AD-777 809

PROCESSING AND BALLISTIC EVALUATION OF DUAL-HARDNESS TITANIUM ARMOR

Roger A. Perkins, et al

Lockheed Missiles and Space Company,
Incorporated

Prepared for:

Army Materials and Mechanics Research Center

April 1974

DISTRIBUTED BY:

NTIS
National Technical Information Service
U. S. DEPARTMENT OF COMMERCE
5285 Port Royal Road, Springfield Va. 22151

ACCESSION for

NTIS	White Section	☑
DDC	Buff Section	☐
UNANNOUNCED		☐

JUSTIFICATION. ....

....

BY...

DISTRIBUTION/AVAILABILITY CODES

Dist.	AVAIL. and/or SPECIAL
A	

1c

REPORT DOCUMENTATION PAGE		READ INSTRUCTIONS BEFORE COMPLETING FORM
1. REPORT NUMBER AMMRC CTR 74-17	2. GOVT ACCESSION NO.	3. RECIPIENT'S CATALOG NUMBER AD 777809
4. TITLE (and Subtitle) PROCESSING AND BALLISTIC EVALUATION OF DUAL-HARDNESS TITANIUM ARMOR		5. TYPE OF REPORT & PERIOD COVERED Final Report August 1972 - February 1974
		6. PERFORMING ORG. REPORT NUMBER
7. AUTHOR(s) Roger A. Perkins and Elliott H. Rennhack		8. CONTRACT OR GRANT NUMBER(s) DAAG46-73-C-0013
9. PERFORMING ORGANIZATION NAME AND ADDRESS Lockheed Palo Alto Research Laboratory Lockheed Missiles & Space Company, Inc. Palo Alto, California 94304		10. PROGRAM ELEMENT, PROJECT, TASK AREA & WORK UNIT NUMBERS D/A Project: 1T062105A328 AMCMS Code: 502E.11.29400 Agency Accession: DA OD4718
11. CONTROLLING OFFICE NAME AND ADDRESS Army Materials and Mechanics Research Center Watertown, Massachusetts 02172		12. REPORT DATE April 1974
		13. NUMBER OF PAGES 67
14. MONITORING AGENCY NAME & ADDRESS(if different from Controlling Office)		15. SECURITY CLASS. (of this report) Unclassified
		15a. DECLASSIFICATION/DOWNGRADING SCHEDULE

16. DISTRIBUTION STATEMENT (of this Report)

Approved for public release; distribution unlimited.

17. DISTRIBUTION STATEMENT (of the abstract entered in Block 20, if different from Report)

18. SUPPLEMENTARY NOTES

Reproduced by
NATIONAL TECHNICAL
INFORMATION SERVICE
U S Department of Commerce
Springfield VA 22151

19. KEY WORDS (Continue on reverse side if necessary and identify by block number)

Armor
Lightweight armor
Composite armor
Dual-hardness armor
Titanium alloys
Fabrication

20. ABSTRACT (Continue on reverse side if necessary and identify by block number)

A program to demonstrate the potential of dual-hardness titanium alloy armor for defeating small-caliber armor-piercing threats has been conducted. The armor consisted of either a Ti-3Si-3Fe-0.5N or a Ti-3Si-2Fe-3Mo-0.75 front-face alloy diffusion/roll-bonded to a Ti-7Al-2.5Mo back-face alloy. The plates were heat treated to a hardness level of R_C 54-59 on the surface of the front face and to R_C 39-43 on the back face. Screening studies of alloy compositions leading to the selection of the two different front-face alloys are discussed. A materials and

1a

process improvement study is described that is designed to improve bond-line strength and overall ballistic performance of the dual-hardness composite. Results of a ballistic evaluation study conducted with 30-cal AP projectiles to determine the effect of variable front-to-rear thickness ratios for areal densities of 6, 8, 10, and 12 lb/ft^2 are presented. Performance in ballistic tests is correlated with alloy hardness, composition, structure, thickness ratios and areal densities to assess the overall potential of the dual-hardness titanium armor. Spall resistance of the hard-face alloy is shown to be governed by alloy composition, processing conditions, and thickness ratio. Excellent spall resistance is demonstrated by the Ti-3Si-3Fe-0.5N alloy homogenized to eliminate segregation of silicides and bonded to yield a final 30:70 front-to-rear thickness ratio. The ballistic limit is governed largely by hardness of the front face, and hardness should be R_c 58 or higher for significantly improved performance compared with homogeneous titanium alloy armor. A recommendation for continuing work on this promising new armor concept is presented.

FOREWORD

This report covers the work performed under Contract No. DAAG 46-73-C-0013, D/A Project 1T062105A328, AMCMS Code 502E.11.29400, Agency Accession DA OD4718. The work was administered by the Army Materials and Mechanics Research Center, Watertown, Massachusetts 02172, with Mr. Dino J. Papetti as Technical Supervisor.

The report was prepared by Roger A. Perkins of the Metallurgy and Composites Laboratory, of the Lockheed Palo Alto Research Laboratory, Lockheed Missiles & Space Company, Inc., Palo Alto, California. Dr. Elliott H. Rennback of the Manufacturing Research Department, Lockheed Missiles & Space Company, Inc., Sunnyvale, California, was coprincipal investigator with Mr. Perkins and assisted in planning and conducting the experimental study. Mr. Earl Montgomery of the Metallurgy and Composites Laboratory was responsible for processing of all materials and for manufacture of the dual-hardness test plates. Metallographic and electron microprobe studies were conducted or supervised by Mr. William C. Coons of the Metallurgy and Composites Laboratory.

CONTENTS

Preceding page blank

ILLUSTRATIONS

TABLES

Section 1

INTRODUCTION

1.1 BACKGROUND

A recently completed program to evaluate the producibility and ballistic characteristics of dual-hardness titanium alloy armor demonstrated technical feasibility for a new armor material (Ref. 1). The hard-faced titanium composite plate defeated 30- and 50-cal AP projectiles by a mechanism similar to that observed in dual-hardness steel. A new class of titanium alloys constaining 3 to 5% Ge, 2 to 3% Si, 2 to 3% Fe or Mn, and 0.5 to 1% N heat treated to hardness levels of Rc 55-62 fractured both 30- and 50-cal AP projectiles on impact. The hard alloy comprised 60% of a roll-bonded composite plate with a Ti-7Al-2.5 Mo rear face. The alloy was sufficiently tough to resist gross fracture and exhibited a good potential for multihit protection. The Ti-7Al-2.5 Mo alloy showed excellent potential for a strong, tough, back face alloy in a dual-hardness composite material. It did not crack, delaminate, or spall on projectile defeat or penetration.

Although the V50 ballistic limit achieved in this initial test was comparatively low (1600 to 1700 fps) for a 14 to 15 lb/ft^2 areal density panel, the results were sufficiently encouraging to warrant further development of a dual-hardness concept for titanium armor. The low ballistic limit appeared to result from a bond line weakness in which the front face separated from the back face on impact at velocities above 1700 fps. Less extensive bond line separation was observed in 10 lb/ft^2 plates with 30:70 to 50:50 front to rear ratio by 30-cal AP projectiles impacted at 2800 fps.

Bond separation is believed to result from an excessive front face thickness and possible bond line embrittlement from interdiffusion of elements across the interface. Considering the low modulus of titanium, the front face most likely should be less than half the total thickness to provide adequate resistance to deformation in the rear face. Ballistic tests at varying thickness ratios are needed to establish the best ratio for a dual-hardness titanium armor.

Bond line embrittlement can be reduced or overcome by the use of diffusion barriers or by improved bonding and rolling practices. Fracture appears to occur in the hard alloy just beyond the bond interface. Diffusion of aluminum into this region from the rear face alloy could reduce toughness. Differential working to produce shear at the interface appeared to be helpful in improving bond strength. This may break up undesirable structures and help to homogenize regions near the interface where interdiffusion has occurred. An evaluation of processing variables is needed to determine the best approach for producing high-strength, tough bonds between the two alloys.

Two other problems need to be addressed for the potential of a dual-hardness titanium armor to be more clearly defined. The first is related to the composition of the hard-face alloy. The alloy contains 3 to 5% germanium, an expensive material in somewhat limited supply. The use of Ge as a hardener could increase the cost of armor by as much as $2 to $3/lb over more conventional titanium alloys. A lower cost material would be preferred for any large-scale application. Other additions such as copper or increased silicon content to achieve high hardness without germanium, or at a lower germanium level, should be considered.

The second problem is related to producibility of the hard-face alloy. Severe cracking was encountered in melting 4 to 6-in.-diameter ingots of 20 to 30 lb weight. This appeared to be the result of an excessive Ge + Si content in some cases, and incomplete solution or segregation of nitrogen (added as Si_3N_4) in others. Alloys with the the lowest Ge + Si + N contents consistent with adequate hardness had the best producibility. The use of master alloys to reduce segregation, particularly of nitrogen, may solve most of the ingot cracking problems. Melting studies are needed to determine the optimum practices for producing sound ingots. Crack-free ingots containing 2% Si, 3% Ge, and 0.5% N have been produced, and the potential for acceptable producibility by balanced composition and improved melting practices appears to be good.

Based on results of the initial study and an assessment of recognized problem areas, further development of the dual-hardness concept for titanium armor is warranted. The probability for the successful development of an improved metallic armor material based on this concept is considered to be favorable.

1.2 OBJECTIVE AND SCOPE

The objective of this program was to demonstrate the potential of dual-hardness titanium armor for defeating a range of ballistic threats from small-caliber armor piercing and fragmented high-velocity projectiles. The armor consisted of either a Ti-3Si-3Fe-O.5N or a Ti-3Si-2Fe-3Mo-0.75N front-face alloy diffusion/roll-bonded to a Ti-7Al-2.5Mo back-face alloy. The component was heat-treated to a hardness level of R_c 57-59 on the front face and R_c 39-43 on the back-face. The program was conducted in two phases: (1) a materials and processes improvement study designed to improve bond line integrity and overall ballistic performance of the composite, and (2) a ballistic evaluation study designed to optimize the front-to-back-face thickness ratios at different areal densities for defeat of a variety of armor-piercing and fragmentation threats.

A range of new alloy compositions was considered for the front-face alloy in the materials and processes study with the aim of reducing or limiting germanium content which was used as a hardening agent in previous work on dual-hardness titanium. Copper and increased silicon were evaluated as substitutes for germanium. Alloy compositions were screened on the basis of hardness and workability to select the two most promising compositions for ballistic evaluation. Twenty-five lb ingots of these alloys (Ti-3Si-3Fe-0.5N and Ti-3Si-2Fe-3Mo-0.75N) were processed to plate, bonded to Ti-7Al-2.5Mo back plates and processed to composite test plates.

The effect of various rolling temperatures on differential rolling, alloy interdiffusion, bond integrity, and surface quality of the composites was assessed. Variations in heat treatment to achieve maximum surface and in-depth hardness were evaluated. Finally, a test panel of each material was evaluated with 30-cal AP projectiles to assess overall ballistic performance. The Ti-3Si-3Fe-0.5N was selected on the basis of all studies and tests as the most promising front-face alloy for dual-hardness titanium armor.

Additional studies were conducted with this alloy to further enhance bond line strength and ballistic performance. The effect of bonding time and temperature, diffusion barriers, and homogenization treatments on bond strength was assessed with 30-cal AP ballistic test panels. Resistance to front-face spall on impact or penetrations by the projectile was used to evaluate performance. Results were correlated with changes in structure in the bond line region. Homogenization of the front-face alloy both before and after bonding and rolling was selected as the best approach for improved spall resistance in the dual-hardness plates.

In the second phase of the program, the optimum front-to-rear thickness ratio was established for different areal densities. Test plates 7 × 7 in. square were prepared at areal densities of 6, 8, 10, and 12 lb/ft^2 with front-to-rear thickness ratios of 30:70, 40:50, and 50:50. These were tested against 30-cal AP threats to establish a V-50 ballistic limit. The 6- and 12-lb/ft^2 plates were prepared from the Ti-3Si-3Fe-0.5N alloy, while the 8- and 10-lb/ft^2 plates were prepared from the Ti-3Si-2Fe-3Mo-0.75N alloy as the front face. The rear face in all cases was Ti-7Al-2.5Mo. Results of ballistic tests were correlated with the hardness, composition, structure, and thickness ratios of the test plates to determine the overall potential of the dual-hardness titanium armor concept.

Section 2

EXPERIMENTAL PROCEDURES

2.1 ALLOY MELTING AND FABRICATION

The hard-face titanium alloys studied in this investigation were arc-melted from sodium-reduced Ti sponge (Reactive Metals Company), < 60 mesh metal powders (Fe, Si, Ge, Cu, Mn, Mo), and either TiN or nitrided Ti sponge. In composition modification studies, small button ingots (50 to 150 gm) were arc-melted in helium by the nonconsumable (tungsten) electrode process. Each button was melted a minimum of four times and was turned over after each melt to aid in homogenization of alloy additions. No problems were encountered with alloy segregation or cracking in any of the experimental button ingots produced. The 50-gm ingots were prepared by the Lockheed Palo Alto Research Laboratories, while the 150-gm ingots were prepared by the Oregon Metallurgical Corporation, Albany, Oregon.

The hard-face alloy ingots used in the investigation were vacuum-arc-melted by the consumable electrode process. All melting was done by the Oregon Metallurgical Corporation. A first melt ingot (0.4-in.-diameter) was prepared from titanium sponge, nitrided titanium sponge (10 to 15% N), and < 60 mesh metal powders (Si, Fe, Mo). This ingot in all cases was very gassy and cracked on cooling to room temperature. Alloy ingredients were highly segregated. The cracked ingot pieces were tack-welded together (heliarc), and a second melt ingot was prepared in a 6- or 8-in.-diameter, water-cooled copper mold. If this ingot cracked, the pieces were tack-welded together and remelted until a sound ingot was produced. Careful hot topping with reduced power inputs and slow cooling with mold water flow reduced or shut off were required to produce crack-free ingots. All finished ingots contained an outer rim of very porous material (similar to rimming steel) with gas holes extending as deep as 1/2-in. below the surface. This material was not removed prior to forging and all surface conditioning was done at the forged-billet stage.

The small button ingots (50- to 150-gm) were rolled directly to sheet from a temperature of 932°C at the Lockheed Palo Alto Research Laboratory. Reductions of 5 to 10%/pass were used in rolling the ingots to the 0.1-in.-thick sheet. The 25- and 50-lb ingots were press-forged to sheet bar prior to rolling. Forging was done by Coulter Steel and Forge, Emeryville, California. The ingots were preheated in a furnace at 400°C for 1 hr, and then were heated to 750°C for 2 hr. They were then transferred to a furnace at 982°C, soaked 1 hr, and draw-forged by pressing with one stroke to a 2- to 3-in.-thick plate. All ingots were coated with Markal paint for oxidation resistance during heating and were rolled in a fiberglass mat before forging to prevent excessive surface chill on contact with forging dies. The forged plates were air-cooled to room temperature.

Press-forged plates were surface-conditioned by grinding (where necessary) and were rolled to the bonding gage (thickness) from 1038 °C. The plates were coated with Markal paint for oxidation protection and were reduced at 10% in thickness/pass on a 2-high 8- × 10-in. rolling mill. The plates were reheated to 1038°C between each pass. The high rolling temperature was selected for breakdown rolling of the hard-face alloys to aid in homogenization and to prevent cracking. Good quality plates. suitable for diffusion bonding with minimum surface conditioning, were produced in all cases.

The back-face alloy plates in all cases were processed from a 2-in. -thick plate previously rolled from a commercial size (1-ton) experiment heat of Ti-7Al-2.5Mo. The plates were rolled from 982°C to final bonding gage at 10% reduction in thickness per pass. Pieces were coated with Markal for rolling and were reheated to 982°C after each pass. All rolling was done by LMSC in a 2-high 8- × 10-in. mill.

Front- and back-face alloy plates were surface-conditioned for bonding by lathe turning the opposing surface flat and parallel. All cracks, checks, pores, and other surface defects were removed from the two surfaces to be bonded together. The outside surfaces were cleaned up to be flat and parallel but were not necessarily free of defects. Carbide tools were used for lathe-turning the hard-face alloys.

2.2 BONDING AND ROLLING OF DUAL HARDNESS PLATES

The front and rear-face alloys were diffusion-bonded in a vacuum hot press to form a composite plate suitable for rolling. A sound metallurgical bond with good interdiffusion of alloy elements was achieved in 1 hr at 760°C with a pressure of 375 lb/in.^{2} in vacuum at 10^{-5} Torr. The plates were pressed between platens of Inconel 718 precoated with an acetone slurry of aluminum oxide to prevent the plates from bonding to the platens. The two platens and plates for bonding were heated by induction using a 15 kW, 9600-cps motor generator set as the power supply. Temperature was measured and controlled by a Pt-Pt-10Rh thermocouple placed in a small hole drilled in one side of the Ti-7Al-2.5Mo back-face alloy plate. The plates were brought to temperature under an alignment holding load of 31 lb/in.^{2} and were held 1 hr at temperature under full load (375 psi). The bonded plates were cooled under full load in a helium atmosphere to room temperature.

The diffusion-bonded composites were press-forged to 2.0-in. -thick sheet bar (if necessary) from 1040°C. Pieces of 2 in. or less were rolled directly without upset foring. (Note: The maximum opening on the 2-high, 8-× 10-in. mill is 1.8-in., allowing for a 10% reduction on the first pass with a 2-in. -thick plate.) All rolling of dual hardness plates was done from a furnace temperature of 982°C using a 10% reduction per pass with a reheat after each pass. The plates were turned 90 deg after each pass to produce a cross-rolled texture as well as to maintain a square shape in the finished plates. Maximum rolled plate size was 9 × 9 in. square. All plates were flattened by cooling from 982°C to room temperature while pressed between flat steel platens at a load of 150 tons. They were grit-blasted to remove oxide scale and were trimmed to 7 × 7 in. square to remove split or delaminated edges.

The test plates were annealed 30 min to 2 hr at 1000°C in air and were water-quenched by tank immersion with rapid agitation. They were fully hardened by a subsequent age for 1 hr at 400°C in air. The hardened plates were grit-blasted to remove oxide scale, and hardness was checked on the front and rear surfaces with a Rockwell (C-scale) tester.

2.3 COMPOSITE EVALUATION

All ballistic evaluations were conducted in test range at the Army Materials and Mechanics Research Center. The plates were impacted with 30-cal AP rounds at varying velocities. Each plate was hit with five rounds, one in the plate center and one in the center of each of four quadrants of the plate. An approximate V-50 ballistic limit was calculated from projectile velocity and penetration behavior. The test plates were held in a steel frame with an aluminum sheet witness plate behind to detect back-face spall. Spent projectiles or fragments that penetrated the armor were collected in a sandbox behind the witness plate. All tests were conducted at 0-deg obliquity from a distance of about 20 ft.

The structure and hardness of test plates through the section and at the bond line was evaluated by metallographic study of small samples prepared from plate edge trimmings. These sections were annealed independent of the test plates and may have a somewhat more rapid cooling rate as a result of smaller sample size. In general, 1- × 3-in. pieces were annealed, and the metallographic sample was sectioned from the center of the piece after it was fully hardened. Hardness through the thickness was measured either with a Rockwell (C-scale) tester or a Leitz Microhardness (DPH) tester at a 500-gm load.

Two different etchants were used to reveal microstructural details. The morphology and distribution of silicide phases were revealed by an electrolytic chromic acid etch. The samples were anodized in a solution of 1% Cr_2O_3 in water at a voltage and a time sufficient to produce a uniform blue film on the surface. This clearly outlined all silicide particles which are not stained by the anodizing treatment. The samples were then polished to remove the anodized film and were etched for about 1 sec in a solution of $7.5HF-2.5HNO_2-90H_2O$. This etch revealed the significant structural features of the matrix. It was not possible to use both the etch and anodized film techniques togehter, and samples had to be evaluated for one microstructural feature or the other.

Section 3

IMPROVEMENT OF MATERIALS AND PROCESSES

3.1 COMPOSITIONAL MODIFICATIONS

In the previous study on dual-hardness titanium, alloys containing 3 to 5% Ge, 2 to 3% Si, 2 to 3% Fe or Mn, and 0.5 to 1.0% N were employed as the hard front face (Ref. 1). These alloys could be hardened to R_C 60–62, had good producibility with respect to melting and fabrication, and appeared to have good potential for dual-hardness armor based on limited ballistic tests. However, at a price of about $114/lb ($250/kg) for Ge suitable for alloying, the addition of 3 to 5% Ge to the alloy would increase the cost by $3.40 to $5.70/lb over that of a conventional Ti base alloy. A 50:50 composite armor would cost a minimum of $1.70 to $2.85/lb more than a germanium-free alloy. Considering processing losses and materials yields, the cost would be considerably higher. The cost of commercial titanium alloys today is too high and limits their use in several high-volume applications. This added cost, if not completely necessary, would severely limit any potential applications for a dual-hardness armor.

An independent study was made of all work leading to the development of the Ti-Ge-Si-Fe(Mn)-N alloys, and additional compositions were tested to assess the possibility of producing a high hardness alloy free of germanium. A summary of pertinent data is presented in Fig. 1.

Ternary alloys of Ti-Ge-N, Ti-Si-N, and Ti-Cu-N all hardened to R_C 50–57 range on water-quenching from 1000°C. None of these alloys exhibited any response to low-temperature aging. The addition of Fe or Mn stablizes the beta phase at the annealing temperature and produces a 2-phase alpha + alpha prime structure on water-quenching which is age-hardenable. Of the quaternary alloys, highest hardness is achieved in the Ti-Si-Fe(Mn)-N system. As shown in Fig. 1, alloys with 3 to 4% Si, 1% N, and 2 to 4 % Fe (or 3 to 4% Mn) could be hardened to R_C 59–60 with a good aging response. Copper and germanium were not particularly effective hardeners in similar quaternary alloys [Ti-Cu(Ge)-Fe(Mn)-N].

The only alloys capable of being hardened to the R_C 60–62 range were those containing both germanium and silicon. As previously established, alloys with 2 to 3% Si, 3 to 7% Ge, 2 to 4% Fe or Mn, and 1% N consistently could be hardened to R_C 60–62 range. Combinations with Cu and Si or Ge, Si, and Cu would not be hardened above R_C 59. Germanium content could not be reduced below 3% if a hardness level above R_C 60 was required.

In view of these findings, the need for hardness levels above R_C 60 in dual-hardness titanium was reassessed. Recent tests with dual-hardness steel indicate

HEAT NO.	COMPOSITION (wt %) Ge	Cu	Si	Fe	Mn	N
	3					1
A7	5					1
A8	10					1
		3				1
A3		5				1
A4		10				1
			2			1
A39			3			1
			4			1
			2		3	1
A41			3		3	1
			4		3	1
A63			3		4	1
A60			3		5	1
A64			3		6	1
A59			3	1		1
A56			3	2		1
A57			3	3		1
A58			3	4		1
	5			2		1
	5			4		1
	5				3	1
	5				7	1
A17	5		2		3	1
A27	5		2		3	1.2
A29	7		2		3	1
A18	3		3		3	1
A21	3		2		3	1.2
A26	3		2		5	1.5
A9		5	2			1
A13		5	2		3	1
A11		5	3		3	1
A14		5	4		3	1
		3	2		3	1
A19		3	3		3	1
A12		3	4		3	1
A16	2	3	2		3	1
A20	1	2	3		3	1
A24	3	3	3		2	1.2

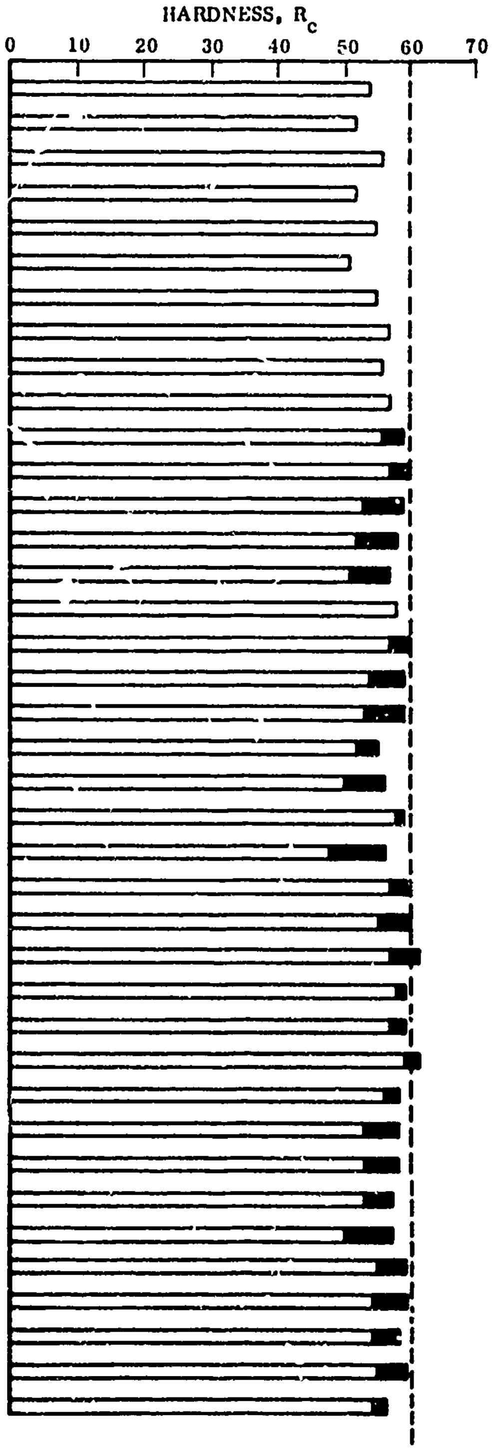


Fig. 1 Effect of Composition on Hardness of Front-Face Alloy (Open Bar – Water-Quenched; Solid Bar – Aged)

that front-face hardness levels in the range R_c 58–60 may be adequate for a projectile breakup defeat mechanism. It is likely that the slight increase in hardness above R_c 60 achieved in titanium by the addition of Ge to the alloy may add little if anything to performance. Considering the large cost penalty involved in achieving this hardness level, it was decided that further work with the Ge-containing alloys was not warranted. A new target of R_c 58–60 was selected for the hard alloy, and the alloy development data were reassessed in the light of this goal.

The basic alloy that will meet this goal is a composition of Ti, 3 to 4% Si, 2 to 4% Fe (Mn), and 1% N, as shown in Fig. 1. The alloy must be balanced in composition to achieve a satisfactory combination of hardness, heat treatment response, and hot workability. Nitrogen has a dual role in the alloy. First, it establishes the base hardness level by interstitial solution hardening. Second, it determines the relative amount of alpha and beta phases present at the annealing temperature. Quench-hardening above the base level is achieved through the martensitic transformation of a silicon-enriched beta phase on rapid cooling. Silicon, in effect, produces a quench-hardening martensite in titanium. The beta stabilizer (Fe or Mn) also controls the amount of beta phase present at the annealing temperature and, in addition, regulates the kinetics of the transformation on cooling. It can stabilize small amounts of beta to room temperature and give an additional hardness increase on aging at low temperature.

The problem of achieving a balanced composition is largely related to the embrittling effects of silicon and nitrogen. As nitrogen is increased above 1%, hot workability is marginal, and the alloys are predominantly alpha phase. Hardening response is poor. At lower nitrogen contents, the base hardness is lower, and large amounts of beta can be retained for quench-hardening. Good quench-hardening, however, requires a high silicon content in the beta phase. Alloys containing over 3% silicon also are difficult to hot-work. However, by adjusting the N and Fe or Mn contents to produce a 50% alpha–50% beta structure at the annealing temperature, a silicon-enriched beta phase with good quench-hardening can be produced. Silicon will partition largely to the beta phase, and with 50% beta in a 3% Si alloy, an effective 5 to 6% Si content can be realized in the beta without destroying hot workability.

As shown in Fig. 1, copper can be used with, but not in place of, silicon in the alloy. Ti-Cu, like Ti-Si, is a fast eutectroid system that will quench-harden the transformed beta phase. Copper, however, is not as effective a hardener as silicon. There was some indication that copper is effective as an age hardener, however. Alloys with Cu + Si containing 3 Mn and 1 N quenched to a lower hardenss value (R_c 50–53) than alloys with Si alone (R_c 55–57) but aged to the same final hardness level (R_c 58–60). The same effect could be achieved by increasing the amount of beta stabilizer (Fe or Mn) above 3%. The lower quenched hardness and larger low temperature againg response might be a distinct advantage with respect to minimizing distortion of the dual-hardness plates during quenching.

Additional studies were made of alloy compositions to achieve the best balance for producibility and performance. A silicon level of 3% was selected as the maximum consistent with good producibility and the minimum consistent with good hardenability. (Excessive cracking was observed in solidification and cooling of small button ingots containing over 3% Si.) In this first series of tests, copper and germanium were added to a 3% Si base to determine their relative effect on producibility. The following three alloys were prepared as 150-gm, arc-melted buttons by Oregon Metallurgical Corporation:

- A41 – Ti-3Si-3Mn-1N
- A42 – Ti-3Si-3Cu-3Mn-1N
- A43 – Ti-3Si-3Ge-3Mn-1N

No problem was encountered in melting the alloys, and Cu and Ge could be used to increase the amount of quench-hardness without promoting excessing cracking. The straight Si alloy (A41), however, had the best hot workability and was rolled to 0.1-in.-thick plate without edge or surface tearing. The copper bearing alloy (A42) had mild edge tearing, and the germanium bearing alloy (A43) had extensive edge tearing on rolling. Both would be marginal or would give low yields on processing. The hardness data for these alloys (Table 1) show that all three harden to R_c 59–60 with the Ge alloy having the highest hardness. The Cu-bearing alloy had the lowest quenched hardness (R_c 53) and the highest hardness increase (6 R_c points) on aging. All three alloys had the same high hardenability as assessed by the difference in hardness for a given section thickness in the air-cooled and aged versus water-quenched and aged conditions. A high ratio of air-cooled to quenched properties signifies a good ability to deep-harden in heavy sections.

Based on these results, a Ti-3Si-3Mn-1N alloy was selected as the best material for further modification. Adjustments in the nitrogen and beta stabilizer contents were made to achieve a proper structural balance for maximum hardness and hardenability. Iron and molybdenum also were evaluated as substitutes for manganese to control hardenability. Molybdenum is particularly attractive for controlling hardenability without the deleterious segregation and embrittlement effects often associated with high iron or manganese contents.

As shown in Table 1, alloys without beta stabilizers have a high hardenability (air-cooled properties 98 to 100% of water-quaneched) but a low hardness and no aging response. A small addition of beta stabilizer gives a small aging response with maximum hardness but with low hardenability (air-cooled properties 86 to 91% of quenched properties). Increasing the beta stabilizer increases the aging response but decreases the maximum hardness. Hardenability, on the other hand, is increased to a high level (98 to 100+%). Iron has a greater effect on increased hardenability than manganese at equal levels of addition, and molybdenum in combination with iron or manganese has a very beneficial effect. Significantly, the required hardeness level (R_c 58–60) can be achieved at nitrogen contents as low as 0.5% with a proper balance of beta stabilizers. The low nitrogen content is particularly desired for good producibility.

Table 1

EFFECT OF BETA STABILIZERS AND NITROGEN ON HARDENABILITY OF Ti-3Si ALLOYS

Heat No.	Nominal Composition (Wt %)							Hardness – Rockwell-C(a)				Hardness Ratios (%)	
	Cu	Ge	Si	Mn	Fe	Mo	N	WQ	WQ+A	AC	AC+A	AC:WQ	AC+A:WQ+A
A41			3	3			1	57	59	53	54	95	92
A42	3		3	3			1	53	59	53	54	100	92
A43		3	3	3			1	57	60	55	56	96	93
A39			3				1	57	57	57	57	100	100
A59			3		1		1	57	59	52	54	91	91
A56			3		2		1	57	60	52	53	91	88
A57			3		3		1	54	59	53	55	98	93
A58			3		4		1	53	59	54	57	102	97
A41			3	3			1	57	59	53	54	93	91
A63			3	4			1	53	59	54	56	102	95
A60			3	5			1	52	58	53	57	102	98
A64			3	6			1	51	57	52	56	102	98
A65			3	3		3	1	50	57	53	58	106	102
A70			3				0.75	53	53	52	53	98	100
A50			3	3			0.75	54	59	49	51	91	86
A68			3	6			0.75	50	57	52	56	104	98
A67			3	3		3	0.75	51	58	53	57	104	98
A76			3		2	3	0.75	53	59	54	57	102	97
A69			3				0.5	49	49	48	48	98	98
A51			3		3		0.5	55	60	49	52	89	87
A49			3	3			0.5	55	59	49	50	89	85
A61			3	5			0.5	51	59	51	54	100	91
A62			3	7			0.5	50	57	50	56	100	98
A66			3	3		3	0.5	51	58	52	55	102	95
A75			3		2		0.5	57	59	48	48	84	81

(a) 0.1-in. thick plate.

WQ – 20 min 1000°C, water quench
AC – 20 min 1000°C, air cool
A – aged 20 min 400°C

Based on an evaluation of the data in Table 1, the following two alloys were selected for more detailed processing studies and ballistic evaluation:

- A51 – Ti-3Si-3Fe-0.5N
- A76 – Ti-3Si-2Fe-3Mo-0.75N

The first alloy will harden to R_c 60, with the air-cooled properties being 87% of the quenched properties. The low hardenability will give a graded hardness decrease through the thickness of heavy sections and may promote better toughness at the bond line. The second alloy can be hardened to R_c 59 with the air-cooled properties 97% of the quenched properties. This alloy should deep-harden in heavy sections on quenching and may give better overall ballistic performance if bond strength is good.

A 50-lb trial ingot of each composition was vacuum arc-melted by Oregon Metallurgical Corporation using the consumable electrode process. The first alloy (A15, Ormet Designation OMC 6553-895-1) was arc-melted in a 6-in.-diameter mold (second melt), and considerable difficulty was encountered with cracking during cooling. The ingot was remelted repeatedly with varying power inputs and cooling rates until a sound crack-free ingot was produced. The conditions that produced a sound ingot were then used to melt the second alloy (A76, Oremet Designation OMC 6553-895-2). This ingot was cast in a 8-in.-diameter mold and melted with ease. No cracking was encountered.

Oremet attributed the problems with the first ingot to inadequate melting practices and concluded that the success with the second alloy resulted from use of proper conditions developed by trial-and-error melting of the first alloy. However, as will be discussed later, the Ti-3Si-3Fe-0.5N alloy continued to present problems in cracking on cooling, even with good melting conditions. Alloy segregation in this composition appears to be a key factor, and repeated melting is necessary to homogenize the ingot before it will cool without cracking.

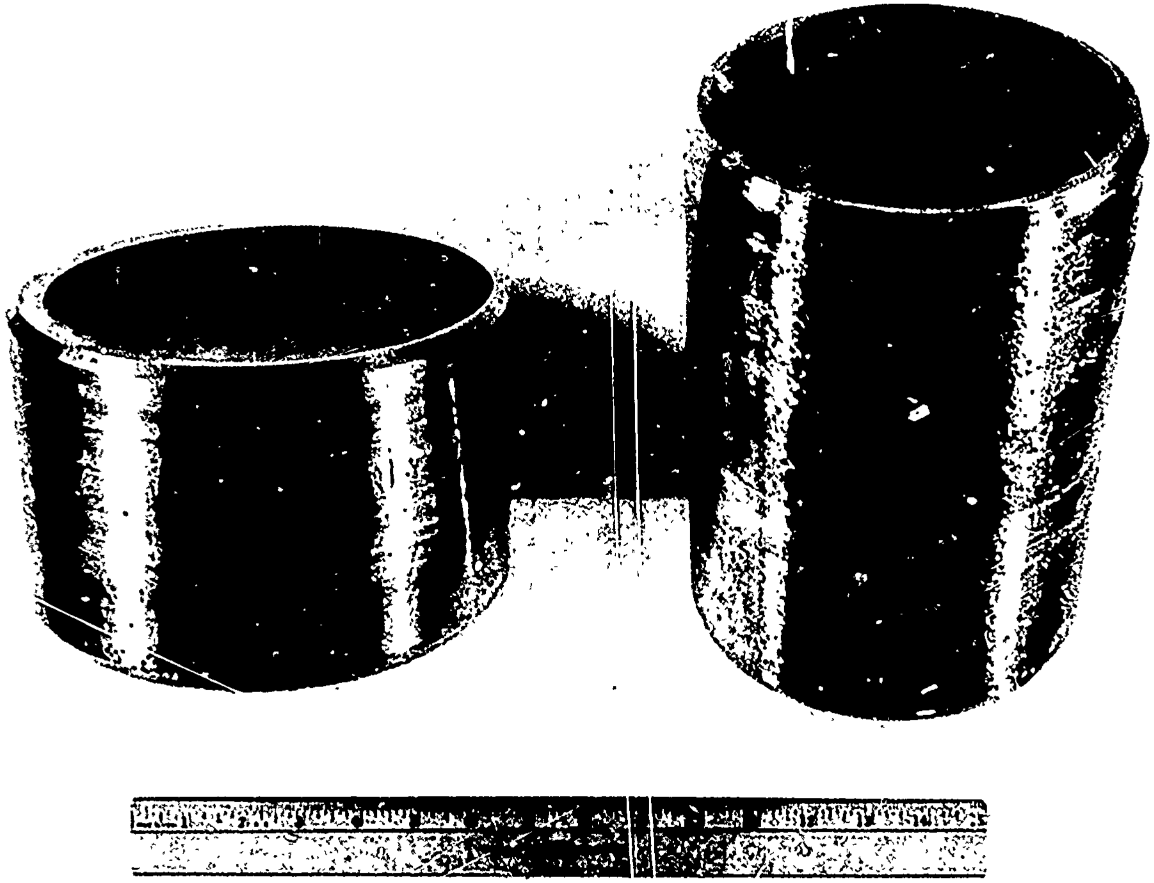
The composition of the two trial ingots produced by Oremet is given in Table 2, and the appearance of the as-cast sound ingots is shown in Fig. 2. A light cut has been taken on the sidewall by lathe-turning to reveal the subsurface porosity. The inots were forged in the condition shown in Fig. 2. Appearance of the 2-in.-thick forged sheet bars is shown in Fig. 3. Both ingots had a large split in the pipe end and a few shallow surface tears associated with regions of excessive porosity. In general, the condition of both alloys was excellent, and over 80% of the forged billets were recovered as useful metal for processing to plate. As shown in Table 2, the actual chemical analyses are close to the aim. The oxygen content is high and reflects the use of Ti sponge that was not screened to remove fine oxygen-rich fraction.

A small piece of each billet was rolled to a 0.5-in. plate and annealed to check hardness levels. As shown in Table 3, the Ti-3Si-3Fe-0.5N alloy had a much lower surface hardness (R_c 57) than originally obtained with annealed and aged 0.1-in.-thick plate (R_c 60). This reflects the low hardenability of this composition and indicates

Table 2

COMPOSITION OF TRIAL ALLOY INGOTS

	OMC-6553-895-1		OMC-6553-895-2	
	Aim wt %	Actual wt %	Aim wt %	Actual wt %
C	–	0.019	–	0.025
O	–	0.34	–	0.41
N	0.50	0.47	0.75	0.74
Fe	3.0	3.08	2.0	1.78
Si	3.0	2.80	3.0	2.83
Mo	–	–	3.0	2.32
	Weight = 47.6 lb		Weight = 37.4 lb	



B3313

× 0.3

895-2
Ti-3Si-2Fe-3Mo-0.75N

895-1
Ti-3Si-3Fe-0.5N

Fig. 2 Appearance of Arc-Melted High Hardness Ti-Alloy Ingots

Fig. 3 Appearance of Press Forged High Hardness Ti-Alloy Ingots

that a minimum hardness (R_c 58) may be difficult to achieve in thick-section armor. The other alloy has a hardness close to that of the thin plate (R_c 58 versus R_c 59) which reflects the good hardenability of this composition. As shown in Table 3. the alloy has uniform hardness through a 0.5-in.-thick section and should provide a high surface hardness of at least the minimum value (R_c 58) on thick armor.

3.2 ROLLING AND DEFORMATION BEHAVIOR

Rolling temperature was believed to be an important factor governing the overall performance of dual-hardness titanium armor. At low temperatures, the Ti-7Al-2.5Mo rear-face alloy is considerably stronger than the front-face alloys, and extensive differential rolling occurs. This rolling could be helpful by improving alloy interdiffusion at the bond line. On the other hand, it could lead to bond separation if insufficient interdiffusion and incomplete development of a good metallurgical bond had occurred in the prior bonding step. The ductility of the hard-face alloy also is reduced at low temperatures, and fractures could be generated in this material. The back-face alloy, on the other hand, will form fine equiaxed alpha-beta structures during work at low temperatures, and toughness of this material may be improved. A rolling temperature that produces the optimum metallurgical condition in each area — front face, bond line, and rear face — must be found for each particular alloy combination.

One dual-hardness composite plate was prepared from each hard-face alloy (895-1 and 895-2) by diffusion bonding a 4 × 4 × 0.75-in. plate to a Ti-7Al-2.5Mo back-face plate of the same size. Each plate was cut into four pieces, each 2 × 2 × 1.5-in. thick. The pieces were press-forged to 1.3 in. thick and rolled to 0.4 in. thick with 10% reduction per pass from 750, 850, 950, and 1050°C. (One piece of each alloy rolled at each temperature.) Samples were sectioned for heat-treatment studies and metallurgical examination of alloy structure and bond-line structure. In addition, thickness ratios and hardness changes across the section were measured. Samples rolled from each temperature were solution-annealed at 950, 1000, 1028, and 1066°C (15-min anneal, water-quench) and aged 30 min at 400°C. Results of all tests and evaluations are summarized in Tables 4, 5, and 6 and Figs. 4 through 11.

High strains are developed in the bond region on rolling at 750° to 850°C. At 750°C, the plates separated at the bond line after several passes, and a composite rolled plate could not be produced. At 850°C, partial separation occurred, and a poor quality plate, bonded only in the center, was obtained.

Excellent quality plates with sound bonds across the whole section were obtained on rolling at 950° to 1050°C. As shown in Table 4, little differential rolling occurs at these temperatures. At 850°C, the hard-face alloy comprises 37 to 40% of the composite as a result of the lower strength of these alloys relative to the Ti-7Al-2.5Mo alloy on rolling at lower temperatures. Differential rolling has a deleterious effect on the bond and should be avoided in processing the composite test plates.

Hardness of the front-face alloy is not particularly sensitive to variations in either rolling or annealing temperature. As shown in Table 5, the hardness at the

Table 3

HARDNESS OF TRIAL ALLOY PLATE

No.	Alloy	Condition	Hardness, R_c(a)	
			Surface	Midsection
895-1	Ti-3Si-3Fe-0.5N	20 min, 950°C, WQ	53	53
		Aged, 20 min, 400°C	57	55
895-2	Ti-3Si-2Fe-3Mo-0.75N	20 min, 950°C, WQ	53	55
		Aged, 20 min, 400°C	58	58

(a) 0.5-in.-thick section.

Table 4

EFFECT OF ROLLING TEMPERATURE ON RATIOS OF FRONT-TO-BACK THICKNESS

Alloy No.(a)	Rolling Temperature (°C)	Thickness (in.)		% Hard Alloy
		Front Face	Rear Face	
1	750	Bond Separation on Rolling		
	850	0.151	0.253	37.5
	950	0.186	0.213	46.6
	1050	0.204	0.200	50.5
2	750	Bond Separation on Rolling		
	850	0.165	0.249	39.8
	950	0.225	0.182	55.3
	1050	0.205	0.190	52.0

(a) Alloy 1 – Ti-3Si-3Fe-0.5N.
2 – Ti-3Si-2Fe-3Mo-0.75N.

Table 5

EFFECT OF ROLLING AND ANNEALING TEMPERATURE ON COMPOSITE PLATE HARDNESS[a]

Alloy[b] No.	Rolling Temperature (°C)	Annealing[c] Temperature (°C)	Front Face		Rear Face
			Surface	Near Interface	
1	850	950	58	57	32.5
	950	950	57.5	57	32
	1050	950	57.5	56	35
	850	1000	59	59	32.5
	950	1000	58.5	58	32
	1050	1000	58	56	35.5
	950	1028	59	59	38
	950	1066	58	54	36.5
2	850	950	56	56	33
	950	950	56	56	32
	1050	950	57	56	33
	850	1000	57	57	35
	950	1000	56	55	31.5
	1050	1000	57	57	34
	950	1028	58	56	39.5
	950	1066	58	56	38.5

(a) 0.4-in.-thick plate, 40–50% hard-face alloy.

(b) Alloy 1 – Ti-3Si-3Fe-0.5N.
2 – Ti-3Si-2Fe-3Mo-0.75N.

(c) Annealed 15 min at temperature, water-quenched, and aged 30 min at 400°C.

surface and near the bond line varies by about ± 1 R_C hardness point for all different rolling and annealing temperatures. No significant trends in hardness changes are indicated. The back-face alloy, on the other hand, hardened significantly on annealing above 1000°C (Table 6). The beta transus for the Ti-7Al-2.5Mo alloy appears to be about 1025°C and quenching from above this temperature produces a hard accicular structure that has significantly reduced toughness.

The hardness data in Table 5 indicate that alloy 895-1 developed a higher surface and in-depth hardness than alloy 895-2. The hardness values are lower than those obtained in earlier studies with thin plate (0.1 in. thick, Table 1), but are similar to those tests on thick plate (0.5 in. thick, Table 3). Alloy 895-2 did not show as good a hardening response in thick plate (0.4 in. thick) in these later tests and appeared to be more sensitive to variations in cooling rate resulting from section size effects. Both alloys showed a good depth of hardening as bonded composite plates.

The effect of rolling temperatures on alloy structures is shown in Figs. 4 through 7. On rolling from 1050°C, silicon is taken into solution in the hard-face alloy and precipitates on cooling as fine particles in prior beta grains and grain boundaries. Although this did not increase hardness significantly, it could have a deleterious effect on toughness. Similarly, rolling from 1050°C produces a coarse accicular structure in the back-face alloy that could have reduced toughness. As previously shown, fine accicular (martensite) structures produced by quenching from above the beta transus have significantly reduced toughness (Table 6). The only significant effect of increased rolling temperatures in the range of 850° to 950°C is a coarsening of the alpha-beta structures in both the front- and rear-face alloys (Figs. 6 and 7). This should not have any deleterious effect on toughness or ballistic performance.

Increased rolling temperatures do have a significant effect on bond-line composition and structure. As shown in Figs. 8, 9, and 10, significant increases in the interdiffusion of alloy constituents occur with increased rolling temperatures. This behavior is summarized in Fig. 11. The width of the diffusion zone (depletion plus enrichment zones) increases from about 33 μ for iron at 850°C to 95 μ at 1050°C (rolling temperature). Similarly, the interdiffusion zone for aluminum increases from 17 μ at 850°C to 101 μ at 1050°C. Silicon shows much less interdiffusion with a maximum penetration of 28 μ at 1050°C.

The excessive diffusion of aluminum into the hard-face alloy on rolling at 1050°C is not desirable. Alloying studies in the development of these compositions revealed that small additions of Al severely embrittle the nitrogen-rich Ti-Si-Fe(Mn)-N alloys. The structure of the bond region of plates rolled from 1050°C also is not particularly attractive, as shown in Figs. 7 and 8. The coarse alpha-beta accicular structures formed in this region by excessive interdiffusion of Al, Fe, and N could result in reduced toughness in the bond-line region. The structures of alloys rolled from 850° and 950°C appear to be more desirable.

3.3 ALLOY SELECTION

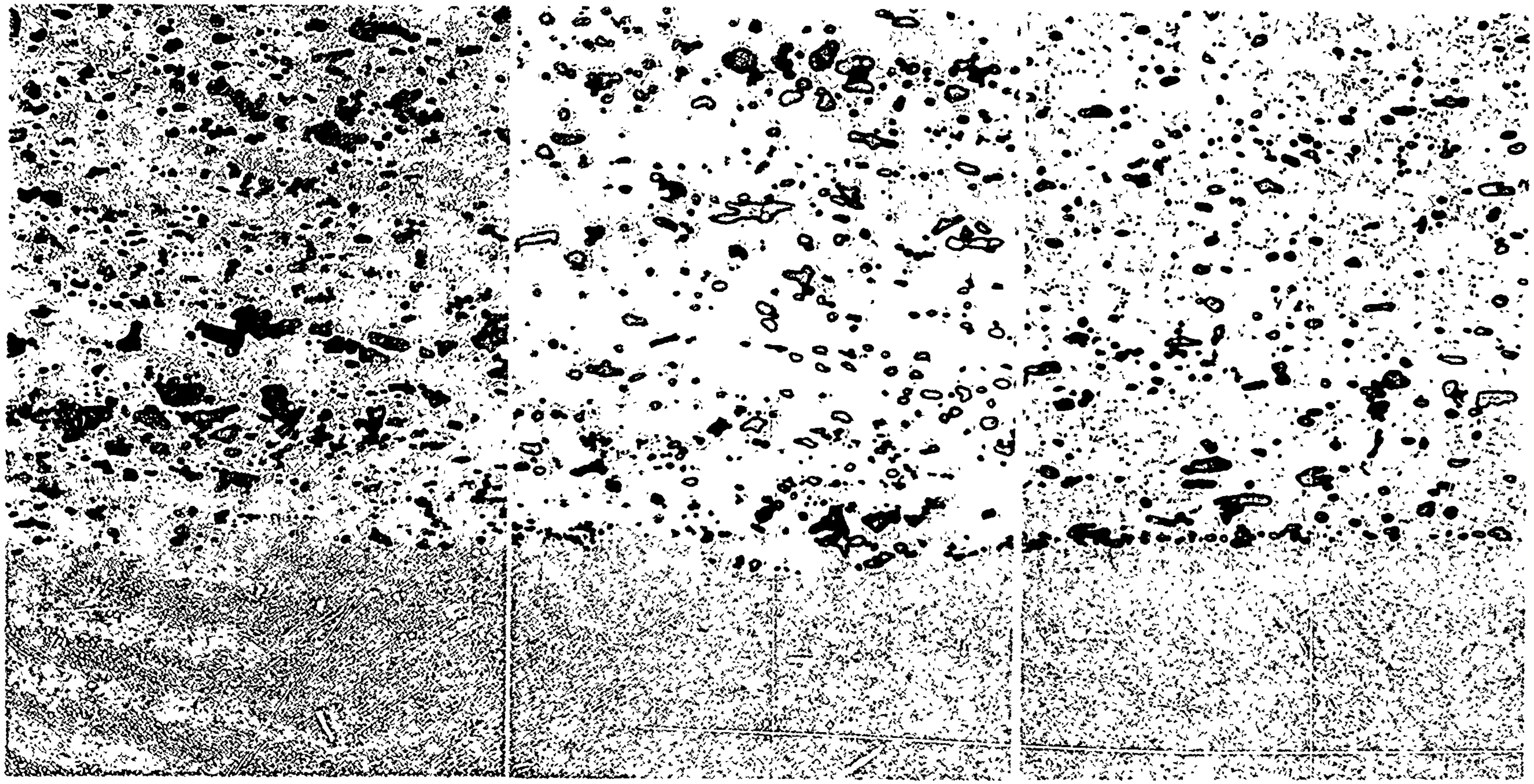
Ballistic tests with 30-cal AP projectiles were used as a final test for the two candidate hard-face alloys. Test plates 7 × 7 × 0.4-in. thick with a 50:50 ratio at

Table 6

EFFECT OF ANNEALING TEMPERATURE ON HARDNESS AND TOUGHNESS OF Ti-7Al-2.5Mo ALLOYS

Solution Temperature (°C)	Hardness, R_c	Impact Toughness[(a)] (ft-lb)
898	32–35	120–122
952	34–36	120–121
1002	41–43	72–109
1050	42–44	57–68
1101	44–46	20–28
1156	46–47	20–23

(a) 3/8 × 3/8 × 2-in. bar, unnotched, room temperature.

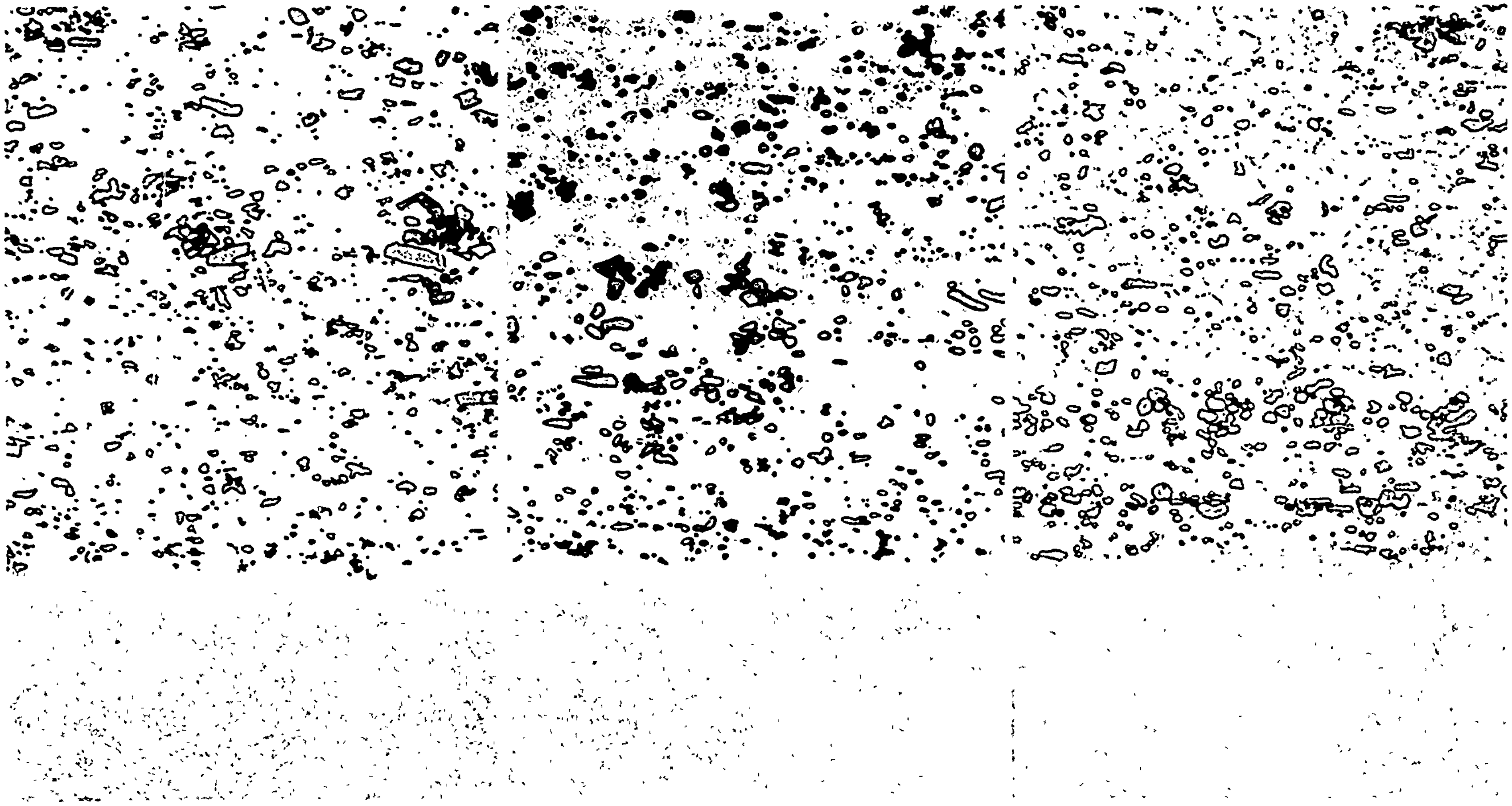


B5886 ROLLED FROM 850°C

B5896 ROLLED FROM 950°C

B5902 ROLLED FROM 1050°C

Fig. 4 Effect of Rolling Temperature on Silicide Distribution in Dual Hardness Plate, Ti-3Si-3Fe-0.5N (895-1) on Ti-7Al-2.5Mo (× 400)

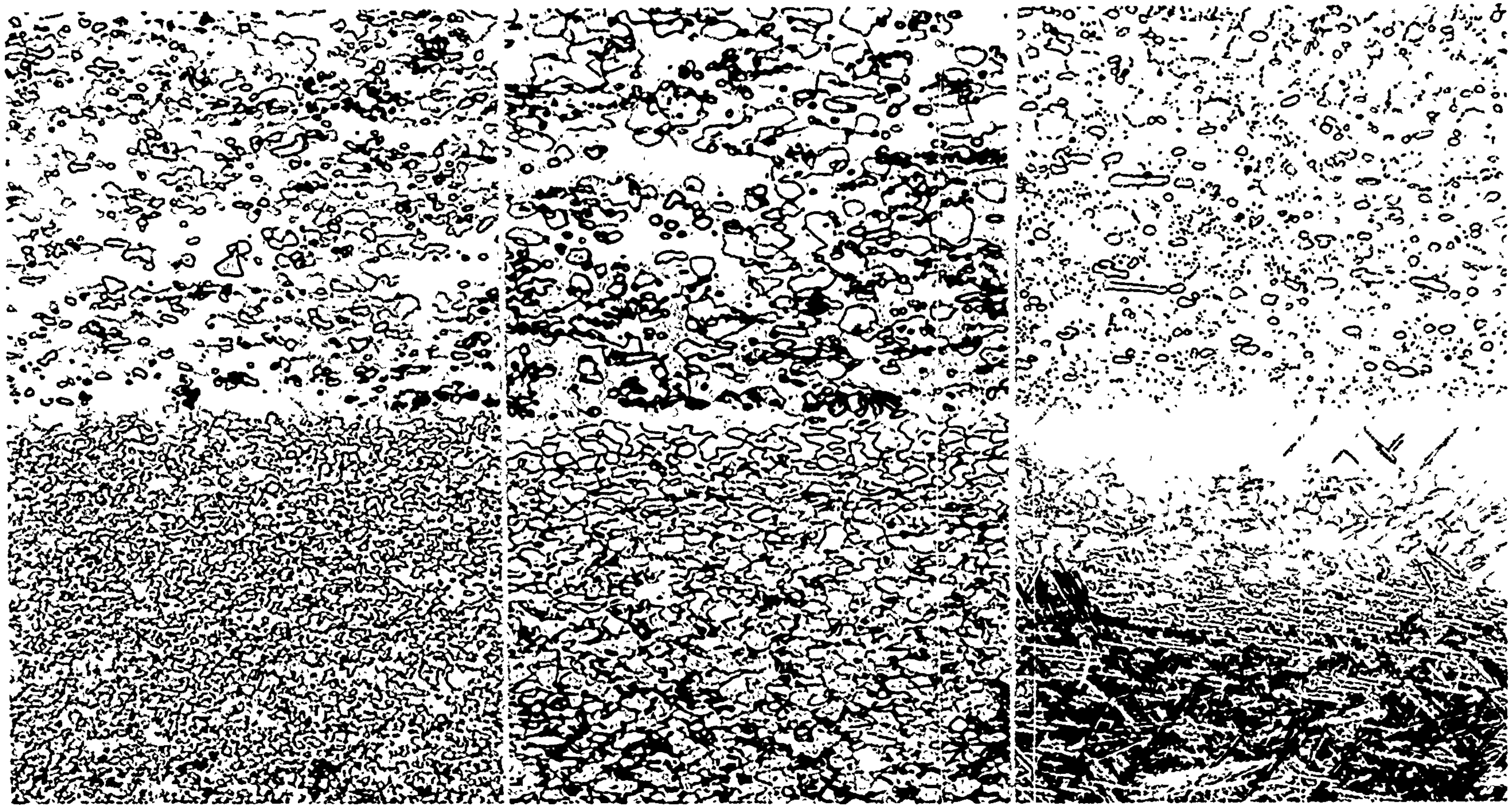


B5894 ROLLED FROM 850°C

B5899 ROLLED FROM 950°C

B5900 ROLLED FROM 1050°C

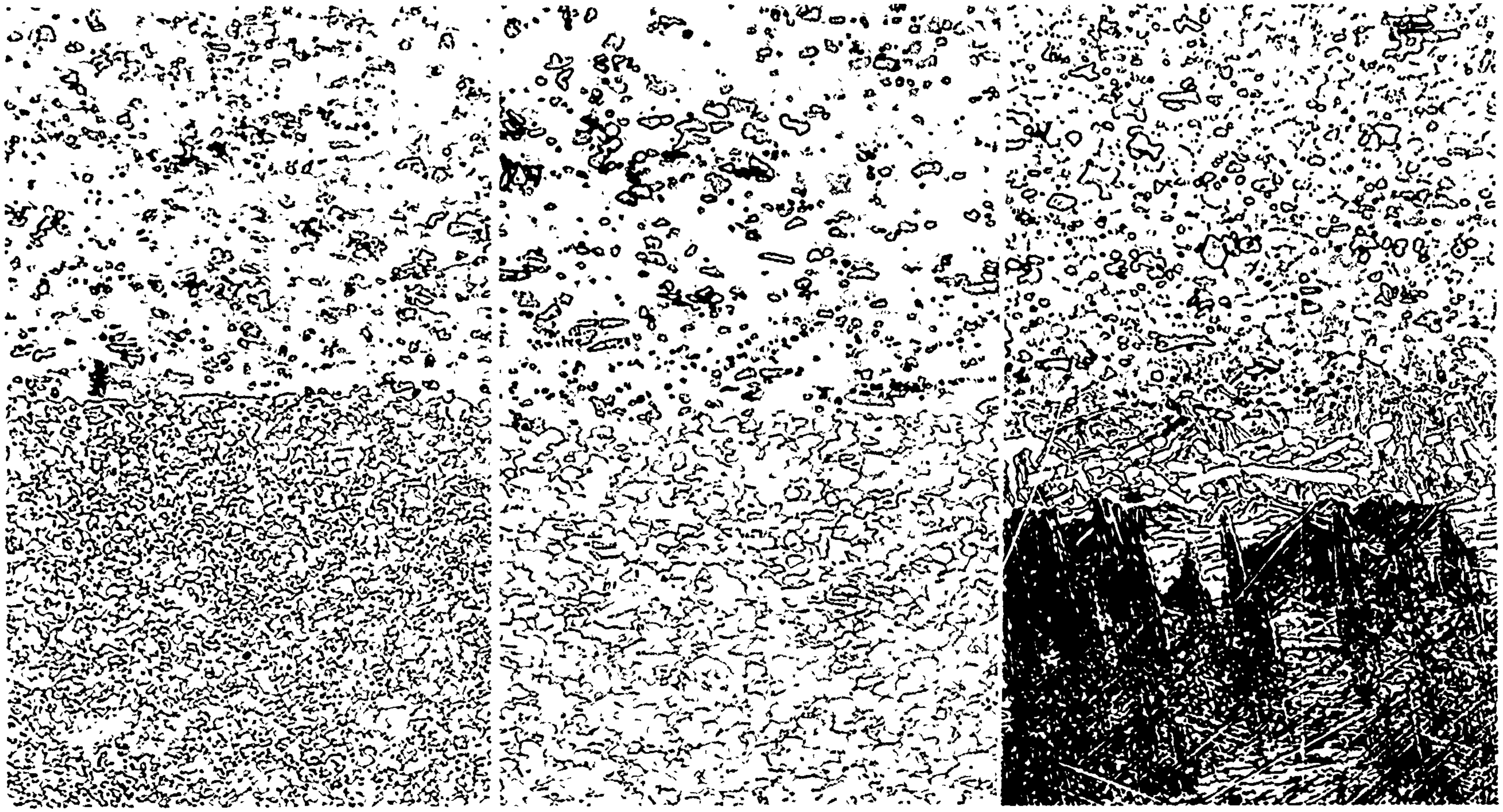
Fig. 5 Effect of Rolling Temperature on Silicide Distribution in Dual Hardness Plate, Ti-3Si-2Fe-3Mo-0.75N (895-2) on Ti-7Al-2.5Mo (× 400)



B5935 B5937 B5940

ROLLED FROM 850°C ROLLED FROM 950°C ROLLED FROM 1050°C

Fig. 6 Effect of Rolling Temperature on Alloy Structure of Dual Hardness Plate, Ti-3Si-3Fe-0.5N (895-1) on Ti-7Al-2.5Mo (× 400)

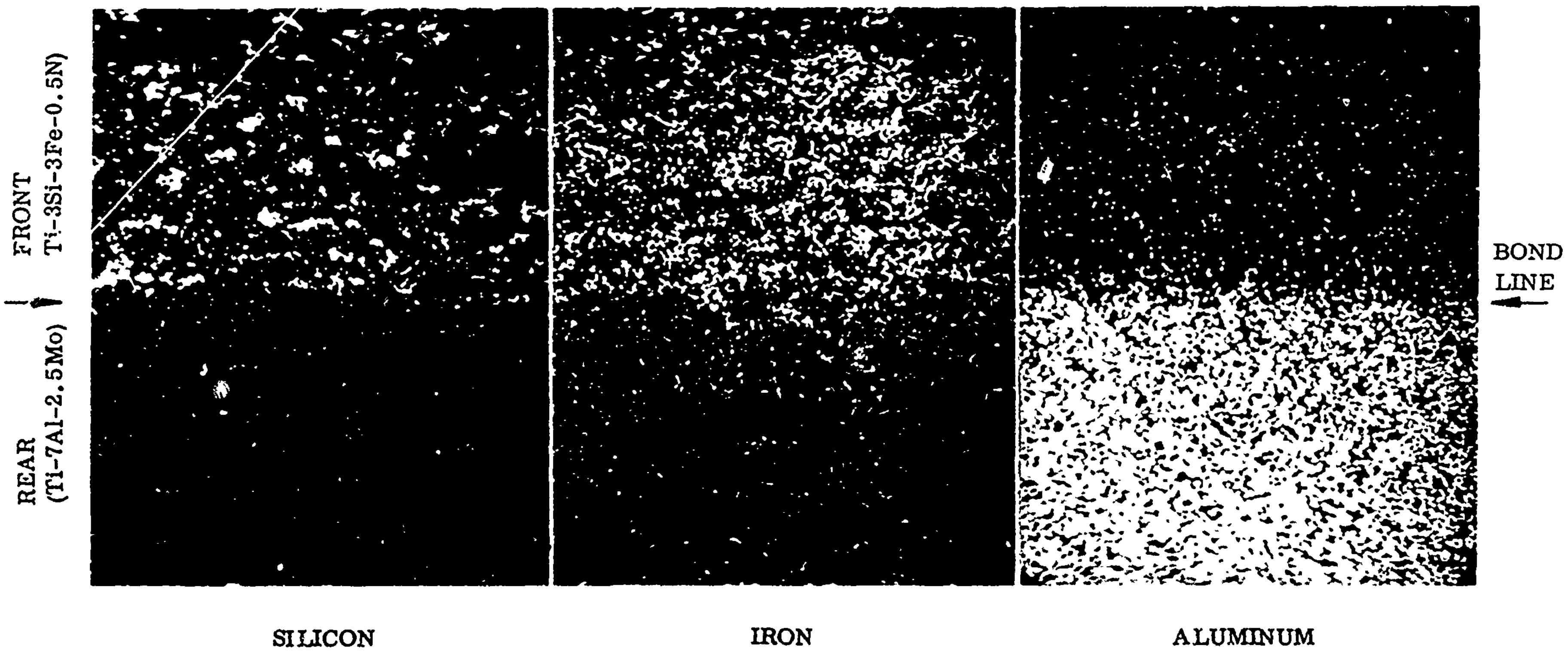


B5936 ROLLED FROM 850° C

B5938 ROLLED FROM 950° C

B5941 ROLLED FROM 1050° C

Fig. 7 Effect of Rolling Temperature on Alloy Structure of Dual Hardness Plate, Ti-3Si-2Fe-3Mo-0.75N (895-2) on Ti-7Al-2.5Mo (× 400)



ROLLED FROM 850°C

Fig. 8 Distribution of Elements at Bond Line, Low Temperature Rolling (× 500)

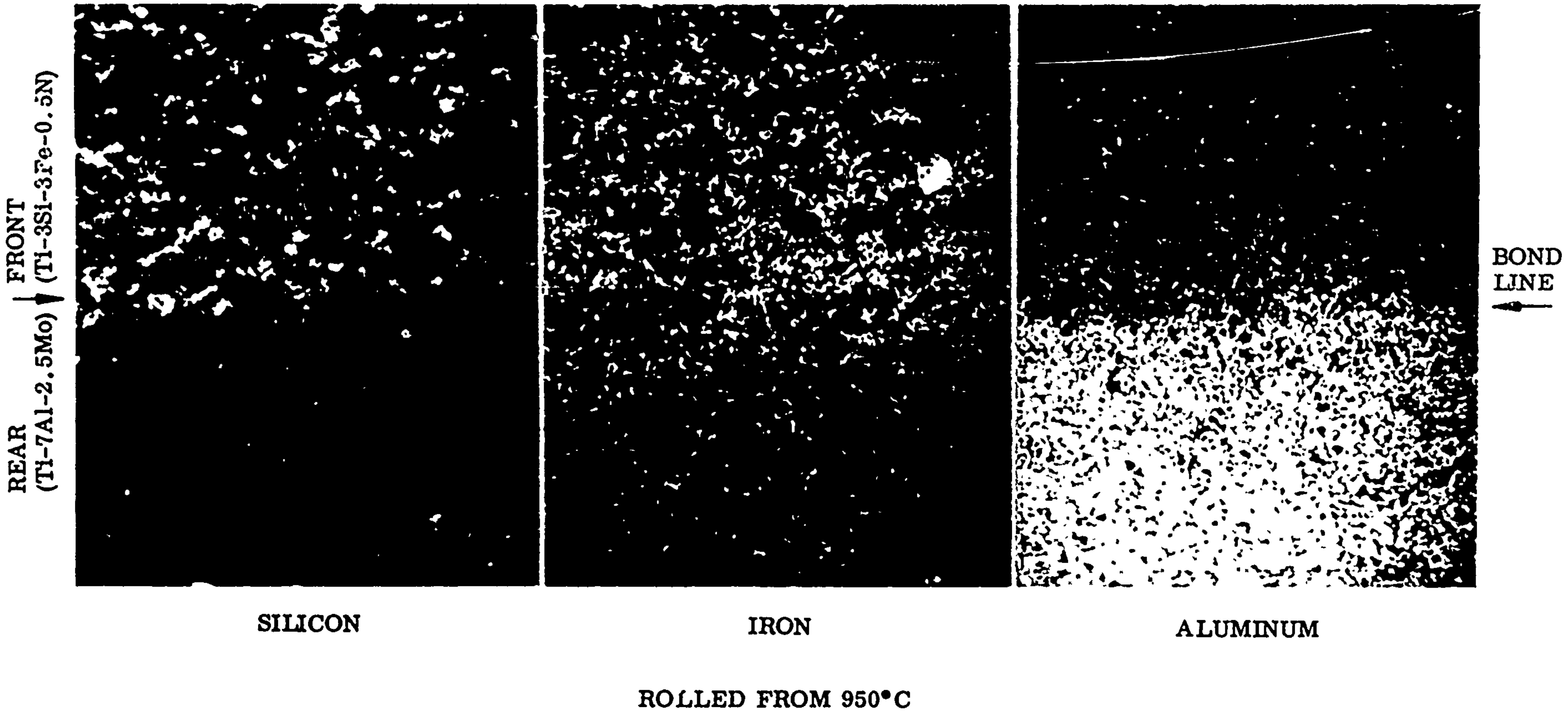
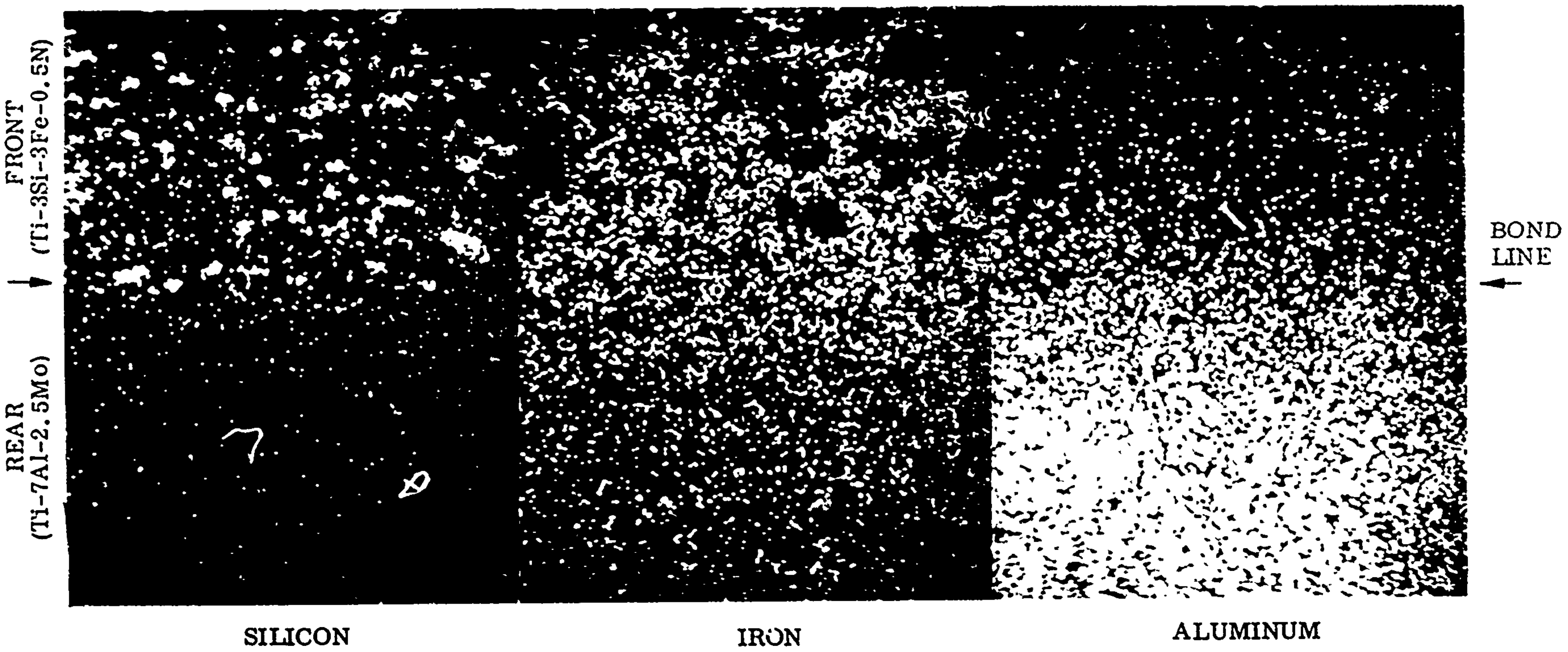


Fig. 9 Distribution of Elements at Bond Line, Intermediate Temperature Rolling (× 500)



ROLLED FROM 1050° C

Fig. 10 Distribution of Elements at Bond Line, High Temperature Rolling (× 500)

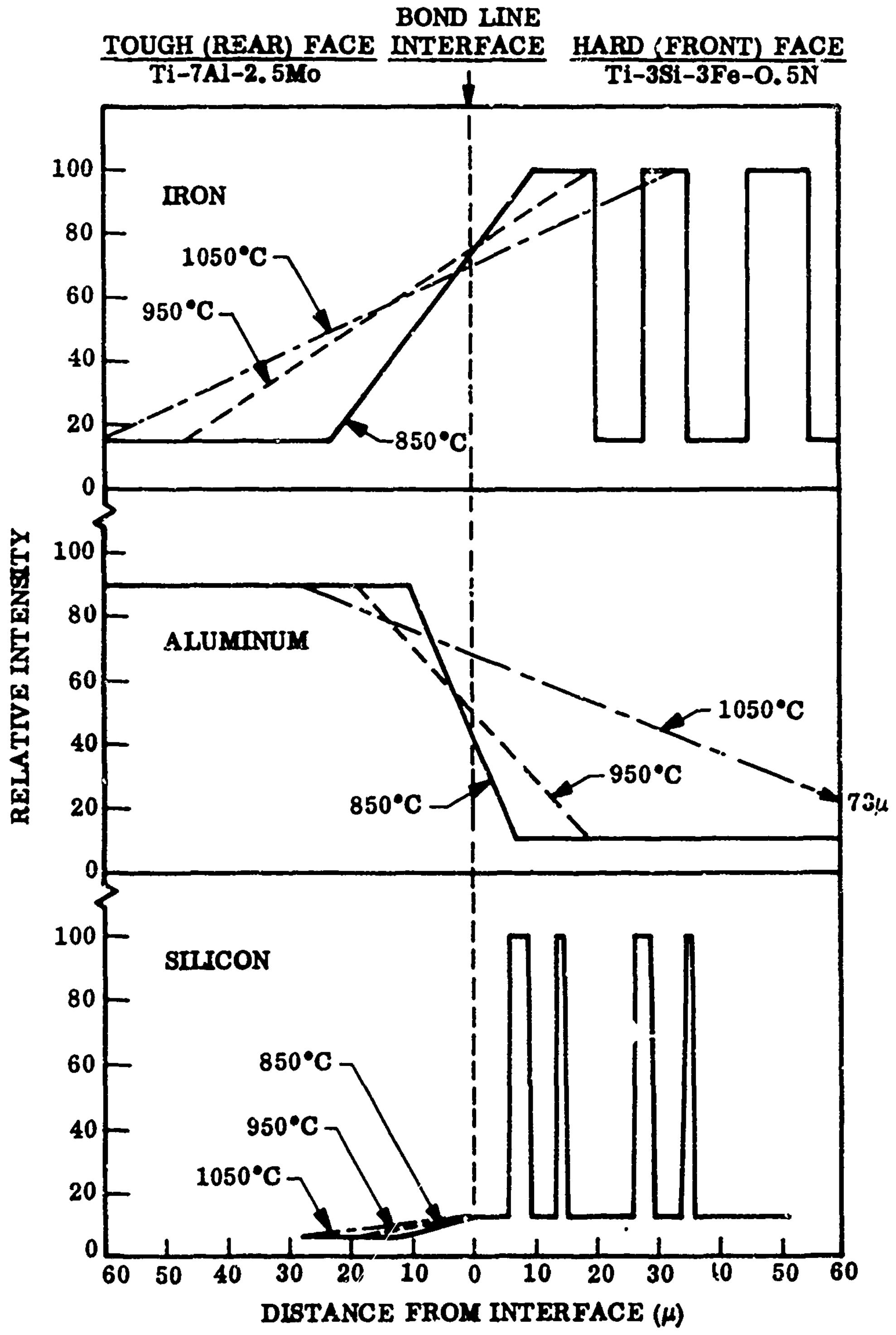


Fig. 11 Effect of Rolling Temperature on Interdiffusion of Elements Across Bond Line

9 lb/ft^2 were impacted with five 30-cal AP projectiles at different velocities. Results are summarized in Table 7, and the appearance of the fired test plates is shown in Fig. 12. Microstructures in the bond zone are shown in Fig. 13. The V-50 limit for both plates was in the same range, and no significent difference in ballistic performance was noted between the two different alloys. The established V-50 limits are classified and can be obtained from AMMRC with appropriate clearance and need to know. The major difference in behavior was in resistance to front-face spall. Alloy 895-1 had the best resistance to front spall and 50 to 60% of the hard face was intact after 5 hits. Alloy 895-2 had poor resistance to spall, and less than 5% of the front face was intact after 5 hits.

A poor bond strength and low surface hardness is believed to have contributed to low ballistic limits for these plates. Plate 895-1 had a surface hardness of R_c 56 which is considerably less than the R_c 59-60 range found for this alloy in thin sections. Plate 895-2 had a higher hardness, R_c 58, but had very poor bond strength. As shown in Fig. 13, silicide segregation existed at the bond line in this test plate. This probably was a factor in the low resistance to front-face spall.

An evaluation of hardness, hardenability, structure, hot workability, and ballistic performance indicated that alloy 895-1 (Ti-3Si-3Fe-0.5N) had the best overall potential for a hard-face alloy material. Considerable weight was given to its resistance to front-face spall in ballistic tests and to its excellent hot workability in reaching this decision. The graded hardness through the section was believed to be an important factor contributing to the spall resistance. The low hardenability of this alloy, however, will limit the maximum hardness that can be achieved in heavy sections.

3.4 BOND STRENGTH ENHANCEMENT

Front-face spall on ballistic impact is a major concern in the development of dual-hardness armor. Spall in the dual-hardness titanium appeared to be related to a weakness in the hard-face alloy just above the bond line. This was considered to be due either to incomplete bonding that initiated cracks in the bond region or to local embrittlement as a result of interdiffusion of alloy elements. As shown in Fig. 11, aluminum from the back face diffuses to a considerable depth into the front face, and aluminum is known to embrittle the alloy. Also, nitrogen, iron, and silicon from the hard face diffuse into the back-face alloy, leaving a deploted zone on one side and an enriched zone on the other side of the bond.

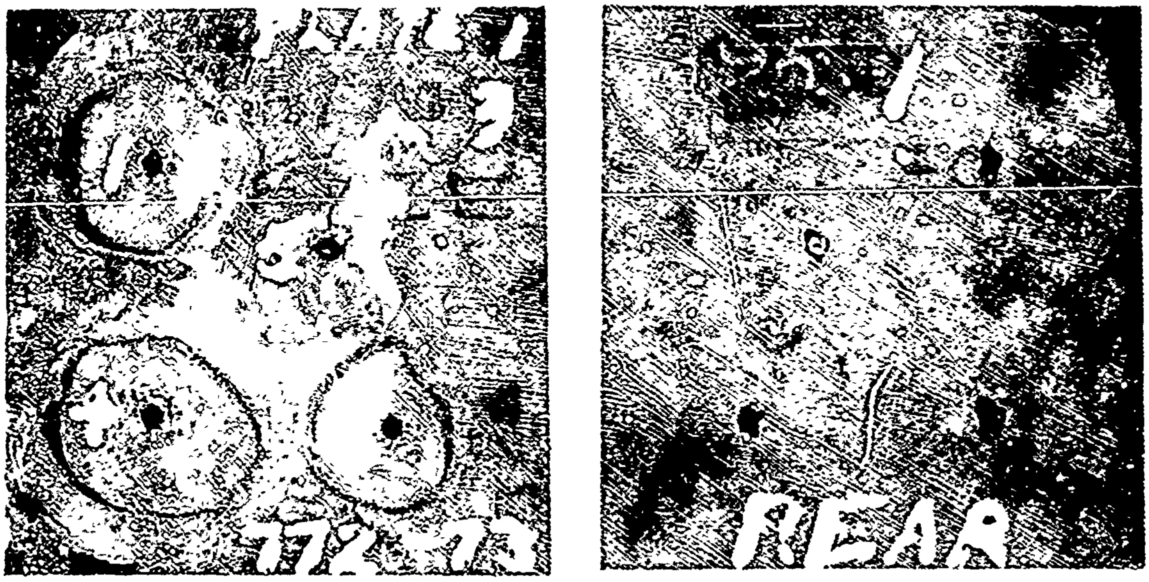
Six ballistic test plates, each 7 × 7 × 0.4-in. of 50% hard and 50% soft dual-hardness titanium armor, were produced for tests. The plates had a Ti-3Fe-3Si-0.5N

Table 7

RESULTS OF BALLISTIC TESTS FOR ALLOY SELECTION[a]

Plate No.	Hard-Face Alloy	Hit No.	Penetration Rating	Front-Face Spall
895-1	Ti-3Si-3Fe-0.5N	1	Complete	1-1/2-in. dia.
		2	Complete	1-1/2-in. dia.
		3	Partial	No spall
		4	Complete	1-1/2- to 2 in. dia.
		5	Complete[b]	1-/2-in. dia.
895-2	Ti-3Si-2Fe-3Mo-0.75N	1	Partial	2-1/2- to 3-in. dia.
		2	Partial	25% front face
		3	Partial	No spall
		4	Complete	Extensive
		5	Complete	Extensive

(a) 7 × 7 × 0.4-in. plate, 9-lb/ft^2, 30-cal AP.
(b) Projectile tip fractured.



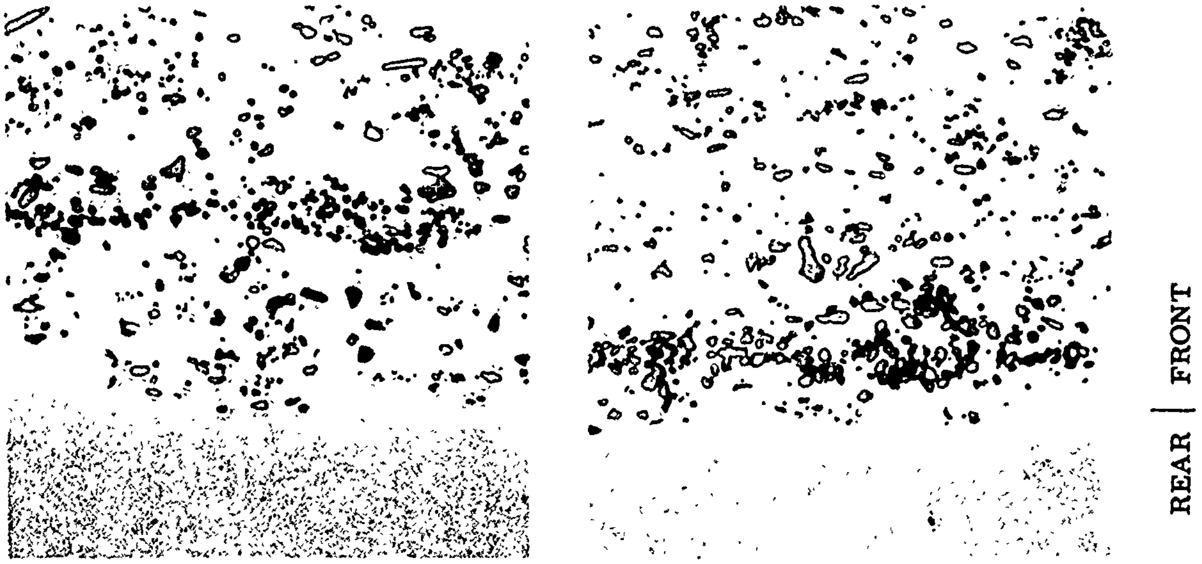
FRONT REAR

TEST PLATE 895-1

FRONT REAR

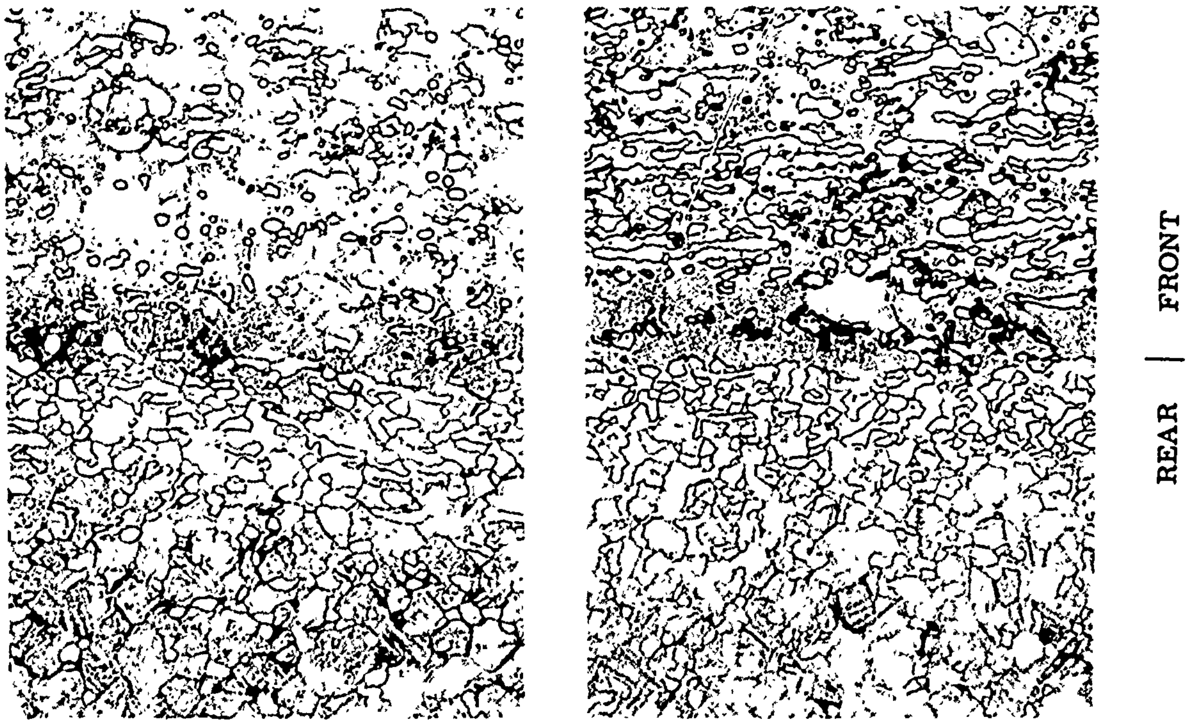
TEST PLATE 895-2

Fig. 12 Alloy Evaluation Test Plates, 9 lb/ft^2, 30-cal AP, (× 0.5)



B5909 B5911

DISTRIBUTION OF SILICIDES



B5942 B5943

ALLOY STRUCTURES

895-1
Ti-3Si-3Fe-05N (FRONT)
Ti-7Al-2.5Mo (REAR)

895-2
Ti-3Si-2Fe-3Mo-0.75N (FRONT)
Ti-7Al-2.5Mo (REAR)

Fig. 13 Structure of Alloy Evaluation Test Plates (× 400) alloy front face hardenable to R_c 59 and a Ti-7Al-2.5Mo back face. The starting 4 × 4 × 0.75-in. thick plates were bonded with the following conditions:

31-1: 760°C-4 hr bond.
31-2: 870°C-1 hr bond.
31-3: Hard-alloy plate homogenized 24 hr at 1200°C in vacuum prior to bonding. Normal 1 hr-760°C diffusion bond.
31-4: 5-mil-thick Ti diffusion barrier, 1 hr-1400°F bond.
31-5: Normal 1 hr-760°C bond with a subsequent 2 hr-1000°C diffusion anneal after rolling to final thickness.
31-6: 10-mil-thick Mo diffusion barrier, 1 hr-1400°F bond.

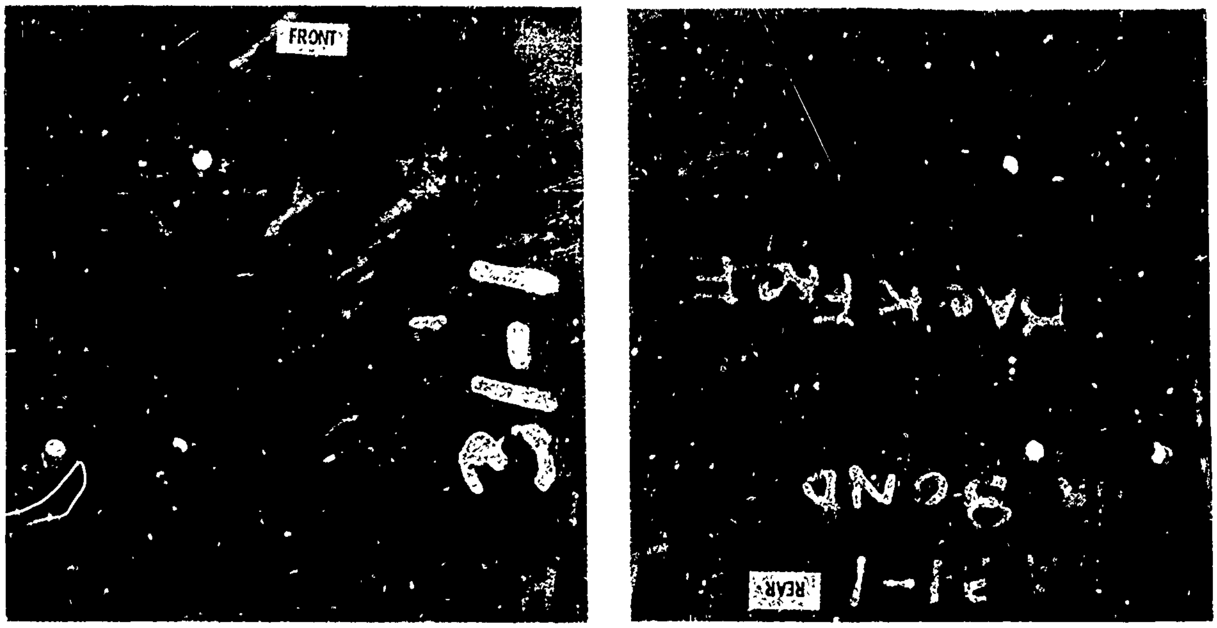
The purpose of these studies is to determine the effect of varying degrees of interdiffusion across the bond line on the tendency for bond-line separation on ballistic impact.

All plates were processed to final gage by the same schedule – 10% reduction per pass from 982°C. Plate 31-6 with the Mo diffusion barrier separated at the bond line and was dropped from the study. Five good-quality ballistic test plates were produced from the other materials. Bonds on all plates appeared to be excellent. Each plate was impacted with 5 rounds of 30-cal AP ammunition to establish an approximate V-50 limit. Ballistic performance and spall resistance were correlated with variations in bonding procedures and structures in the bond zone. Results are presented in Figs. 14 through 20 and Table 8. The V-50 limits are classified and can be obtained from AMMRC with appropriate clearance and need to know.

The highest ballistic limit was developed in the plate where the front-face alloy had been homogenized 24 hr at 1200°C prior to bonding to the back face. This plate also had the best resistance to surface spall (Fig. 15) with only one small area out of four areas where projectiles penetrated both plates. The plate that was homogenized 2 hr at 982°C after bonding and rolling also showed good spall resistance but had a ballistic limit about 100 fps less than that of the high-temperature homogenized plates.

All of the plates had a ballistic limit within ± 70 fps of the average value and no large variation with changes in processing conditions was found. The major effect of bonding variables was on front-face spall resistance. The plate with the titanium diffusion layer had the poorest performance and over half the front spalled in five hits (Fig. 16).

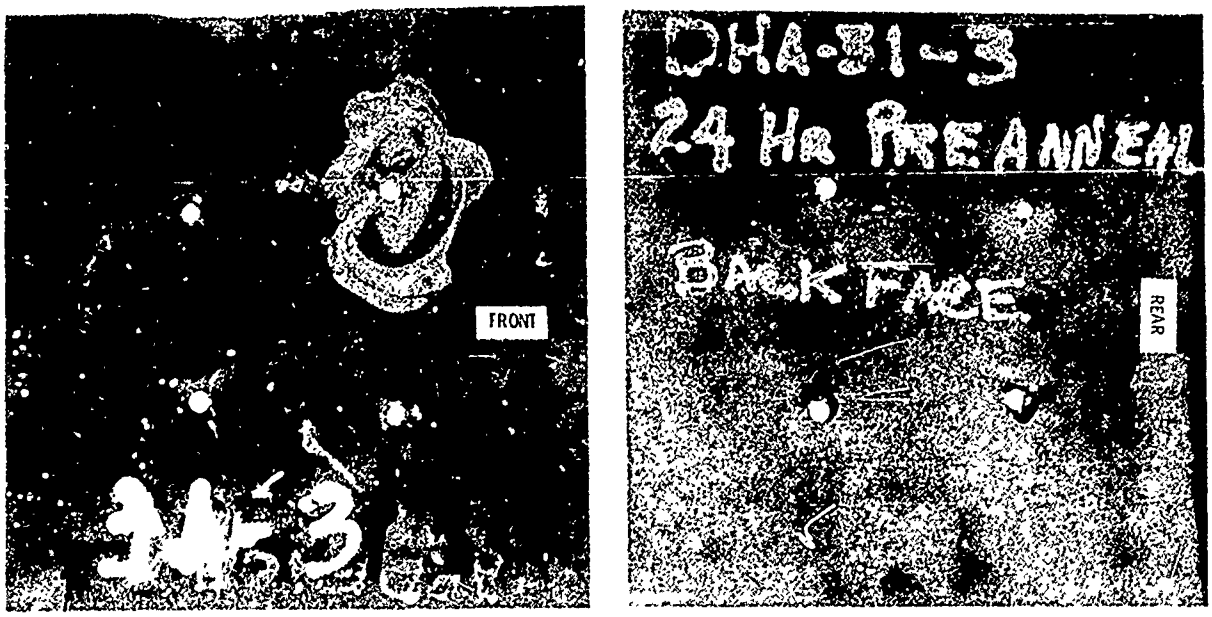
Microstructures at the bond line indicate that segregation of silicides in the hard alloy may be largely responsible for front-face spalling. As shown in Figs. 17, 18, and 20, silicides tend to occur as banded segregates parallel to the plate surface, particularly near the bond line. These produce planes of weakness in the alloy where fractures can initiate and propagate readily on ballistic impact. The 24 hr-1200°C homogenization treatment coarsens the silicide particles and eliminates banded segregates (Figs. 18 and 20). The darkened particles near the interface in these photos are silicides that have been modified by interdiffusion with the substrate. It is not known what change in composition has occurred to alter their etching behavior in this way.



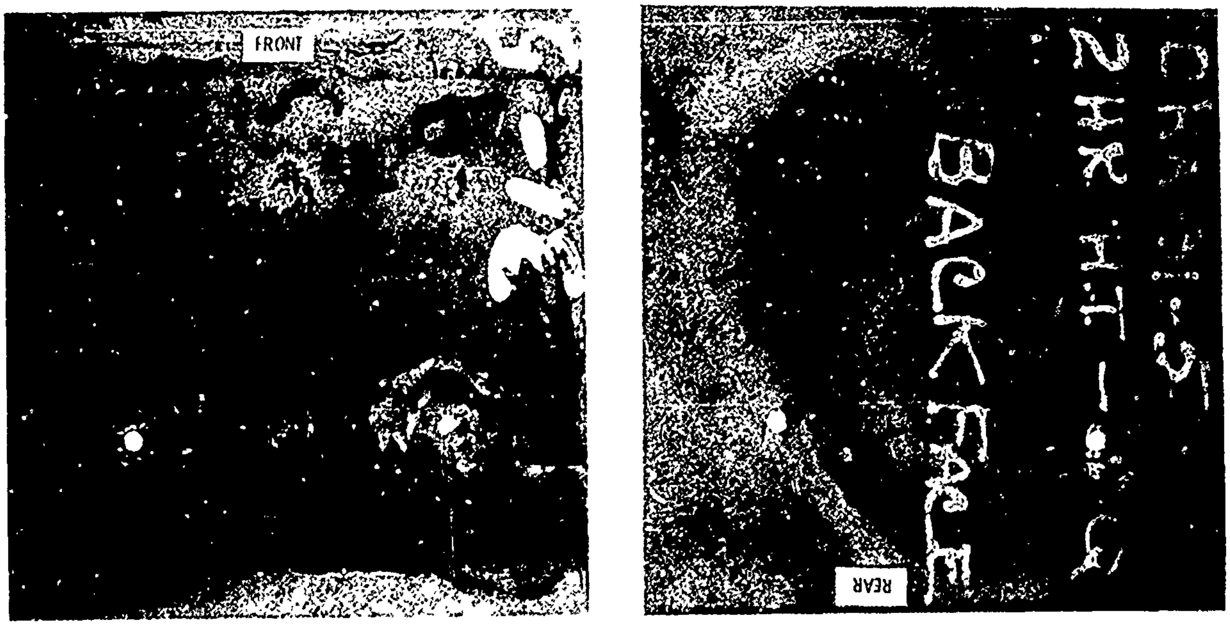
TEST PLATE 31-1
4-HR-760°C BOND

TEST PLATE 31-2
1-HR-870°C BOND

Fig. 14 Bond Evaluation Test Plates, 9 lb/ft^2, 30-cal, AP, (× 0.5)

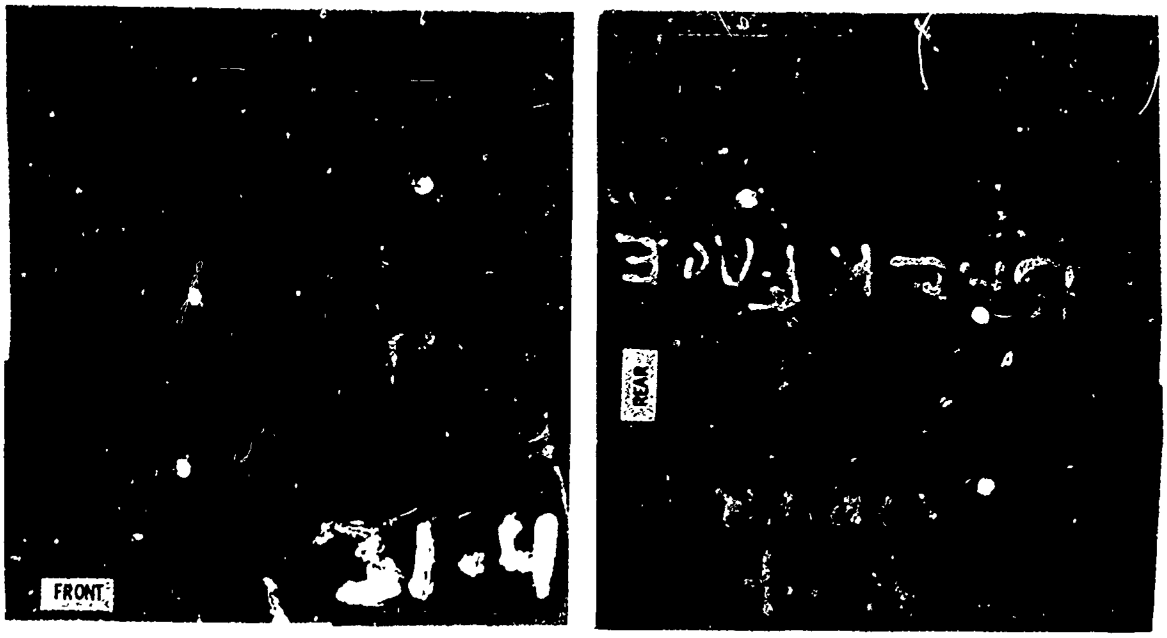


TEST PLATE 31-3
24-HR-1200°C HARD ALLOY HOMOGENIZED



TEST PLATE 31-5
2-HR-1000°C TEST PLATE HOMOGENIZED

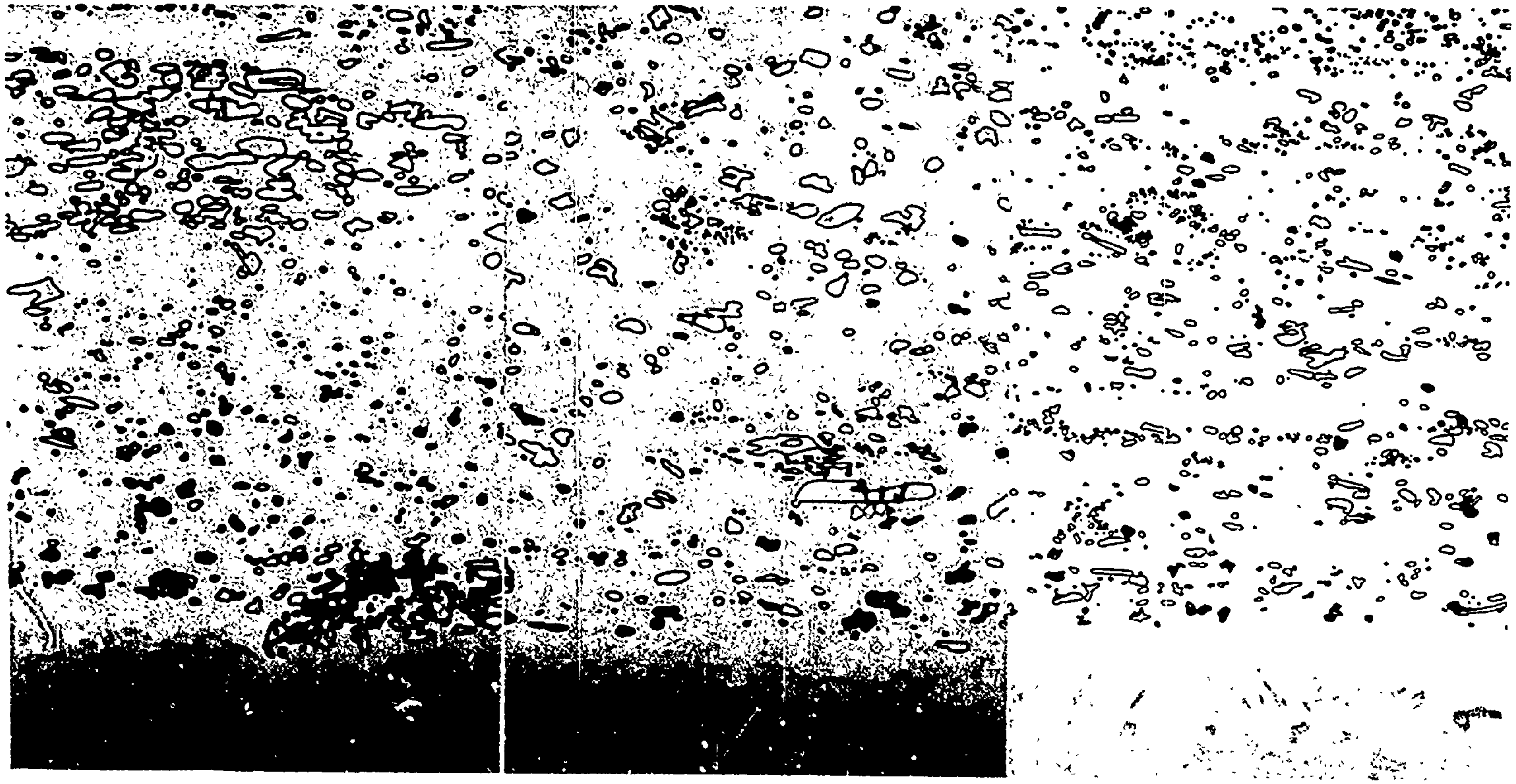
Fig. 15 Bond Evaluation Test Plates, 9 lb/ft^2, 30-cal AP. (× 0.5)



TEST PLATE 31-4

TITANIUM DIFFUSION BARRIER

Fig. 16 Bond Evaluation Test Plate, 9 lb/ft^2 30-cal AP



B5762

PLATE 31-1

4-HR-760°C BOND

B5764

PLATE 31-2

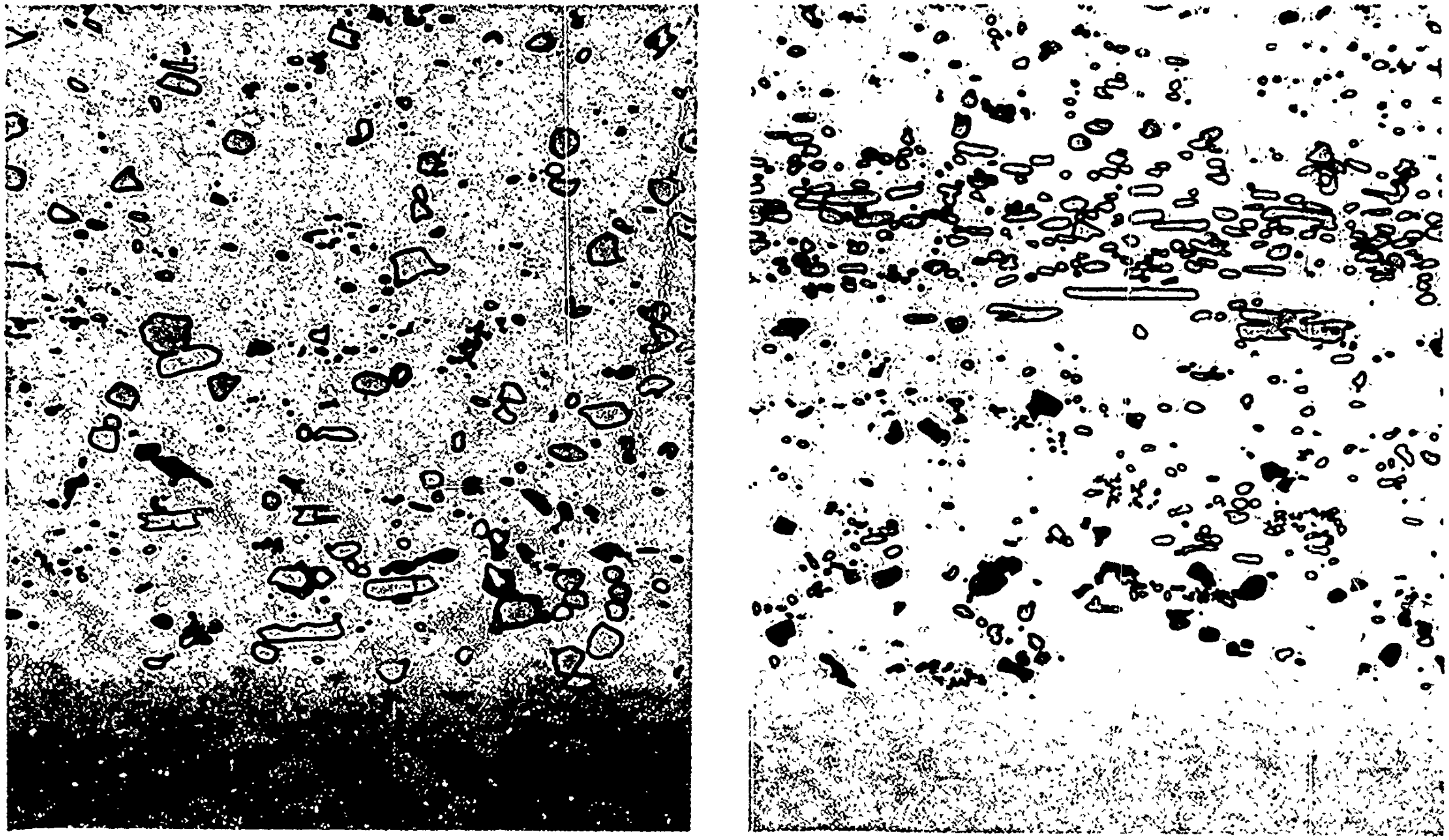
1-HR-870°C BOND

B5768

PLATE 31-4

1-HR -760°C BOND
WITH TITANIUM
BARRIER

Fig. 17 Effect of Bonding Conditions on Silicide Distribution in Dual Hardness Test Plates, Ti-3Si-3Fe-0.5N (895-1) on Ti-7Al-2.5Mo (× 400)



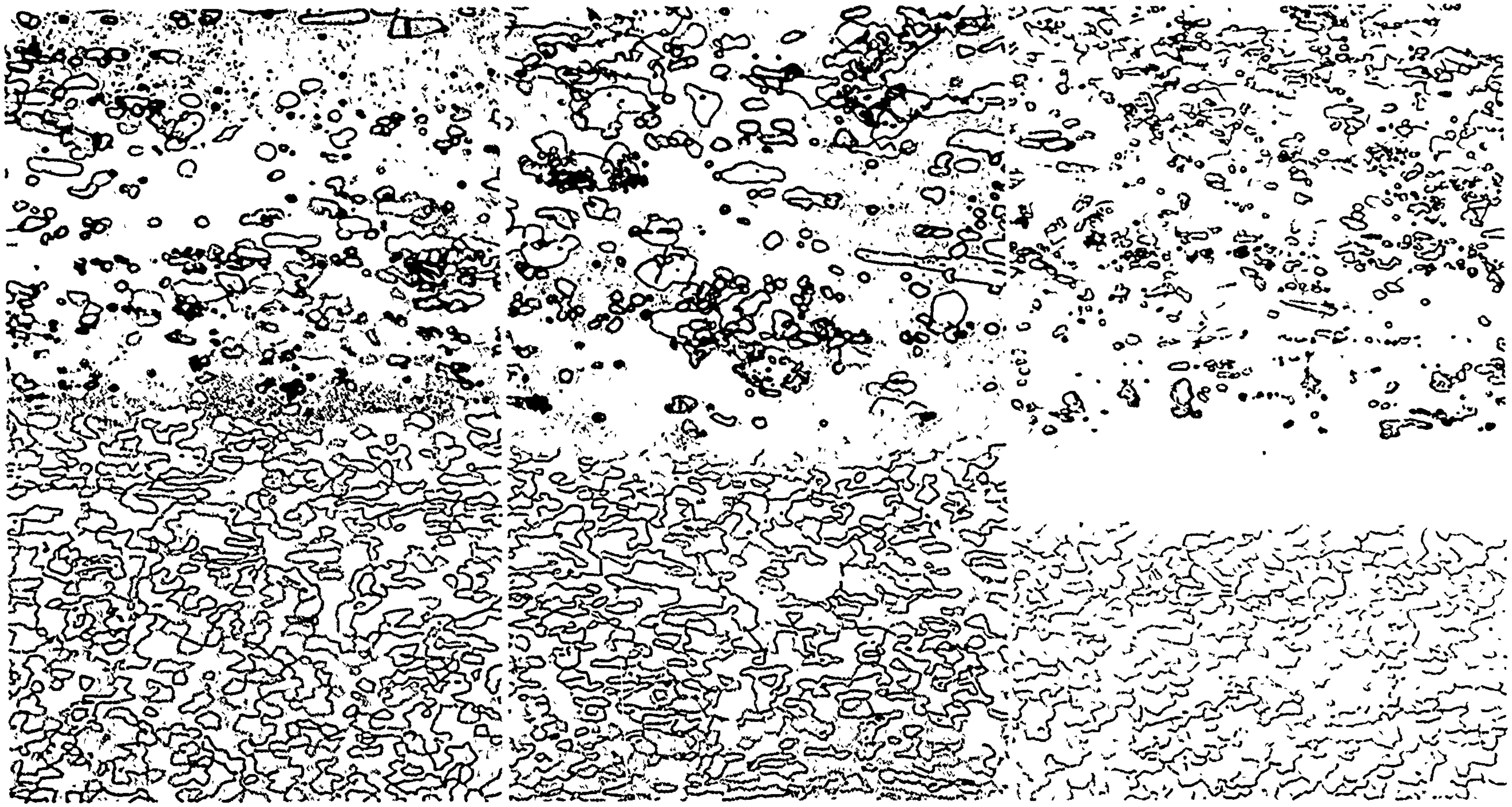
B5767 PLATE 31-3

24-HR-1200°C HOMOGENIZATION BEFORE DIFFUSION BONDING

B5770 PLATE 31-5

2 HR-1000°C HOMOGENIZATION AFTER ROLLING

Fig. 18 Effect of Homogenization Anneals on Silicide Distribution in Dual Hardness Test Plates, Ti-3Si-3Fe-0.5N (895-1) on Ti-7Al-2.5Mo (× 400)

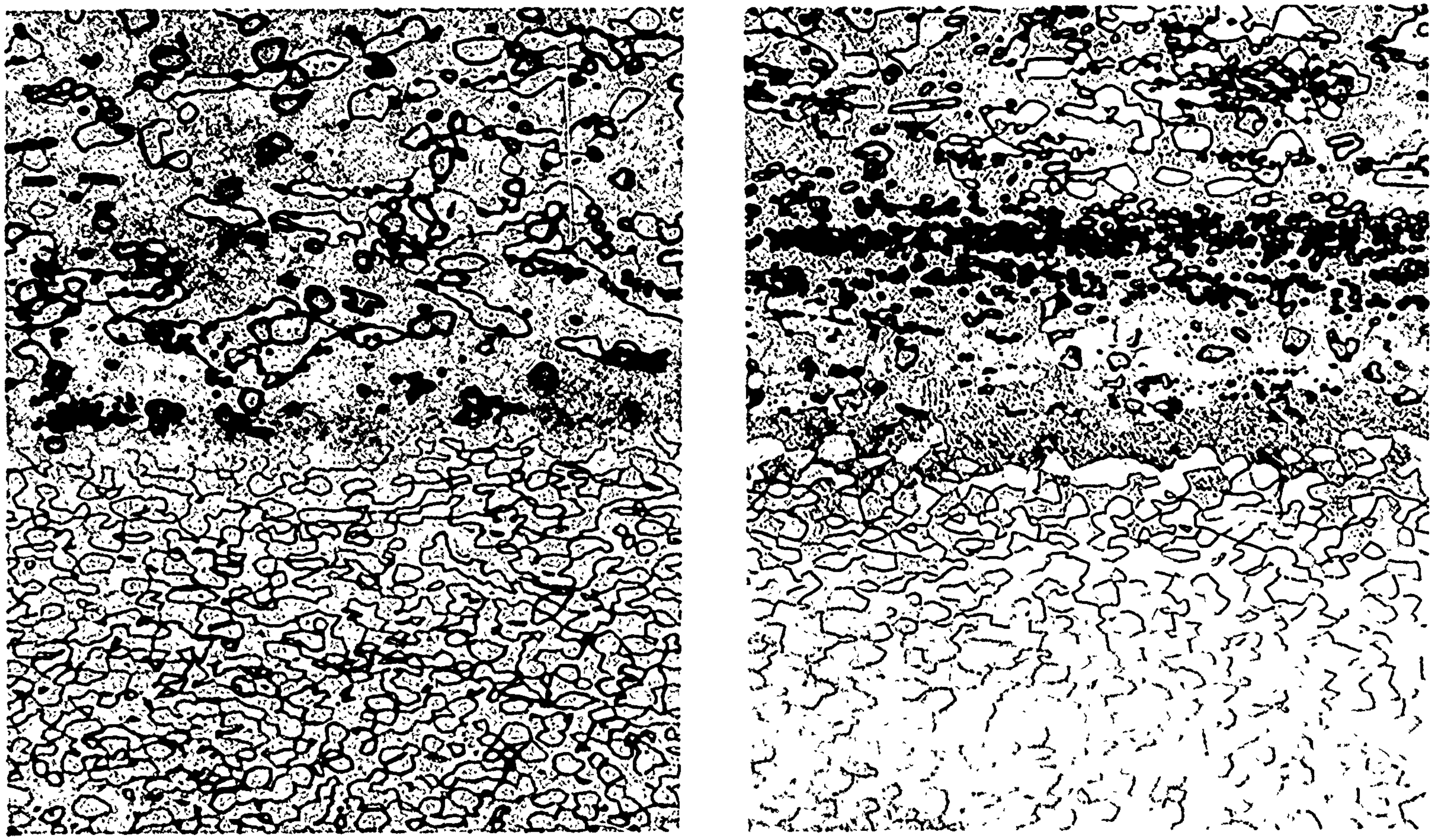


B5755 PLATE 31-1
4-HR-760°C BOND

B5756 PLATE 31-2
1-HR-870°C BOND

B5759 PLATE 31-5
1-HR-760°C BOND

Fig. 19 Effect of Bonding Conditions on Alloy Structure of Dual Hardness Test Plates, Ti-3Si-3Fe-0.5N (895-1) on Ti-7Al-2.5Mo (× 400)



B5757 PLATE 31-3

24-HR-1200°C HOMOGENIZATION
BEFORE DIFFUSION BONDING

B5761 PLATE 31-5

24-HR-1000°C HOMOGENIZATION
AFTER ROLLING

Fig. 20 Effect of Homogenization Anneals on Alloy Structure of Dual Hardness Test Plates, Ti-3Si-3Fe-0.5N (895-1) on Ti-741-2.5Mo (× 400)

Table 8
RESULTS OF BALLISTIC TESTS FOR BOND LINE EVALUATION(a)

Plate No.	Processing Condition	Hit No.	Penetration Rating	Spall
31-1	4 hr-760°C bond	1	Complete	Yes, 2-in. dia. (on largest spall area)
		2	Complete	Yes
		3	Partial	Small crater
		4	Partial	crater
		5	Complete	Yes
31-2	1 hr-870°C bond	1	Partial	2-1/2-in.-dia. spall
		2	Complete	2 round 1-in. dia. spall ea., no spall on 2 other rounds – total spall 25% of front face
		3	Complete	
		4	Complete	
		5	Complete	
31-3	Hard-face alloy homogenized 24 hr 1200°C before bonding. 1 hr-760°C bond	1	Partial	1/2-in.-dia. spall at edge
		2	Complete	None
		3	Complete	None
		4	Complete	None
		5	Complete	1-in.-dia. spall
31-4	Ti diffusion barrier 1 hr-760°C bond	1	Complete	55% of front face spall
		2	Complete	
		3	Complete	
		4	Partial	
		5	Partial (void)	
31-5	1 hr-760°C bond + 2 hr-982°C homogenization anneal after rolling	1	Complete	1-in. dia. spall
		2	Partial	None
		3	Complete	1-in. dia. spall
		4	Partial	None
		5	Partial	None

(a) 30 Cal. AP, AMMRC range.

The etched structures (Figs. 19 and 20) reveal no significant differences in back-face alloy structure as a function of different bonding conditions. The back face in all cases has an equiaxed alpha-beta structure with close to a 50:50 ratio of phases. The use of a titanium diffusion barrier produced a wide accicular single-phase alpha region at the bond line. This region probably was embrittled by nitrogen from the front face and as a result formed a plane of weakness at the bond line.

3.4 SELECTION OF PROCESSING PARAMETERS

An analysis of all test data and structural evaluations indicates that elimination of silicide segregation and excessive strain in the bond region should lead to significantly improved performance. The hard-alloy plates should be homogenized at the highest temperature possible (1200° to 1300°C) to break up banded segregates in the alloy. At these temperatures, significant amounts of silicon can be dissolved in solid solution, and an effective redistribution of silicides can be realized. The treatment must be done before bonding since the back-face alloy would be severely grain-coarsened, and excessive interdiffusion would occur under these conditions. A short homogenization anneal at 950° to 1000°C would be helpful after bonding and rolling to break up any residual silicide or alloy segregation in the bond area and to relieve any residual stress from differential rolling.

A diffusion bonding treatment of 1 hr at 760°C under a load of 345 psi produces a sound metallurgical bond which will remain intact on pressing or rolling at 950° to 1000°C. No improvement is realized from a higher temperature or longer time bonding treatment.

Rolling should be done at a temperature where little if any differential rolling of the two alloys occurs. The temperature should be below 1050°C to avoid the formation of intergranular silicides in the hard alloy and accicular structures in the back-face alloy. Rolling at 950° to 1000°C gives excellent results and produces fine-grained equiaxed alpha-beta and alpha-alpha prime structures in the rear- and front-face alloys, respectively.

Finally, the solution anneal for hardening is best done at 1000°C. Higher temperatures will embrittle the back-face alloy and low temperatures produce insufficient hardness and strength in the back-face alloy. Properties of the front-face alloy are not critically dependent on the solution temperatures. Annealing time can be as short as 10 min or as long as 2 hr. The longer times aid in homogenization and stress relief in the bond area. The best aging temperature as previously determined is 400°C, and times can range from 30 min to 4 hr with no significant change in hardness after a 30-min exposure.

In summary, the optimum processing conditions for the dual-hardness titanium plates are as follows:

(1) Homogenize hard-face alloy 24 hr at 1200°C in vacuum.
(2) Vacuum diffusion bond hard alloy to Ti-7Al-2.5Mo back-face alloy for 1 hr at 760°C under a load of 350 to 500 psi.

(3) Roll the bonded composite from 950° to 1000°C with 10% reduction per pass. Reheat and turn plate 90 deg between each pass (cross roll).
(4) Flatten plates by cooling from 950° to 1000°C while under sufficient pressure to hold flat.
(5) Anneal 2 hr at 1000°C, water-quench.
(6) Age 1 hr at 400°C, air-cool.

Section 4
BALLISTIC EVALUATION OF DUAL-HARDNESS ARMOR

4.1 MANUFACTURE OF TEST PLATES

Two ingots of a nominal Ti-3Si-3Fe-0.5N hard-face alloy were melted by the Oregon Metallurgical Corporation. The intent was to melt a 150-lb ingot, 8-in. dia. by 2 to 3 ft long. In both cases, thermal stresses from steep temperature gradients in the ingot generated a fracture after about a 1-ft-long section had solidified. The upper third of the ingot separated from the lower two-thirds along a fracture plane inclined about 45 deg to the vertical axis as shown in Fig. 21. In each case a dense, sound ingot stub 55 to 65 lb in weight was produced. The analysis of the two ingot pieces is given below:

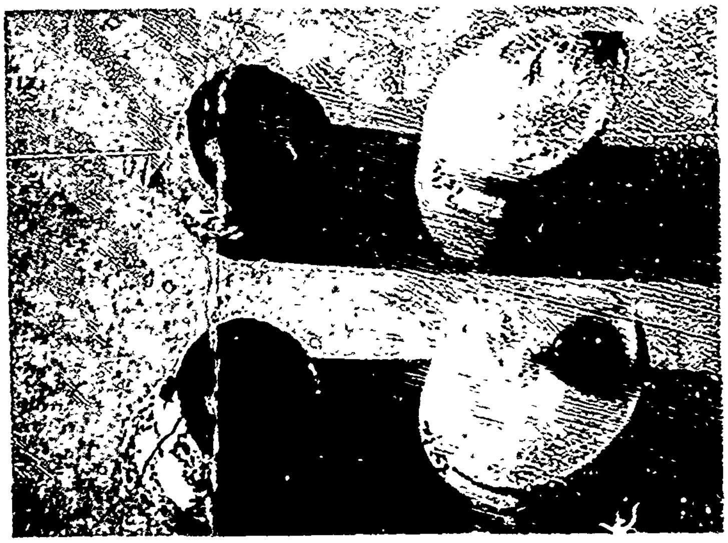
Table 9

COMPOSITION OF FINAL HARD-FACE ALLOY INGOTS
(wt %)

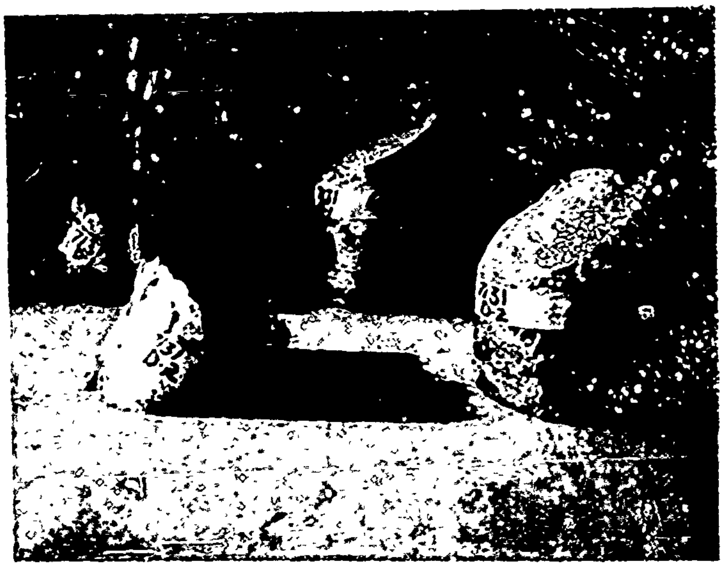
	C	H	O	N	Fe	Si
Heat 7131D1	0.017	0.0072	0.21	0.47	3.14	3.5
Heat 7131D2	0.032	0.0032	0.22	0.52	2.60	3.5

The iron content of Heat D2 is on the low side but this may be the result of limited sampling and alloy segregation. Both ingots were melted from the same first melt ingot stock and should have had essentially the same composition. Alloy segregation in the ingots was evident on the fractured surfaces. Extensive subsurface porosity at a depth of over 0.5 in. was observed in both alloys. In general, the ingot quality was poor, and low yields were anticipated. Extensive remelting appeared to be required to eliminate alloy segregation. Even then, there was no assurance that this alloy would be melted successfully as large ingots using the best techniques developed to date. Further research on melting and alloying practices for high silicon alloys is required, and continued attempts to produce crack-free ingots by trial and error do not appear to be warranted.

The two ingot stubs were processed to sheet bar by Coulter Steel and Forge Company, Emeryville, California. Ingot D2 shattered in the preheat furnace as a result of high thermal stress. The largest piece was forged to plate successfully. However, the plate cracked severely on rolling, and no useful material was obtained.



TOP VIEW



SIDE VIEW

Ti-3Si-3Fe-0.5N

Fig. 21 Appearance of Fractured 8-In. Diameter Ingots of High Hardness Ti-Alloy

Ingot D1 did not crack on heating but developed deep cracks and fissures on press-forging to 2-in.-thick plate. One-quarter of the billet appeared sound and was processed further. This piece was rolled successfully to 1-in.-thick plate from 982°C. The rolled plate was large enough to prepare two test plates for ballistic evaluation. After homogenization annealing at 1200°C, extensive alloy segregation in both plates was evident. It is doubtful whether test plates produced from these materials will be representative of the performance of this alloy composition.

Sufficient material remained from the two alloy evaluation heats, 895-1 and 895-2, to make 9 additional 7 × 7-in. test plates, and a total of 11 ballistic tests were prepared successfully. A summary of the plate sizes and alloys used is presented in Table 10. All plates were processed by the schedule presented in Section 3.5. Excellent quality plates with the exception of T11 were produced. This plate, made from the badly segregated D1 material, was cracked and poorly bonded. All other plates were essentially crack-free and had excellent bonds. As shown in Table 10, front-face hardness was in the R_c 58–59 target range for all but plates T8, 10, and 11. These are the thicker plates that are more difficult to harden fully. In addition, plates T10 and 11 are from the segregated D1 material and are not necessarily representative of this class of hard-face alloys.

Table 10

FINAL BALLISTIC TEST PLATES
(7 × 7-IN.)

Plate No.	Hard-Face Alloy	Composite Thickness (in.)	Areal Density (lb/ft^2)	Ratio of Front-to-Rear Thickness	Hardness, R_c	
					Front	Rear
T1	895-1	0.26	6	30:70	58	38
T2	895-1	0.26	6	40:60	58	39
T3	895-1	0.26	6	50:50	58	40
T4	895-2	0.34	8	30:70	59	42
T5	895-2	0.35	8	40:60	58	41
T6	895-2	0.35	8	50:50	58	42
T7	895-2	0.43	10	30:70	58	43
T8	895-2	0.44	10	40:60	57	41
T9	895-2	0.43	10	50:50	58	41
T10	D1	0.52	12	30:70	56	42
T11	D1	0.53	12	40:60	54	42

(a) 895-1, D1 – Ti-3Si-3Fe-0.5N.
895-2, – Ti-3Si-2Fe-3Mo-0.5N.

Structures at the bond line of representative test plates for each areal density are shown in Figs. 22 and 23. The silicide phase is uniformly distributed as a result of the homogenization anneals. A random distribution of large silicide particles existed throughout the hard front-face alloy as shown in Fig. 22. Freedom from stringers or gross segregates of silicides should give maximum resistance to front-face spall on ballistic impact.

Structure of the alloy matrix as revealed by etching is shown in Fig. 23. No significant differences for the different areal densities was noted. The Ti-7Al-2.5Mo back-face alloy has a fine-grained equiaxed alpha-beta structure with about a 50:50 alpha-beta ratio in all materials. The hard-face alloy has a fine martensitic structure in the transformed prior beta grains and a precipitate of what may be small silicide particles in some of the larger alpha islands. The structure is rich in the transformed beta phase (α') with an estimated $\alpha:\alpha'$ ratio of 30:70. No concentration of α' at the bond line was observed, and the alloy phases were distributed uniformly throughout the hard-face material. The only significant microstructural difference is a coarsening of the α' platelet size with increased areal density. The thicker sections cool more slowly, and the beta transformation product is coarser. This results in a decreased hardness of the alloy as indicated in Table 10. The back-face alloy, on the other hand, becomes harder with the slower cooling rate and has the maximum hardness in the sections over 0.3 in. thick.

4.2 RESULTS OF BALLISTIC TESTS

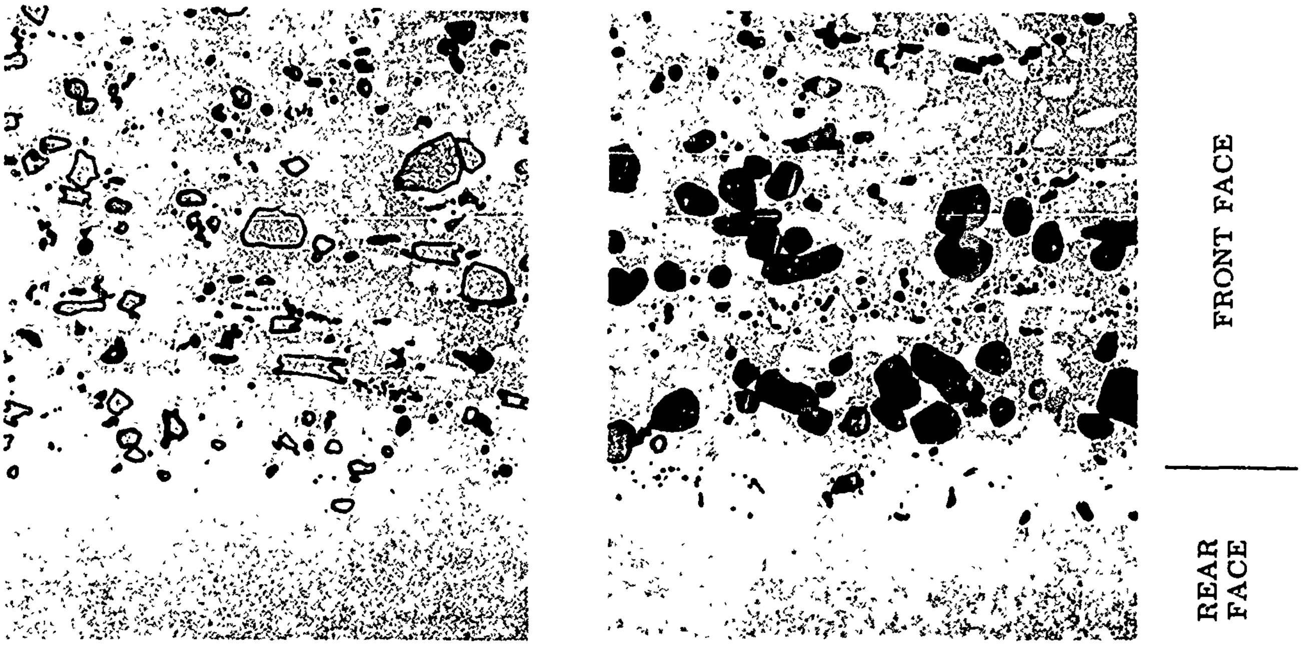
The 11 panels were test-fired with 30-cal AP projectiles at the Army Materials and Mechanics Research Center Range. The V-50 ballistic limits established by these tests are classified and have not been incorporated in this report to permit an unrestricted broad distribution of the overall study of dual-hardness titanium. The ballistic data can be obtained from AMMRC on request with appropriate clearance and need-to-know.

General results in terms of test-plate appearance and front-face spall characteristics are summarized in Figs. 24 through 27 and Table 11. A schematic comparison of V-50 ballistic limits with those characteristic of homogeneous Ti-6Al-4V as function of areal density is presented in Fig. 28.

The low nitrogen alloy (Ti-3Si-3Fe-05N) used in the 6- and 12-lb/ft^2 test plates exhibited exceptionally good resistance to front-face spall. As shown in Figs. 24 and 27, only small chips spalled in the immediate impact region, even with complete penetration of the test plate. The plates exhibit a good multihit capability.

A slight effect of variations in front- to rear-face thickness ratio on spall resistance for this alloy is indicated. As shown in Table 11, spall resistance decreased as the front-to-rear thickness ratio increased. Best spall resistance was exhibited by plates with a 30:70 front:rear thickness ratio.

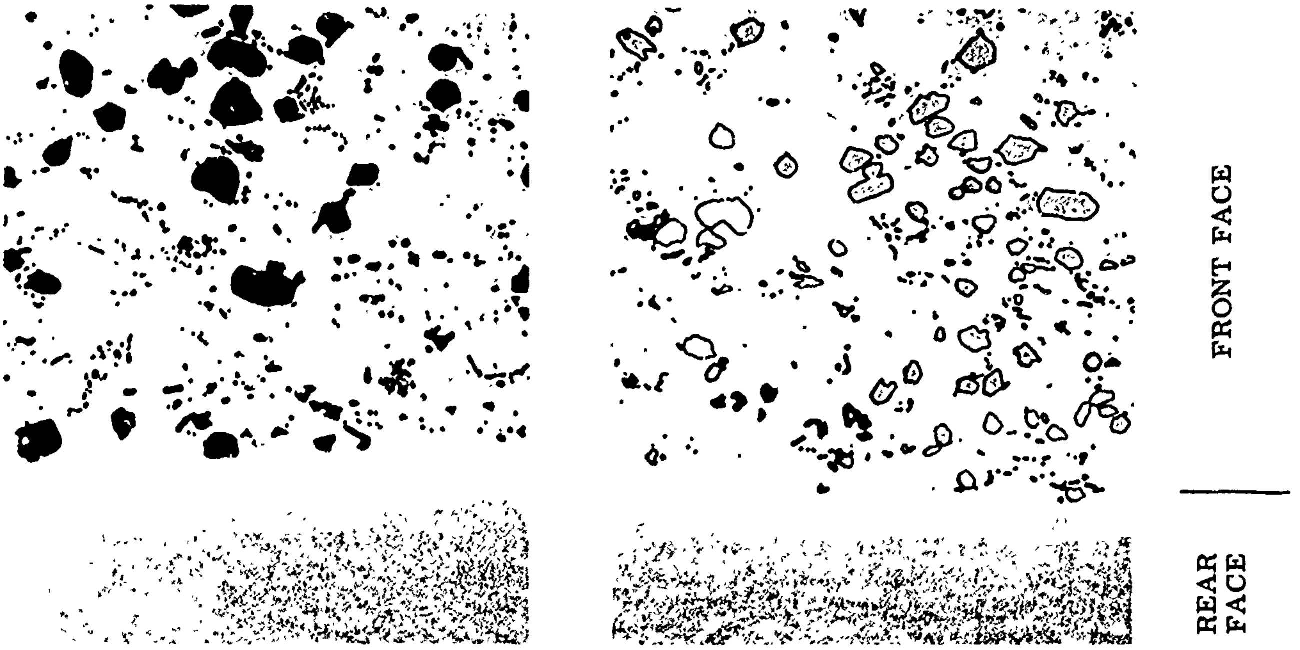
The high nitrogen alloy (Ti-3Si-2Fe-3Mo-0.75N) used to prepare the 8- and 10-lb/ft^2 test plates had very poor resistance to front-face spall. As shown in Figs. 25 and 26,



B6256 B6262

T-2 6 lb/ft^2 T-11 12 lb/ft^2

Front-Face Alloy – Ti-3Si-3Fe-0.5N

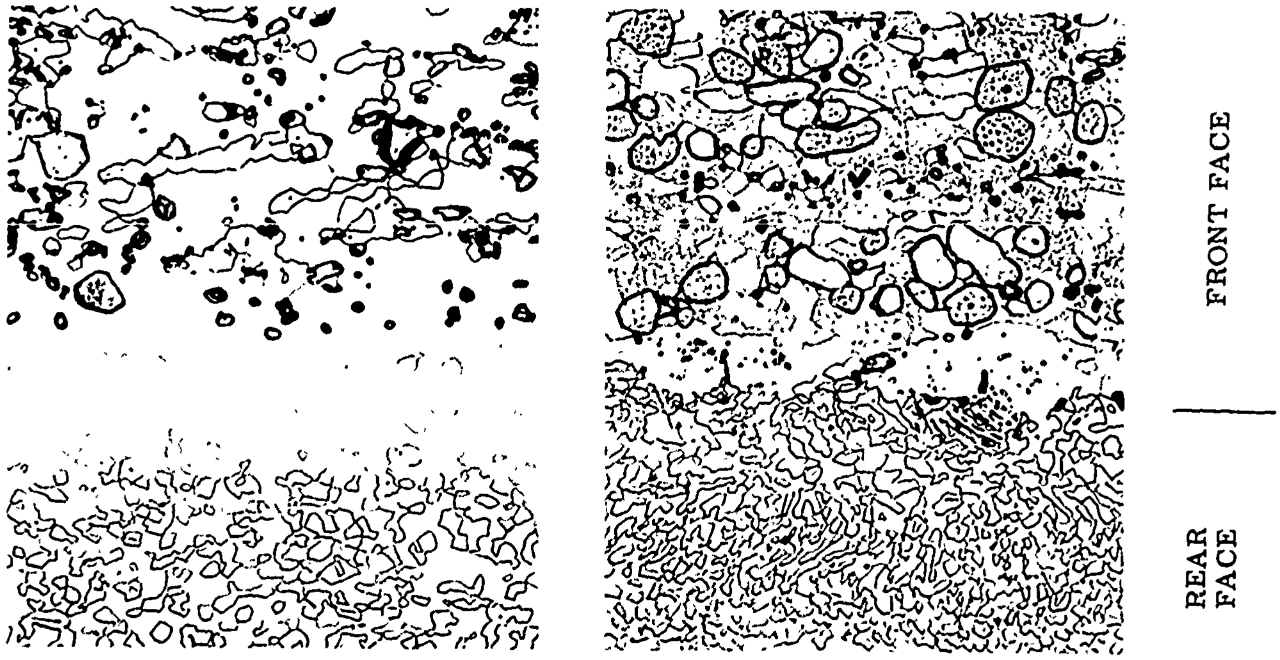


B6258

T-5 8 lb/ft^2 T-8 10 lb/ft^2

Front-Face Alloy – Ti-3Si-2Fe-3Mo-0.75N

Fig. 22 Silicide Distribution in Final Test Plates, 40:60 Front-To-Rear Thickness Ratio, ×400



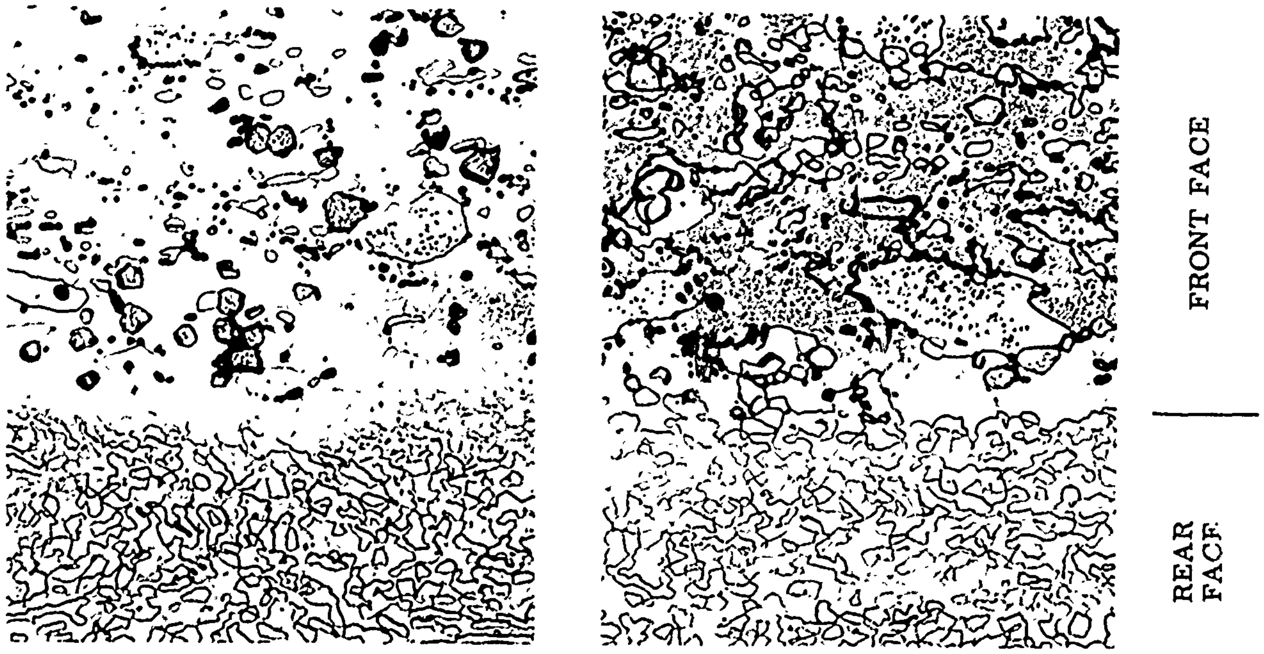
B6257

T-2 6 lb/ft^2

B6263

T-11 12 lb/ft^2

Front-Face Alloy – Ti-3Si-3Fe-0.5N



B6259

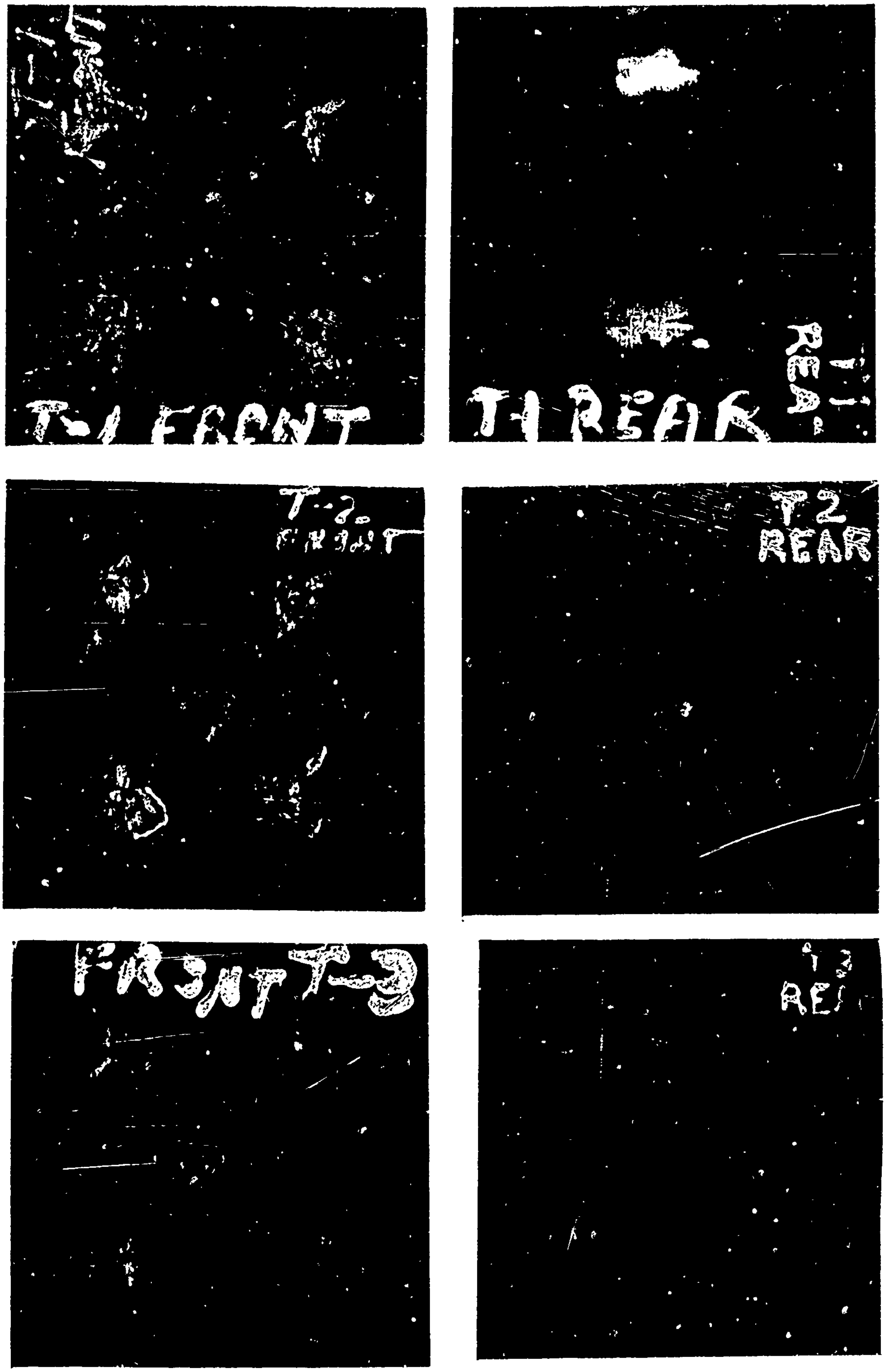
T-5 8 lb/ft^2

B6262

T-8 10 lb/ft^2

Front-Face Alloy – Ti-3Si-2Fe-3Mo-0.75N

Fig. 23 Structure of Final Test Plates, 40:60 Front-To-Rear Thickness Ratio, ×400



30:70 RATIO

40:60 RATIO

50:50 RATIO

Fig. 24 6 lb/ft^2 Ballistic Test Plates, ×0.45

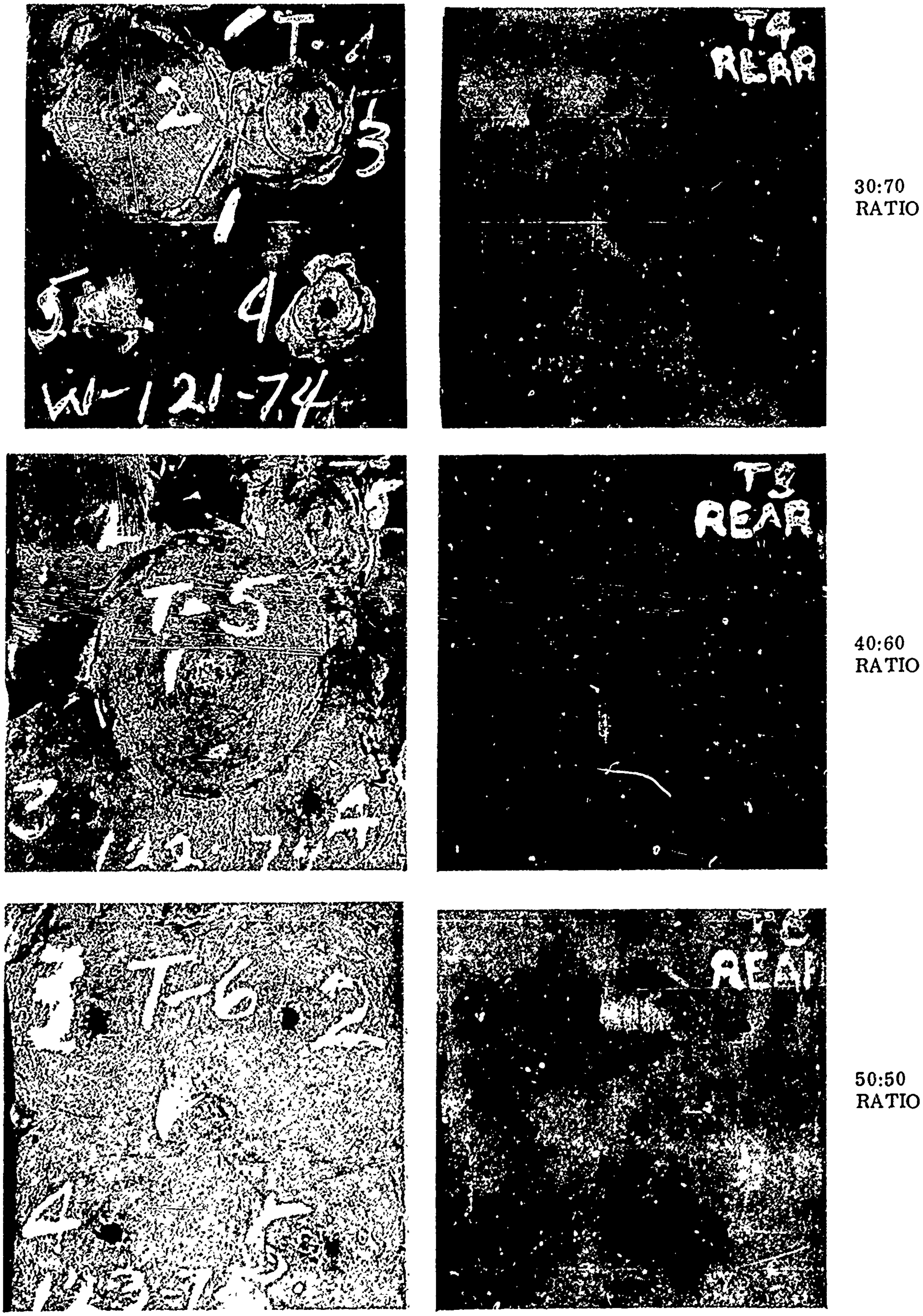
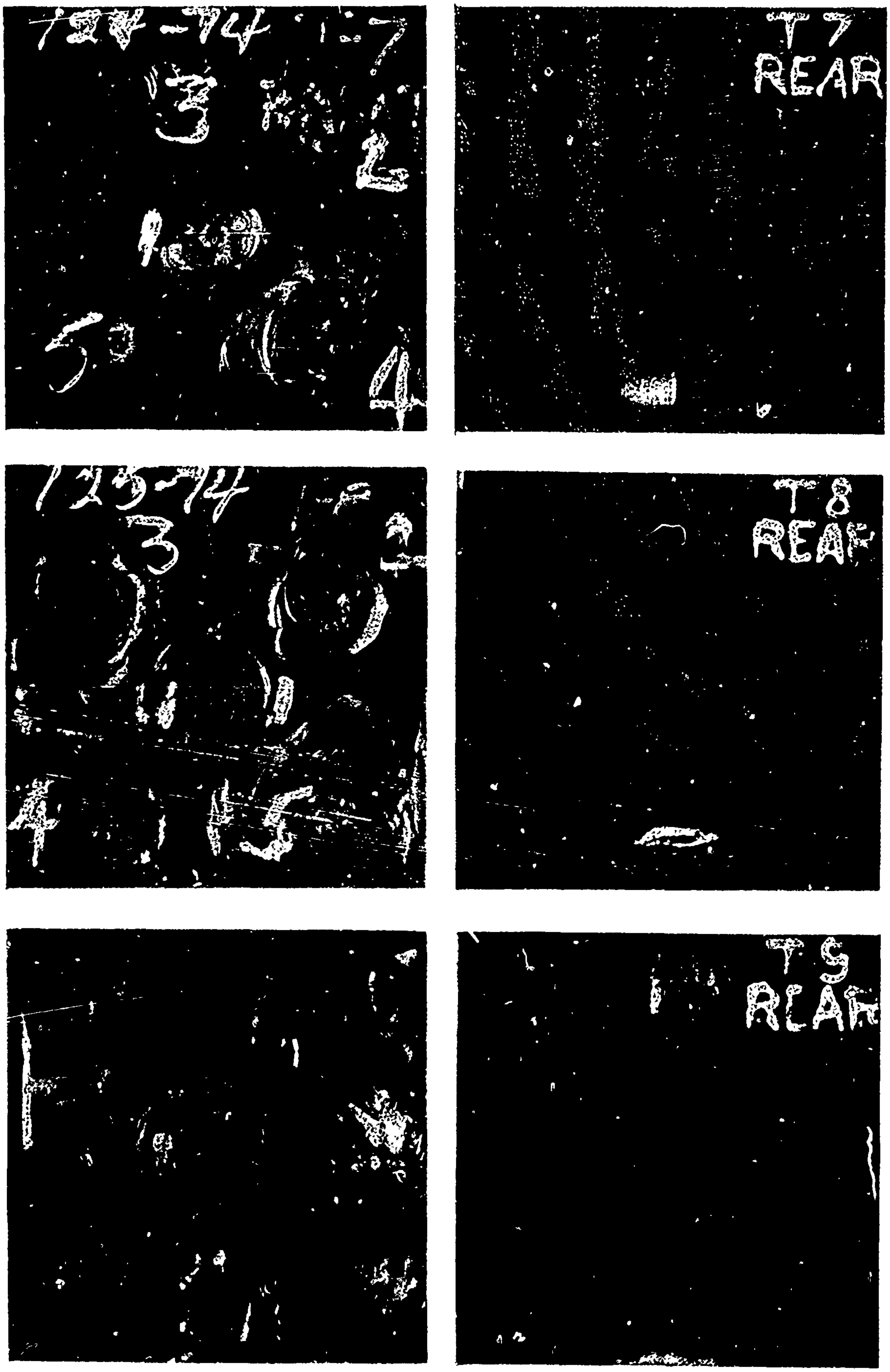


Fig. 25 8 lb/ft^2 Ballistic Test Plates, ×0.45

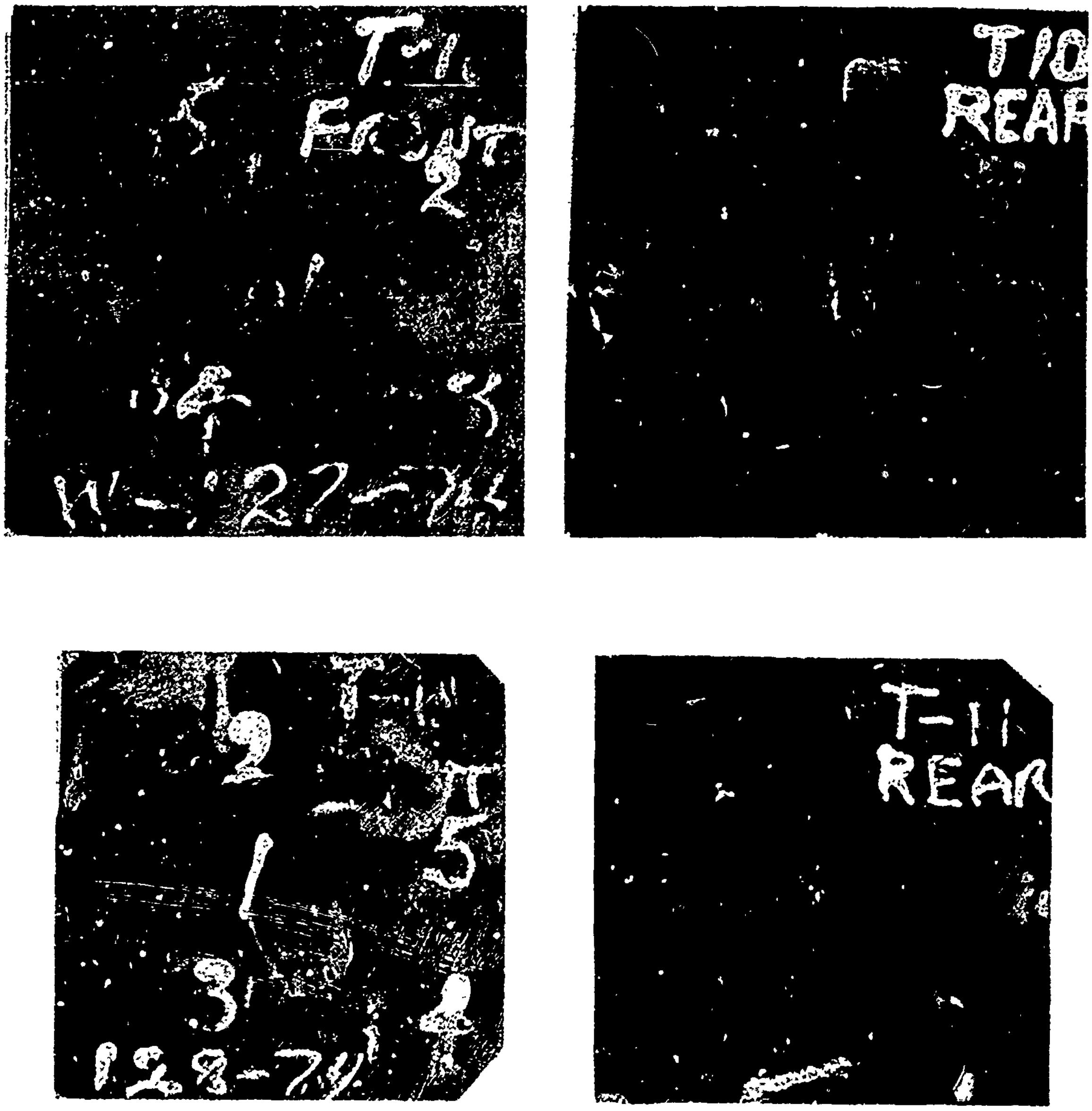


30:70
RATIO

40:60
RATIO

50:50
RATIO

Fig. 26 10 lb/ft^2 Ballistic Test Plates, ×0.45



30:70
RATIO

40:60
RATIO

Fig. 27 12 lb/ft^2 Ballistic Test Plates, ×0.45

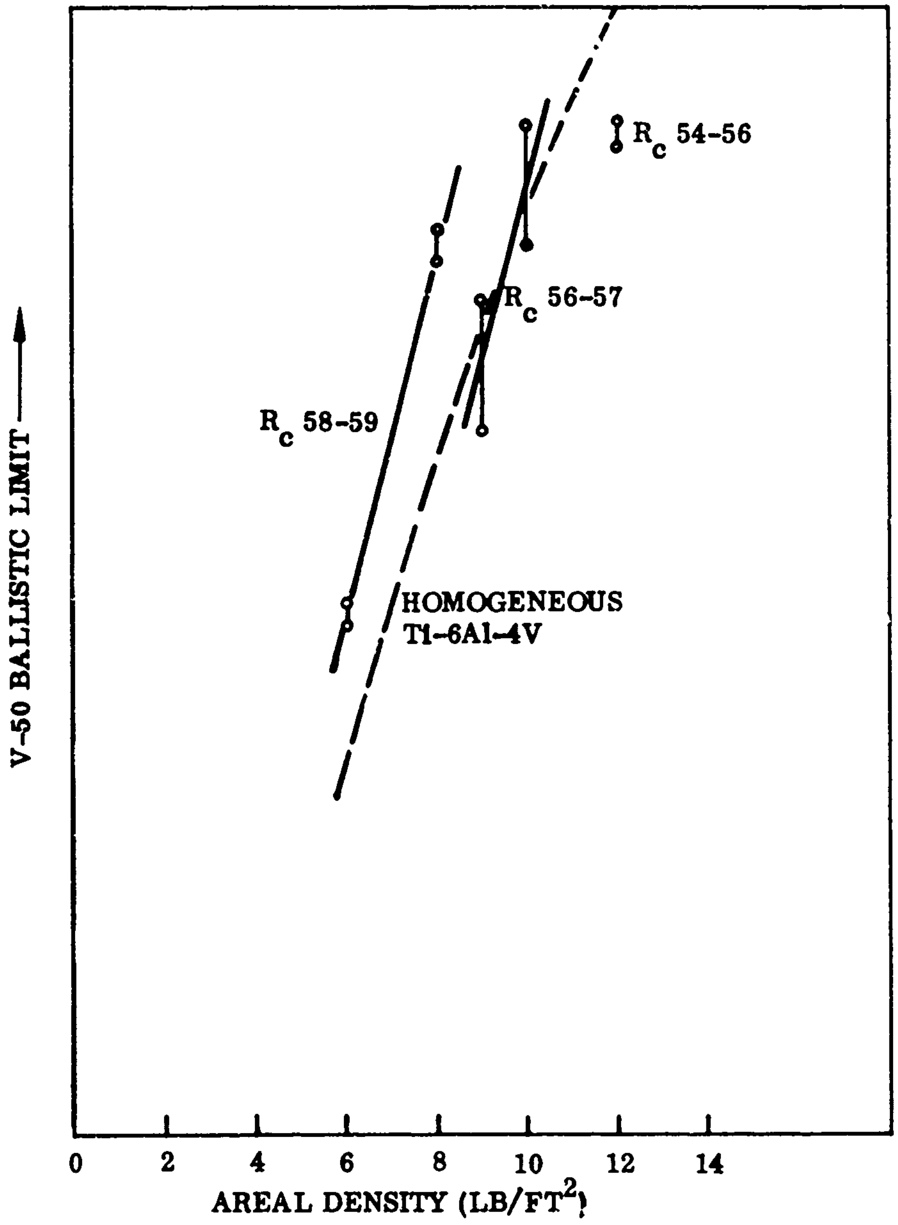


Fig. 28 Effect of Areal Density and Front-Face Hardness on V-50 Ballistic Limit of Dual Hardness Titanium Armor.

Table 11

SPALL CHARACTERISTICS OF FINAL BALLISTIC TEST PLATES

Plate No	Hard Alloy	Total Thickness (in.)	Areal Density (lb/ft^2)	Front:Rear Thickness Ratio	Hit No.	Penetration Rating	Front-Face Spall Area
T1	895-1	0.26	6	30:70	1	C	Nil
					2	P	Nil
					3	P	Nil
					4	P	Nil
					5	C	Nil
T2	895-1	0.26	6	40:60	1	C	Nil
					2	P	0.5-in. dia.
					3	P	Nil
					4	P	Nil
					5	P	0.5-in. dia.
T3	895-1	0.26	6	50:50	1	P	Nil
					2	C	Nil
					3	P	Nil
					4	P	1-in. dia.
					5	C	1-in. dia.
T4	895-2	0.34	8	30:70	1	P	Nil
					2	P	3-in. dia.
					3	C	1.5-in. dia.
					4	C	1.5-in. dia.
					5	P	0.5-in. dia.
T5	895-2	0.35	8	40:60	1	P	4-in. dia.
					2	C	3-in. dia.
					3	C	2-in. dia.
					4	C	3-in. dia.
					5	P	2-in. dia
T6	895-2	0.35	8	50:50	1	P	2-in. dia.
					2	C	3-in. dia.
					3	C	3-in. dia.
					4	C	3-in. dia.
					5	P	3-in. dia.
T7	895-2	0.43	10	30:70	1	C	1-in. dia.
					2	C	1-in. dia.
					3	C	1-in. dia.
					4	P	1.5-in. dia.
					5	P	Nil
T8	895-2	0.44	10	40:60	1	P	2-in. dia.
					2	C	1-in. dia.
					3	C	1.5-in. dia.
					4	C	3-in. dia.
					5	P	3-in. dia.
T9	895-2	0.43	10	50:50	1	C	3-in. dia.
					2	P	Nil
					3	P	3-in. dia.
					4	C	3-in. dia.
					5	P	2-in. dia.
T10	D-2	0.52	12	30:70	1	C	Nil
					2	P	Nil
					3	P	Nil
					4	C	Nil
					5	P	Nil
T11	D-2	0.53	12	40:60	1	C	1.5-in. dia.
					2	P	1.5-in. dia.
					3	P	Nil
					4	C	Nil
					5	P	1-in. dia.

(a) 895-1, D1 – Ti-3Si-3Fe-0.5N hard alloy.
895-2 – Ti-3Si-2Fe-3Mo-0.75N hard alloy.

(b) P – Partial, C – Complete

large spall areas occurred around virtually all hits. Again, spall resistance was best at the 30:70 front-to-rear thickness ratio and decreased with increasing front-face thickness. The plates with a 50:50 ratio were virtually destroyed by five hits with 30-cal AP projectiles.

The V-50 ballistic limits were not materially changed by variations in front-to-rear thickness ratios. The data as plotted in Fig. 28 show the spread in values (highest to lowest) of the V-50 limits for each areal density. Although the scales are missing from these plots, it can be seen that the spread is not very great considering the slope of the curves and is probably well within any scatter band for one thickness ratio. In addition, there are no trends in which low ballistic limits occur at low thickness ratios, or vice versa. The ballistic limits scatter on a random basis and do not correlate directly with thickness ratios.

As shown in Fig. 28, there does appear to be some correlation between front-face hardness and ballistic limit. The thinner plates (6 to 8 lb/ft^2) cool more rapidly on quenching and have the highest surface hardness (R_c 58-59). These plates also have the highest ballistic limit and are superior to homogeneous Ti-6Al-4V. The thicker plates which cool more slowly have a lower hardness and the ballistic limit is comparable to that of Ti-6Al-4V. Data from the 9 lb/ft^2 alloy and bond evaluation test plates are included in this plot to indicate more clearly the influence of surface hardness on ballistic performance. The results indicate that a surface hardness in excess of R_c 60 is required for a major increase in the performance of titanium armor.

The back-face alloy (Ti-7Al-2.5Mo) performed exceptionally well in all tests. No back-face spall occurred on any of the hits and the ductile tearing which occurred on complate penetration was highly localized.

Section 5
CONCLUSIONS AND RECOMMENDATIONS

5.1 CONCLUSIONS

1. A dual-hardness titanium armor consisting of a Ti-3Si-3Fe-0.5N hard-face alloy bonded to a Ti-7Al-2.5Mo back-face alloy provides good protection against 30-cal AP threats at low areal densities (6 to 8 lb/ft^2). The armor has higher ballistic limit than homogeneous titanium alloy armor in sections up to 0.35 in. thick. It has a multihit capability with good resistance to bond-line separation and to front- and rear-face spall.

2. The dual-hardness armor has reduced capabilities at higher areal densities (10 to 12 lb/ft^2) because of low hardenability of the hard-face alloy in heavy sections. The ballistic limit is no better than that of homogeneous titanium alloys in sections greater than 0.35 in. thick.

3. Variations in the front-to-rear-face thickness ratios between 30:70 and 50:50 have little effect on ballistic limit at areal densities between 6 to 12 lb/ft^2. Resistance to front-face spall, however, increases with decreasing thickness of the hard-face alloy. Maximum spall resistance with a good ballistic limit is achieved at a 30:70 front-to-rear thickness ratio in all cases.

4. Surface hardness is the key factor governing the ballistic limit of the dual-hardness titanium armor. A surface hardness of R_c 58-59 is required to exceed the capabilities of homogeneous titanium alloy armor. Hardness above R_c 60 appears to be required to achieve a ballistic performance superior to their commercial steel armors.

5. The Ti-3Si-3Fe-0.5N alloy is the optimum composition for the hard face in terms of cost, producibility, and ballistic performance. The alloy has excellent hot workability at 700° to 1000°C and can be hardened to R_c 59 in thin sections (up to 0.35 in. thick). Hardness can be increased to R_c 60-61 by adding 3 to 5% Ge to the alloy at a cost penalty of \$2 to \$3/lb.

6. The silicon and nitrogen additions are the most important factors governing producibility and front-face spall resistance. Silicon segregation in melting promotes cracking at all stages of processing and contributes to poor spall resistance. Homogenization by repeated melting and high-temperature solution anneals is essential to good producibility and spall resistance. Nitrogen contents above 0.5% aggravate the problem of silicon segregation and contribute directly to poor workability and spall resistance.

7. This study has demonstrated the feasibility of producing a dual-hardness titanium armor with a ballistic performance superior to that of commercial titanium alloys. Further development and evaluation of this new class of materials as lightweight armor for a broad range of threats are warranted.

5.2 RECOMMENDATIONS

It is recommended that the development and evaluation of dual-hardness titanium armor based on a Ti-3Si-3Fe-0.5N hard-face alloy bonded to a Ti-7Al-2.5Mo back-face alloy be continued. Specifically, it is recommended that:

1. The Ti-3-Si-3Fe-0.5N alloy armor be evaluated at low areal densities (2 to 8 lb/ft) for resistance to fragmentation-type threats.
2. The same alloy modified with 3 to 5 Ge to increase hardness to R_c 60-62 be evaluated at intermediate areal densities (6 to 8 lb/ft^2) against 30-cal AP threats.

This additional work would not require any further materials or process development. The primary purpose would be to characterize more completely the ballistic performance and ultimate potential of this new class of materials. The evaluation at the higher hardness level (Ge-modified alloy) would provide sufficient data to determine the feasibility of exceeding the ballistic performance of steel armors in a titanium base system. All work should be restricted to low areal densities (less than 0.4 in. thick) where good hardening of the front-face alloy can be realized. On completion of this evaluation, a thorough assessment of the ultimate potential of dual-titanium hardness armor for small arms and fragmentation-type threats would be possible.