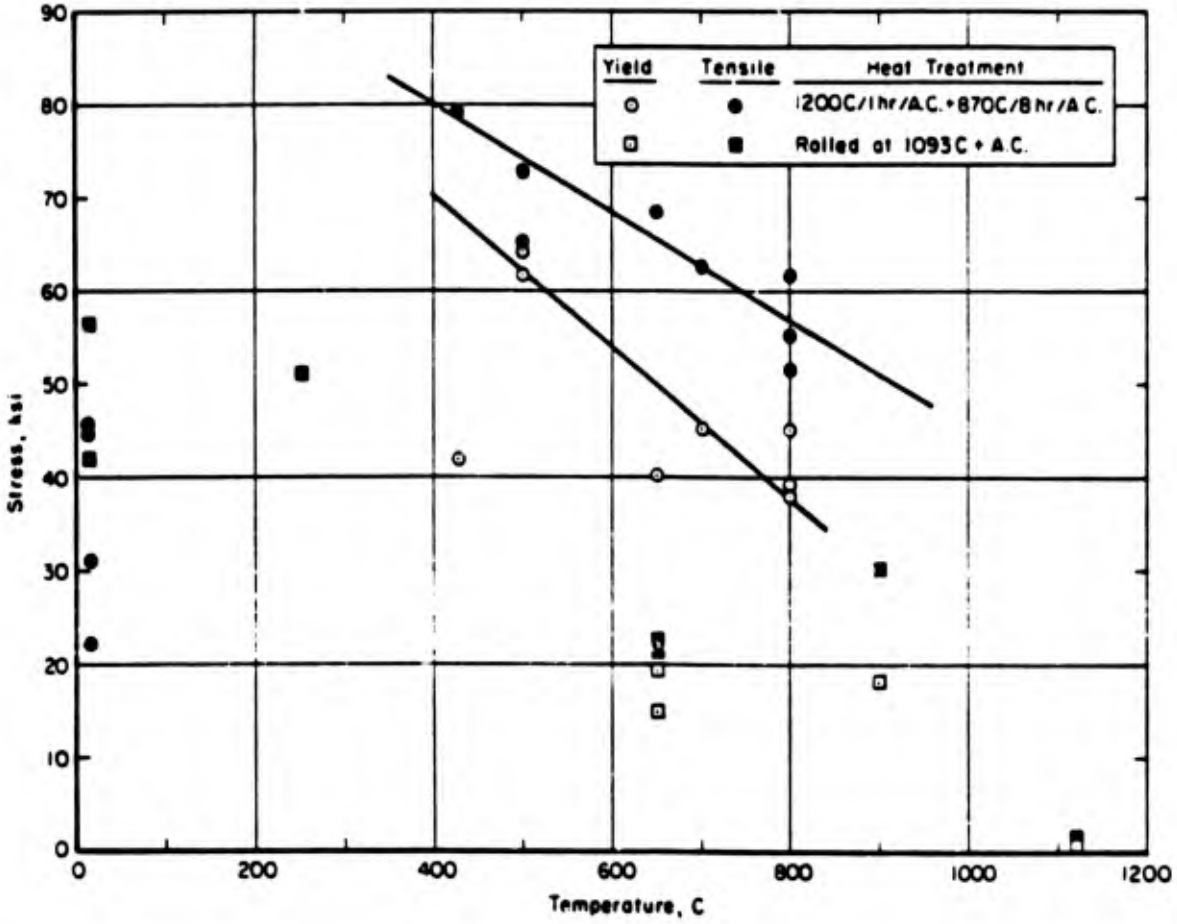

Figure 51. Stress Dependence on Temperature of Ti-25Al-5Nb Sheet

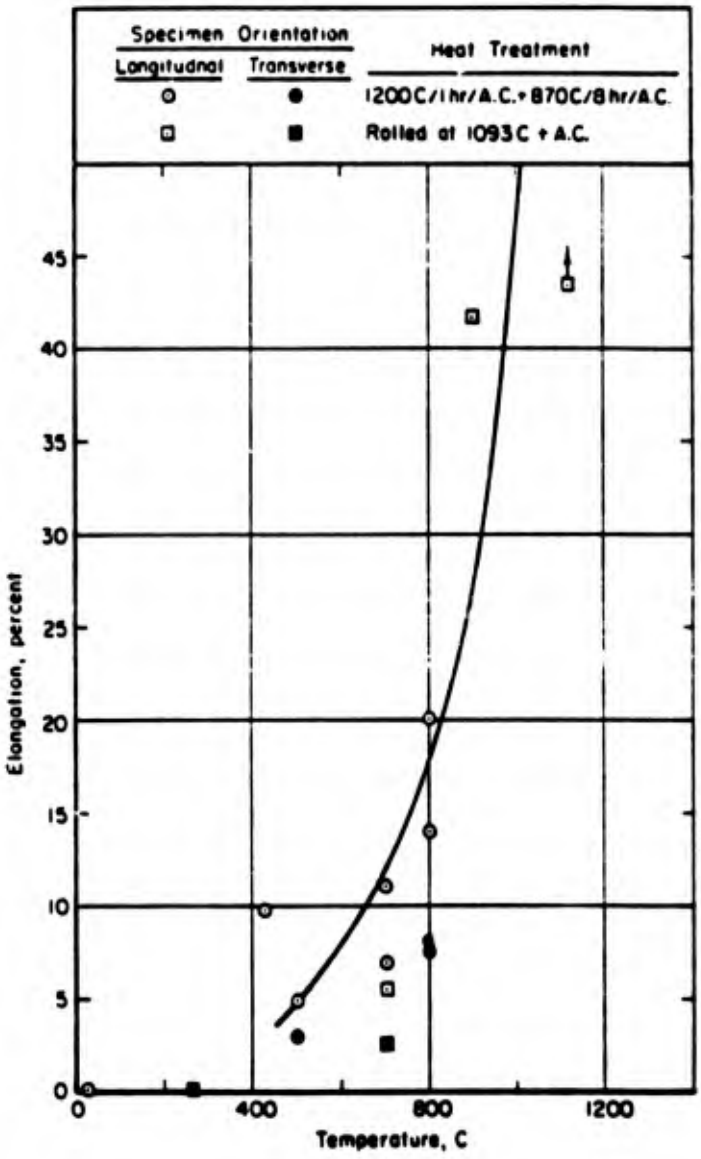

Figure 52. Elongation Dependence on Temperature for Ti-25Al-5Nb Sheet

tested at 649 C in the as-rolled condition. These specimens exhibited very low yield and tensile stresses which appeared to be associated with recrystallization during testing. A 50 percent increase of the grain size was measured at the fracture area after testing.

Fatigue test results for an R = 0.1 and at temperatures of 427 and 649 C are presented in Table 11 for Ti-25Al-5Nb after various heat treatments. All specimens which failed after 4000 cycles exhibited multiple cracks in the gage sections. These cracks were usually visible with a low power microscope or visually.

The fatigue stress was established using 0.6, and, subsequently, 0.75 and higher, times the tensile yield stress at 649 C to achieve an acceptable number of cycles to failure, in agreement with published data. However, it was found that the ratio of fatigue stress to yield stress at equivalent cycles to failure varied significantly with heat treatment, the highest being the solution treated and aged material. However, runout (1.4×10^7 cycles) was obtained with the material (Heat Treatment 3) annealed for 8 hours at 870 C at a stress (35 ksi) at which all other heat treatments produced failure in less than 6.1×10^4 cycles.

Creep test results at 649 C in Table 12 were obtained on Ti-25Al-5Nb rolled at 1093 C and given the heat treatments indicated on the table. The solution treated and aged material (Heat Treatment 1), which were run at the highest fractions of the tensile yield stress, did not rupture at times in excess of 400 hours and, therefore, were terminated. This material had a fully transformed acicular α_2 microstructure of relatively fine grain size, which was difficult to measure. The other heat treatments had coarser, more equiaxed structures as shown in Figures 30 and 31. The microstructure providing the best creep performance (time to 1 percent) was obtained with Heat Treatment 3 (as-rolled + 1600 F/8 hrs. + A.C.).

TABLE 11. FATIGUE TEST RESULTS FOR Ti-25Al-5Nb SHEET ROLLED AT 2000 F (Roll Temperature 1550 F)

Specimen Number[a]	Test Temperature F (C)	Applied Stress (ksi) Max.	Min.	Cycles to Fracture	Remarks
(1) Heat treatment: 2200 F for 1 hour and air cooled + 1600 F for 8 hours and air cooled:					
5892BFL5	800 (427 C)	35.0	3.50	60,870	
(2) Heat treatment: 2000 F for 1 hour and air cooled + 1600 F for 8 hours and air cooled:					
5892AT2	1200 (649 C)	26.6	2.66	79,144	Broke in the grips while adjusting the thermocouple
5892A15	800 (427 C)	35.0	3.50	11	Possible presence of a mechanical defect
5892AL6	800 (427 C)	35.0	3.50	73	Broke in the grip
(3) Heat treatment: 1600 F for 8 hours and air cooled:					
589081F12	800 (427 C)	43.5	4.35	16,45	
589081FL3	800 (427 C)	35.0	3.50	14×10^6	Test discontinued
5890 B2T3	800 (427 C)	43.5	4.35	4,077	

(a) L and T in Specimen Number denote longitudinal and transverse specimens, respectively.

TABLE 12. CREEP TEST RESULTS FOR Ti-25Al-5Nb SHEET ROLLED AT 2000 F (Roll Temperature 1550 F)

Specimen Number	Test Temperature F (C)	Tensile Yield Stress at 1200 F (ksi)	Applied Stress (ksi)	Time to 1% Creep (hours)	Creep Rupture Strain (%)	Time to Ruptuer (hours)	Remarks
(1) Heat treatment: 2200 F for 1 hour and air cooled + 1600 F for 8 hours and air cooled (acicular α_2):							
5892BFL3	1200 (649 C)	40.2	31.0	267.6	2.02*	617*	Test discontinued
5892BFL4	1200 (649 C)	40.2	34.0	1.6	11.37*	428*	Test discontinued
(2) Heat treatment: 2000 F for 1 hour and air cooled + 1600 F for 8 hours and air cooled (13.5 microns average grain size):							
5892AL3	1200 (649 C)	44.4	29.5	--	--	--	Broke within 5 minutes; no measurable strain
5892AL4	1200 (649 C)	44.4	34.0	30.8	10.00	342	
(3) Heat treatment: 1600 F for 8 hours and air cooled (12 microns average grain size):							
5890B2L3	1200 (649 C)	57.8	41.5	100.0	6.0	246.1	
5890B1FL1	1200 (649 C)	57.8	43.5	16.4	10.0	255.0	

* Test discontinued at indicated strain and time.

CONCLUSIONS

The results of this program on the titanium aluminide alloys based on Ti_3Al and TiAl demonstrated the following:

(1) Both cast and HIP-powder compacts of the Ti_3Al-base alloys can be rolled successfully to sheet with the same degree of success.

(2) The limited ductility of these alloys necessitates special considerations for HIP-can material selection and material preparation to avoid cracking during the HIP cycle.

(3) Success in rolling sheet to 1.3 mm from HIP-powder compacts occurred with the alloys which exhibited significant grain growth at and above the HIP temperature. Failure during rolling was consistently observed with the TiAl+Nb compacts for which the average grain size could not be made to exceed the range of the powder-particle size. This failure always occurred early in the rolling sequence, but was nearly completely averted by rolling within 20 C of the melting temperature. Isothermal forging of the TiAl+Nb was nearly successful except for porosity at grain-boundary triple points. The improved ductility observed with the TiAl+Nb alloy after HIP and annealing at 1400 C may have been caused by stability of the oxide film on the REP powder. Further evaluation of this possibility and procedures to avert complications should be pursued before it is concluded that sheet cannot be rolled from TiAl+Nb powders.

(4) The range of forces for rolling of the Ti_3Al-base alloys at temperatures above 1000 C (with heated rolls) is within current commercial practices with other alloys.

(5) The Ti-24Al-11Nb alloy containing 0.25 percent Si possessed the finest grain size and the best reported superplastic behavior. Although finer grain sizes appeared more easily attainable with the Ti-24Al-11Nb alloy, the processing of this alloy was not fully evaluated because Ti-25Al-5Nb and the TiAl-base alloys were the primary program alloys. The finest average equiaxed grain size obtained with the Ti-25Al-5Nb sheet was 11 microns, although a finer acicular structure was obtained after heat treatment and low temperature (below 1100 C) rolling. The two phase $\alpha + \beta$ microstructure of the Ti-24Al-11Nb was probably most critical in achieving a fine grain size and superior superplastic behavior.

(6) The principles for working of the current generation of the titanium-aluminide alloys are similar to, but less complex, than other alloys which must be hot worked. Success was consistently achieved by working at temperatures and deformation rates where recrystallization appeared to occur during working. Typically, the temperatures were in the range from high in the α range to high in the $\alpha + \beta$ range for the Ti_3Al-base alloys. Higher temperatures (e.g., well into the β field) limited success by causing rapid grain growth. However, slightly higher temperatures than evaluated in this study could be beneficial for conventional working, e.g., breakdown of the cast structure, by forging or rolling at moderate strain rates to avert cooling to unacceptably low temperatures, below about 950 C. These alloys are simpler to hot work than others because of the lack of precipation of high-temperature strengthening phases.

(7) The TiAl+Nb+W alloy can be successfully rolled consistently at 1350 C with heated rolls if light reductions are used for the initial passes. However, the demonstrated successful procedures were limited and may have been necessitated largely by the large grain size obtained after the 1400 C anneal, which was used as a pretreatment for rolling these alloys.

(8) The Ti-25Al-5Nb alloy can be conventionally roll formed at preheat temperatures as low as 870 C and minimum part temperatures of approximatley 500 C during forming.

(9) Mechanical testing of the Ti-25Al-5Nb alloy showed that the brittle-to-ductile transition temperature was in the range of 500 to 600 C, but depended on prior treatment and the orientation (longitudinal or transverse) of the specimen. With the exception of ductility, the best performance during tensile, creep and fatigue testing was achieved with the α-rolled material given a low temperature (870 C) aging or stabilization treatment.

APPENDIX

REPRESENTATIVE ROLLING MEASUREMENTS AND ANALYSIS FOR THE TITANIUM-ALUMINIDE ALLOYS BASED ON Ti_3Al AND TiAl

TABLE A-1. MEASUREMENTS DURING ROLLING OF TITANIUM ALUMINIDE ALLOY, Ti-25Al-5Nb

Billet/Sheet Number	Sheet Thickness (in)		Temperature (C)			Roll Speed (fpm)	Separating Force (10^3 lbs/in)			Flow Stress (Ksi)	Strain Rate (sec^{-1})
	Initial	Final	Preheat	Rolls	Final		Front	Middle	Rear		
5891											
-1	0.702	0.590	1175	815	1010	50		6.4		9.1	2.5
-2	0.590	0.500	1175	815	1020	50		4.3		6.8	2.6
-3	0.500	0.420	1175	815	1004	50					
-4	0.420	0.340	1175	815	1018	50		5.7		8.5	3.5
-5	0.340	0.260	1175	815	1020	50		5.7		7.8	4.5
5891-1											
-1	0.260	0.240	1175	815	950	50		8.0		23.2	2.8
-2	0.240	0.190	1175	815	950	50		6.7		11.0	5.0
-3	0.190	0.160	1175	815	950	50		6.7		13.9	4.8
5891-11	Conventionally Rolled										
-1	0.160	0.140	1175	125	800	100		21.6		25.9	6.6
-2	0.140	0.095	1175	125	800	100		23.2		15.4	11.5
-3	0.095	0.070	1175	125	800	100		17.6		14.0	10.8
5891-2											
-1	0.260	0.240	1175	815	925	50		7.3		21.5	2.8
-2	0.240	0.200	1175	815	930	50		5.2		10.1	4.4
-3	0.200	0.155	1175	815	925	50		8.4		13.5	5.6
5891-22											
-1	0.163	0.134	1125	815	954	105		33.6		48.7	12.3
-2	0.134	0.120	1125	815	900	105		32.8		58.9	9.1
-3	0.120	0.096	1125	815	952	105		30.4		41.2	14.2
-4	0.096	0.086	1125	815	960	105		25.6		48.4	10.6

TABLE A-1. (Continued)

Billet/Sheet Number	Sheet Thickness (in)		Temperature (C)			Roll Speed (fpm)	Separating Force (10^3 lbs/in)			Flow Stress (Ksi)	Strain Rate (sec^{-1})
	Initial	Final	Preheat	Rolls	Final		Front	Middle	Rear		
5892											
-1	0.680	0.510	1175	815	1032	50		4.9		5.3	3.3
-2	0.510	0.390	1175	815	1026	50		3.8		4.7	3.7
-3	0.390	0.305	1175	815	1010	50		4.1		6.1	3.9
5892-1											
-1	0.305	0.280	1093	815	968	50		8.3		33.8	1.8
-2	0.280	0.250	1093	815	977	50		20.1		39.8	3.3
-3	0.250	0.220	1093	815	960	50		20.0		35.2	4.2
-4	0.220	0.175	1093	815	960	50		23.0		34.8	5.2
-5	0.175	0.151	1093	815	963	50		17.2		34.1	4.9
-6	0.151	0.135	1093	815	966	50		15.2		33.5	4.8
-7	0.135	0.115	1093	815	960	50		19.8		35.7	6.0
-8	0.115	0.096	1093	815	960	50		31.4		46.2	6.5
5892-2											
-1	0.305	0.280	1093	815	977	50		9.9		33.9	2.2
-2	0.280	0.255	1093	815	954	50		24.0		51.9	3.0
-3	0.255	0.220	1093	815	963	50		24.0		42.9	4.0
-4	0.220	0.165	1093	815	966	50		22.4		32.3	5.4
-5	0.165	0.158	1093	815	960	50		5.3		23.9	2.0
-6	0.158	0.145	1093	815	966	50		14.4		38.1	3.9
-7	0.145	0.120	1093	815	954	50		27.2		42.5	6.0
-8	0.120	0.098	1093	815	954	50		33.6		46.3	6.8

TABLE A-1 (Continued)

Billet/Sheet Number	Sheet Thickness (in)		Temperature (C)			Roll Speed (fpm)	Separating Force (10^3 lbs/in)			Flow Stress (Ksi)	Strain Rate (sec^{-1})
	Initial	Final	Preheat	Rolls	Final		Front	Middle	Rear		
5893-1											
-1	0.290	0.250	1066	815	965	63	23.6	26.2	28.2	43.3	4.5
-2	0.250	0.200	1066	815	940	63	34.0	35.2	49.6	47.7	5.9
-3	0.200	0.147	1066	815	960	63	28.6	38.4	47.0	45.3	7.5
-4	0.147	0.108	1066	815	960	105	27.2	38.6	19.2	45.1	14.5
-5	0.108	0.097	1066	815	955	105	16.0	24.0	13.2	46.8	10.4
-6	0.097	0.091	1066	815	945	105	9.6	17.6	8.0	45.0	9.9
5893-2											
-1	0.290	0.253	1066	815	927	73	20.8	32.6	9.6	55.7	5.0
-2	0.253	0.205	1066	815	943	73	17.6	38.0	8.8	51.9	6.6
-3	0.205	0.145	1066	815	966	73	19.2	34.6	24.0	38.0	9.2
-4	0.145	0.112	1066	815	954	73	23.4	32.0	11.2	41.3	9.7
-5	0.112	0.096	1066	815	949	73	17.6	23.4	19.6	40.9	8.6
-6	0.096	0.086	1066	815	949	73	10.2	16.6	7.2	38.0	7.6

TABLE A-2. MEASUREMENTS DURING ROLLING OF TITANIUM ALUMINIDE ALLOY, Ti-24Al-11Nb

Billet/Sheet Number	Sheet Thickness (in)		Temperature (C)			Roll Speed (fpm)	Separating Force (10^3 lbs/in)			Flow Stress (Ksi)	Strain Rate (sec^{-1})
	Initial	Final	Preheat	Rolls	Final		Front	Middle	Rear		
4A											
-1	0.516	0.481	1150	815	966	59					
-2	0.481	0.400	1150	815	941	59	11.6	11.6	11.6	17.7	3.6
-3	0.400	0.330	1150	815	921	80	12.6	13.0	13.4	20.2	5.5
-4	0.330	0.257	1150	815	921	80	17.6	17.8	24.2	24.3	6.9
-5	0.257	0.202	1125	815	900	80	23.0	23.2	29.0	32.6	7.7
-6	0.202	0.158	1125	815	900	80	25.6	26.6	31.4	36.9	8.8
-7	0.158	0.131	1125	815	932	80					
-8	0.131	0.124	1125	815	954	80	22.4	22.4	25.6	56.3	3.2
-9	0.124	0.109	1066	815	949	80	21.6	21.8	27.2	42.4	9.1
-10	0.109	0.100	1066	815	943	80	15.4	15.0	20.6	38.3	7.1
4B											
-1	0.522	0.485	1093	815	910	59	12.0	11.0	12.2	26.8	2.2
-2	0.485	0.426	1093	815	843	59	23.4	21.4	25.6	37.2	3.0
-3	0.426	0.355	1093	815	882	80	30.2	31.6	34.4	45.2	5.2
-4	0.355	0.285	1093	815	877	80	34.2	33.6	36.2	44.7	6.2
-5	0.285	0.225	1066	815	893	80	37.6	35.4	40.0	46.1	7.2
-6	0.225	0.176	1066	815	938	80	37.8	36.0	40.6	46.5	8.3
-7	0.176	0.140	1066	815	932	80					
-8	0.140	0.130	1066	815	938	80	33.4	30.4	33.4	62.2	5.8
-9	0.130	0.113	1066	815	943	80	28.8	29.4	34.4	49.7	8.3
-10	0.113	0.106	1066	815	949	80	20.0	22.4	24.6	52.7	6.0

TABLE A-2. (Continued)

Billet/Sheet Number	Sheet Thickness (in)		Temperature (C)			Roll Speed (fpm)	Separating Force (10^3 lbs/in)			Flow Stress (Ksi)	Strain Rate (sec^{-1})
	Initial	Final	Preheat	Rolls	Final		Front	Middle	Rear		
5A											
-1	0.528	0.458	1125	815	950	80	13.4	13.8	15.8	23.3	4.1
-2	0.458	0.389	1125	815	960	80	13.4	14.6	16.0	23.7	4.8
-3	0.389	0.310	1125	815	954	80	18.6	18.6	24.0	25.7	5.7
-4	0.310	0.240	1125	815	960	80	19.2	20.8	20.4	27.9	7.2
-5	0.240	0.160	1125	815	950	80	21.8	25.0	29.8	26.2	10.2
-6	0.160	0.125	1125	815	954	80	23.4	27.2	36.6	37.7	9.9
-7	0.125	0.098	1066	815	960	80	30.2	32.4	36.6	42.2	11.1
-8	0.098	0.080	1066	815	940	80	26.6	29.8	31.4	41.5	11.4
-9	0.080	0.070	1066	815	950	80	19.8	25.0	24.0	41.6	10.3
5B											
-1	0.542	0.492	1066	815	910	59	19.0	19.6	19.6	48.6	2.4
-2	0.492	0.422	1066	815	971	59	23.6	12.6	27.2	21.0	3.2
-3	0.422	0.354	1066	815	860	80	15.8	16.0	20.2	25.3	5.1
-4	0.354	0.280	1066	815	877	80	42.2	40.6	26.0	51.5	6.4
-5	0.280	0.234	1010	815	900	80	43.2	42.0	48.0	59.8	6.3
-6	0.234	0.189	1010	815	900	80	40.3	40.0	46.0	53.6	7.6
-7	0.189	0.151	1010	815	877	80					
-8	0.141	0.129	1010	815	588	80	30.6	33.6	37.8	52.8	8.1
-9	0.129	0.112	1010	815	893	80	29.4	31.4	35.0	51.5	8.3
-10	0.112	0.108	1010	815	888	80	19.4	20.2	25.8	56.1	4.5

TABLE A-3. MEASUREMENTS DURING ROLLING OF THE TITANIUM ALUMINIDE ALLOY, TiAl+Nb+W

Spec. No.	Thickness (inches)		Temperature (C)			Roll Speed (fpa)	Separating Force for TiAl+Cladding (10^3 lbs/in)			Corrected Separating Force for TiAl Only (10^3 lbs/in)	Flow Stress (Ksi)	Strain Rate (sec^{-1})
	Initial	Final	Preheat	Rolls	Final		Front	Middle	Rear	Middle		
5691B												
-1	1.060	0.955	1343	815	1204	59	0.9	4.8	0.9	9.5	15.3	1.8
-2	0.955	0.854	1343	815	1200	59	2.4	5.2	2.0	8.6	13.8	2.0
-3	0.854	0.760	1343	815	1210	59	2.8	5.2	2.4	8.1	13.4	2.2
-4	0.760	0.646	1343	815	1180	59	1.7	5.6	2.9	10.4	14.8	2.7
-5	0.647	0.542	1343	815	1163	59	2.2	5.0	2.2	8.4	12.1	3.1
-6	0.542	0.473	1343	815	1149	59	2.4	5.3	2.4	8.8	15.6	2.9
-7	0.473	0.418	1343	815	1149	59	1.4	3.1	1.4	5.2	10.3	3.0
-8	0.418	0.355	1343	815	1121	59	1.4	4.1	1.4	7.4	12.9	3.7
-9	0.355	0.308	1343	815	1104	59	1.7	4.1	2.0	7.0	13.9	3.7
-10	0.308	0.246	1343	815	1093	59	1.3	4.1	1.3	7.5	11.8	5.0
-11	0.246	0.201	1343	815	1093	59	2.0	4.9	1.6	8.4	14.8	5.3
-12	0.201	0.134	1343	815	1060	59	1.6	4.3	1.6	..6	9.1	8.2

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