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EVALUATION OF DUAL-HARDNESS TITANIUM ALLOY ARMOR

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Lockheed Missiles and Space Company
Palo Alto, California

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Final Report--Contract DAAG46-69-C-0047

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Prepared for

ARMY MATERIALS AND MECHANICS RESEARCH CENTER
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Dual Hardness Titanium Alloy Armor Study
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FOREWORD

This report covers the work performed under Contract No. DAAG 46-69-C-0047, D/A Project 21X2040 96A 7022, AMCMS Code 5025.11.294. The work was administered by the Army Materials and Mechanics Research Center, Watertown, Massachusetts 02172, with Mr. D. Papetti as Project Officer.

The report was prepared by Roger A. Perkins and Elliott H. Rennhack of the Materials Sciences Laboratory of the Lockheed Palo Alto Research Laboratory, Lockheed Missiles & Space Company, Palo Alto, California.

ABSTRACT

A program to evaluate the producibility and ballistic characteristics of dual-hardness titanium utilizing a new class of high-hardness titanium alloys developed by the Lockheed Palo Alto Research Laboratory has been conducted. The results of studies to optimize the composition and heat-treatment schedules for the hard alloys are presented. The approach to melting and fabrication of 25-lb ingots of these alloys is discussed. Factors influencing the selection of a suitable back-face alloy are reviewed, and methods for preparing dual-hardness composite test plates from the optimum materials are detailed. Results of ballistic tests with 0.30- and 0.50-cal AP/M2 projectiles at AMMRC are presented and correlated with the metallurgical structure and properties of the composites. An assessment of the technical feasibility of producing an effective dual-hardness titanium armor based on this concept is made, and recommendations for continuing work are presented.

CONTENTS

ILLUSTRATIONS

TABLES

Section 1

INTRODUCTION

1.1 BACKGROUND

The ideal armor material is one that will (1) shatter projectiles on impact, (2) resist penetration by the projectile or its fragments, and (3) absorb the projectile's energy without complete fracture or back-face spall. This requires a material that has the dual characteristics of high hardness and good toughness. Materials that are sufficiently hard to shatter projectiles generally do not have sufficient toughness to resist fracture or spalling. A composite approach in which a hard front face material is coupled with a strong, tough backup material offers a promising solution to the development of high-performance armor.

Sufficient data have been generated to demonstrate that composites with a sharp variation in properties between front and rear faces have the best potential as armor materials. The superiority of metallurgically bonded dual-hardness steel over high-hardness, homogeneous alloys has been clearly demonstrated by extensive ballistic tests and service in the field. Boron carbide and other ceramics backed up with tough plastic laminates also provide superior protection. Both concepts are in use today for high-performance systems.

From a producibility and service point of view, the ideal armor material also must possess several other important characteristics. First, the material must be fabricable into desired shapes at reasonable cost. Dual-hardness steel has a marked advantage over ceramic armor in both cases. Although recent advances in technology have removed major limitations on the fabrication of ceramic armor, such composites are high in cost. Second, it is desirable for the material to have a multihit capability and to resist damage by shock, vibration, and impact. Again, dual-hardness steel has a marked superiority over ceramic armor in this regard. As a third consideration, the armor material should contribute a minimum of weight to a vehicle for maximum protection. This is important for all types of vehicles – air, ground, and water. In this case, ceramic armor has a marked advantage over dual-hardness steel. In the current state of development, ceramic armor generally will handle an equivalent threat at a much lower overall weight compared with steel. In many applications, this advantage far outweighs high cost, multihit restrictions, and limitations on design and manufacture.

An analysis of other potential armor materials suggests that dual-hardness titanium conceivably could combine the attractive features of both steel and ceramic armors. A significant weight advantage could be gained over steel while still retaining the fabricability, design flexibility, multihit capability, and load-carrying ability of an all-metallic system. For equivalent armor thicknesses, titanium will have about 60% the weight of steel, which gives a 40% weight savings in armored systems. Although somewhat heavier than ceramic armor at equivalent thicknesses, a titanium armor system conceivably could have less total weight since the ceramic armor must be used as an overlay on a load-carrying structure. The metallic armor, on the other hand, can be used as the load-carrying member itself. Although more expensive than steel, titanium armor materials will be significantly less expensive than ceramic armors. This will be particularly true when producibility, design, and manufacturing requirements are taken into consideration.

Although considerable effort has been devoted to the development of titanium alloys for armor, none of the materials disclosed to date have been capable of defeating high-velocity projectiles by a fracture mechanism. Intermediate-hardness alloys have been used to defeat projectiles by resisting penetration. Titanium alloys of sufficiently high hardness to shatter projectiles have not been produced. Based on the current theory of defeat mechanisms in dual-hardness steel and composite ceramic armor, the hardness of the front face should approach the hardness of the steel penetrator (i.e., R_c 65 – 70). The highest values reported for titanium alloys are in the range of R_c 50 – 55. In addition, alloy compositions that can be hardened to this range tend to be very brittle and are subject to cracking as a result of high stress generated from thermal gradients encountered in melting and fabrication. The inability to produce titanium alloys at hardness levels above R_c 55 has precluded the development of an effective dual-hardness titanium armor.

In 1968, the Lockheed Palo Alto Research Laboratory undertook a study of new approaches to the design of high-hardness titanium alloys suitable for dual-hardness armor. The result was the development of a new class of alloys containing Ge, Si, Mn, and N which were hardened by a combination of mechanisms, including (1) solid solution alloying, (2) phase transformation, (3) formation of a hard-phase dispersion, and (4) intermetallic compound precipitation. The alloys could be solution annealed to a semihard condition (R_c 45 – 55) for fabrication and subsequently aged to full hardness (R_c 55 – 65). This is the highest level of hardness that has been reported for a wrought, heat-treatable titanium-base alloy. A small plate of one alloy was diffusion bonded to commercial Ti-8Al-1Mo-1V alloy plate and the composite hot rolled 70% into 0.1-in.-thick plate with a 50:50 ratio of hard-to-soft face thicknesses. The hardness after heat treatment was R_c 62 – 64 for the front face and R_c 46 – 48 for the soft face. A good metallurgical bond was obtained with a gradual change in hardness across the interface.

The discovery of this new class of alloys and their utilization in producing a dual-hardness composite that approaches the hardness level of steel armor represent potentially significant advances in titanium armor technology. It appears that the development of an effective dual-hardness titanium armor may now be feasible. An evaluation of the producibility of these alloys in ingot form and their ballistic performance as composite test panels is required to define more clearly the potential of this approach. The present investigation was designed to meet this need.

1.2 OBJECTIVE AND SCOPE

The objective of this program was to establish the technical feasibility of producing an effective dual-hardness titanium armor utilizing a new class of high-hardness titanium alloys. Alloys in the system Ti, 3 – 5% Ge, 2 – 3% Si, 3 – 7% Mn (or Fe), and 0.5 – 1.25% N were screened with respect to hardness and heat treatment response to permit selection of front face compositions that could be hardened to three different R_C levels: 55 – 60, 60 – 65, and >65. Methods for arc melting 4- to 6-in.-diameter ingots weighing 18 to 25 lb and for press forging and rolling these ingots into plate were studied. The compatibility of working and heat-treatment procedures with commercial back-face alloys also was considered. Back-face alloy candidates were screened for similar compatibility characteristics as well as for toughness and for ballistic performance. In addition, processing schedules for diffusion and roll bonding the hard- and soft-face alloys into composite plate were established.

Materials and procedures selected on the basis of these studies were used to prepare eight ballistic test panels measuring 5 by 5 by 0.625 in. for testing and evaluation. The plates were subjected to ballistic tests using 0.30- and 0.50-cal AP-M2 projectiles at the Army Materials and Mechanics Research Center, Watertown, Mass. Results were correlated with hardness, structure, and composition of both the front and rear faces and the nature of the bond interface to serve as a guide for future development of these materials. The results of this program indicate good technical feasibility for the manufacture of an effective dual-hardness titanium armor, and recommendations for further development of this concept are presented.

Section 2
MATERIALS AND PROCESS DEVELOPMENT

2.1 HARD-FACE ALLOYS

2.1.1 Technical Background

Significant hardening in titanium is developed by (1) interstitial alloying with C, O, and N, (2) the omega transition phase, and (3) precipitation of coherent, intermetallic compound particles. In the first method, nitrogen is the most potent hardener. The degree of hardening achieved with this element increases proportionately with increasing nitrogen content and reaches a maximum at the saturation value. A saturated solid solution of nitrogen in alpha titanium has a hardness of R_c 66 to 68; the saturated solid solution of beta which transforms to alpha on cooling has a hardness of R_c 52 – 55 (Ref. 1). Nitrogen also reduces the ductility of titanium and increases the ductile-to-brittle transition temperature. In general, nitrogen hardening occurs only at low temperatures; high nitrogen alloys are soft and ductile at elevated temperatures. At sufficiently high nitrogen contents, the ductile-to-brittle transition temperature exceeds the normal range for hot working titanium, and the alloys cannot be fabricated. Studies at the Lockheed Palo Alto Research Laboratory have shown that up to 2.0 wt% nitrogen can be added to unalloyed titanium without complete loss of ductility. Such alloys have marginal hot workability. At a level of 1.0 to 1.5 wt% N in titanium, the alloys have good ductility at high temperatures and can be readily hot worked and formed. The base hardness of binary titanium-nitrogen alloys in the range of 1.0 – 1.5 wt% N is R_c 45 to 50.

The hardening of titanium by heat treatment depends primarily on the rate of decomposition of the beta phase (method 2). Control of this reaction is facilitated by the addition of appropriate substitutional-type solute elements (beta stabilizers), which retard the $\beta \rightarrow \alpha$ transformation. During this reaction, a hard transition phase known as omega has been observed to form (i.e., $\beta \rightarrow \omega + \alpha$) in a number of alloy systems including Ti-Fe, Ti-Cr, Ti-V, Ti-Mo, and Ti-Ni (Ref. 2). Hardness levels of alloys containing this phase have been reported in the literature to approach 57 R_c (Ref. 3), the highest values being developed in the Ti-Fe system (Ref. 4). Alloys hardened by this mechanism also tend to be extremely brittle. This hardness is still well below the level of R_c 60 – 65 desired for a hard-face alloy.

Compound precipitation (method 3) has not been particularly effective in hardening titanium alloys. Studies have been made with both hypoeutectoid and hypereutectoid alloys in the Ti-Si, Ti-Ge, Ti-Ni, Ti-Cu, Ti-Co, and Ti-Cr systems. Work performed at the Lockheed Palo Alto Research Laboratory has shown that hardness levels of 35 to 47 R_c can be achieved in Ti-20 Cr, Ti-15 Co, and Ti-10 Ge alloys when isothermally aged following solution treatment. Hardening in this instance is derived from the development of a coherent, Widmanstätten distribution of $TiCr_2$ and Ti_2Co platelets in a beta matrix and Ti_5Ge_3 precipitates in an alpha matrix. The latter alloy was particularly effective and hardened from R_c 31 to R_c 47 on isothermal aging at 925°C.

Although none of the aforementioned methods by itself yields a hardness in the target range of R_c 60 to 65, the potential for achieving such a goal through a combination of methods appeared promising; such an approach in other alloy systems has indicated that the effects derived from different hardening mechanisms are generally additive. To test this hypothesis, the additive effects of interstitial, omega, and precipitation hardening were studied. Titanium-germanium alloys were selected as a base because of their good age-hardening response. Nitrogen was selected as the interstitial hardener, and Mn, Fe, V, and Cr were evaluated as beta stabilizers. Alloys ranging in hardness from R_c 47 to 64 were produced. The aging response, however, was lower than desired, and the hardest alloys (high germanium) were excessively brittle.

In seeking to improve the balance between toughness and hardness and to increase the response to aging, silicon was added as a replacement for part of the germanium. Of the many alloys studied, only two compositions (Ti-5Ge-2Fe-2Si-1N and Ti-5Ge-3Mn-2Si-1N) exhibited the desired combination of hardness and resistance to cracking. The hardness of these alloys ranged from R_c 56 – 59 in the solution-annealed condition (water quenched from 1000°C) to R_c 60 – 64 in the fully aged condition (2 hr at 400°C). The alloys were not sensitive to thermal shock and demonstrated relatively good resistance to fracture on impact.

The microstructure of these alloys as quenched from 1000 – 1100°C and aged at 400°C consisted of nearly equal amounts of equiaxed alpha and alpha prime (martensite). The latter is an extremely fine transformed beta structure which can be resolved at a magnification of 3000×. Typical microstructures of these alloys are shown in Figs. 23, 24, and 28. The average diameters of the alpha and prior beta grains were $10 - 20\mu$. The ultrafine structure appeared to result from growth resistrictions imposed on these phases by intermetallic compound particles. A third phase, most likely $Ti_5(Ge,Si)_3$, with an average particle size of 4μ was randomly distributed in the prior beta grains and in the boundaries between the alpha and alpha prime (martensite) regions. It is believed that the fineness of the structure, coupled with the fact that the hard and brittle compound phases do not form a continuous network, accounts for the reasonably good toughness of the alloy.

As a result of these studies, a new and unique class of wrought titanium alloys has been developed. For the first time, a wrought titanium alloy has been produced with a hardness approaching that of steel armor piercing projectiles and dual-hardness steel armor plate. The materials also have reasonable toughness. The best alloys developed to date contain less than 10% total alloy content with the compositions Ti/3–5%Ge/1–2%Si/2–4%Mn or Fe, and 0.5–1.5%N. Measured densities are in the range of 5.0 to 5.1 gm/cc (0.18–0.185 lb/in.^{3}). At a thickness of 0.3125-in. (1/2 of a 5/8-in.-thick dual-hardness plate), the alloy weighs 8.14 to 8.28 lb/ft^2. If backed with an equivalent thickness of a tough titanium alloy such as commercial Ti-8Al-1Mo-1V, a dual-hardness armor plate, 5/8 in. thick would weigh 15.3 to 15.45 lb/ft^2. This is only 3.5 to 4.75% more than unalloyed titanium and is 59 to 60% the weight of dual-hardness steel. Thus, if effective as an armor material, up to a 40% weight saving could be realized.

These initial alloy development studies conducted by the Lockheed Palo Alto Research Laboratory in 1968 were exploratory in nature and merely served to define a new class of alloys, potentially useful for dual-hardness armor. The range of compositions within which the desired combination of properties is produced and the optimum processing procedures for melting, fabrication, and heat treatment were not defined. The work required to optimize compositions and processing variables was conducted under the current contract and the results are presented in the following sections of this report.

2.1.2 Alloy Selection

The effect of compositional variations on the hardness and hardenability of the basic Ti-5Ge-2Si-3Mn(Fe)-1N alloy was studied to aid in the selection of the candidate hard-face alloys. The Ge:Si ratio was investigated first to determine the extent to which germanium could be replaced by silicon and to establish optimum ratios. Six alloys were prepared with Ge:Si ratios (based on weight percent) ranging from zero to infinity. The total alloy content ranged from 4 to 10 wt%. Substitution was made on the basis of 1.0 wt% Ge being equivalent to 2.5 wt% Si, which is the direct ratio of molecular weights of the two elements. Starting with a Ti-10Ge-3Mn-1N base, silicon was substituted for Ge on the basis of this equivalence. The alloy at zero Ge:Si ratio was Ti-4Si-3Mn-1N. Samples of rolled 0.150-in.-thick plate were solution annealed 10 min at 1100° and/or 1200°C and aged 2 hr at 400°C. The results of hardness tests are presented in Table 1.

Preliminary studies had indicated that at least 2 wt% Si was needed in a Ti-5Ge-3Mn-1N alloy for optimum aging response and high hardness. The data shown in Table 1 confirm these earlier findings. Alloys containing only Ge or Si had the lowest hardness and aging response. Maximum hardness on quenching from 1100°C occurred at a Ge:Si ratio of 2.5:1, and the maximum aging response occurred at this ratio when the alloys were quenched from 1200°C. The data indicate that a 1:1 Ge:Si ratio is capable of retaining acceptable properties in these alloys. The germanium content can be lowered to about 2.5 wt% and the silicon increased to 3.0 wt% without major detriment to the alloy. Better properties are achieved, however, at a level of 5 wt% Ge and 2 wt% Si.

A complete substitution of Si for Ge does not appear to be desirable from the standpoint of either producibility or hardness. The Ti-4Si-3Mn-1N alloy was very susceptible to thermal shock cracking on rapid heating or cooling and was difficult to hot roll. The distribution of silicides was not favorable for good ductility at high temperature. The addition of Ge to the alloy either modifies the solubility relationships or changes the morphology and distribution of the compound phase to a point where producibility is greatly improved. Silicon in excess of about 3 wt% does not appear to be desirable.

Table 1

EFFECT OF Ge:Si RATIO ON HARDNESS

Composition (wt%)(a)		Ratio Ge:Si	Hardness, R_c			
			Solution Anneal – 1100°C		Solution Anneal – 1200°C	
Ge	Si		As-Quenched	2 hr at 400°C	As-Quenched	2 hr at 400°C
0	4	0	58.8	60.0	59.2	60.4
2.4	3	0.8	59.8	60.4	60.0	61.0
5	2	2.5	61.2	62.5	57.8	60.8
6.9	1.25	5.5	–	–	58.0	60.0
7.5	1	7.5	–	–	59.4	–
10	0	Inf	58.0	58.6	–	–

(a) Base alloy, Ti-xGe-ySi-3Mn-1N.

The effect of nitrogen content variations in alloys with optimum Ge:Si ratios was subsequently studied to establish the hardness and hardenability ranges attainable. Earlier tests had indicated that nitrogen content and the solution-treatment temperature largely govern the ratio of alpha: alpha prime, which in turn controls the hardness and response to heat treatment. The data obtained are presented in Table 2 and are summarized in Fig. 1. All tests were conducted on a 0.125-in.-thick plate hot rolled from 60-g arc-melted buttons.

With low Ge, Si, and N, the alloys showed large degrees of quench hardening and little aging response. This behavior will lead to hardness gradients in thick plates. Increasing the nitrogen to 1.0 – 1.25% reduces the amount of quench hardening and increases the response to aging. Increasing the Ge content from 3 to 5% tended to decrease the aging response of the high-nitrogen alloys. Increasing the silicon from 2 to 3%, on the other hand, appeared to enhance aging at all nitrogen levels. Alloys rich in Ge and Si showed the least amount of quench hardening and the greatest amount of age hardening. As such, they should through-harden in thick sections. High hardenability will be an important factor in the development of an effective dual-hardness armor. Properties of Ge- and Si-rich alloys also seem to be less sensitive to variations in nitrogen content. This is an important factor in producibility, since tight controls over chemistry are not desirable. From all points of view, alloys with Ge and Si on the high side appear to have the best potential.

Table 2

EFFECT OF COMPOSITION ON HARDNESS AND HARDENABILITY OF EXPERIMENTAL HARD-FACE ALLOYS

Alloy No.	Composition (wt %)					Hardness, R_c (a)		
	Ge	Si	Mn	Fe	N	20 min at 950°C Air Cool	20 min at 950°C Water Quench	20 min at 950°C WQ + 2 hr at 400°C WQ
1	3	2	3	–	0.50	49.7	56.0	56.3
2	3	2	3	–	0.75	50.8	56.6	55.9
3	3	2	3	–	1.00	53.3	54.7	58.6
4	3	2	3	–	1.25	53.6	57.1	60.2
5	5	2	3	–	0.75	51.7	57.0	58.0
6	5	2	3	–	1.00	55.1	57.7	60.0
7	5	2	3	–	1.25	56.3	59.6	59.8
8	3	3	3	–	0.75	53.6	56.9	59.0
9	3	3	3	–	1.00	52.2	57.0	59.9
10	3	3	3	–	1.25	53.2	56.2	60.9
11	5	3	3	–	1.00	56.0	56.8	62.2
12	5	3	3	–	1.25	57.2	58.1	60.7
13	5	3	–	3	1.00	56.1	56.2	61.0
14	3	2	–	3	1.00	54.2	55.0	59.9
15	3	2	4	–	1.00	51.3	54.0	59.8
16	3	2	7	–	1.00	53.0	50.5	56.0

(a) Surface hardness, 0.125-in.-thick plate.

NOMINAL COMPOSITION-W/O

Ge	Si	Mn	Fe	N
3	2	3		0.5
3	2	3		0.75
3	2	3		1.0
3	2	3		1.25
5	2	3		0.75
5	2	3		1.0
5	2	3		1.25
3	3	3		0.75
3	3	3		1.0
3	3	3		1.25
5	3	3		1.0
5	3	3		1.25
3	2		3	1.0
5	3		3	1.0
3	2	4		1.0
3	2	7		1.0

SOLUTION ANNEAL – 20 MIN. 950°C
AGED – 2 HR 400°C

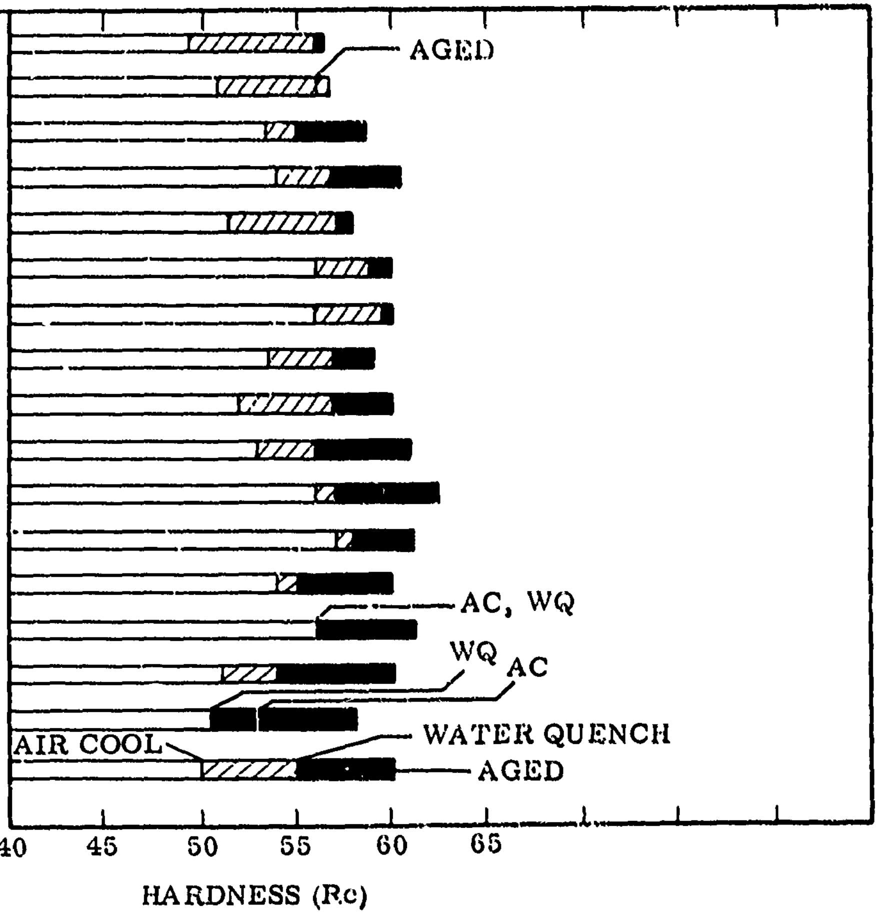


Fig. 1 Effect of Composition on the Heat-Treatment Response of Candidate Hard-Face Alloys

The substitution of iron for manganese also appears to be desirable. The iron-stabilized alloys exhibited much less quench hardening for a given Ge, Si, and N level and are expected to have superior hardenability. A slight increase in Mn content (to 4%) did not have the same effect. Increasing the Mn to 7%, on the other hand, depressed the M_s to below room temperature and produced an air-hardening alloy. The formation of omega by beta decomposition on slow cooling is believed to contribute to hardening in this case. The phase relationships in these alloys are very complex, and the response to heat treatment is difficult to rationalize in terms of composition and structure. Far more detailed studies (beyond the scope of the current program) will be required to develop an adequate understanding of their behavior.

Based on the results of these studies, three alloy compositions were selected for the manufacture of test plates for ballistic evaluation:

(1) Ti-3Ge-2Si-3Mn-0.5N
(2) Ti-5Ge-3Si-3Mn-1.0N
(3) Ti-5Ge-3Si-3Fe-1.0N

The low-nitrogen alloy is a promising candidate for a hard-face material in the R_c 55 – 60 hardness range. The alloy softens to R_c 45 – 50 on slow cooling and quench hardens to R_c 56 – 57. Little if any aging response is indicated. This alloy would be expected to have the best producibility and highest toughness by virtue of its lower nitrogen and silicon contents. It will not be particularly deep hardening and should exhibit a hardness gradient from surface to center in heavy sections. The high-nitrogen alloys are the best candidates for a hard-face material in the R_c 60 – 65 range. They have the highest surface hardness of all the alloys and have the least differential between air-cooled and water-quenched properties. As such, they should be deep hardening and have a minimum hardness gradient in quenched heavy sections. The iron-bearing alloy shows the greatest hardenability.

None of the alloys achieved hardness levels above R_c 65. It was anticipated that increased nitrogen or a higher beta stabilizer content would lead to hardness values in this range. Increasing the nitrogen from 1 to 1.25% did not have a significant effect on hardness, however, and resulted in excessive brittleness. Alloys with 1.25% nitrogen tended to crack on melting and could not be produced satisfactorily. Increasing the manganese content to 4% had a very slight hardening effect, but further additions (up to 7%) softened the alloy. Beta was retained on quenching, and the alloy hardened on slow cooling as a result of beta decomposition. It had good aging response, but the loss of quench hardening through the depression of the M_s below room temperature lowered the base hardness to an unacceptably low level.

2.1.3 Melting

The basic melting practices for the hard-face alloys were developed by nonconsumable electrode arc melting of 60-g buttons. Initially, Japanese titanium sponge (~1/4 – 1/2 mesh), refined germanium ingot, lump silicon, electrolytic manganese, and titanium nitride (18% N) powder were used as raw materials. The titanium portion of each charge was premelted into a button, sectioned in half, and each half drilled to permit incorporation of the fine TiN powder master alloy. This procedure was necessary to prevent the TiN from blowing away under the arc during melting. All charge components were then placed in the melting cavity of a water-cooled copper hearth and melted under helium at a pressure of about 0.75 – 0.8 atm. The button was melted for 2 min on each side and was then cut in half, placed on end, and remelted. This process was repeated until a button of uniform composition free of cracks was produced.

Alloying with nitrogen proved to be a major problem. Titanium nitride was difficult to dissolve, and segregated regions rich in nitrogen were formed. These regions tended to crack on cooling or on remelting and had to be eliminated. Up to eight remelts with sectioning of the button between melts were required to produce homogeneous alloys free of cracks. It was apparent that sound ingots could not be produced using this approach. Consideration was given to the preparation of a lower nitrogen content master alloy to solve this problem. However, preparation of a homogeneous master alloy presented as many difficulties. As a result, other sources of nitrogen for alloying with titanium were considered.

The other ingredients of the alloy (Si, Ge, Mn, or Fe) form lower melting alloys and/or compounds with nitrogen and were considered to be a better source for introducing this element. Since silicon nitride (Si_3N_4) is a commercial product available in a standard high-purity grade (39% N), it was selected for initial studies. Button melting tests similar to those described previously revealed that silicon nitride alloyed readily with titanium. Premelting of the titanium for alloying with the nitride, as previously discussed, was not required when Si_3N_4 was used as the nitrogen addition agent. In one test, no segregation was detected after the initial double melt (from two sides) of a 60-g charge of a Ti-5Ge-2Si-3Mn-1N alloy. Problems with ingot cracking and the necessity of repeated melting for homogenization were eliminated. Silicon nitride, therefore, was selected as the addition agent for nitrogen in preparing the alloys. No further problems with cracking as a result of nitrogen segregation were encountered in preparing 60-g buttons for the remainder of the program.

The use of silicon nitride resulted in a drop of hardness of about 3 points on the Rockwell C (R_c) hardness scale for all alloy compositions that had previously been made with a titanium nitride addition. As shown below, a Ti-5Ge-2Si-3Mn-1N alloy was notably softer in all conditions when Si_3N_4 was used as the source of nitrogen.

	Hardness – R_c		
		0.1-in. Plate	
Addition	As-Cast	Aged 20 min, 950°C WQ	Aged 2 hr, 400°C WQ
TiN	56	58	59
Si_3N_4	53.5	55	58
Si_3N_4 + Si	55.5	58	60

Since silicon is known to be an important hardener, it was postulated that a low recovery of silicon might be responsible for this effect. The charge calculation for silicon was based on full recovery of all the silicon from the nitride. If the recovery were low, the hardness would drop as shown. In one test, 50% excess silicon was added on the assumption that none of the silicon from the Si_3N_4 would be recovered. As shown above, the hardness of this alloy was identical to that of earlier alloys prepared with a TiN addition.

Samples were submitted to the Oregon Metallurgical Corporation, Albany, Ore., for chemical analysis to check the recovery of alloy additions before proceeding with preparation of the large-size ingots. As shown in Table 3, full recovery of the silicon from the nitride and from the elemental addition was realized in button melts. The silicon content was 50% above the aim which corresponds to nearly full recovery (97 – 98%) of the amount added. The assumption of a low silicon recovery obviously was wrong. The germanium recovery was low and erratic (45 – 68%), and the manganese recovery was low (69%). Nitrogen recovery was fairly good (80%).

The lower hardness attained with a Si_3N_4 addition appears to be the result of a number of factors, and far more detailed studies will be needed to determine the exact cause of the observed effect. However, since the introduction of excess silicon will offset the hardness loss without detriment to the alloy, this practice was followed in all subsequent work. With the exception of the first 18-lb trial ingot melt, no allowance was made for silicon recovery from Si_3N_4 in making the charge calculations. This leads to 50% excess silicon addition.

A 25-lb charge of Japanese Sponge, germanium ingot, silicon metal, electrolytic manganese, and silicon nitride was shipped to the Oregon Metallurgical Corporation, Albany, Ore., for trial melting of a Ti-3Ge-2Si-3Mn-0.5 N alloy. Nineteen hexagonal 2-in. by 5-in.-long compacts with alloy ingredients cored in the center were cold pressed. The charge could not be blended effectively due to the wide range of particle sizes. The compacts were MIG welded under argon into three electrodes using titanium wire without difficulty.

The three electrodes were vacuum melted into a 3-1/2-in.-diameter water-cooled copper mold to form a first-melt ingot. The first electrode was melted at the standard power input used for titanium alloys. It was difficult to form a good pool of molten metal, and the sidewall condition was poor. The core tended to melt first as a result of the low melting point alloy concentration. The melt appeared to be more viscous than Ti-6Al-4V melted under similar conditions. The second and third electrodes were melted with 50% higher power input. The pool was much larger, and the process was easier to control. The melt still appeared to be viscous. A sound ingot weighing 24.3 lb was produced.

The first [illegible] ingot was remelted into a 4-1/2-in.-diameter mold under the same conditions (50% higher power). The melt was very viscous. The ingot was hot-topped with a lower power input to reduce the pipe and sidewall flash. The ingot cracked into two pieces at the base of the pipe on cooling. The upper piece

Table 3
ALLOY RECOVERY IN MELTING OF HARD-FACE ALLOYS

Material		Composition (wt%)					
		Ge	Si	Mn	Fe	N	O
LMSC-1[(a)]	Aim	5	3	3	–	1.0	–
	Added	5.00	4.50	3.00	–	1.00	–
	Obtained	2.29	4.38	2.08	0.072	0.88	0.20
LMSC-2[(a)]	Aim	5	3	–	3	1.00	–
	Added	5.00	4.50	–	3.00	1.00	–
	Obtained	3.40	4.36	<0.10	3.13	0.80	0.19
Heat OMC 4484[(b)]	Aim	3	2	3	–	0.5	–
	Added	3.00	2.00	3.00	–	0.50	–
	Obtained	1.90	1.70	3.06	–	0.56	0.25
Heat OMC 4599-D1[(c)]	Aim	5	3	3	–	1.0	–
	Added	5.00	4.50	3.00	–	1.00	–
	Obtained	5.27	3.50	2.37	–	0.94	0.16
Heat OMC 4599-T3[(d)]	Aim	5	3	3	–	1.0	–
	Added	5.00	4.50	2.20	–	1.00	–
	Obtained	5.13	3.16	2.20	–	0.99	0.28
Heat OMC 4599-D2[(c)]	Aim	5	3	–	3	1.0	–
	Added	5.00	4.50	–	3.00	1.00	–
	Obtained	5.46	3.63	–	2.25	1.32	0.29

(a) Triple nonconsumable electrode arc melt in helium, 60-g button.
(b) Double vacuum consumable electrode arc melt ingot, 25 lb, no excess silicon added.
(c) Double vacuum consumable electrode arc melt ingot, 25 lb, excess silicon added.
(d) Triple vacuum consumable electrode arc melt ingot, 25 lb, excess silicon added.

was 2-1/2 to 3-1/2 in. long and contained all the pipe. It weighed 7.9 lb. The lower piece was 5-1/2 to 6-1/2 in. long and weighed 15.7 lb. This section was crack free and appeared to be quite sound. The total ingot weight was 23.6 lb, and the recovery of useful material was 67%. The heat was designated as OMC 4484.

The ingot sidewall was conditioned by lathe turning to a diameter of 4 in. Small voids were exposed on conditioning, and the outer portion of the ingot seemed to be rather porous. The appearance of the upper section, with an as-cast sidewall, and the lower section, with a conditioned sidewall, is shown in Fig. 2. The ingot chemistry was determined by an analysis of the final sidewall turnings, and the results are shown in Table 3 (heat OMC4484). The silicon recovery was low (85%) compared with the 60-g button melts, and the nitrogen recovery was high (100%). Manganese recovery also was good. Germanium recovery was similar to that attained in button melts. In general, the soundness, chemistry, and overall quality of the ingot were very good. It had an extremely fine equiaxed grain structure. Hardness was uniform at R_c 48 – 49 from surface to center and top to bottom.

Based on these results, it was concluded that production of the alloys in ingot form was feasible. Cracking at the pipe was most likely the result of local segregation combined with high stress resulting from rapid cooling of a heavy section. Modifications of charge ingredients and melting practices to alleviate these difficulties were made in melting the remaining alloys.

2.1.4 Fabrication

The hard-face alloys initially were rolled to plate and sheet at temperatures of 1100 – 1200°C. In this range, reductions of 15 – 20% per pass were made with ease. The small button ingots were coated with a low-melting silicate glass (MARKAL) to minimize oxidation. The high rolling temperatures, however, resulted in excessive surface contamination, and a tendency for surface checking was observed. These temperatures also are not compatible with the processing of commercial alloys that would be used in making dual-hardness composites. Much lower temperatures were expected to be required for working the composite test plates.

Rolling studies were conducted to establish the temperature and reduction schedule that could be used successfully in processing the hard-face alloys. The alloys were found to have good ductility at all temperatures above 700°C. Reductions of up to 20% per pass were made at metal temperatures ranging from 700 to 1000°C. Rolling below 700°C in a black heat range was not studied in detail. Cracking was observed on several occasions; however, rolling in these cases was made at too low a temperature. It was established that the hard-face alloy could be processed to plate and sheet at rolling temperatures and schedules normally used for processing commercial titanium alloys.

Press-forging studies were subsequently conducted to provide a basis for the initial breakdown of ingot structures by upset or barrel forging. Sixty-gram buttons were successfully reduced 50-% in thickness by upset forging on a 150-ton hydraulic press. Excessive friction between the pressure plates and the titanium alloy caused

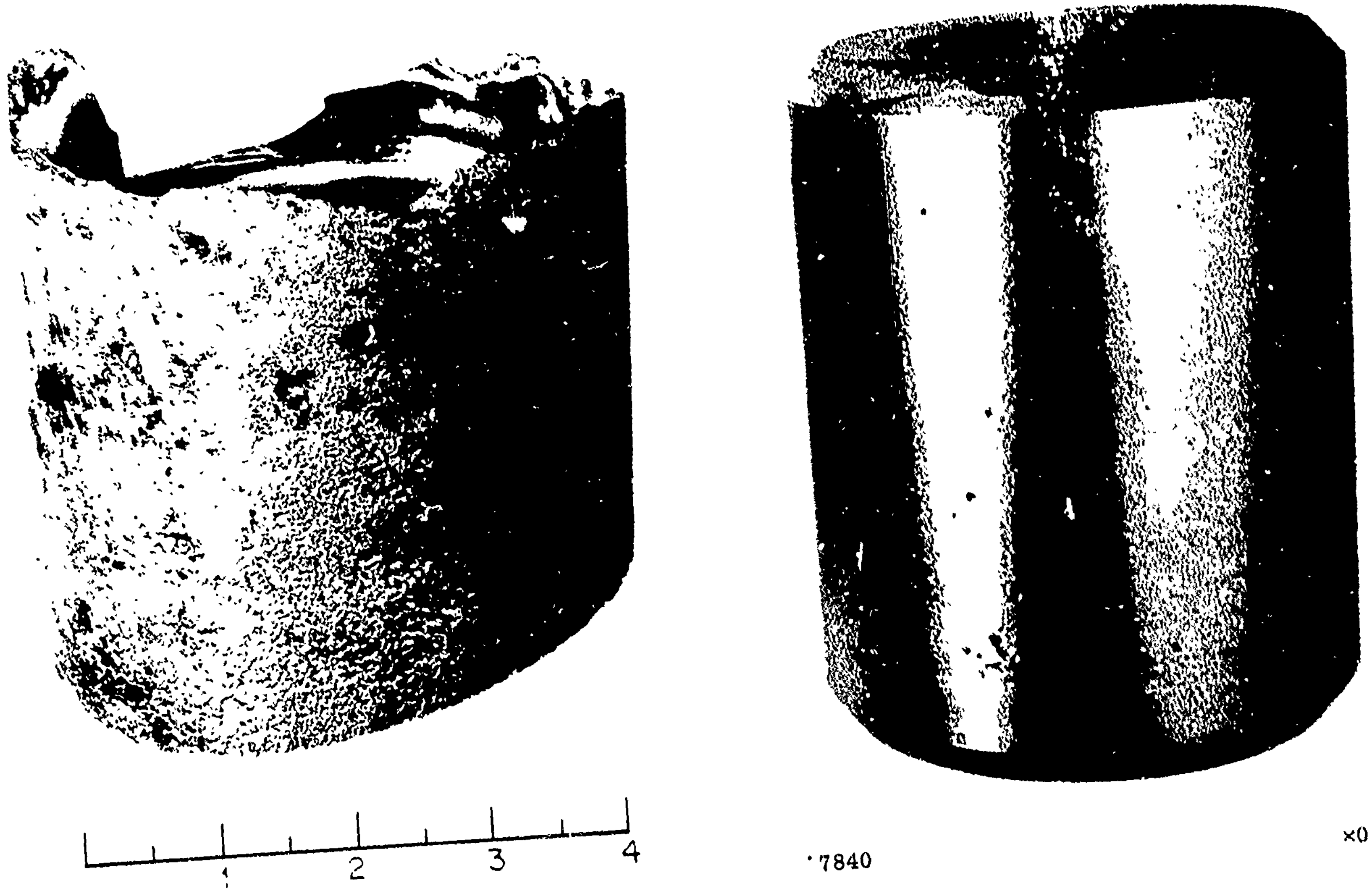


Fig. 2 Consumable Electrode Arc Melted Ingot of a Ti-3Ge-2Si-3Mn-0.5N Alloy. Heat OMC 4484. (Left: Ingot Top, as-cast. Right: Conditioned Ingot.)

deep surface tears to form in all samples. This problem was solved by using fiberglass cloth separation layers between the mating surfaces. The cloth served both to insulate and to lubricate the composite surfaces from the pressure plates. A low melting (900°C) fiberglass cloth gave best results. Samples coated with fiberglass were upset forged 50% at 1000 – 1100°C without forming surface cracks or tears. The forged plates also were free of edge tears.

The use of fiberglass cloth for surface insulation and lubrication was extended to rolling of the hard alloys. One layer of cloth placed on each side of a billet was fused onto the surface in the first pass through the mill. The material remained on the surface during all subsequent passes. Plates rolled by this practice had an excellent surface finish. Contamination was greatly reduced, and surface tears again were virtually eliminated.

Forging and rolling practices developed on the small button ingots were used in processing the large Ti-3Ge-2Si-3Mn-0.5N alloy ingot (OMC 4484). The ingot was sectioned into two billets, each 4-in. diameter by 2-3/4-in. long. The billets were coated with Markal soft glass and placed in a cold furnace. They were heated slowly to a dull red heat (about 700°C), after which, they were transferred to a second furnace held at 1100°C. This procedure was used to minimize thermal gradients that might result in cracking. The billets were soaked one hour at 1100°C and then press forged to 1-3/16-in. thick plate at a load of 125 tons on a hydraulic press. Fiberglass cover sheets were used for surface insulation and lubrication. The billets were barrel forged (axis parallel to the platens) with a total reduction of 70% in thickness. Forgeability was excellent and no edge or surface tearing occurred.

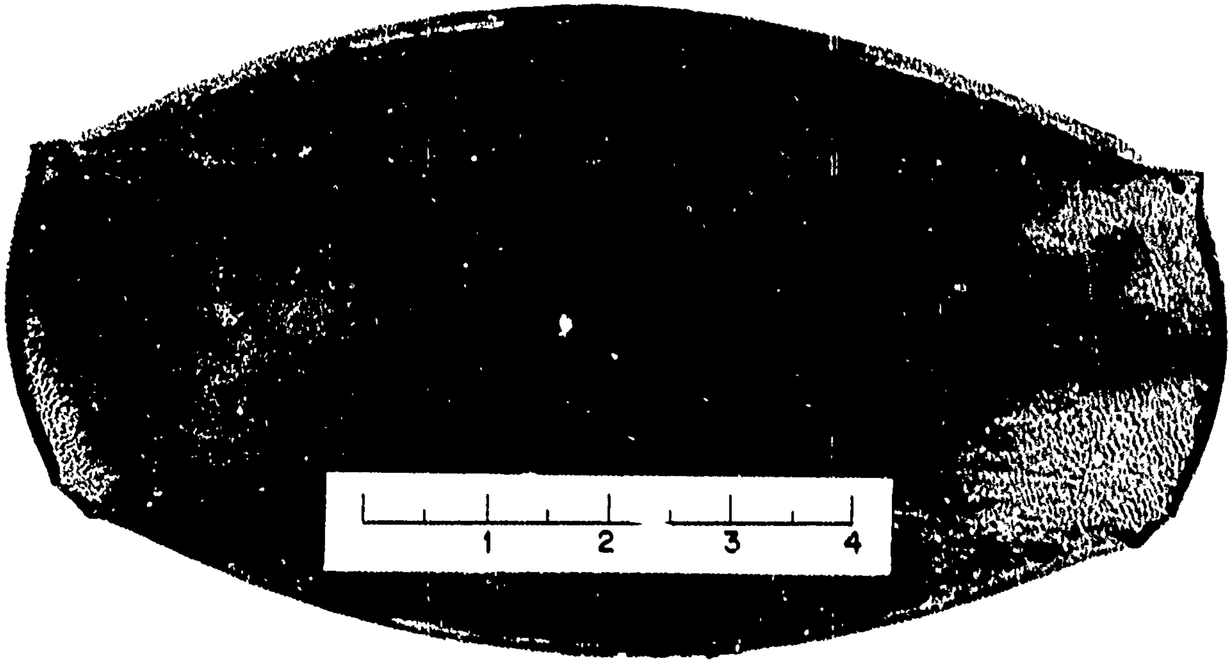
The forged plates were reheated to 1100°C and rolled to 0.83-in. thick plate on a 2-high Fenn rolling mill using reductions of 10 – 15% per pass. The billets were reheated after each pass and fiberglass cover sheets were used for surface insulation and lubrication. The plates were unidirectionalty rolled to a finish size of 5 by 10 by 0.83-in. After the last pass, they were reheated to 1100°C and straightened by pressing between flat steel plates on a 125 ton hydraulic press. The finished plates were cleaned by sand blasting. They had an excellent surface condition and were free of edge and surface tears. The appearance of the as-rolled plates is shown in Fig. 3.

The results of these studies indicate that large ingots of the hard face alloys can be processed to plate by forging and rolling using conventional equipment and procedures. The working temperature (1100°C) is higher than that used in processing commercial alloys but experience in rolling button ingots and the excellent response of the large ingot indicate that lower temperatures most likely can be used.

2.1.5 Heat Treatment

Solution anneal and aging studies were conducted with sheet and plate of the candidate hard face alloys to establish the optimum treatments. Materials prepared from both small button ingots and the large trial ingot (OMC 4484) were used in these studies. As previously mentioned, the hardness of these alloys is dependent largely on

7864 EDGE ×1



7863 TOP SURFACE ×0.65

Fig. 3 Forged and Rolled Plate of a Ti-3Ge-2Si-3Mn-0.5N Alloy, As-Rolled the nitrogen content and the solution annealing temperature. These factors govern the alpha:beta ratio in the alloy which in turn governs the base hardness as well as the aging response. The results of a study of the effect of the solution annealing temperature at three different nitrogen levels and two different germanium levels are shown in Fig. 4. The data are set forth in Table 4.

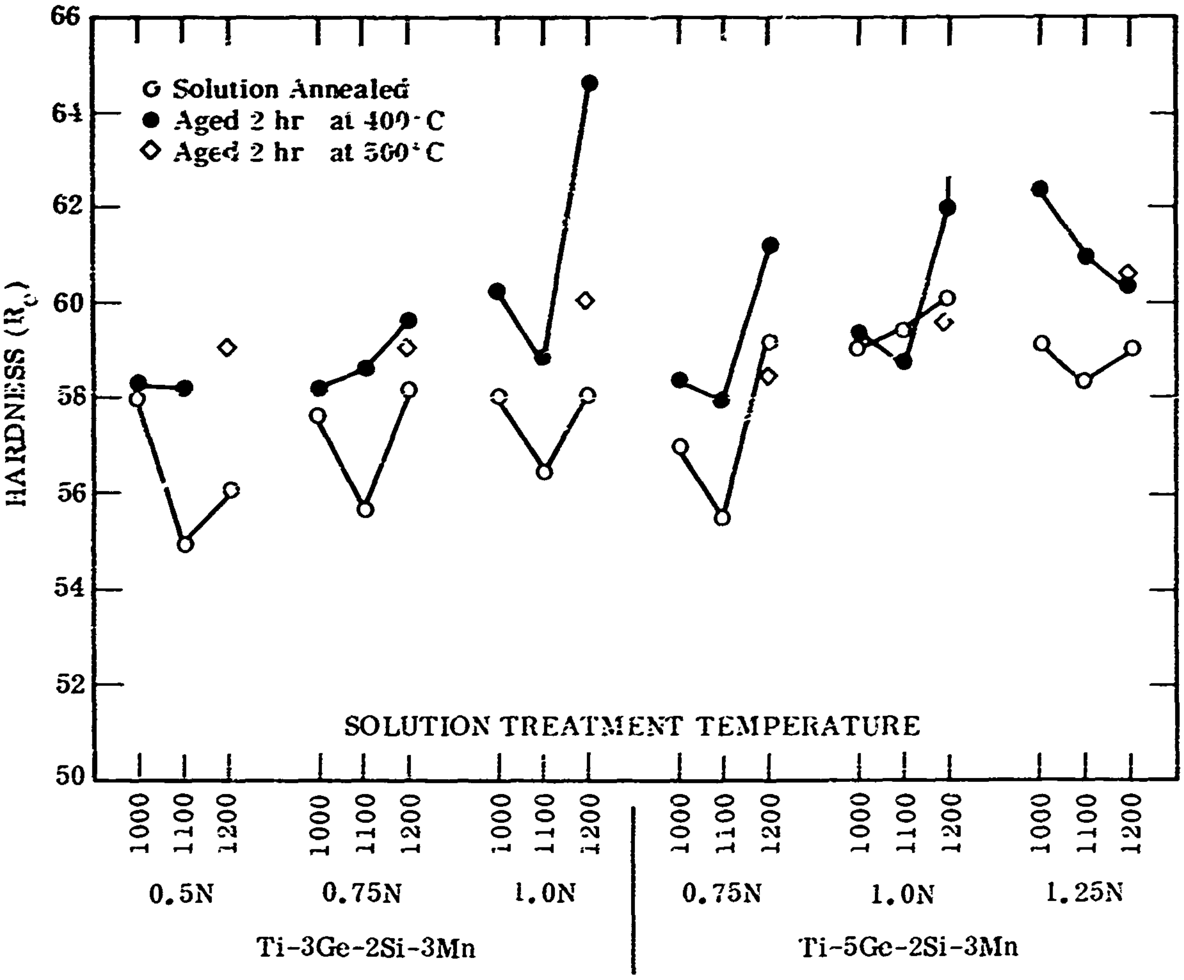


Fig. 4 Heat-Treatment Response on Hard-Face Alloys as a Function of Ge and N Content

The maximum hardness and response to aging tended to occur with a solution anneal at 1200°C. This produced a beta-rich structure. The lowest aging response tended to occur in alpha-rich structures produced by solution annealing at lower temperatures. Base hardness, however, reached a minimum on annealing at 1100°C and tended to increase with either higher or lower solution temperatures. Since the high temperatures (above 1000°C) probably will not be compatible with the back face alloys to be used in dual-hardness composites, only the lower temperature range was considered for processing the test plates. Additional tests revealed that the best combination of properties were obtained by water quenching from a solution temperature of 900 to 1000°C.

Table 4

EFFECT OF Ge AND N CONTENT ON THE HEAT-TREATMENT RESPONSE OF EXPERIMENTAL HARD-FACE ALLOYS

Alloy No.	Composition (wt %)				Solution Temperature (°C)	Hardness, R_c		
	Ge	Si	Mn	N		Solution Annealed(a)	Aged 2 hr at 400°C	Aged 2 hr at 500°C
1	3	2	3	0.5	1000	58.0	58.2	–
					1100	54.9	58.2	–
					1200	56.8	–	59.0
2	3	2	3	0.75	1000	57.6	58.2	–
					1100	55.6	58.6	–
					1200	58.4	59.6	59.4
3	3	2	3	1.00	1000	58.0	60.2	–
					1100	56.4	58.0	–
					1200	58.6	64.6	60.2
5	5	2	3	0.75	1000	57.0	58.4	–
					1100	55.4	57.9	–
					1200	59.0	61.2	58.4
6	5	2	3	1.00	1000	59.0	59.2	–
					1100	59.4	58.7	–
					1200	59.3	>62	59.5
7	5	2	3	1.25	1000	59.0	62.4	–
					1100	58.3	60.9	–
					1200	59.0	60.3	60.5

(a) 10 min at temperature, water quench.

The data shown in Fig. 4 indicate a complex interplay between composition and solution treatment. Although the hardness changes are not large, the base solution annealed hardness does increase both with increasing germanium and nitrogen content. Hardness after aging, however, largely depends on nitrogen content and the added germanium does not appear to be very effective. Detailed studies of the microstructure as influenced by compositionand heat treatment are needed to provide an understanding of this behavior.

Aging studies were conducted to find the optimum time and temperature for treating the hard face alloys. As shown in Table 5, a typical alloy achieved maximum hardness after 1 – 2 hr at 400°C. A slightly longer time was required for full hardening at 300°C (2–4 hr). At 500°C, the alloy showed no hardening after 1 to 7 hr and apparently overaged very rapidly. The results of this study confirm earlier findings that 2 hr at 400°C is a good treatment for aging alloys of this type.

Table 5

AGING RESPONSE OF Ti-3Ge-3Mn-2Si-1N ALLOY

Aging Temperature (°C)	Hardness, R_c					
	As-Quenched[a]	Aging Time (hr)				
		1	2	4	7	8
300	53.6	55.5	56.6	57.3	–	57.7
400	53.6	56.6	58.8	55.3	–	57.6
500	53.6	53.8	53.1	53.8	52.8	–

(a) Solution annealed 10 min at 900°C.

Samples of rolled plate from the large ingot of the Ti-3Ge-2Si-3Mn-0.5N alloy (OMC 4484) were solution annealed and aged by the optimum practices to check the hardenability of these alloys in heavy sections. The results of studies of 0.1-, 0.4-, and 0.8-in. plate from this alloy are summarized in Fig. 5. Base hardness on slow cooling was R_c 45 – 46. The alloy quench hardened to R_c 56 – 57 on the surface and exhibited a pronounced section thickness effect. Hardness is very dependent on cooling rate, and thick sections had a softer core on water quenching than thin sections. At 0.4 in. thick, the center was about R_c 50 – 51 when the surface was R_c 56 – 57. Hardness on aging 2 hr at 400°C increased to R_c 57 – 58 on the surface and R_c 52 – 53 in the center. Although surface hardness is in the target range (R_c 55 – 60), most of the section is below the required hardness. As previously discussed,high nitrogen alloys exhibit less quench hardening and more age hardening. It is believed that high nitrogen alloys will show a deeper hardening in thick sections. Alloys with iron as a substitute for manganese also would be expected to have far better hardenability.

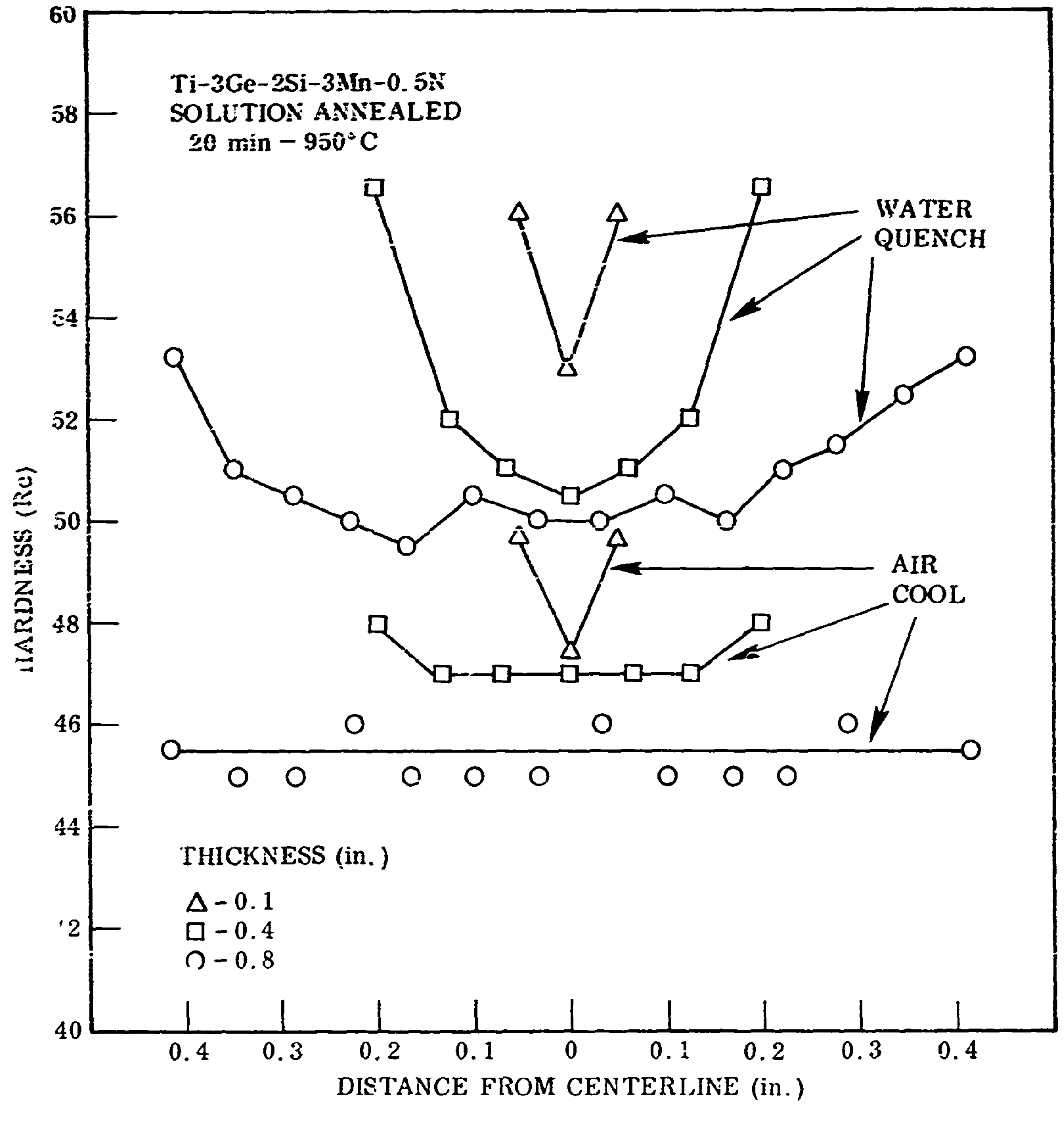


Fig. 5 Hardness Profiles in Ti-3Ge-3Mo-2Si-0.5N Alloy Plate (Heat OMC 4484) as a Function of Section Thickness and Cooling Rate

A study of the effect of solution temperature on the hardness and hardenability of the low nitrogen alloy plate (heat OMC 4484) was made to determine if the depth of hardening could be improved by a different treatment. As shown in Table 6, the hardness on slow cooling was minimum with a solution temperature of 1000°C. The alloy had a higher hardness on slow cooling from 1200°C. In addition, the aging response was greater in a sample water quenched from 1200°C. The data indicate that a very high temperature solution anneal would be required to develop high hardness and deep hardening in this alloy. Unfortunately, this type of heat treatment is not compatible with the commercial alloys considered for the back face and cannot therefore, be used to improve properties of the hard-face alloy in a dual-hardness composite.

Table 6

HEAT-TREATMENT RESPONSE OF A
Ti-3Ge-3Mn-2Si-0.5N ALLOY

Solution Temperature (°C)	Cooling Rate	Hardness, R_c	
		Solution Annealed	Aged 2 hr at 400°C
900	Air Cool	50.6	52.1
1000		49.4	50.4
1100		50.7	52.8
1200		52.8	53.9
900	Water Quench	57.1	58.0
1000		57.0	57.2
1100		50.7	52.8
1200		56.9	60.5

2.2 BACK-FACE ALLOYS

2.2.1 Alloy Selection

The alloys selected for evaluation are shown in Table 7. Three classes of alloys were considered: all alpha, near alpha (beta lean), and alpha-beta. The ideal back face alloy should combine high strength with good toughness. As shown in Table 7, the strength of the alloys increases with increasing beta content. Heat treatment response also is greatest. Toughness, on the other hand, decreases with increased beta, particularly after aging. It was believed that the optimum alloy would be found in the near alpha (beta lean) class of materials.

With the exception of the 7Al-2.5Mo alloy, all of the candidates shown in Table 7 are standard commercial alloys. The 7Al-2.5Mo alloy is a modification of the commercial 7Al-4Mo alloy designed to have greater toughness. The impact is equal to or better than that of the all alpha alloys, while the strength level is considerably higher. It was included for evaluation on the basis of its excellent combination of strength and toughness.

Table 7

CANDIDATE ALLOYS FOR BACK FACE[a]

Alloy	Heat-Treat Range		Hardness, R_c	Tensile Properties				V-Notch Impact (ft-lb)
	Solution Anneal (°C)	Age (°C)		YS (ksi)	UTS (ksi)	El. (%)	Modulus (psi × 10^{-6})	
Alpha								
Ti-5Al-2.5Sn	845	–	36	115	120	10	16–17	20–25
Near Alpha								
Ti-8Al-1Mo-1V	900	–	30	120	130	10	17–18	20–30
	900	595	–	130	140	5–10	17–18	–
Alpha-Beta								
Ti-7Al-2.5Mo	925	595	–	138	156	6	17–18	40
	985	595	–	143	160	4	17–18	15–20
Ti-6Al-4V	925–955	–	30	120	130	10	16–17	15–20
	925–955	535	–	145	160	10	16–17	10
Ti-6Al-6V-2Sn	845–915	–	32	140	150	10–20	16–17	15
	845–915	535–620	–	160	175	6–8	16–17	–

(a) Typical properties.

The desired characteristics of the back face alloy for dual-hardness armor are (1) compatibility with the hard face alloys on rolling, (2) retention of strength and toughness after heat treatments required for the front face alloy, and (3) ability to withstand ballistic impact without interfacial separation and front face spall, or back face cracking (fracture). The first and third factors were evaluated for each alloy by manufacture and ballistic testing (30 Cal-AP/M2) of 4 by 4 by 0.4-in. test panels using a Ti-3Ge-3Mn-2Si-0.5N alloy (heat OMC 4484) as the front face material. The second factor was studied by hardness and unnotched impact tests of heat treated samples (0.4 × 0.4 × 2-in.) of the back face alloys. Results of these studies are summarized in Table 8.

Table 8

EFFECT OF ROLLING TEMPERATURE AND HEAT TREATMENT ON HARDNESS AND TOUGHNESS OF BACK-FACE ALLOYS

Alloy	Rolling Temp. (°C)	Solution Temp.(a) (°C)	Aging Temp.(b) (°C)	Rockwell Hardness, R_c	Unnotched Impact (ft-lb)
Ti-5Al-2.5Sn	960	1000	–	–	128+
	960	1000	400	36	128+
	960	1030	–	–	128+
	960	1030	400	42	90
	960	1100	–	–	58,51
	960	1200	–	–	25
Ti-8Al-1Mo-1V	960	960	–	40,44	120,128+
	960	960	400	46	128+
	960	1030	–	45	58
	960	1030	400	47	54
	960	1100	–	50	20
	1000	1030	–	44	93
	1000	1030	400	44	96
	1030	1000	–	45	128+
	1030	1030	–	46	55
	1070	1000	–	43	105
	1070	1030	400	43	43
Ti-7Al-2.5Mo	960	960	400	41	128+
	960	1000	–	44	74
	960	1000	400	44	74
Ti-6Al-4V	960	960	–	44	42
	960	960	400	44	60
Ti-6Al-6V-2Sn	960	960	–	46	48
	960	960	400	50	16

(a) 20 min, water quench.
(b) 2 hr, water quench.

As anticipated, the all-alpha or beta-lean alloys had the best retention of toughness after solution annealing at 960° – 1000°C and aging at 400°C. These treatments are designed to give optimum hardness in the front face alloy. The alpha-beta Ti-7Al-2.5Mo alloy also had excellent toughness and was not embrittled by the 2 hr age at 400°C. The 6Al-6V-2Sn had significantly reduced toughness after aging as a result of omega embrittlement. It can be seen from Table 8 that the solution anneal temperature is high and the aging temperature low in comparison to the normal treatments used for alpha-beta alloys. This can result in excessive beta formation, grain coarsening, and beta embrittlement. The all alpha or beta lean alloys are far more compatible with front face alloy heat treat schedules. With the exception of the Ti-7Al-2.5Mo alloy, the high strength, alpha-beta alloys lose toughness on annealing at 960°C. The alpha or alpha-rich alloys, on the other hand, can be heated to 1030°C before toughness begins to decrease. The higher beta transus temperature most likely accounts for this behavior. If these alloys are embrittled by heating to high temperatures, toughness cannot be regained by reannealing at a lower temperature. As shown in Table 8, the Ti-8Al-1Mo-1V alloy did not regain toughness lost by heating to 1100°C on reheating to 960°C.

The results of these studies indicate that the Ti-8Al-1Mo-1V and the Ti-7Al-2.5Mo alloys are the best candidates for the back face material. Although the Ti-5Al-2.5Sn alloy has good toughness and resists embrittlement, it will have a significantly lower strength for any given heat treatment. In addition to higher strength, the other two alloys also have a higher modulus (Table 7) which is believed to be a critical factor in dual-hardness armor materials. These alloys appear to have the best combination of strength, modulus, toughness, and latitude in processing for use as the back face. They can be rolled and annealed at temperatures to 1000°C without serious degradation and are most compatible with the working and annealing practices established for the hard face alloys.

2.2.2 Composite Preparation

The final selection of the back face alloy was based on an evaluation of compatibility with the front face alloys in processing of a dual-hardness composite and on the resistance to cracking and bond separation in ballistic impact tests. Ballistic test plates, 4 by 4 by 0.4 in., were prepared from Ti-3Ge-2Si-3Mn-0.5N alloy plate (heat OMC 4484), and all of the candidate back face alloy plates (Table 7). The composites had a 50:50 ratio of hard to soft face alloys and were prepared from 0.75-in.-thick plates of each material. The plates were surface ground on each side to aid in the formation of a uniform bond.

The initial bond between the hard and soft alloys was formed by pressure bonding in a vacuum hot press. Two pressure platens, each 6 in. in diameter, were machined from a billet of Inconel 718 alloy. The plates to be bonded were positioned between the platens in a water-cooled stainless steel cylindrical tank that could be evacuated to a pressure of less than 10^{-5} Torr. A slurry of alumina powder in acetone was applied to the surface of each platen to prevent bonding to the titanium alloy plates. The platens were attached to water-cooled stainless steel push rods which extended through the vacuum chamber in O-ring sealed sliding glands. A hydraulic press was used to apply the load and a calibrated load cell was used to measure and regulate the load.

The two platens and alloy plates to be bonded were heated by induction using a 10,000 cps, 30 kW Tocco motor generator set. Temperature of the platens was measured by an L&N optical pyrometer. A Pt/Pt-10 Rh thermocouple was used to measure the temperature of the titanium alloy plates. The couple was inserted into a 0.1-in.-diameter by 0.25-in.-deep hole drilled in one side of the back-face alloy.

The bonding operation consisted of evacuating the chamber to 10^{-5} Torr and heating the pieces rapidly to the bonding temperature without an applied load. Normal heat up time was 30 min during which the system was thoroughly outgassed. When the titanium pieces reached the desired temperature, the load was applied and steady state conditions were held for one hour to produce the bond. The bonded pieces were slow cooled to room temperature (about 3 hr) in a helium atmosphere before being removed from the chamber.

Previous work on diffusion bonding of titanium alloys at these laboratories has shown that sound metallurgical bonds could be produced when held 1 hr at 790°C under a load of 800 psi. Bond strength was equal to that of the base metal, and tensile coupons containing a transverse bond failed outside the bond area. These parameters were used in producing the dual-hardness composites. The plates were heated to 760 – 790°C and held in contact for 1 hr at a pressure of 600 – 1000 psi. The platens attained a temperature of 900 – 930°C as a result of better coupling with the induction coil. A sound metallurgical bond was attained as indicated by the fact that the diffusion bonded plates could be press forged and hot rolled in air to a total reduction in thickness of 73% without bond separation.

The pressure bonded composites were reduced 30% in thickness by press forging from 1000°C. The forged plates were immediately re-heated to 700, 960, or 1000°C depending on the back face alloy and were reduced an additional 62% in thickness by rolling. The best compatibility in forging and rolling was found for the Ti-6Al-4V alloy. There was no tendency for differential working and the finished plate retained a 50:50 balance of hard and soft alloys. The Ti-6Al-6V-2Sn alloy was slightly weaker than the hard face alloy during working but in general compatibility with this material also was excellent. The all alpha or alpha rich alloys were stiffer than the hard face alloy and tended to roll differentially. Lowering the working temperature to 700°C improved compatibility of the Ti-5Al-2.5Sn alloy. The most undesirable compatibility was found for the Ti-7Al-2.5Mo alloy rolled at 1000°C. This material was very stiff and resisted deformation relative to the hard alloy. The finished composite was 29% hard face and 71% soft face as a result of the differential rolling. The data for this study with respect to relative deformation characteristics are summarized in Table 9.

Table 9

SUMMARY OF 30-cal AP/M2 BALLISTIC EVALUATION STUDIES[(a)]

Test No.	Backface Alloy	Rolling Temp. (°C)	Solution Temp.(b) (°C)	Aging Temp.(c) (°C)	Rockwell Hardness		Unnotched Impact		Thickness			Aerial Density (lb/ft²)	Comments
					Front R_c	Rear R_c	Rear(d) (ft-lb)	Composite(e) (ft-lb)	Front (in.)	Rear (in.)	Total (in.)		
D2	Ti-5Al-2.5Sn	700	960	400	56	36	128+	–	0.167 48% Hard	0.180	0.347	8.16	Complete front spall. Severe rear deformation Mild rear delamination. Projectile undamaged
D1	Ti-8Al-1Mo-1V	960	960	400	56	41	128+	68	0.165 46.5% Hard	0.190	0.355	8.08	Moderate front spall. 2 in. diameter bond separation. Moderate rear deformation. Major rear delamination. Projectile undamaged.
E3	Ti-8Al-1Mo-1V	1000	1000	400	56	42	100–128+	61, 72, 67	0.175 42.7% Hard	0.235	0.410	9.50	Moderate front spall. 2 in. diameter bond separation. Moderate rear deformation. Rear delamination. Projectile undamaged.
E4	Ti-7Al-2.5Mo	1000	1000	400	56	44	70–80	28, 49	0.120 28.9% Hard	0.295	0.415	9.60	Slight front spall. No bond separation. Little rear deformation. Mild rear delamination. Projectile undamaged.
E1	Ti-6Al-4V	960	960	400	56	44	40–60	–	0.200 50% Hard	0.200	0.403	9.38	Complete front spall. Severe rear deformation. Mild rear delamination. Projectile undamaged.
E2	Ti-6Al-6V-2Sn	950	960(f)	None	57	50	40–50	23	0.215 53.3% Hard	0.188	0.403	9.47	Through fracture, plate shattered. Extensive bond separation. Projectile fractured.

(a) Front face heat OMC 4484. Ti-3Ge-3Mn-2Si-0.5N. Standard 30 cal AP/M2, ~ 2800 fps, 0°.
(b) 20 min, water quench.
(c) 2 hr, water quench.
(d) Range based on studies of plate material, 0.4-in. square section.
(e) Impact on front face, 0.4-in. square section.
(f) Quench in brine at -2°C.

The roll bonded composites (except E-2) were solution annealed by water quenching from 960 – 1000°C as shown in Table 9. No cracking occurred. Front face hardness was 54 – 55 R_c. The plates were aged 2 hr at 400°C, and hardness increased to 56 – 57 R_c on the front face. Hardness at the interface was 52 R_c, and the back face hardness is given in Table 9. The Ti-6Al-6V-2Sn alloy was quenched from 960°C into iced brine (-2°C) in an effort to maximize back face toughness. The sample was not aged.

2.3 COMPOSITE EVALUATION

The six test plates were test fired with 30 cal AP/M2 projectiles at full load and a range of 30 ft at a Lockheed Missiles & Space Company rifle range. Velocity was estimated to be about 2800 ft/sec. The plates were supported in a wood frame at the edges and were hit at the center with 0-deg obliquity. The results are summarized in Table 9 and in Figs. 6 – 8.

The worst performance was shown by composites with Ti-6Al-4V, Ti-5Al-2.5Sn, and Ti-6Al-6V-2Sn alloys as the back face. The first two had a bond separation on impact and a complete front face spall. The back plates were excessively deformed. The composite with a Ti-6Al-6V-2Sn back face shattered on impact due to low toughness of the back face alloy. This plate, however, did fracture the projectile. The composite with a Ti-8Al-1Mo-1V alloy gave somewhat better results. A 2-in.-diameter arc around the impact point underwent bond separation and spalled. The backup plate was cracked in the plane of the sheet (delamination). The best performance was exhibited by the plate with a Ti-7Al-2.5Mo backup. No interfacial separation occurred and only a small 1-in. diameter chip spalled from the front face (Fig. 6). The back face was undeformed and no delamination in the plane of the plate was observed.

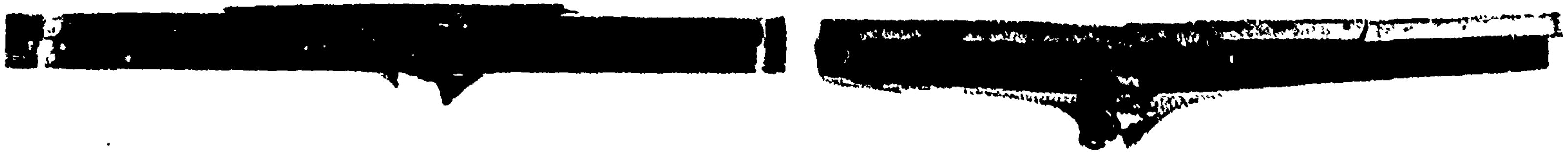
Impact bars (0.4 by 0.4 by 2) were cut from the test panels and broken by hitting the hard face in a Manlabs compound pendulum machine. Results are summarized in Table 9. The toughness of the composite was roughly half that of the back face alloy tested in the same manner. In all cases, cracks initiated in the hard face alloy terminated at the bond line.

The results indicated the Ti-7Al-2.5Mo alloy to be the best backup material with the Ti-8Al-1Mo-1V alloy a second choice. The superiority of these materials is believed to be due to the large amount of differential working that occurs during press forging and rolling. The bond integrity and ability to resist spall appears to increase in direct proportion to the amount of differential working that occurs. The high modulus of these two alloys relative to other titanium alloys also may be a contributing factor.

Two additional composite ballistic test panels containing a hard front face of Ti-3Ge-3Mn-2Si-0.5N and a back face of Ti-7Al-2.5Mo were fabricated to confirm the results of these studies. Both composites were initially formed by diffusion bonding a 0.7-in.-thick hard face plate to a 0.6-in.-thick backup plate (54% hard) in

×1

8591



×1

8592

D-1 Ti-3Ge-2Si-3Mn-0.5N on Ti-8Al-1Mo-1V D-2 Ti-3Ge-2Si-3Mn-0.5N on Ti-5Al-2.5Sn

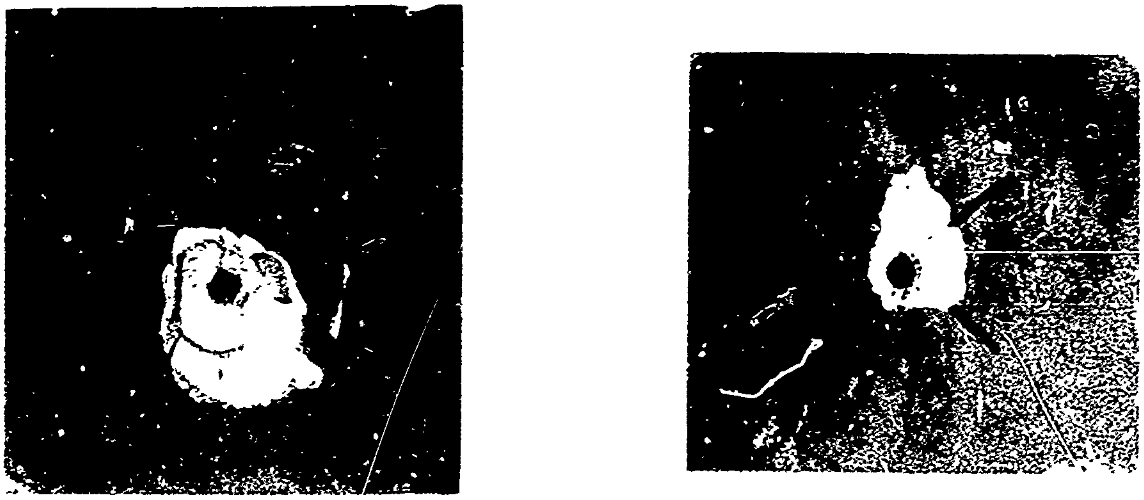
Fig. 6 Results of Bond Integrity and Backface Alloy Ballistic Tests, Alpha and Near-Alpha Backface Alloys

8750 ×0.6 8750 ×0.6

E-1. Ti-3Ge-3Mn-2Si-0.5N
on Ti-6Al-4V

E-2 Ti-3Ge-3Mn-2Si-0.5N
on Ti-6Al-6V-2Sn

Front (Hard) Face



8750 ×0.6 8750 ×0.6

E-3 Ti-3Ge-2Si-3Mn-0.5N
on Ti-8Al-1Mo-1V

E-4. Ti-3Ge-2Si-3Mn-0.5N
on Ti-7Al-2.5Mo

Front (Hard) Face

Fig. 7 Results of Bond Integrity and Backface Alloy Ballistic Tests
Near-Alpha and Alpha-Beta Backface Alloys

8751 ×0.6

E-1. Ti-3Ge-2Si-3Mn-0.5N
on Ti-6Al-4V

8751 ×0.6

E-2. Ti-3Ge-2Si-3Mn-0.5N
on Ti-6Al-6V-2Sn

Backface

8751 ×0.6

E-3. Ti-3Ge-2Si-3Mn-0.5N
on Ti-8Al-1V-1Mo

8751 ×0.6

E-4. Ti-3Ge-2Si-3Mn-0.5N
on Ti-7Al-2.5Mo

Backface

Fig. 8 Results of Bond Integrity and Backface Alloy Ballistic Tests. Near-Alpha and Alpha-Beta Backface Alloys vacuum for 1 hr at 780°C under an applied pressure of 500 psi. Following bonding, one composite was press forged 30% at 950°C, while the other composite was press forged a similar amount at 1050°C. Both composites were subsequently cross rolled approximately 53% at the same temperature used for forging into 0.4-in.-thick plates. These plates were then surface conditioned to yield 4 by 4 by 0.4-in. ballistic test panels. Based on thickness measurements, front to back face thickness ratios of 0.73:1 and 1:1 were obtained in the plates rolled at 950° and 1050°C, respectively. The starting ratio was 1.17:1. The greatest amount of differential working occurred at 950°C where the back face alloy is predominantly high strength alpha.

The composite panels were heat treated in accordance with practice used for the composite with a Ti-6Al-6V-2Sn back face to develop maximum hardness in the front face. Heat treatment in this case consisted of solution treating 20 min at 1000°C, quenching in CO_2 cooled brine at -2°C and aging 2 hr at 400°C. The front face, however, only hardened to R_c 53 which is 3 – 4 points lower than expected. The ballistic tests were conducted with the standard 0.30 cal AP/M2 projectile (i.e., muzzle velocity of ~2800 ft/sec) at 0-deg obliquity and a range of 30 ft. Table 10 summarizes the results of the rolling compatibility and ballistic evaluation studies, and cross sections of the ballistic test panels are presented in Fig. 9. For comparison, a cross section from a similar composite test panel containing a back face alloy of Ti-8Al-1Mo-1V is also shown in Fig. 10.

The data and test panel cross sections indicate that the Ti-3Ge-3Mn-2Si-0.3N/Ti-7Al-2.5Mo composite which had the greatest differential working (i.e., composite processed at 950°C) exhibited the best performance. The front face showed only a slight degree of spall and minor bond separation, while little or no deformation or delamination was detected in the back face. In contrast to this behavior, the composite with a 1:1 thickness ratio (worked from 1000°C) sustained considerably more front face spall and bond separation. Furthermore, the back face showed a slight tendency to delaminate.

The ballistic performance of the Ti-7Al-2.5Mo composite rolled from 950°C also was superior to that for a differentially rolled composite fabricated previously with a Ti-8Al-1Mo-1V back face alloy having the same ratio of front to rear face thickness (plate E-3). Here again, the former composite demonstrated less front face spall and bond separation and significantly less back face deformation and delamination (Fig. 10). Results of the rolling compatibility and ballistic evaluation studies confirmed earlier findings which indicated that differential rolling enhanced bond integrity and spall resistance and that the Ti-7Al-2.5Mo back face alloy was ballistically superior to the Ti-8Al-1Mo-1V composition under equivalent processing and test conditions. No difficulty was foreseen for incorporating differential rolling into the processing schedule for manufacturing the final ballistic test panels. In this case, a front to back face thickness ratio of 1.5:1 was anticipated to yield a 1:1 ratio in a 0.625-in.-thick test panel following 75% reduction of the diffusion bonded composite by forging and rolling.

Table 10

EVALUATION DATA FOR Ti-3Ge-3Mn-2Si-0.5N/Ti-7Al-2.5Mo DUAL-HARDNESS COMPOSITE ARMOR

Test No.	Rolling Temperature (°C)	Thickness (in.)			Front to Back Face Thickness Ratio	Aerial Density (lb/ft^2)	Hardness (R_c)				Ballistic Behavior$^{(c)}$
							Solution Treated$^{(a)}$		Aged$^{(b)}$		
		Front Face	Back Face	Total			Front Face	Back Face	Front Face	Back Face	
F-1	950	0.175 (42% hard face)	0.240	0.415	0.73:1	9.7	53	42	57	44	Slight front spall and bond separation. Little rear deformation. No rear delamination, projectile undamaged.
F-2	1050	0.204	0.209	0.413	1:1	9.7	53	44	56	44	Moderate front spall. Extensive bond separation. Moderate rear deformation. Mild rear delamination, projectile undamaged.

(a) 20 min at 1000°C followed by iced brine quench (-2°C)

(b) 2 hr at 400°C

(c) Standard 0.30 cal AP/M2 projectile, ~2800 fps, 0° obliquity

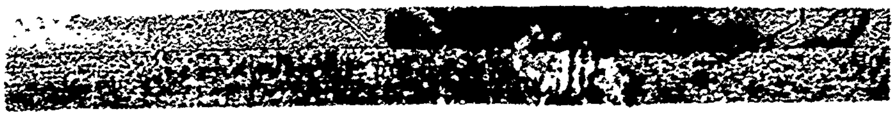
E-3. Ti-3Ge-2Si-3Mn-0.5N on Ti-8Al-1Mo-1V

E-4. Ti-3Ge-2Si-3Mn-0.5N on Ti-7Al-2.5Mo

Fig. 9 Cross Section of Bond Integrity and Backface Alloy Ballistic Test Panels E-3 and E-4

8801 ×1.25

E-3. Ti-3Ge-2Si-3Mn-0.5N on Ti-8Al-1Mo-1V
Rolled From 950°C



8801 ×1.25

F-1. Ti-3Ge-2Si-3Mn-0.5N on Ti-7Al-2.5Mo
Rolled From 950°C

8801 ×1.25

F-2. Ti-3Ge-2Si-3Mn-0.5N on Ti-7Al-2.5Mo
Rolled From 1050°C

Fig. 10 Cross Section of Bond Integrity and Backface Alloy Ballistic Test Panels E-3, F-1, and F-2

Section 3

MATERIALS PREPARATION

3.1 HARD-FACE ALLOYS

Three different compositions were selected as front-face alloys for ballistic test plates:

(1) Ti-3Ge-2Si-3Mn-0.5N
(2) Ti-5Ge-3Si-3Mn-1.0N
(3) Ti-5Ge-3Si-3Fe-1.0N

The first alloy is the composition prepared as the trial melt ingot (heat OMC 4484) for use in development studies of processing and properties. Sufficient 0.8-in. thick material remained from which to make one 6 by 6 by 0.625-in. test plate of a 50:50 dual-hardness composite for final ballistic evaluation. The second and third alloys were selected as more promising compositions for the hard-face, and sufficient material was prepared for the manufacture of twelve 6 by 6 by 0.625-in. ballistic test panels.

It was planned to produce four test plates of alloy (2) in the solution annealed condition (R_C 55 – 60) and four test plates in the fully aged condition (R_C 60 – 65). Four test panels also were to be produced from alloy (3) in the fully aged condition (R_C 60 – 65). This material was expected to have the greatest depth of hardness.

The material selected for the back-face alloy in all 13 proposed test plates was Ti-7Al-2.5Mo. The work conducted in preparing these materials and in producing the dual-hardness composites is described in this section of the report.

3.1.1 Melting

The raw materials used in preparing three 25-lb ingots of the two additional hard-face alloys were sodium-reduced Ti sponge (Reactive Metals Company), electrolytic manganese, silicon metal, Armco ingot iron, refined germanium ingot, and lump silicon nitride. All the alloy additions were crushed to about 1/4-in. mesh which was the approximate size range of the titanium sponge. The charge weights were calculated on 100% recovery of all additions except for silicon from the silicon nitride. No allowance was made for recovery of silicon from this source; consequently, each melt contained approximately 50 percent excess silicon.

The alloys were melted by the Oregon Metallurgical Corporation, Albany, Oregon. The procedures used, basically, were those described in Section 2 for the preparation of the first 25-lb trial melt. An improved hot-topping practice was employed, and the cooling water to the copper mold was turned off on completion of the melt to reduce the cooling rate in an effort to minimize ingot cracking.

The first ingot to be melted was the Ti-5Ge-3Si-3Mn-1N alloy, heat OMC 4599 D1. The pressed electrode was melted into a 3.5-in. diameter mold. The ingot cracked into eight pieces on cooling. As shown in Fig. 11, the outer rim of the ingot was extremely porous and contained many large gas holes. Large pieces of unmelted silicon nitride also were visible as inclusions in the ingot. The ingot pieces were subsequently tack welded together successfully and remelted in a 4.5-in.-diameter copper mold. The second melt ingot appeared to be crack free. When the outersurface was machined, however, the ingot sidewall was found to be extremely porous.

An additional 1/2 in. was removed from the diameter by surface grinding; however, the extensive porosity remained. It appeared that the porosity was very deep and that the ingot would not be suitable for processing. Consequently, the ingot was returned to Oregon Metallurgical Corporation for remelting.

The third melt of this ingot was made in a 6-in.-diameter mold. The ingot again cracked into several pieces on cooling. The pieces were tack welded together and remelted a fourth time in a 4.5-in. mold. The ingot was carefully hot topped and cooled very slowly to room temperature. It appeared to have far less porosity, and no cracks were exhibited on the external surfaces. The upper "piped" section was cropped off, and the remainder of the ingot was cut into two sections, each 4.5 in. in diameter by 3 in. long. A large "center-burst" (crack) was found extending from the top to the bottom and from side to side in both halves. The rest of the ingot appeared to be crack free. Some porosity was still evident, but it was very shallow (less than 0.25 in. deep). It was estimated that two to four test plates could be prepared from this ingot, depending upon how the large "center-burst" propagated in forging.

The second ingot to be melted was the Ti-5Ge-3Si-3Fe-1N alloy, heat OMC 4599 D2. The ingot was double vacuum arc melted using the procedures described in Section 2. During the second melt, the electrode broke and dropped into the molten pool shortly after the start of melting. More than a third of the ingot had to be scrapped. The remaining sound material was side laced with unalloyed titanium for added strength and was melted successfully into a 4.5-in.-diameter ingot.

The surface of the ingot was found to be cracked and very porous when conditioned by lathe turning. The ingot was sectioned longitudinally into two halves to determine the extent of the cracking and the porosity. Results are shown in Fig. 12. The porosity extended to a depth of 0.5 in. and a deep pipe extended to about one-third of the total ingot length. Cracks emanate transversely and diagonally from the "piped" area. Examination indicated that sufficient material might be salvaged for the manufacture of two test plates from this alloy.

The third ingot to be melted was another heat of the Ti-5Ge-3Si-3Mn-1N alloy, heat OMC 4599 T3. The alloy was double vacuum arc melted into a 4.5-in.-diameter mold by the basic practice previously described (Section 2). The second melt ingot was found to be cracked, and a third melt was made in a 6-in.-diameter mold. This ingot had a deep pipe and a large centerline crack extending from top to bottom and side to side (Fig. 13). Shallow surface porosity also was in evidence.

Fig. 11 First Melt Ingot of Ti-5Ge-3Mn-3Si-1N Alloy
Heat OMC 4599 D1

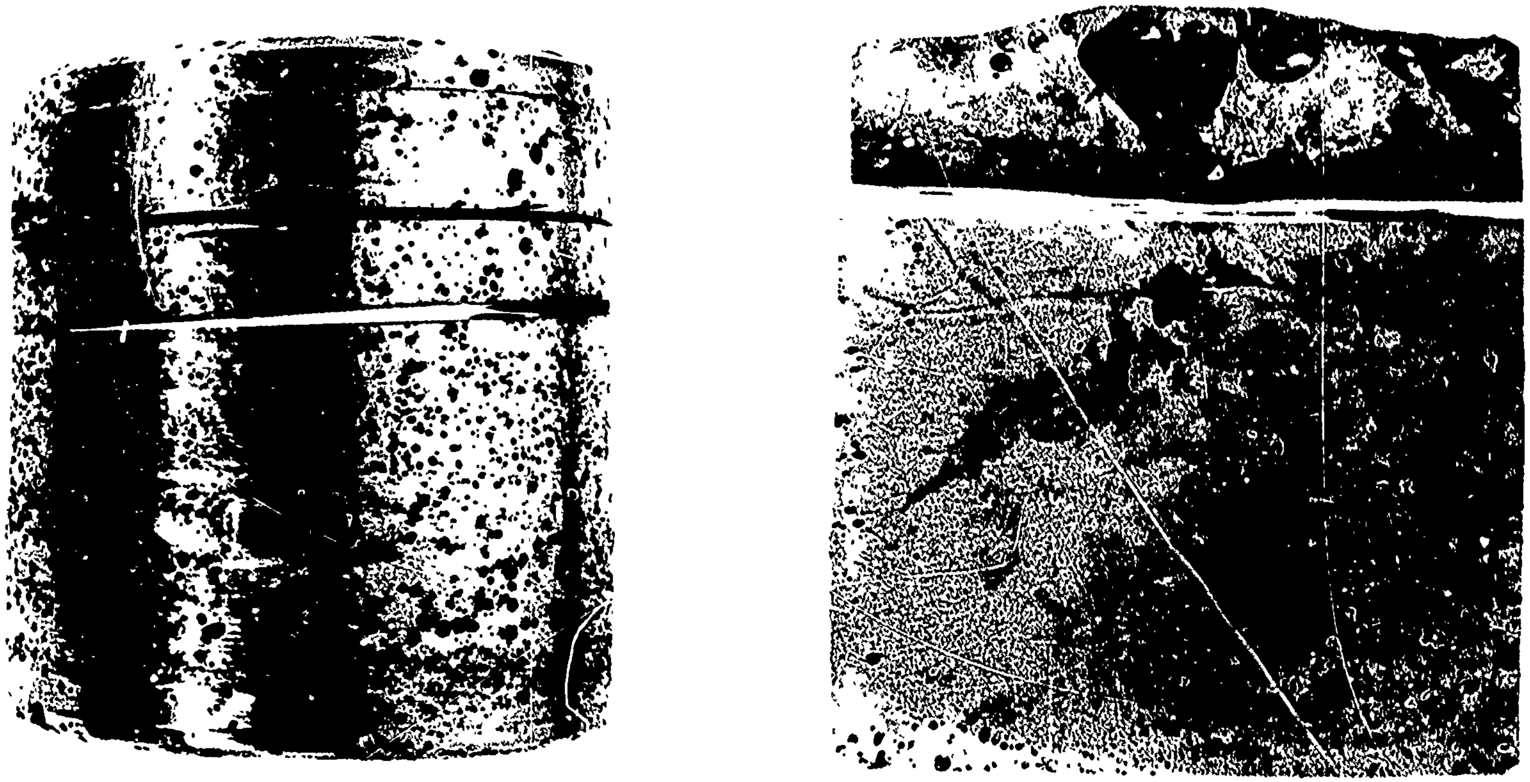
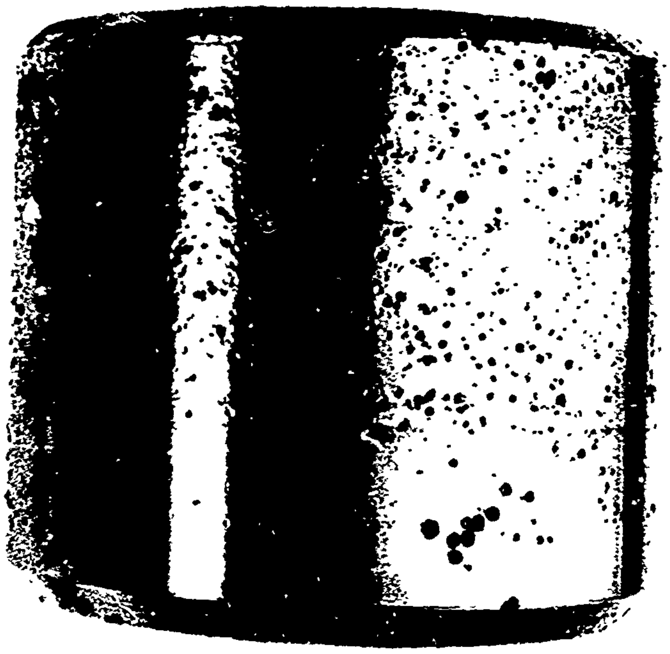


Fig. 12 Machined Sidewall and Cross Section of Double Arc Melted Ti-5Ge-3Si-3Fe-1N Ingot, Heat OMC 4599 D2



9167

×0.67

9167

×0.67

Fig. 13 Machined Sidewall and Cross Section of Triple Arc Melted Ti-5Ge-3Si-3Mn-1N Ingot. Heat OMC 4599-T3

The ingot was sectioned into two halves longitudinally along the centerline crack. Each half was found to be sound and free of secondary cracks. The porosity was very shallow, and the lower two-thirds of the ingot was quite sound. The upper "piped" section was cropped off, and sufficient sound material remained for the preparation of four test plates.

The final chemical analysis of the three ingots is given in Table 3. It can be seen that the germanium recovery was much higher than that observed in previous melts (100%), while the silicon recovery was much lower. The actual silicon and germanium residuals were very close to the desired nominal composition. The residual germanium content, however, appeared to be much higher than that obtained in the button ingots and in the first trial melt ingot. Some of these variations may be due to analytical difficulties in determining the residual germanium. The nitrogen content was very close to nominal in the two manganese-bearing alloys (D1, T3) but was abnormally high in the iron-containing alloy (D2). No explanation for this result can be made at this time.

3.1.2 Forging

The sound pieces from each ingot were coated with Markal soft glass and heated to 1050°C for press forging to plate. Fiberglass cloth cover sheets were used for surface insulation and lubrication. The two pieces from heat D1 (Ti-5Ge-3Si-3Mn-1N) containing the deep centerline crack were barrel forged with the crack in a vertical position. This was done to prevent the crack from forming a flaw in the center of the finished plate.

The upper half of the ingot which contained most of the crack, split badly on press forging. No useful material could be salvaged for rolling to sheet. The lower half of the ingot, which contained only a portion of the crack, forged well. The pieces were reduced to 2-in.-thick plate in one pressing (55% reduction). Two 3 by 3 by 2-in. pieces, which appeared suitable for further processing, were cut from the forged plate. These pieces were used to form test plates TP-7 and TP-8.

The two halves of ingot T3 (Ti-5Ge-3Si-3Mn-1N) were barrel forged to 1.75-in. thick plate (42% reduction). This material worked very well, and two plates of good quality were produced. The plates were cut into four pieces, each 3 by 3 by 1.5 in. These sections were used in fabricating test plates TP-1 through TP-4.

The two ingot halves of heat T3 (Ti-5Ge-3Si-3Fe-1N) were upset forged to 1.75-in.-thick plate (45% reduction). The workability of this alloy was poor, and extensive cracking occurred. Two plates measuring 2 by 4 by 0.75 in. thick were all that could be gained from this heat for further processing. These plates were used to form test plate TP-6.

The surfaces of all the forged test plates were conditioned by Blanchard grinding to remove all cracks and porosity. They were ground flat and parallel to prepare the pieces for diffusion bonding to the back-face alloy. An excellent finish was obtained on all surfaces, and no problems were encountered in grinding the alloys. The edges of all plates were conditioned by polishing on a water cooled, SiC abrasive belt. Deep cracks and porosity were conditioned locally by spot grinding. All visible cracks and defects were removed before bonding the forged plates to the back-face alloy plates. The thicknesses of the front-face alloy pieces after conditioning are set forth in Table 11.

3.2 BACK-FACE ALLOYS

Hot-rolled plate from an experimental 1 ton heat of Ti-7Al-2.5Mo produced for the Lockheed Missiles & Space Company was used for the back-face alloy of each test plate. Eight pieces 4 by 4 by 1.25 in. were cut for bonding to the hard-face alloy plates. The 4 by 4 surfaces were machined flat and parallel. The thickness of each piece was adjusted to the front-face alloy thickness in such a way as to provide an approximate balance of 60% hard face and 40% soft face.

The exact thicknesses of the rear-face pieces and the front-to-rear thickness ratios are listed in Table 11. The 60:40 ratio was selected on the basis of anticipated differential rolling behavior of the two alloys at 950 – 1000°C, as discussed in Section 2. It was estimated that the hard-face alloy would deform more than the backup alloy and that the final ratio would be 50:50 in the 0.625 in. thick composite test plates.

3.3 DUAL-HARDNESS TEST PLATES

3.3.1 Diffusion Bonding

The front- and rear-face alloy plates were pressure diffusion bonded using the equipment and procedures described in Section 2 of this report. The pieces were pressed between heated Inconel platens at pressures of 500 to 1000 psi for 1 hr at temperatures of 750° to 800°C. The exact conditions used in bonding each composite are summarized in Table 11. The appearance of an as-bonded alloy composite (TP-2) is shown in Fig. 14. All the pieces bonded well, and no lack of bonding at the edges was indicated. The back-face alloy was trimmed to conform exactly to the front-face alloy dimensions after bonding. This facilitated inspection of the bond line and produced a material more suitable for forging and rolling.

The two pieces of the Ti-5Ge-3Si-3Fe-1N alloy (4599 D2) were not of sufficient thickness to make two test plates. A double diffusion bond was made with these plates. The two pieces were simultaneously pressure bonded to each other and to a Ti-7Al-2.5Mo back-face plate in one operation. The bonded composite (TP-6) contained two bond lines, one of which was in the center of the hard-face alloy and appeared to be of excellent quality.

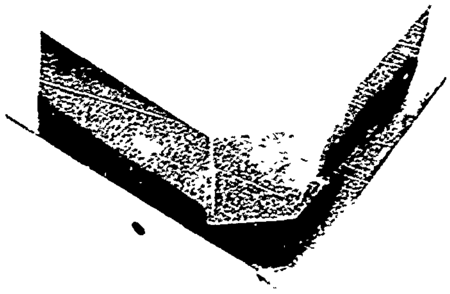
3.3.2 Forging and Rolling

With the exception of TP-5 (Ti-3Ge-2Si-3Mn-0.5N), all diffusion-bonded composites were reduced about 40% in thickness by press forging from a temperature of 1050°C.

Table 11

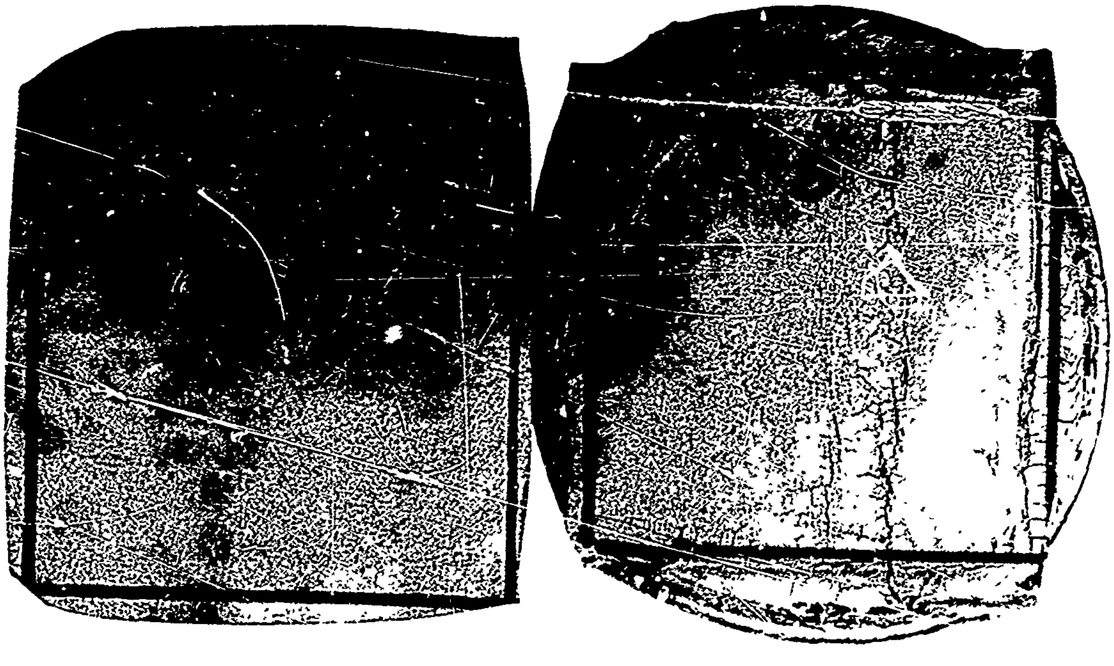
SUMMARY PROCESSING DATA OF DUAL-HARDNESS TITANIUM ALLOY COMPOSITE BALLISTIC TEST PANELS

Nominal Front Face Alloy Composition (wt%)	Heat No.	Test Panel No.	Starting Thickness (in.)			Front to Rear Face Thickness Ratio	Diffusion Bonding		Press Forging			Rolling			Comments
			Front Face	Rear Face	Total		Temp. (°C)	Pressure (psi)	Temp. (°C)	Thick. (in.)	Red. in Thick (%)	Temp. (°C)	Thick. (in.)	Red. in Thick. (%)	
Ti-3Ge-3Mn-2Si-0.5N	4484	TP-5	0.671	0.562	1.233	1.19:1(54% Hard)	790	500	-	-	-	950	0.668	45.7	No edge or surface defects.
Ti-5Ge-3Mn-3Si-1.0N	4599-T3	TP-1	1.232	0.920	2.152	1.34:1(57% Hard)	794	890	1050	1.275	39.9	950	0.667	47.6	Some edge and surface cracking.
Ti-5Ge-3Mn-3Si-1.0N	4599-T3	TP-2	1.230	0.920	2.150	1.34:1(57% Hard)	777	790	1050	1.275	40.6	950	0.672	47.3	Some edge and surface cracking.
Ti-5Ge-3Mn-3Si-1.0N	4599-T3	TP-3	1.405	0.801	2.206	1.76:1(64% Hard)	765	805	1050	1.370	37.9	950	0.676	50.6	Edge and surface cracks.
Ti-5Ge-3Mn-3Si-1.0N	4599-T3	TP-4	1.402	0.801	2.203	1.75:1(64% Hard)	794	818	1050	1.310	40.5	950	0.675	48.4	Edge and surface cracks.
Ti-5Ge-3Mn-3Si-1.0N	4599-D1	TP-7	1.450	1.125	2.575	1.29:1(56% Hard)	780	920	1050	1.500	41.7	950	0.667	55.5	Bond separation at edges. Edge and surface cracking.
Ti-5Ge-3Mn-3Si-1.0N	4599-D1	TP-8	1.460	1.125	2.585	1.30:1(56% Hard)	765	935	1050	1.500	42.0	950	0.659	56.1	Bond separation at edges. Edge and surface cracking.
Ti-5Ge-3Fe-3Si-1.0N	4599-D2	TP-6	1.585	1.125	2.710	1.41:1(59% Hard)	755	1000	1050	1.500	44.8	950	0.662	55.9	Edge and surface cracks.



9713 ×0.5

TP-2, Ti-5Ge-3Mn-3Si-1N ON Ti-7Al-2.5Mo
As-Diffusion Bonded



9400 ×0.5

TP-5, Ti-3Ge-2Si-3Mn-0.5N on Ti-7Al-2.5Mo

TP-8, Ti-5Ge-3Si-3Mn-1N on Ti-7Al-2.5Mo

As-Rolled

Fig. 14 Appearance of Ballistic Test Plates, As-Diffusion Bonded and As-Rolled

The final thickness of the composites and exact reductions are shown in Table 11. Composite TP-5 was not forged but was rolled directly to finished size. Fiberglass cover sheets were employed during forging for surface lubrication and heat insulation purposes.

Composites TP-1 through TP-4 were forged between mild steel anvils. It was found that the anvils deformed and prevented lateral movement along the contact surfaces. This caused rather deep surface tears to form in a square pattern that outlined the original shape before forging. This condition was eliminated in the three remaining forgings (TP-6, 7, 8) by the use of tungsten pressure plates. The tungsten did not deform, and the surface condition was vastly improved.

In general, all the diffusion-bonded composites forged well, and no major cracking was encountered. Composites TP-7 and 8 were found to have a poor bond along the edges, and some bond separation occurred on deformation.

None of the alloys showed any tendency toward differential working on forging. Unlike earlier experience with the Ti-3Ge-2Si-3Mn-0.5N alloy, the 5Ge-1N alloys used to produce these composite test plates appeared to deform to about the same extent as the Ti-7Al-2.5Mo backup alloy when press forged from 1050°C. As a result, it was decided to roll from a lower temperature in order to obtain the differential working required to finish up with a 50:50 composite test plate.

All the forged composite plates were reduced 45 – 55% in thickness to 0.650 to 0.675-in.-thick plate by rolling from 950°C. The exact reductions and final thicknesses are given in Table 11. Fiberglass cover sheets were again used for surface insulation and lubrication during rolling. All plates were reduced to about 1.0 in. thick by rolling in one direction, whereupon they were rotated 90 deg and finish rolled to a thickness of 0.650 to 0.675 in. The amount of cross-rolling was regulated to produce a finished plate about 6 by 6 in. square. The plates were reheated to 950°C after the last pass and straightened by pressing between steel pressure plates on a 125-ton hydraulic press.

All the composite plates rolled well, and no major problems were encountered. Plates TP-7 and 8 experienced further bond separation along one edge on rolling and exhibited the least desirable surface appearance of all the plates. Plate TP-5 rolled without the formation of any surface defects on either surface and had a surface quality equal to or better than that of commercial titanium alloys.

The appearance of this plate is shown in Fig. 14. The appearance of a typical high Ge, high N alloy test plate, TP-3, also is shown for comparison. Small surface tears tended to form on the hard-face alloys during rolling. This was due to the fact that the fiberglass mats did not soften sufficiently at 950°C to provide adequate lubrication during rolling. Better surfaces were obtained with these alloys when a thin coating of viscous glass remained on the surface during rolling.

Test plate TP-5 behaved exactly as expected during rolling with respect to differential working. At 950°C, the front-face alloy was weaker than the back-face alloy and rolled to a greater extent. As shown in Table 12 the final ratio of hard to soft alloys was 50:50 as predicted. None of the other hard-face alloys behaved in this manner. The high Ge, high N alloys were found to have the same deformation behavior as the back-face alloy at 950°C. These materials either have a higher ratio of alpha to beta than the low Ge, N alloy or have a greater degree of solid solution hardening at high temperatures. As a result, no differential rolling occurred, and the final front to rear face ratios were about the same as the initial values (60:40). The exact values are given in Table 12.

All the test plates were cropped to 5 by 5-in.-square, and local defects were removed from the front and rear surfaces by belt grinding. Most of the grinding was done on the hard face to remove surface tears introduced in forging and rolling. In addition, sufficient material was removed to reduce the total thickness to 0.620 – 0.635 in. (0.625 nominal). The plates were slightly bowed in two directions and tended to be somewhat thicker in the center than on the edges after grinding. The bow was such that the hard face was slightly convex and the back face slightly concave. This is believed to be largely the result of residual stress from differential thermal expansion of the front and rear face alloys.

3.3.3 Heat Treatment

All the test plates were solution annealed 20 min at 1000°C, quenched in iced brine at -2°C, and aged 2 hr at 400°C. The solution temperature was selected on the basis of results of a heat-treatment study which established the optimum temperature for developing the best combination of hardness and toughness in the rear face alloy (Section 2). This study was based on the use of a 0.7-in.-thick Ti-7Al-2.5Mo back-face alloy. In fabricating the actual composite test plates, however, a different lot of the back-face alloy measuring 1.25 in. in thickness was employed. Consequently, before heat treating the composite test plates, sections of the thicker back-face alloy were rolled to a thickness of 0.4 in. according to the rolling schedule used in processing the test plates and heat treated to confirm the results of the earlier study.

As shown in Fig. 15, the unnotched impact value of the Ti-7Al-2.5Mo alloy plate in the solution annealed and aged conditions did not drop significantly until the solution temperature exceeded 1000°C. Hardness increased to R_C 44 – 45 in the aged condition with a solution anneal at 1000°C. This annealing temperature appeared to impart the best combination of high strength (high hardness) and toughness to the back-face alloy.

All the test plates were subsequently annealed successfully. No cracking or bond separation occurred on quenching the large plates in iced brine from 1000°C. The brine was cooled with dry ice and was agitated by CO_2 bubbles. The samples were coated with Markal glass for oxidation protection, and the quench medium had a large amount of suspended scale and glass fragments. These factors combined to produce a low quenched hardness in the manganese-bearing hard-face alloys. The hardness values were 2 to 3 R_C points below anticipated values for these alloys based on prior work. The back-face alloy hardness was exactly as expected. however.

Table 12

RESULTS OF 0.50- AND 0.30-CALIBER AP/M2 BALLISTIC TESTS ON DUAL-HARDNESS TITANIUM ALLOY COMPOSITE ARMOR PERFORMED AT AMMRC

Nominal Hard-Face Alloy Composition (wt %)	Test Panel No.	Thickness (in.)			Front to Rear Face Thickness Ratio	Aerial Density (lb/ft²)	Hardness (R_c)		Projectile Velocity (ft/sec)		Ballistic Behavior
		Front Face	Rear Face	Total			Front Face	Rear Face	0.30 cal	0.50 cal	
Ti-3Ge-3Mn-2Si-0.5N	TP-5	0.315	0.310	0.625	1:1(50% Hard)	14.75	56.4(b)	45.6	–	1740	Partial front face spall and bond separation. Projectile fractured but penetrated composite.
									–	1585	Little front face spall and no bond separation. Projectile shattered and defeated.
Ti-5Ge-3Mn-3Si-1.0N	TP-1	0.351	0.274	0.625	1.28:1(56% Hard)	14.85	60.5(b)	44.8	–	1685	Complete front face spall and bond separation. Projectile fractured but penetrated composite.
Ti-5Ge-3Mn-3Si-1.0N	TP-7	0.385	0.235	0.620	1.64:1(62% Hard)	14.45	56.0(c)	43.0	–	1665	Complete front face spall and bond separation. Projectile fractured but penetrated composite.
Ti-5Ge-3Mn-3Si-1.0N	TP-2	0.377	0.248	0.625	1.52:1(60% Hard)	14.90	60.6(b)	44.2	2630	–	Complete front face spall and bond separation. Projectile fractured but penetrated composite.
Ti-5Ge-3Mn-3Si-1.0N	TP-3	0.403	0.233	0.636	1.73:1(63% Hard)	14.92	59.8(b)	45.8	2440	–	Complete front face spall and bond separation. Projectile fractured but penetrated composite.
Ti-5Ge-3Mn-3Si-1.0N	TP-4	0.400	0.225	0.625	1.78:1(64% Hard)	14.80	60.8(b)	46.0	2270	–	Complete front face spall and bond separation. Projectile fractured but penetrated composite.
Ti-5Ge-3Fe-3Si-1.0N	TP-6	0.379	0.246	0.625	1.54:1(61% Hard)	14.70	60.9(c)	43.8	1820	–	Little front face spall and no bond separation. Some edge cracking. Projectile fractured and defeated.
									1975	–	Complete front face spall and bond separation. Projectile fractured and defeated.
Ti-5Ge-3Mn-3Si-1.0N	TP-8	0.376	0.249	0.625	1.51:1(60% Hard)	14.52	56.6(c)	43.2	1765	–	Complete front face spall and bond separation. Projectile fractured but penetrated composite.

(a) Obliquity, 0 deg.
(b) Solution treated 20 min at 1000°C, water quenched and aged 2 hr at 400°C.
(c) Solution treated 20 min at 1000°C, quenched in dry ice-water mixture and aged 2 hr at 400°C.

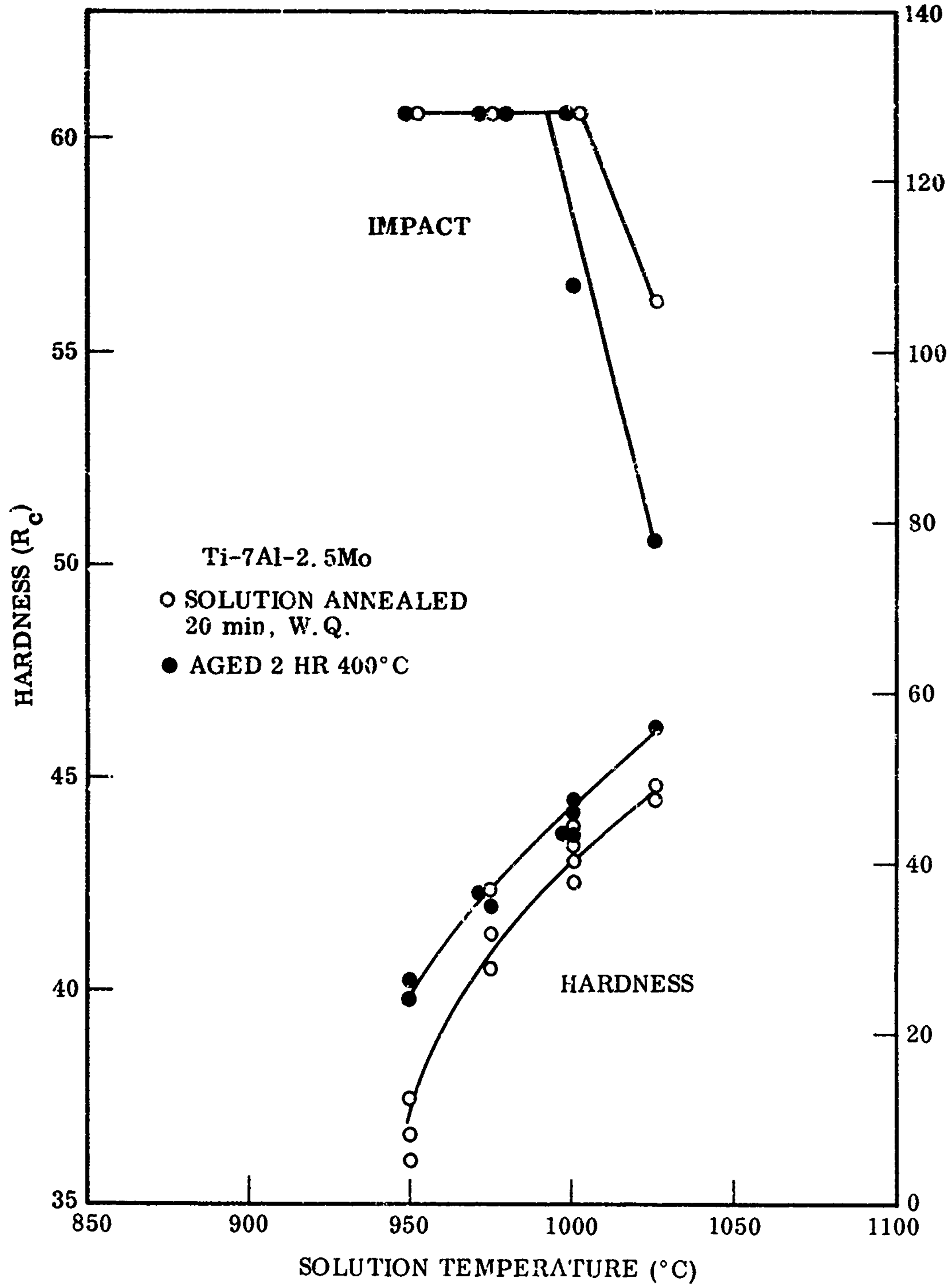


Fig. 15 Effect of Heat Treatment on the Hardness and Toughness of Ti-7Al-2.5Mo 0.4-in.-Thick Plate

It was postulated that dissolution of CO_2 and suspended dirt in the quench water as well as the Markal coat on the surface reduced the quench rate to a level considerably below that attained in the prior work. The iron modified alloy, TP-6, is less sensitive to cooling rate and attained the expected hardness (R_c 61) in this quench medium.

It was decided to reanneal the manganese-bearing alloy test plates and quench in clean water maintained at ambient temperature. Hardness after aging (i.e., 2 hr at 400° C) increased to R_c 60 – 61 for the Ti-5G -3Si-3Mn-1N alloys (TP-1 to 4) and to R_c 56 for the Ti-3Ge-2Si-3Mn-0.5N alloy (TP-5). Plates TP-7 and TP-8 were not reannealed because of the possibility that the relatively weak bond established at the edges might have promoted complete separation on repeated thermal cycling and quenching. These plates were left at a hardness level of R_c 56 – 57. Their behavior in ballistic tests, therefore, might be compared with that of a lower Ge-N alloy (TP-5) at the same hardness level. Plate TP-6 also was not reannealed, since it attained the expected hardness on the initial quench.

Hardness values for the front and rear faces of each test plate in the finished condition are presented in Table 12.

Section 4

RESULTS

4.1 BALLISTIC TESTS

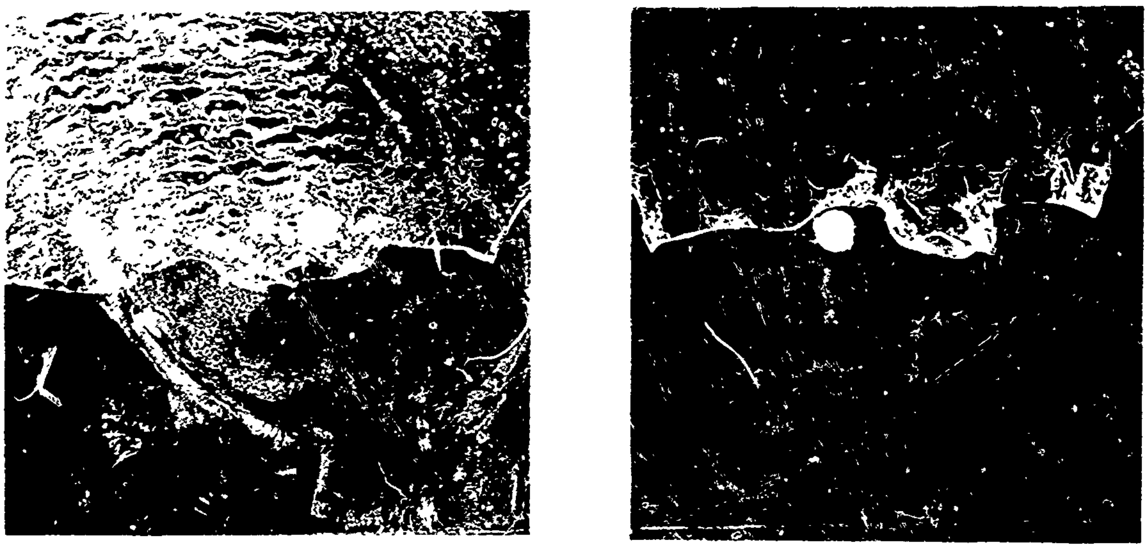
The eight composite panels were test fired with either 0.30- or 0.50-cal AP/M2 projectiles at the ballistic test range of the Army Materials and Mechanics Research Center, Watertown, Mass. The plates were held in a steel frame for a 0-deg obliquity hit at a range of about 20 ft. An aluminum witness plate was placed behind the plates to detect back-face spall, and a sandbox was used to collect spent projectiles or fragments that penetrated the armor. Projectile velocity for each shot was measured with a chronometer. The material characteristics, test conditions, and test results are summarized in Table 12.

The first test was conducted with plate TP-5 and 0.50-cal projectiles. At a velocity of 1740 ft/sec, the projectile penetrated both the front and rear face and caused about two-thirds of the front face to spall. The projectile was fractured, however. A second hit was made near the corner of the plate at a velocity of 1585 ft/sec. In this case, the projectile shattered on impact and did not penetrate the front face. There was neither front face spall nor bond line separation. The surface was cratered in the hit region and was very similar to dual-hardness steel in its appearance after ballistic test. This test panel is shown in Figs. 16 and 17.

The approximate V50 ballistic limit based on the first test was calculated to be 1662 ft/sec. This value was selected as the starting velocity for the next test, plate TP-1. This plate had complete bond line separation on impact. A test was conducted at about the same velocity on plate TP-7, with identical results. In both cases, the projectiles were broken and the nose penetrated the rear plate; the tail end rebounded backward. The tested plates are shown in Fig. 18.

It was decided to test the remaining plates with 0.30-cal AP/M2 projectiles, since the aerial density (14.5 – 15 lb/ft^2) was probably too low for 0.50-cal protection. As shown in Table 12, all the remaining test plates with one exception had complete bond separation on impact at velocities of 1765 – 2630 ft/sec.

Plate TP-6 defeated two projectiles at 1820 and 1975 ft/sec and was not penetrated. Complete front face spall on the second hit precluded further tests to establish a V50 ballistic limit. The limit, however, is in excess of 1975 ft/sec. This test plate contained the deep-hardened iron modified alloy with a front-face hardness of R_c 61. As with the 0.50-cal AP/M2 tests, all the 0.30-cal projectiles fractured on impacting the hard surface. The tested plates are shown in Figs. 19 – 21.



9499 Front Face ×0.6 9502 Back Face ×0.6

9499 ×1.35

Enlarged View of Hit Region and Fragmented Projectile, 1,585 ft/sec

Fig. 16 Appearance of 50-cal AP Test Plate TP-5 Hit at 1,740 (Center) and 1,585 (Lower Left) ft/sec. Hardness $-R_c$ 56−57

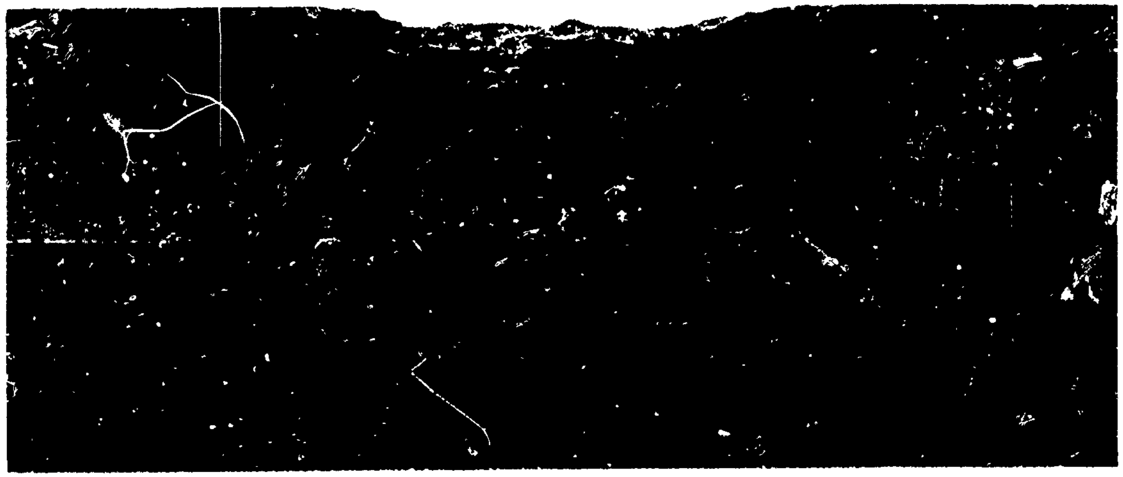
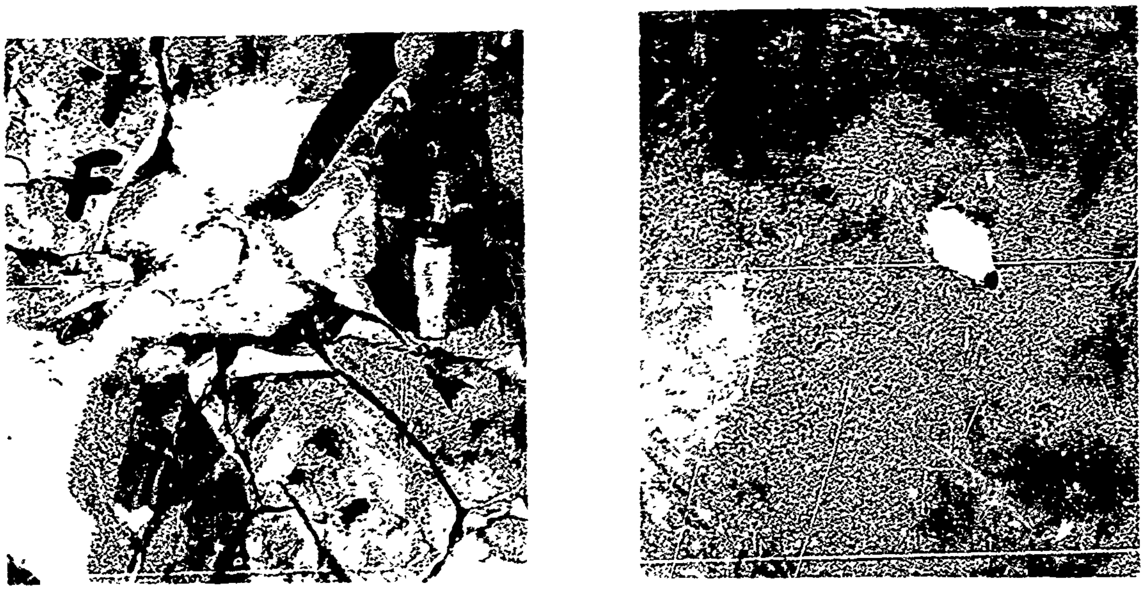
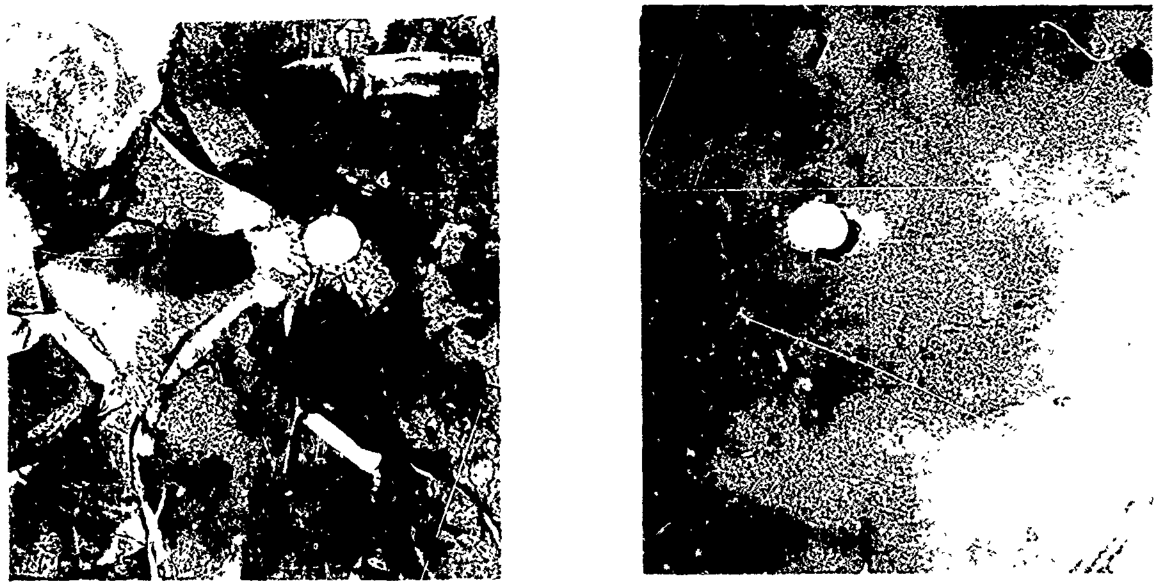


Fig. 17 Cross Section of 50-cal AP Test Plate TP-5 Through Impact Crater of Hit at 1,585 ft/sec



9507 Front ×0.6 9508 Rear ×0.6

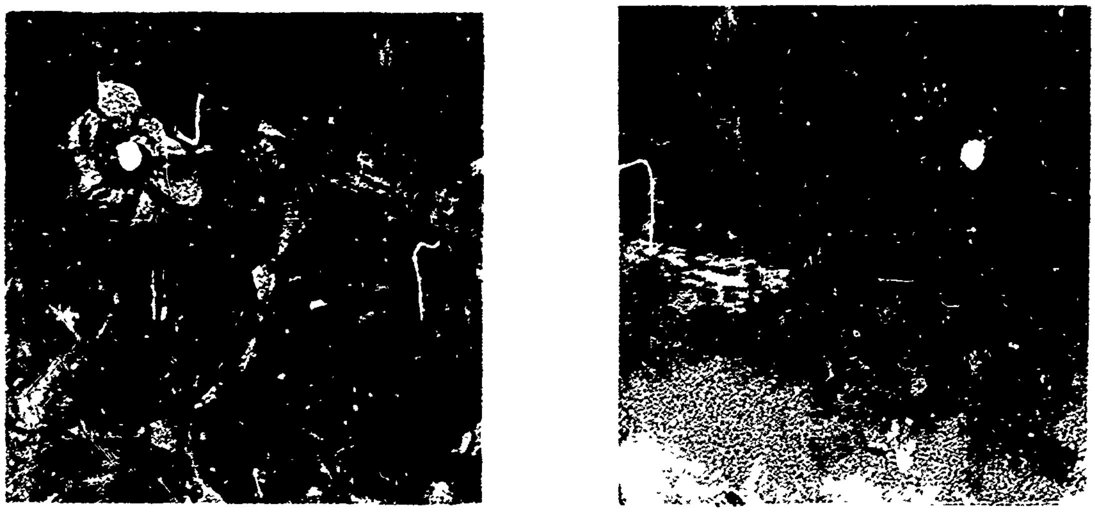
TP-1, 1,685 ft/sec, Ti-5Ge-3Si-3Mn-1N, 60.5 R_c



9507 Front ×0.6 9508 Rear ×0.6

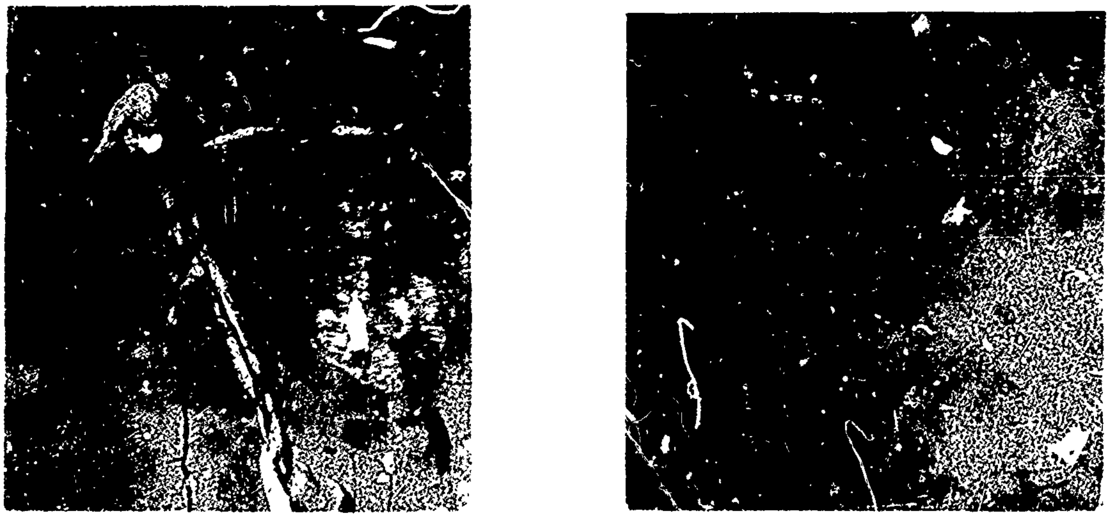
TP-7, 1,665 ft/sec, Ti-5Ge-3Si-3Mn-1N, 56 R_c

Fig. 18 Appearance of 50-cal AP Test Plates TP-1 and TP-7



9495 Front ×0.55 9496 Rear ×0.55

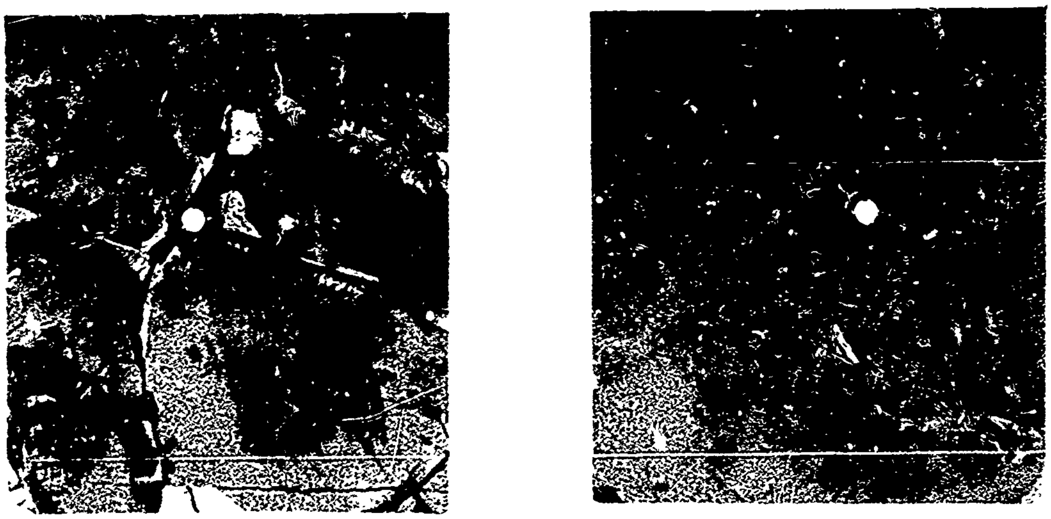
TP-2, 2,630 ft/sec, Ti-5Ge-3Si-3Mn-1N, 60.6 R_c



9495 Front ×0.55 9496 Rear ×0.55

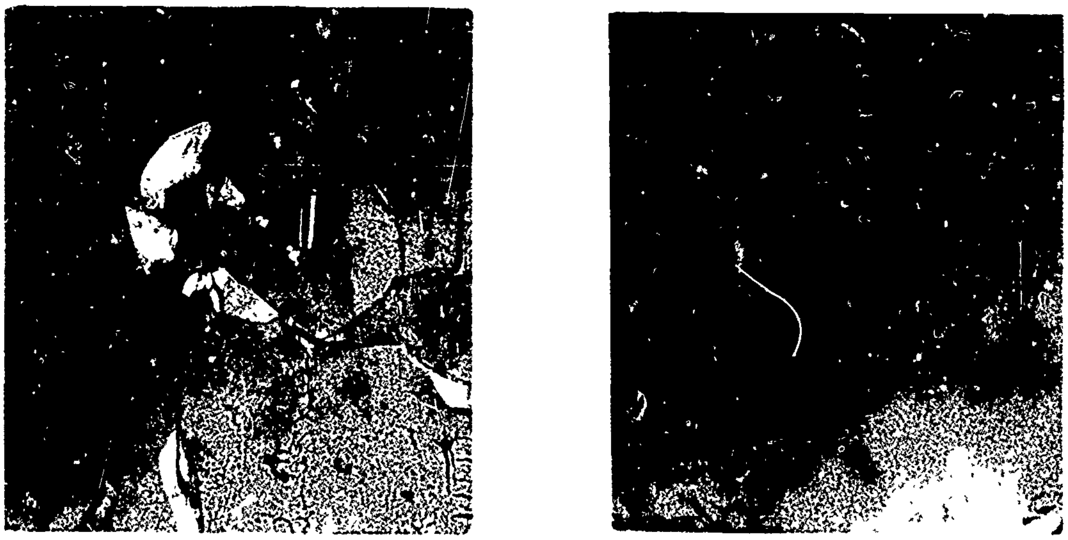
TP-3, 2,440 ft/sec, Ti-5Ge-3Si-3Mn-1N, 5 .8 R_c

Fig. 19 Appearance of 30-cal AP Test Plates TP-2 and TP-3



9497 Front ×0.55 9498 Rear ×0.55

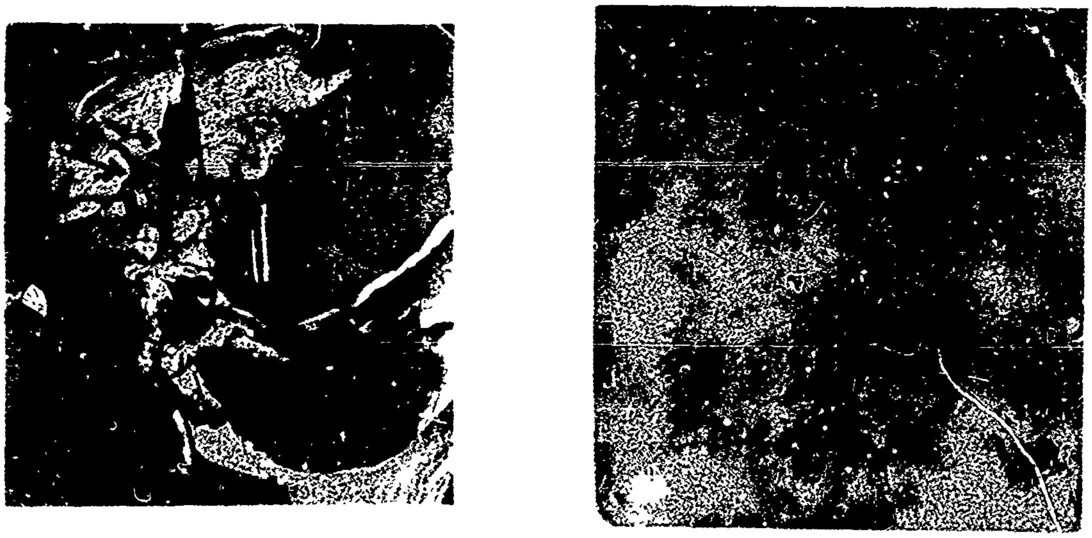
TP-4, 2, 270 ft/sec, Ti-5Ge-3Si-3Mn-1N, 60.8 R_c



9497 Front ×0.55 9498 Rear ×0.55

TP-8, 1, 765 ft/sec, Ti-5Ge-3Si-3Mn-1N, 56.6 R_c

Fig. 20 Appearance of 30-cal AP Test Plates TP-4 and TP-8



9505 Front ×0.55 9506 Rear ×0.55

TP-6, 1 – 1,820 ft/sec,
2 – 1,975 ft/sec

Ti-5Ge-3Si-3Fe-1N, 61 R_c

Fig. 21 Appearance of 30-cal AP Test Plate TP-6

The back-face alloy performed remarkably well in all tests. No cracking or delamination was observed in any of the test plates. Deformation of the plates were negligible, and a clean hole was punched out when the projectile penetrated the armor. The front-face alloys also performed remarkably well. Damage was confined to the immediate area of the hit, and no tendency for major breakup of the alloy was indicated. If fracture had not occurred at the bond line, multihit capability within several calibers would have been feasible.

4.2 METALLURGICAL EVALUATION

Three test plates were selected for metallurgical study:

TP-5 – Ti-3Ge-2Si-3Mn-0.5N
TP-4 – Ti-5Ge-3Si-3Mn-1.0N
TP-6 – Ti-5Ge-3Si-3Fe-1.0N

These plates were considered to be representative of the three classes of materials evaluated in this program.

Plates TP-4 and TP-6 were completely destroyed by bond-line separation on ballistic impact. Large pieces of edge trim from the as-rolled plates were used for metallurgical evaluation in this case. Samples measuring 1 by 1 by 0.625 in. were solution annealed and aged by the same schedule used for the test plates. Plate TP-5 was partially destroyed, and a section of undamaged material was cut from one corner for study. In addition, a sample was cut through the region subjected to ballistic impact (0.50 cal) shown in Fig. 17.

The samples were cross sectioned, mounted in bakelite, and polished with Cer-Cre abrasive for metallographic examination. Hardness surveys were made across the section thickness with a Lietz microhardness tester. Diamond pyramid hardness readings with a 500-g load were taken at 0.050-in. intervals.

After hardness testing, the samples were etched for 15 min in a solution of 1% HF in methyl alcohol to reveal the structure of the hard-face alloy. This was followed by a quick dip (< 1 sec) in a 7-1/2 HF-2-1/2 HNO_3-90 H_2O solution to bring out the structure of the back-face alloy. Standard etch solutions rapidly attacked the compound particles in the hard-face alloys, and a duplex etch for a long time in weak solution plus a flash high-strength etch was found to be best for these materials. Results of the hardness survey and metallurgical studies are presented in Table 13 and Figs. 22 – 28.

Hardness gradients in the three materials were as expected based on the results of the preliminary alloy studies (Section 2). The low Ge-N alloy (TP-5) had a marked hardness gradient through the front-face section as a result of low hardenability. As shown in Fig. 22, hardness dropped from DPH 606 (R_c 57) at the surface to DPH 427 (R_c 43) at 0.15 in. below the surface. A rise in hardness to DPH 518 (R_c 50) at 0.3 in. below the surface was found. This behavior probably is the result of either a change

Table 13

MICROHARDNESS OF TITANIUM ALLOY COMPOSITE BALLISTIC TEST PANELS(a)

No. TP-4 (Ti-5Ge-3Mo-3Si-1N)		No. TP-5(b) (Ti-3Ge-3Mo-2Si-0.5N)		No. TP-6 (Ti-5Ge-3Fe-3Si-1N)	
DPH	Depth Below Hard Face Surface (in.)	DPH	Depth Below Hard Face Surface (in.)	DPH	Depth Below Hard Face Surface (in.)
740	0.003	606	0.003	715	0.005
736	0.050	508	0.050	708	0.050
732	0.100	464	0.100	704	0.100
715	0.150	427	0.150	732	0.150
677	0.200	447	0.200	700	0.200
677	0.250	510	0.250	736	0.250
677	0.300	518	0.300	715	0.300
670	0.340	465	0.314	715	0.347
616	0.345	431	0.317	642	0.358
423	0.352	354	0.350	591	0.365
362	0.400	362	0.400	444	0.368
380	0.450	376	0.450	396	0.400
362	0.500	368	0.500	396	0.450
354	0.550	360	0.550	368	0.500
412	0.594	378	0.660	364	0.550
		406	0.620	368	0.600
				386	0.617

(a) All measurements made with 500-g load.

(b) Data represent average of two separate sets of readings.

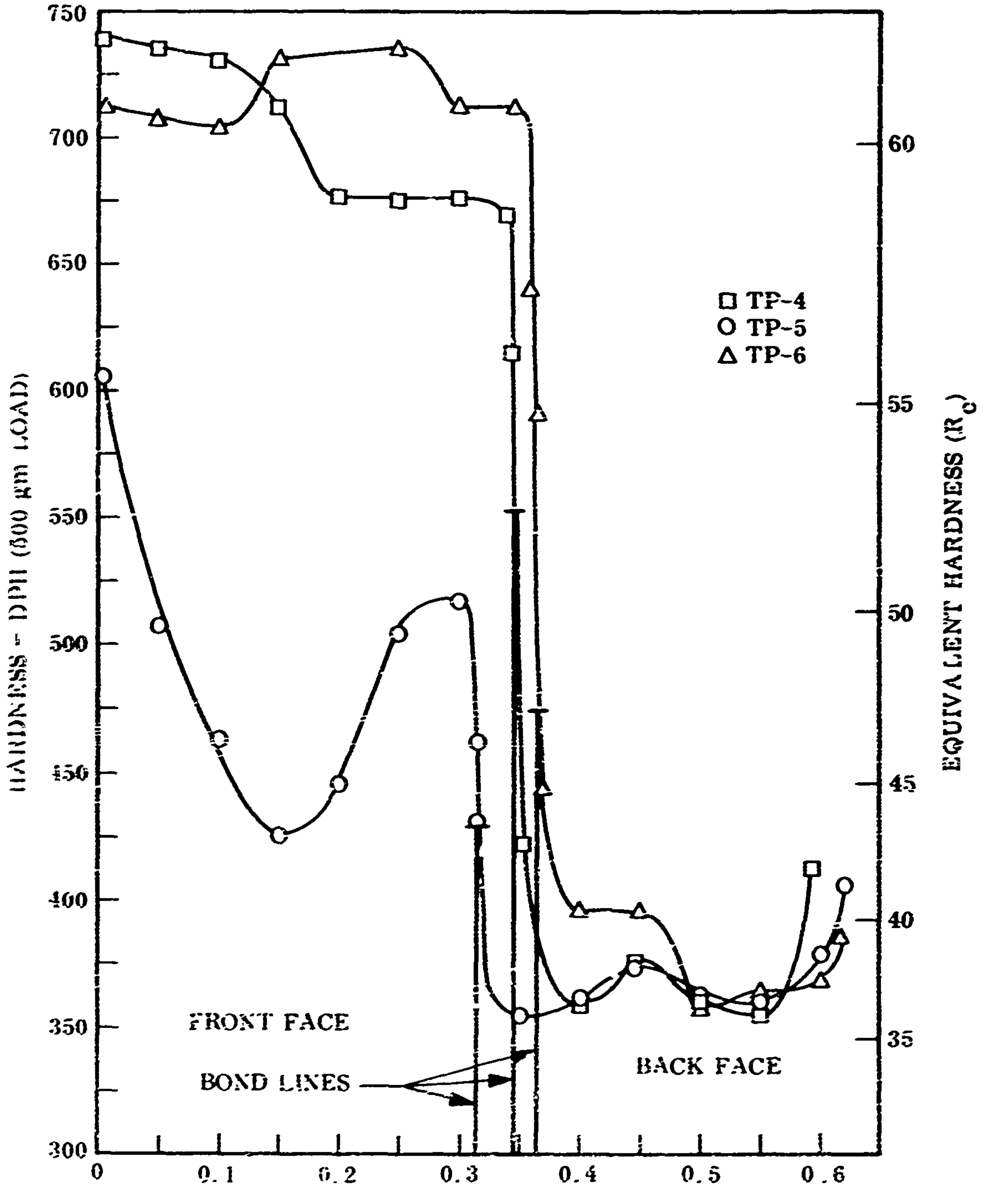
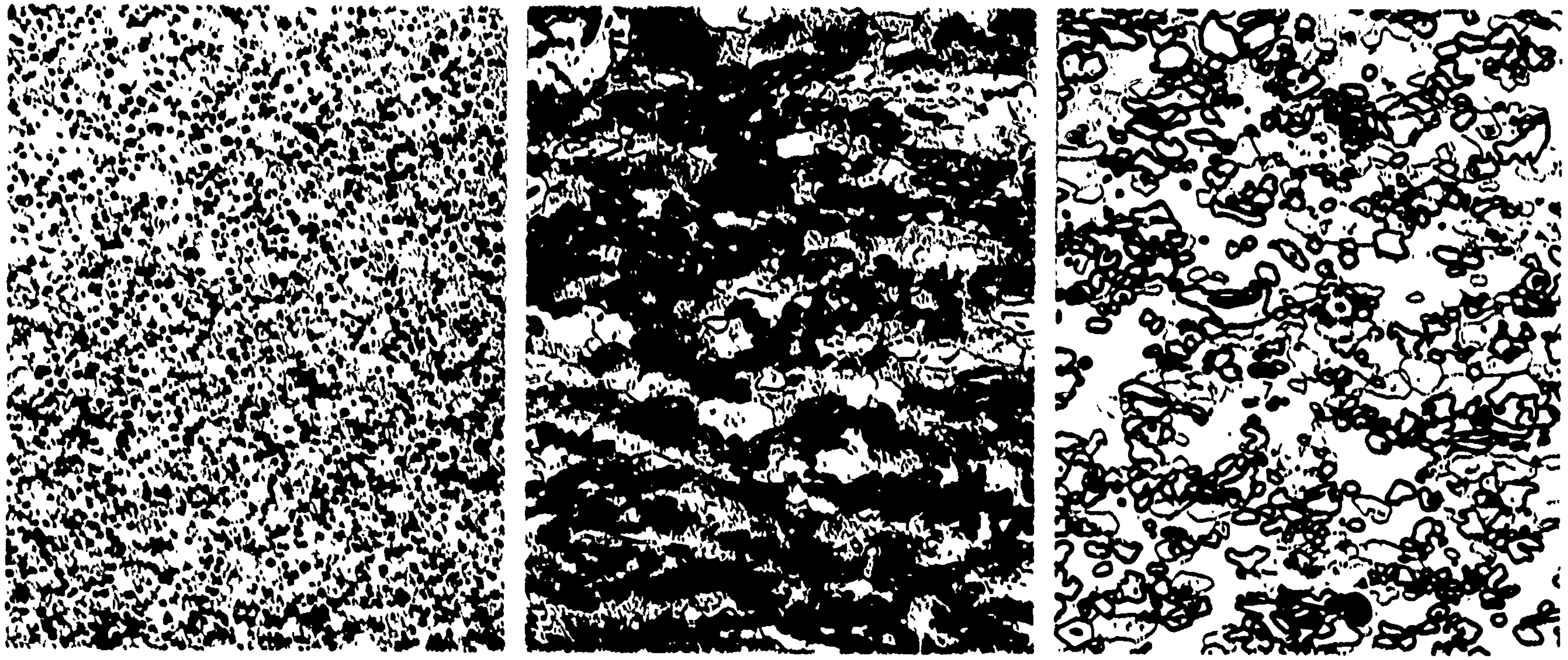


Fig. 22 Hardness Profiles of Dual-Hardness Titanium Alloy Ballistic Test Panels



9788 ×400 9633 ×400 9627 ×400

Ti-3Ge-2Si-3Mn-0.5N
56.4 R_c

Ti-5Ge-3Si-3Mn-1N
60.8 R_c

Ti-5Ge-3Si-3Fe-1N
61.0 R_c

Fig. 23 Microstructure of Hard Face Alloys in Heat-Treated Ballistic Test Plates

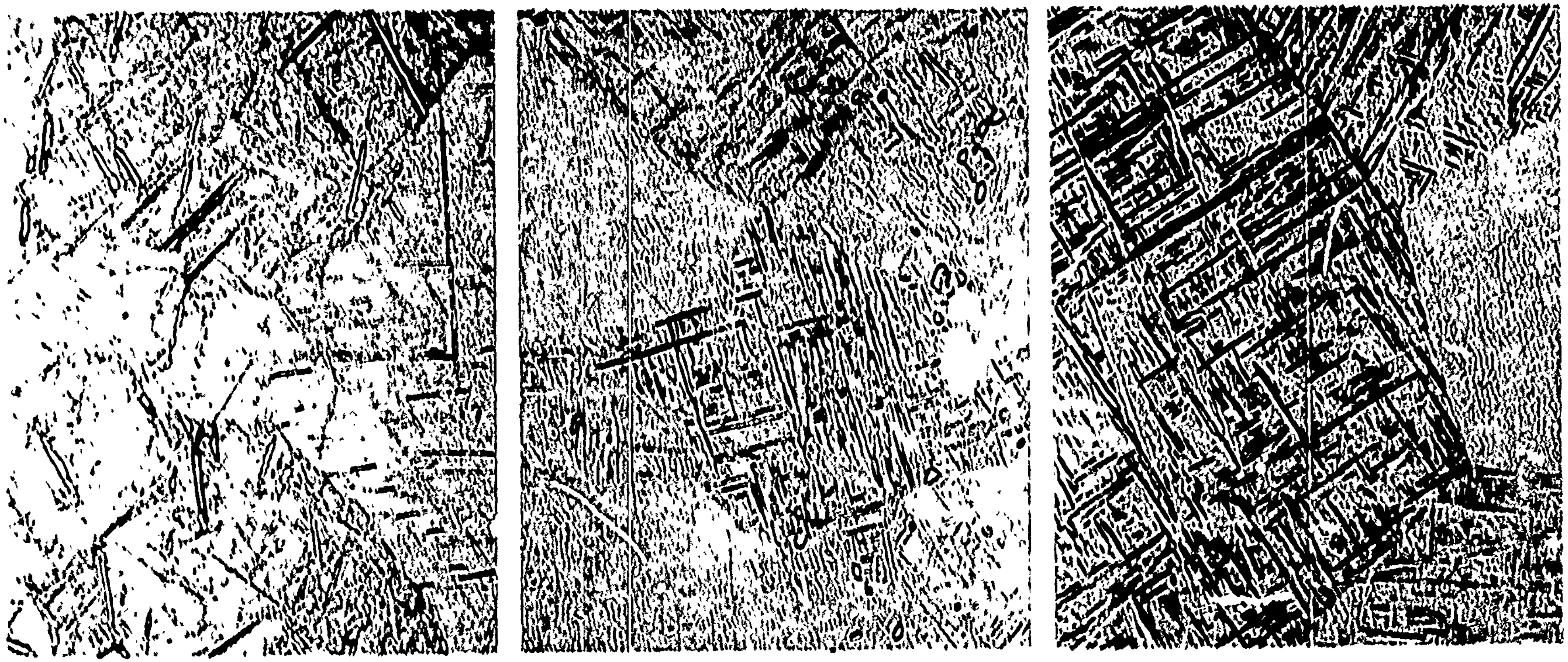
9552 ×3000 9635 ×3000 9629 ×3000

TP-5, Ti-3Ge-2Si-3Mn-0.5N
56.4 R_c

TP-4, Ti-5Ge-3Si-3Mn-1N
60.8 R_c

TP-6, Ti-5Ge-3Si-3Fe-1N
61.0 R_c

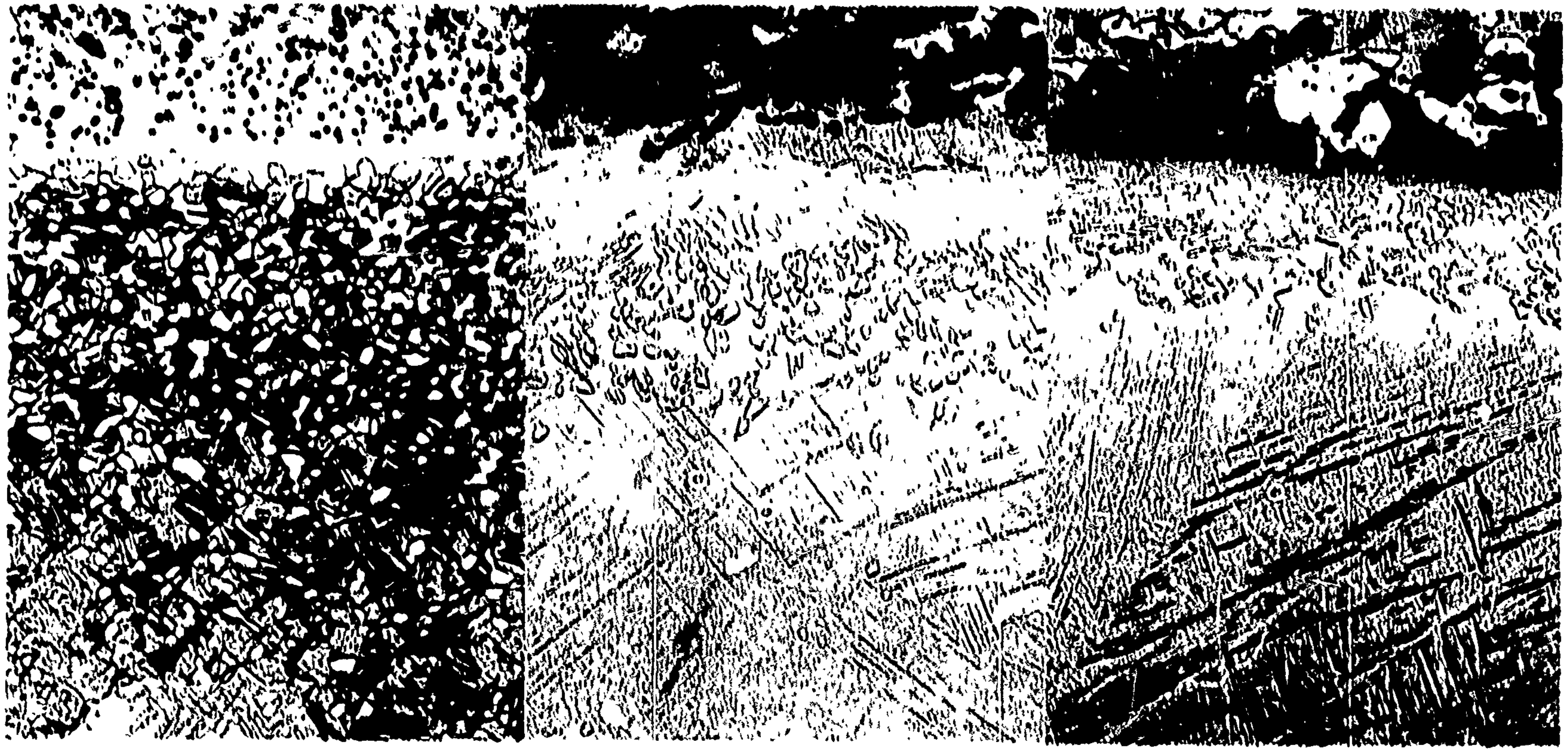
Fig. 24 Microstructural Details of Hard Face Alloys in Heat-Treated Ballistic Test Plates



9556 ×400 9634 ×400 9328 ×400

TP-5, 45.6 R_c TP-4, 46.0 R_c TP-6, 43.8 R_c

Fig. 25 Microstructure of Ti-7Al-2.5Mo Back Face Alloy in Heat-Treated Ballistic Test Plates



9556 ×400 9632 ×400 9626 ×400

TP-5, Ti-3Ge-2Si-3Mn-0.5N
on Ti-7Al-2.5Mo

TP-4, Ti-5Ge-3Si-3Mn-1N
on Ti-7Al-2.5Mo

TP-6, Ti-5Ge-3Si-3Fe-1N
on Ti-7Al-2.5Mo

Fig. 26 Microstructures at the Bond Line of Heat-Treated Ballistic Test Plates

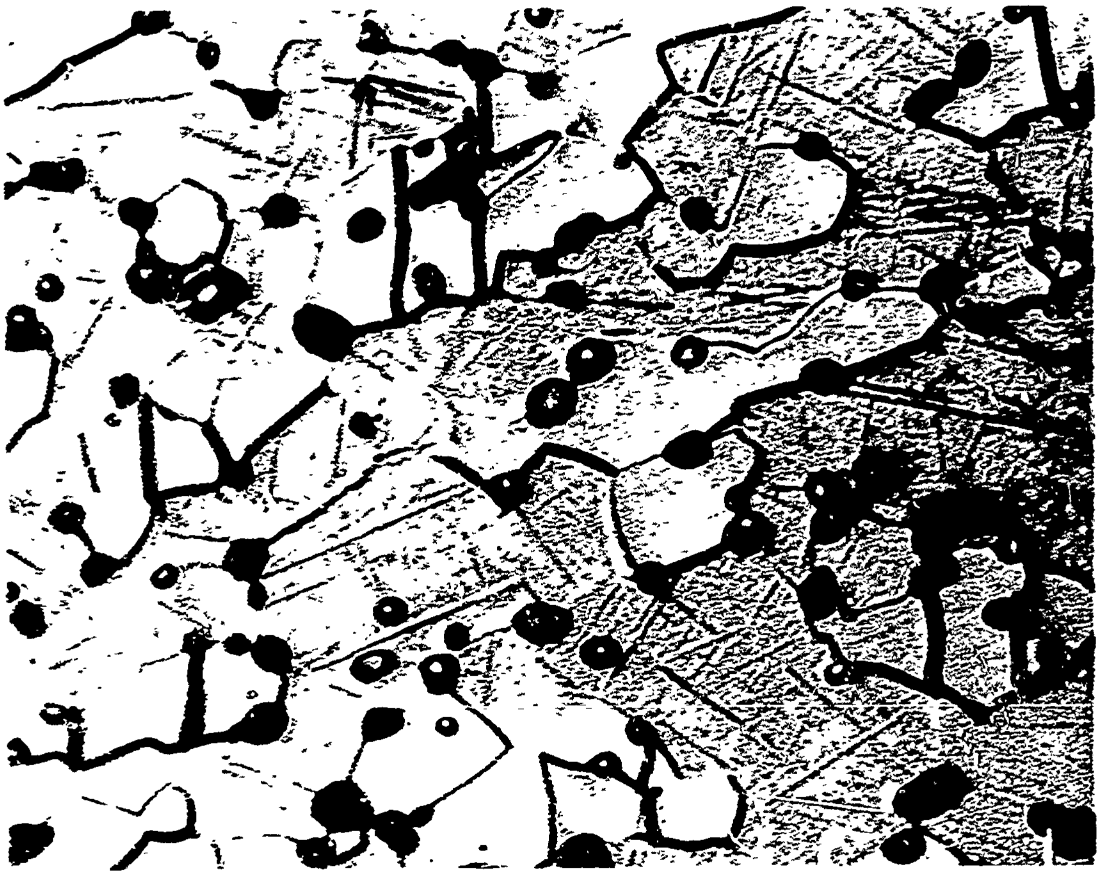
9550 ×3000 9536 ×3000 9630 ×3000

TP-5, Ti-3Ge-2Si-3Mn-0.5N
on Ti-7Al-2.5Mo

TP-4, Ti-5Ge-3Si-3Mn-1N
on Ti-7Al-2.5Mo

TP-6, Ti-5Ge-3Si-3Fe-1N
on Ti-7Al-2.5Mo

Fig. 27 Microstructural Details of the Bond Line in Heat-Treated Ballistic Test Plates



9554 ×3000

Fig. 28 Selective Cracking of the Alpha Phase Near the Surface Adjacent to a 50-cal AP Hit on Test Plate TP-5 in alloy composition due to the diffusion of nitrogen into the back-face alloy or the existence of a critical cooling rate with respect to hardening in this alloy. The exact cause of this unusual behavior was not established, however.

The two high Ge-N alloys had much better hardenability. The Mn-bearing alloy (TP-4) had a slight drop in hardness from DPH 740 (R_c 62) at the surface to DPH 677 (R_c 59) at 0.2 in. below the surface. As expected, the Fe-bearing alloy (TP-6) had the best hardenability with virtually no hardness gradient from the surface to the bond line. Hardness through the front face varied from DPH 700 to 736 (R_c 60-61). The measured DPH values for the front-face alloys correlated very well with measured R_c hardness values based on conversion tables derived for high-hardness steel.

A very pronounced drop in hardness occurred across the bond line for all test plates, as shown in Fig. 22. This behavior is ideal for a dual-hardness armor. The back-face alloy hardness tended to be uniform. A slight increase in hardness was observed at the outer (back) surface. This may be a hardenability effect or may be the result of interstitial contamination during processing. No major hardening as a result of nitrogen diffusion across the interface from the hard-face alloy is indicated. The DPH measurements on the back-face alloy indicate a lower R_c value from hardness coversion tables than the measured R_c values.

The general microstructure of the three front-face alloys is shown in Figs. 23 and 24. The alloys have about a 1:1 ratio of alpha and alpha prime (martensite) as a fine grained, equiaxed structure. The martensite in the Fe-modified alloy (TP-6) is submicroscopic and x-ray diffraction studies revealed that none of the three alloys contained retained beta. The third phase (dark etching) in all alloys is a Ti-Ge-Si compound. It appears in a random distribution, with a tendency to occur at the boundaries between the alpha and alpha prime (martensite) grains. The presence of this phase is believed to be responsible for the fine grain size retained in these alloys after high-temperature working and annealing.

The most significant difference in structure between the three alloys is the ultrafine grain size exhibited by plate TP-5. The compound particles are much smaller in this alloy as a result of its low Ge (3%) and Si (2%) contents. The alloys with 5Ge-3Si contain large particles of the compound and have a much coarser grain structure. These differences indicate that a lower total Ge-Si content may be preferable for optimum refinement of the microstructure.

The most significant structural feature of all the alloys is the fact that none of the three phases form a continuous hard, brittle network. The martensite phase appears to be the matrix phase and has the greatest continuity. Previous work revealed that this phase has a hardness of about 680 DPH in heat-treated alloys of this class. The alpha phase, on the other hand, has a hardness of about 900 DPH. It is believed that the overall hardness and toughness is largely controlled by the properties of the martensitic matrix phase.

Typical structures of the back-face alloy (Ti-7Al-2.5Mo) in each test plate are shown in Fig. 25. Each alloy was annealed at a temperature very close to the beta transus. The structure consists of martensite with a trace of alpha. A large equiaxed prior beta grain size and slight differences in the amount and grain size of the alpha phase are indicated.

Plate TP-6 contained the least amount of alpha and the coarsest structure. This plate was annealed slightly above the beta transus. Plate TP-5 had the highest percentage of alpha and the finest structure. This plate was annealed below the beta transus. These differences in the back-face structure probably reflect slight variations in processing which promoted a small shift in the beta transus. All plates were solution annealed at the same indicated temperature, however (1000° C).

The microstructural details at the bond line between the hard and soft faces are shown in Figs. 26 and 27. An excellent metallurgical bond is indicated in all cases. As shown in Fig. 27, the continuity of the martensite matrix phase in both the front- and back-face alloys across the bond line is excellent. No voids or unbonded areas were detected. The most significant feature in the bond area is the structural change effected on both sides of the bond line as a result of nitrogen diffusion from the front to rear face. The hard-face alloy is depleted in nitrogen above the interface and has a martensite plus compound structure. Martensite is the continuous phase in this region. The back-face alloy is nitrogen-enriched below the interface and shows the presence of a considerable amount of alpha (Fig. 26).

The amount and distribution of alpha in the back face varied significantly for the three alloys. The depth of nitrogen penetration appears to bear some relation to the grain size and structure of the back-face alloy. Plate TP-5, which has a fine-grained structure with some alpha as a result of annealing just below the beta transus, exhibited the greatest depth of nitrogen penetration. A large amount of grain boundary area may have provided an easy path for diffusion in this case. The other two plates with a very coarse prior beta grain structure showed much less penetration. Bulk diffusion through the grains may be significantly slower. The alpha phase in this case also tended to form as platelets along preferred crystallographic planes in the matrix. This may produce a weak zone in the alloy, since the nitrogen-enriched alpha phase will tend to be extremely brittle.

Examination of plates TP-5 and TP-6 indicated that the separation of the front and rear faces on ballistic impact occurred in the hard-face alloy just above the bond line. No cracking or delamination was observed in either the bond line or the back face alloy. Unfortunately, no partially cracked areas could be found to delinate clearly the location and path of the cracks which promoted separation of the two faces. All observations were made in areas where complete separation occurred. Details of the failure mode could not be resolved. No cracks were found in the region of the interface where good bond integrity was maintained.

Compositional and microstructural changes in the hard-face alloy adjacent to the bond as a result of interdiffusion may play important roles with respect to failure. The alloy in its normal condition (away from the bond area) has good resistance to crack propagation. As shown in Fig. 28, cracks generated in the hard alpha phase on ballistic impact are arrested at the interface with the alpha prime (martensite) phase. The region shown in this figure is immediately below the surface of the impact crater on plate TP-5, where a 0.50-cal AP/M2 projectile was defeated. In all areas examined, cracking was confined to the alpha phase. The softer martensitic matrix tended to resist fracture such as to cause blunting of the cracks at the interphase boundary. The high resistance to crack propagation displayed by the martensite phase in these regions strongly suggests that fracture at the interface may be related to compositional changes (interdiffusion) as opposed to structural differences.

Section 5

DISCUSSION OF RESULTS

The results of this study clearly indicate that dual-hardness titanium armor which will function with a defeat mechanism characteristic of dual-hardness steel or ceramic composite armor is feasible technically. A new family of hard titanium alloys in the Ti-Ge-Si-Mn(Fe)-N system which can be heat treated to hardness levels of R_C 55 – 65 has the ability to fracture both 0.30- and 0.50-cal AP projectiles on impact. The alloys have sufficient toughness to resist gross fracture and possess an excellent capability for multihit protection. They can be bonded readily to commercial titanium alloys and processed by conventional techniques into high-quality dual-hardness composites. A Ti-7Al-2.5Mo alloy provides an excellent high-strength, tough backup material for the hard-face alloys in a dual-hardness composite.

Two major problems must be solved before these results can be translated into an effective new armor material. The first problem is related to producibility and concerns the melting and casting of sound crack-free ingots of the hard-face alloy. The second problem is related to ballistic performance and concerns the tendency for bond-line separation on ballistic impact. Results of the study indicate that both problems are soluble and suggest a variety of approaches for their solution.

Porosity and cracking of the ingots and low recoveries during working of the hard alloys appear to be the result of (1) excessive Ge + Si content and (2) incomplete solution and segration of nitrogen alloy additions. A careful study of all results indicates that a level of 3% Ge and 2% Si is adequate from a hardness and hardenability point of view if the alloy contains 1.0% N. Larger additions result in only a slight improvement in these properties and at the expense of producibility. The size and distribution of the compound phase are critical, and a fine random structure appears best for good melting and working characteristics. The one heat (OMC 4484) in which the Ge and Si were retained at these lower levels had excellent producibility.

Although the substitution of silicon nitride for titanium nitride as the nitrogen master alloy solved major problems in alloy segregation and ingot cracking, it did not completely eliminate these difficulties. Part of the remaining problem is related to the relative size effects of the titanium sponge and the nitride particles. The use of a finer particle sodium-reduced sponge and a finer size nitride addition would do much to improve ingot quality. Silicon nitride, however, may still not be the best addition agent. Excessive gas evolution in melting appears to occur as a result of dissociation of the nitride at high temperature. The gas holes, in turn, act as internal stress raisers and contribute to cracking.

The use of titanium-nitrogen master alloys (i.e., Ti/5–10% N) has been suggested by Cremet and others as a possible solution to this problem. Such alloys, however, are difficult to prepare by virtue of the same problems encountered in trying to add 1% N to Ti alloys. A more feasible solution would be to use nitrided titanium sponge for the nitrogen addition. This material could be produced readily as a very uniform product and should solve the problem of segregation and cracking in melting.

Another approach would be to use more stable complex nitrides (master alloys) based on combinations of Ti–Fe–or Mn–Si and N. A third approach would be to use higher pressure (argon or helium) rather than vacuum arc melting to reduce volatilization tendencies. The Oregon Metallurgical Corporation has expressed the belief that these problems can be solved by experimental studies in melting 6-in.-diameter ingots.

The problem of bond-line separation on impact is believed to be related to compositional changes in the hard-face alloy as a result of interdiffusion with the back face or a stress concentration in this region as a result of geometrical considerations. Metallurgical examinations indicate that the martensitic matrix phase in the hard-face alloy effectively resists crack propagation on ballistic impact. Cracking in this phase near the interface may be the result of property changes due to compositional shifts. Nitrogen from the hard phase was observed to diffuse into the back face. This changed the structure and composition of the hard face and may have produced a more brittle martensite. Aluminum from the back-face alloy also may be diffusing into the front face. It is known that aluminum severely embrittles the hard-face alloys. Electron probe studies would be required to identify the changes that actually occur.

In either case, the effects could be overcome by the use of diffusion barriers at the interface. In early work on nitrided titanium armor, thin molybdenum sheet was used at the interface to prevent nitrogen diffusion into the back face. After bonding and rolling, the molybdenum interdiffused to form a Ti-Mo alloy layer. A similar approach using Mo, V, or even unalloyed Ti might be effective with the current alloy systems.

Differential working to break up alloy structures near the interface may offer an alternate solution to the problem. Early ballistic studies indicated that test plates subjected to a large amount of differential working were less susceptible to bond separation on impact. This was also the case in the present test plates where TP-5 experienced the greatest amount of differential working and subsequently exhibited the best resistance to interface separation. Most of the interdiffusion that may be responsible for weakness in this area probably occurs during the long holding period used to effect pressure diffusion bonding. Shear along the interface during forging and rolling may help to break up undesirable structures and homogenize regions where compositional changes have occurred. The amount of differential working is controlled by the working temperature.

At low temperatures, where the commercial alloys have an alpha-beta structure in which the alpha is strengthened by aluminum in solid solution, the hard-face alloy is weak in comparison and deforms more readily. At higher temperatures, the commercial alloys become rich in the softer beta phase and the strength is less than that of the hard alloys. This cross-over behavior with temperature permits good control of differential working behavior.

Geometrical considerations offer still another solution to the problem of bond-line separation. In all cases where the front-face thickness exceeded 50% of the composite thickness, separation at the bond line appeared to occur more readily. It is possible that the back-face alloy in this case does not have sufficient strength to resist deformation (bowing) on impact. This would result in a very high tensile stress at the bond line. Increasing section thickness should improve behavior in this regard. It is likely that the front-face thickness should be considerably less than 50% of the composite in order to achieve needed deformation resistance. The optimum thickness for dual-hardness steel is in the range of 35 to 50%. With titanium, however, it may be even less because of its lower strength and modulus compared with steel. The seven dual-hardness titanium alloy test plates that had the least resistance to separation all had a 60:40 ratio of front to rear face materials. The one plate with the best performance, on the other hand, had a 50:50 ratio.

Studies of the effect of varying thickness ratios are needed to determine the optimum ratios for titanium. A slightly greater total section thickness also may be required to provide needed stiffness. The test plates used in this study were 0.625 in. thick, which is suitable for dual-hardness steel armor but may be somewhat on the light side for titanium. A slightly higher aerial density than that used in these tests may be desirable.

One of the critical components of the hard-face alloy is germanium. Additions of 3 to 5% appear to be required to produce the desired combination of properties. A cursory survey was made of the germanium industry to determine if cost or availability would preclude the use of this element as an alloy addition to titanium for use in armor. Germanium is a byproduct of zinc smelting and is readily available as purified metal suitable for alloying with titanium. The domestic supply is ample to meet current demands for the next 10 years and can be increased significantly, if the need arises, by imports from central and South Africa. The largest processing plant in the world is in Belgium and has a capacity of 135,000 lb of oxide (65 wt% Ge) per year. Belgium is the major European producer. The major producer in the United States is Eagle Picher Company. Other U.S. producers include American Zinc Company, Sylvania Electric Products Company, and American Metal Climax.

Production and consumption figures are difficult to find and have not been published in recent years. In 1963, U.S. industry consumed 20,000 lb of germanium. Total production amounted to 45,000 lb of primary metal and reprocessed scrap. Imports amounted to 30,748 lb of metal and oxide in 1961 but only 5,000 lb in 1963. At present, the metal is being stockpiled owing to a decline in demand. Engineering economists at Battelle Memorial Institute have stated that up to 5000 lb per year would be available for any new venture without a significant effect on the current market. At a 5% alloy addition, 180,000 lb of the hard-face armor alloy could be produced. Much larger quantities of germanium could be made available and imports could be used to bolster the supply if necessary. It would appear that the germanium supply is adequate to support a major program of use in titanium alloys.

Refined germanium metal suitable for alloying is available today at $75 to $77 per pound. At a 5% alloy level, germanium would add $3.75 to $3.85 to the cost of 1 lb of a titanium-base (or any other) alloy. This cost might be reduced by the development of a special alloying grade, since ultrahigh purity, other than interstitial content, is not essential.

Section 6

CONCLUSIONS AND RECOMMENDATIONS

6.1 CONCLUSIONS

(1) A dual-hardness titanium armor that is capable of defeating 0.30- and 0.50-cal. AP projectiles by the penetrator fracture mechanism characteristic of dual-hardness steel and ceramic composite armor is feasible technically.

(2) A new class of titanium-base alloys containing 3 – 5% Ge, 2 – 3% Si, 3% Mn or Fe, and 0.5 to 1.0% N developed by the Lockheed Palo Alto Research Laboratory has excellent potential as a heat-treatable, high-hardness alloy for the front face of a dual-hardness titanium armor.

(3) The hard alloy can be manufactured and processed with equipment and techniques used for commercial titanium alloys. The potential for successful scaleup to large ingots is considered to be favorable. Problems with ingot cracking during melting are related to alloy segregation and can be eliminated by improved control of raw materials and melting feed stock.

(4) The hard alloy has excellent compatibility with commercial high-strength alpha and lean-beta titanium alloys. It bonds readily to these materials and can be processed by the same working and annealing schedules. The deformation characteristics at high temperature are comparable to those of Ti-6Al-4V or Ti-8Al-1Mo-1V.

(5) Hardness of the front-face alloy can be adjusted from R_c 52 to 62 by heat treatment. Hardenability in thick sections is good and is greatly enhanced by the substitution of iron for manganese.

(6) The hard alloy exhibits an excellent potential for multihit capability. The alloy does not shatter on ballistic impact and behaves in a manner similar to that of the hard-face alloy on dual-hardness steel.

(7) A Ti-7Al-2.5Mo alloy is believed to have the best potential as a back-face alloy for dual-hardness titanium armor. The alloy has an exceptionally good combination of strength and toughness and has a significantly higher modulus than other high-strength alpha-beta alloys. The alloy does not crack, delaminate, or spall on ballistic impact.

(8) A tendency for fracture to occur at the bond line of ballistic test plates is largely due to compositional changes effected by interdiffusion and to insufficient composite strength. It is believed that this problem can be overcome by the use of suitable diffusion barriers and by optimizing the front-to-back-face thickness ratio and total thickness for this type of composite material.

(9) Based on results of this study, further development of a dual-hardness 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.

6.2 RECOMMENDATIONS

It is recommended that work on the development of a dual-hardness titanium armor consisting of a Ti-Ge-Si-Mn or Fe-N alloy front face and a Ti-7Al-2.5Mo alloy back face be continued. Further demonstration of producibility and evaluation of performance against 0.30- and 0.50-cal AP threats is required. Specifically, the following factors should be considered in any continuing program:

(1) A study of compositional modifications and changes in melting practices is needed to demonstrate producibility of sound ingots of the hard-face alloy. The study should be conducted on 6 to 8-in.-diameter ingots of 25 – 50 lb weight. Variables to be considered should include a reduction in total Ge + Si, use of nitrided sponge, and blending and sizing of charge ingredients.

(2) A study of processing variables for the dual-hardness composites is required to demonstrate that bond separation on ballistic impact can be prevented. The study should include the use of diffusion barriers to minimize compositional modifications on either side of the interface and the use of refined rolling practices to control structure at the interface. Ballistic tests with both 0.30- and 0.50-cal AP/M2 projectiles should be used to evaluate bond integrity.

(3) A study of ballistic performance in 0.30- and 0.50- cal AP/M2 tests is needed to establish optimum total composite thickness and front-to-rear face thickness ratios for this type of material.

(4) It is recommended that a program be conducted in three sequential tasks in the order outlined above. The results from each task should be carefully evaluated and assessed with respect to technical feasibility of the overall concept before proceeding to the next task.

Section 7

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