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DEVELOPMENT OF TRIP STEELS FOR ARMY APPLICATIONS

Interim Technical Report covering 1 July 1970 to 30 June 1971

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December 1971

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13. ABSTRACT

Development of techniques for air-induction melting of TRIP steel alloys was investigated. Sound ingots of Type A-2 TRIP steel were produced, processed, and are currently undergoing mechanical testing. Several methods for monitoring martensitic transformation were evaluated. A ferromagnetic technique remains to be appraised. Various homogenization treatments were investigated in order to optimize microstructure prior to thermomechanical processing. This is the first interim report on the development of TRIP steels covering the period 1 July 1970 to 30 June 1971. (Authors)

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I. INTRODUCTION

The advent of modern technology has been successful in producing steels with strength levels up to 300 ksi However, these strengths are usually attainable only at the expense of ductility and toughness. The development of TRIP (Transformation Induced Plasticity) steels by V. F. Zackay et al.[1] provides a family of alloys having the capabilities of achieving high yield strengths without corresponding losses in ductility. These high yield strengths are the results not only of alloy composition, but of heavy reductions at warm-working temperatures. These materials differ from conventional ausformed steels, for they remain austenitic at and below room temperature. Improved ductility at the high strength levels is the consequence of a strain-induced martensitic transformation which occurs in metastable austenitic steels. The increased ductility is also accompanied by an increase in fracture toughness and resistance to fatigue crack growth.[2-4]

A wealth of information has been generated from laboratory studies, and it is apparent that this class of alloys represents a significant improvement over currently available high-strength steels. However, considerably more research and development work is required before these alloys can become competitive candidates for high-strength army applications.

The program was divided into integrated segments, namely (1) to develop air-induction melting practice for TRIP alloys, (2) to determine the effects of alloy composition on martensitic transformation, and (3) to determine the effects of process variables on metallurgical and physical properties. Each segment will be discussed independently.

II. DEVELOPMENT OF AN AIR-INDUCTION MELTING PRACTICE

Up to the present, a satisfactory air-induction melting procedure guaranteeing reproducibility of structure and composition for TRIP steel alloys has not been developed. Currently, most TRIP steel alloys are vacuum melted. However, this procedure tends to restrict ingot size as well as increase ingot cost.

To produce quality reliable stock at reasonable cost for forming into end items, a reproducible air-melting practice should be established for the various TRIP steel alloy series or types. However, there are several problems involved in air-induction melting of these alloys. For instance, alloys not containing manganese or silicon will undoubtedly boil out most of their carbon before pouring and still contain detrimental oxides.

[1] ZACKAY, V. F., PARKER, E. R., FAHR, D., and BUSCH, R. *The Enhancement of Ductility in High-Strength Steels.* Trans. ASM, v. 60. 1967, p. 252-259.

[2] GERBERICH, W. W., HEMMINGS P. L., MERZ, M. D., and ZACKAY, V. F. *Preliminary Toughness Results on TRIP Steel.* Trans. of the ASM, v. 61, 1968, p. 843-847

[3] GERBERICH, W. W., HEMMINGS, P. L., ZACKAY, V. F. and PARKER, E. R. *Interactions Between Crack Growth and Strain-Induced Transformation.* Fracture 1969: Proceedings of the 2nd International Conference on Fracture, London, England, April 1969, Chapman & Hall, 1969, p. 288-305.

[4] BIRAT, J P, and GERBERICH, W W *A Mestastable Austenite with Plane Stress Fracture Toughness Near 500,000 PSI-In.½*; October 1970, UCRL-20341.

Alloys containing large amounts of chromium will also suffer losses of this element and contain excessive amounts of chromium oxides. In addition, absorption of nitrogen, hydrogen, and oxygen must be controlled and limited to prevent embrittlement of the ingot.

Alloy Composition

As an initial undertaking, it was decided to cast heats of the Zackay et al.[1] A-2 composition. This composition in weight percent is as follows:

C	Mn	Si	Cr	Mo	Ni
0.25	2.08	1.96	8.88	4.04	7.60

Alloying Materials

Cabot coke was used for the carbon addition, with pure alloys (99.9% purity) being used for the manganese, silicon, chromium, molybdenum, nickel, and aluminum additions. Electrolytic iron was used for the iron addition.

Procedure

A zirconia basic lined air-induction furnace with a 333 kva power source was used to melt the heats of TRIP steel A-2 alloy.

In an effort to minimize the loss of chromium and prevent the excessive absorption of embrittling gases such as hydrogen, oxygen, and nitrogen during melting, a special shield was designed for the furnace and used to cover the molten surface of the metal with a blanket of argon gas. This shield consisted of a simple cylinder open at the ends with hoses attached to an argon supply. This easily removable shield was of sufficient diameter to encircle the open top of the furnace. A continuous flow of argon at a rate of 60 cubic feet per hour was maintained over the molten surface until just prior to pouring.

To compensate for anticipated furnace losses during melting, the percentages of elements charged were as follows:

C	Mn	Si	Cr	Mo	Ni
0.30	2.28	2.11	9.00	4.04	7.60

The Cabot coke, molybdenum, chromium, and nickel were placed in the furnace prior to putting the power on. Once the power was turned on the meltdown occurred, the temperature was brought up to 3100 F and manganese and silicon were then added (Figure 1). After a brief period, the power was turned off, and a final furnace temperature reading was taken (Figure 2). The melt was then poured into a clay graphite crucible (Figure 3) with 0.10 percent aluminum added to the molten stream in the ladle for final deoxidation. A chemical test block and the casting were then poured (Figure 4). The casting mold used for the first two heats was a machined graphite ingot mold, and for the remaining heats, a skin-dried green sand mold.

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Figure 1. Air-Induction Melting
of TRIP Steel Alloy Using Argon
Shield. Alloy Addition Being
Made.
19-066-824/AMC-71

Figure 2. Air-Induction Melting
of TRIP Steel Alloy Using Argon
Shield. Temperature Reading
Being Taken.

19-066-825/AMC-71

Figure 3. Air-Induction Melting of
TRIP Steel Alloy. Metal Being
Poured Into Ladle.

19-066-826/AMC-71

Figure 4. Air-Induction Melting of
TRIP Steel Alloy. Metal Being
Poured Into Mold.

19-066-827/AMC-71

After the casting had cooled, it was removed from the mold, and the sprue and ingate were removed. A chemical analysis of the heat and radiographs of the ingot were taken. Subsequently, the riser was removed from the ingot and specimens taken for examining the microstructure. In addition, one ingot was sectioned longitudinally, surface ground, and etched for examination of the macrostructure.

III. RESULTS AND DISCUSSION

The procedure for melting the TRIP steel A-2 alloy in air using the argon shield proved satisfactory. The loss of chromium during the melt was indeed minimized.

The chemical analysis, weight, and mold composition of the five heats poured are shown in Table I.

The argon blanket covering the molten surface in the furnace during the melting procedure did successfully prevent the excessive absorption of embrittling gases.

As seen in the table, hydrogen contents ranged from 4 parts per million (ppm) (Heat B) to 7 ppm (Heat A). Oxygen contents ranged from a low of 54 ppm (Heat D) to a high of 83 ppm (Heat B), while nitrogen contents ranged from 230 ppm (Heat A) up to 490 ppm (Heat D).

The soundness of the Heat E ingot is shown in Figure 5. The radiograph reveals a solid casting with no evidence of shrinkage or porosity.

The as-cast macrostructure of Heat E is shown in Figure 6. Extremely large grains are shown growing from the outside of the ingot toward the center. Columnar dendrites are in evidence throughout. The structure shows a highly alloyed material with a large amount of segregation resulting from the slow solidification of the metal in the sand mold.

Typical as-cast microstructures of a heat cast in a machined graphite ingot mold are shown in Figures 7a and 7b. These figures show the structure of Heat B in both the etched and unetched conditions. In the etched condition (Figure 7a), a fairly coarse eutectic microstructure is evident with a great deal of segregation, and the grains are quite large. The unetched specimen (Figure 7b) clearly shows a eutectic condition has resulted.

Typical as-cast microstructures of a sand mold ingot are shown in Figures 8a and 8b. In this case (Figure 8a), the structure is extremely coarse due to the very slow solidification of the metal in the sand mold. Spacing between dendritic arms is very large. In addition, there is a large amount of precipitate at the grain boundaries which is quite evident in the unetched condition as shown in Figure 8b.

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Figure 5. Radiograph of TRIP Steel
A-2 Alloy Ingot. Heat E, As-Cast
Condition.
19-066-803/AMC-71
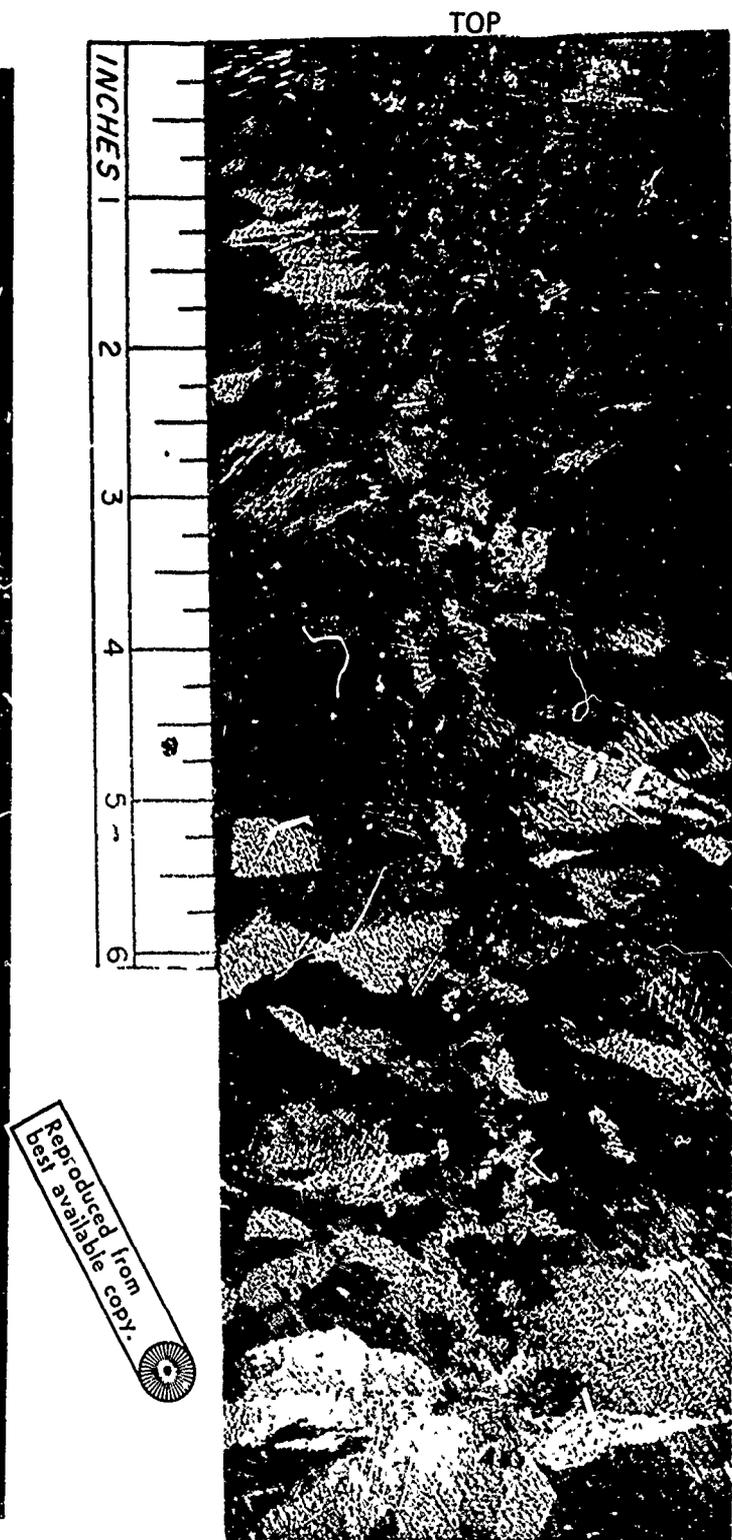
Figure 6. Macroetched Longitudinal Section of TRIP Steel
A-2 Alloy Ingot. Heat E, As-Cast Condition.
19-066-801/AMC-71

a. Etch: 5 g $CuCl_2$, 100 ml HCl, 100 ml $C_2H_5(OH)$, 100 ml H_2O.

b. Unetched

Figure 7. Microstructure of TRIP Steel A-2 Alloy. Heat B, As-Cast Condition, Mag. 50X.
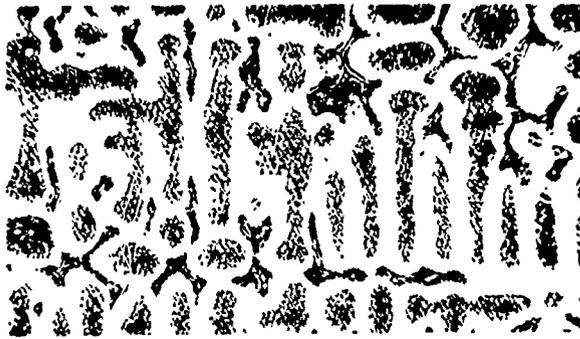
a. Etch: 5 g $CuCl_2$, 100 ml HCl, 100 ml $C_2H_5(OH)$, 100 ml H_2O.

b. Unetched

Figure 8. Microstructure of TRIP Steel A-2 Alloy. Heat E, As-Cast Condition, Mag. 50X.

Table I. TRIP STEEL A-2 ALLOY

Heat	Weight (lb)	Mold Type	Element (wt%)											
			C	Mn	Si	Cr	Mo	Ni	Al	P	S	H	O	N
A	40	Machined Graphite	0.22	2.07	2.04	8.84	4.25	7.64	0.04	0.004	0.008	0.0007	0.0062	0.0230
B	40	Machined Graphite	.28	2.03	2.10	9.06	4.17	7.72	.03	.005	.008	.0004	.0083	.0320
C	95	Sand	.28	2.12	1.98	8.80	4.02	7.38	.04	.003	.008	.0006	.0062	.0260
D	95	Sand	.32	2.07	2.08	8.81	4.01	7.41	.04	.003	.006	.0006	.0054	.0490
E	95	Sand	.27	2.15	2.03	8.97	3.99	7.81	.05	.005	.006	.0005	.0055	.0470

IV. EFFECTS OF ALLOY COMPOSITION ON MARTENSITIC TRANSFORMATION

The composition of TRIP steels have varied greatly. However, no attempt appears to have been made to optimize alloy composition with respect to mechanical properties. To do this, the effect of composition on both transformation temperatures M_s and M_d* and on the ausformed microstructure must be determined.

Transformation strengthening of a TRIP steel at any temperature requires that the M_d be above and the M_s below that temperature. Thus the alloy is normally austenitic after ausforming and before plastic deformation at application or test temperature. The martensite produced upon subsequent deformation increases the toughness and the rate of work-hardening of the steel.

A basic composition was established for determining the effects of variations in alloy composition on M_s and M_d (Table II). The effects of compositional changes were adjusted empirically by balancing the nickel content to maintain a constant M_s. The amount of each element in weight percent was determined in part by the following equation:[5,6]

$$M_s(°F) = 1042-30Ni-30(1+0.1Ni)Cr-60Mn-30Mo-20Si+30Co-853(1+0.11Ni)C \quad (1)$$

Initially twenty 900-gram button heats were cast utilizing a nonconsumable electrode, inert atmosphere melting technique in accordance with the desired chemical compositions listed in Table II. The first entry in the table is identified as the basic TRIP steel composition. Each compositional element was varied along with the nickel content in order to maintain the M_s predicted by Equation 1 at 0 F. The influence of chemical composition on M_d could be then determined independently on the basis of the shifting of the entire transformation (M_s-M_d) range. The effect of alloying elements could also be determined.

The carbide-forming constituents, Mo and C, were varied a few percent in an effort to enhance carbide strengthening during warm working. A range of Cr and Si additions were included to test for M_d alterations. Cobalt was also added to selected compositions in order to investigate the potential of solution hardening.

*M_s Highest temperature at which martensite will form on cooling.
M_d Highest temperature for the initiation of strain-induced martensite.

[5]STEVEN, W. and HAYNES, A. G The Temperature of Formation of Martensite and Bainite in Low-Alloy Steels. J. of the Iron and Steel Institute, August 1956, p. 349-359.

[6]YEO, R. B G. The Effects of Some Alloying Elements on the Transformation of Fe-22.5% Ni Alloys. Trans. AIME, v. 227, August 1963, p. 884-890.

Table II. CHEMICAL COMPOSITION OF TRIP STEELS*

Designation	Ni	Mn	Mo	C	Cr	Si	Co
Basic	10.52	2.00	4.00	0.25	-	-	-
C-20	12.37	2.00	4.00	0.20	-	-	-
C-22	11.59	2.00	4.00	0.22	-	-	-
C-28	9.55	2.00	4.00	0.28	-	-	-
Mo-2	11.77	2.00	2.00	0.25	-	-	-
Mo-3	11.14	2.00	3.00	0.25	-	-	-
Mo-5	9.89	2.00	5.00	0.25	-	-	-
Mo-7	8.64	2.00	7.00	0.25	-	-	-
Si-0.5	10.10	2.00	4.00	0.25	-	0.50	-
Si-1	9.69	2.00	4.00	0.25	-	1.00	-
Si-2	8.86	2.00	4.00	0.25	-	2.00	-
Si-4	7.21	2.00	4.00	0.25	-	4.00	-
Cr-1	9.45	2.00	4.00	0.25	1.00	-	-
Cr-2	8.48	2.00	4.00	0.25	2.00	-	-
Cr-4	6.81	2.00	4.00	0.25	4.00	-	-
Cr-8	4.24	2.00	4.00	0.25	8.00	-	-
Co-0.5	10.79	2.00	4.00	0.25	-	-	0.50
Co-1	11.06	2.00	4.00	0.25	-	-	1.00
Co-2	11.62	2.00	4.00	0.25	-	-	2.00
Co-4	12.72	2.00	4.00	0.25	-	-	4.00
†W-2	4.39	1.00	4.00	0.22	10.00	-	-
†W-3	3.90	1.00	4.00	0.22	10.00	2.00	-
†W-4	4.27	0.50	4.00	0.22	10.00	2.00	-
W-5	5.00	0.50	4.00	0.22	10.00	2.00	2.00
W-6	5.19	0.25	4.00	0.22	10.00	2.00	2.00
W-7	4.77	0.25	5.00	0.22	10.00	2.00	2.00
†W-8	5.90	0.25	4.00	0.20	10.00	2.00	3.00

*Chemical composition is in weight-percent, balance is
Fe.
†Not forged.

Processing

Initially all compositions listed in Table II with the exception of the W
series were processed. Each button was hammer forged at 1900 F to a final thick-
ness of approximately 0.300 inch. The forgings were then homogenized at 2250 F
for one hour followed by water quenching. Selected forgings were subsequently
rolled at 840 F to a total reduction of 50 percent in cross-sectional area.
Close examination of these forgings revealed that the actual M_s values were higher
than predicted, and all the material at least partially transformed above room
temperature.

An unsuccessful effort was made to reclaim the transformed experimental
material by lowering the M_s below room temperature. It had been noted by Zackay
et al.[1] that the 840 F warm working depressed the M_s temperature to a considerable
degree. The alloys, martensitic at room temperature, were therefore carried
through a second processing schedule whereby the warm working was accomplished
directly from the solutionizing temperature as follows:

(a) 2250F - 1 hr.
(b) Furnace cool to 850 F - stabilize.
(c) Roll at 850 F - 10 percent reduction per pass (70 percent total).

A new basic composition, as represented in the W series in Table II, was established based on a calculated martensitic transformation temperature of -40 F (Equation 1). This material behaved as predicted and did not transform at room temperature.

This series was cast in 300-gram buttons utilizing the previously described melting practice. Three buttons (W5, W6, and W7) were processed as follows:

(a) Forge at 1980 F to 0.316 inch
(b) Homogenize - 2250 F - 1 hr - water quench
(c) Roll at 850 F - 10 percent reduction per pass (70 percent total), final thickness 0.100 inch

Test samples, 1/4 inch by 2 inches, were sheared from specimen W5 and TRIPed by bending approximately 15 degrees in a sheet metal brake. Macrographs of an etched bend sample at 6X (Figure 9a) and 10X (Figure 9b) demonstrate that martensite formed in the local high deformation regions of the bend. It is interesting to note that a greater martensite volume fraction appears to be formed on the tension rather than on the compression side of the specimen. Furthermore, the remaining untransformed region in the deformed cross section lies on the compression side of the neutral axis. Hardness measurements confirm this observation.

Vickers hardness readings were taken across the bend specimen (Figure 10). Table III shows the resulting Vhn and equivalent Rockwell numbers. Estimated tensile strengths based on comparative hardness values are listed. It can be seen that an increase of approximately 30 percent in tensile strength was realized.

It is of passing interest that one could produce a sheet material having a hardness gradient from edge-to-edge if so desired. This could conceivably be achieved by a technique such as bending or corrugating a sheet above the M_d and then straightening below the M_d temperature. This effect can be observed in Figures 11a and 11b.

Determination of Martensitic Transformation Temperature

A survey of potential techniques that could be utilized for monitoring martensitic transformation was conducted. Potential techniques based on the following material characteristics were investigated.

1. Resistivity: Identified here as that change in the specimen resistance due to the austenite-to-martensite transformation occurring during tension testing.

Measurements during tension testing of a steel with TRIP characteristics (AISI 303) revealed that the resistance change during the austenite- -martensite transformation was insignificant in comparison to that calculated for a specimen undergoing constant volume length change (Appendix Figure A-1).
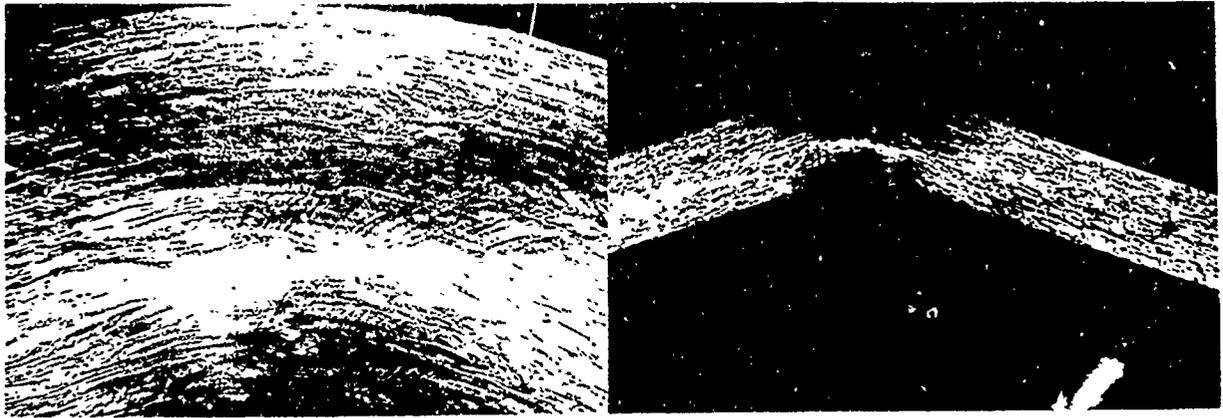
Mag. 6X   b. Mag. 30X

Figure 9. Macrograph of Bend Sample W5.
Etch: 5 g $CuCl_2$, 100 ml HCl, 100 ml $C_2H_5(OH)$, 100 ml H_2O.

Figure 10. Bend Specimen W5, Showing Hardness Reading Locations, Unetched, mag 6X.

Table III. HARDNESS SURVEY OF BEND SAMPLE, W5

Location	Vhn	R_c	TS*(ksi)
1	532	51	254
2	476	47	227
3	584	54	279
4	598	55	293

*Tensile strength based on comparative hardness data.

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2. Indenting: Inducing martensitic transformation at selected test temperatures in sheet TRIP steel using a Brinell hardness tester and subsequently measuring the permeability of the bulged region.

This technique yielded a gradual change in slope when plotting magnetic attraction versus indenting temperature. Therefore no accurate determination of the transformation temperature on deformation could be made (Appendix Figures A-2 and A-3).

3. Ferromagnetic: Utilizing the change in AC permeability to determine the volume fraction martensite during tension testing.

The ferromagnetic technique is yet to be appraised. Preliminary design and fabrication of the electromechanical components has been completed; actual application under test conditions will be evaluated.

V. PROCESSING OF TRIP STEEL

The development of high strength in a TRIP steel is dependent upon a preliminary thermomechanical treatment. Reductions of 50 to 80 percent in a temperature range of 500 to 850 F produce a fine carbide dispersion in the austenite which gives rise to high yield strength. Many problems associated with this heavy, warm-temperature working are yet to be resolved. The homogenization of alloy carbides and the control of austenite grain size prior to warm working are significant considerations because of the decreased high temperature gamma-to-delta

a. Mag. 6X   b. Mag. 30X

Figure 11. Specimen W5, Bent Above M_d and Straightened Below M_d.
Etch: 5 g $CuCl_2$, 100 ml HCl, 100 ml $C_2H_5(OH)$, 100 ml H_2O.

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transformation with increasing alloy content. The achievement of a uniform warm-working effect in forming operations other than rolling has been a problem in conventional ausforming, and remains a difficulty in producing high-strength TRIP steels.

In order to provide base-line data and minimize lead time in evaluating process variables, work was initiated on a vacuum-melted ingot. The information generated would then be applied to air-melted ingots when they became available.

Processing of a Vacuum-Melted Ingot

A 40-lb ingot of the A-2 alloy was produced by vacuum-induction melting. The chemical composition in weight percent was as follows:

C	Mn	Si	Cr	Mo	Ni	P	S	H	O	N
0.27	2.10	2.30	8.90	4.00	7.7	0.005	0.006	0.0005	0.0039	0.0120

This ingot was subsequently hammer forged at 1980 F from an initial diameter of 3-1/2 inches to final diameters of 3/8, 5/8, 1, 1-1/4, and 1-1/2 inches.

Radiographic inspection revealed that all forged diameters of less than 1 inch displayed centerline cracking. The probable cause for the behavior will be discussed later.

Warm swaging (i.e., 850 F) was attempted with a 1-inch diameter bar, the maximum reduction in cross-sectional area achievable was 30 percent. Inadequate holding fixtures prevented higher reductions.

In order to fix proper homogenization time and temperature, material was selectively sectioned from a 5/8-inch-diameter forged bar. Sections 1/2-inch long were processed in a temperature range of 2200 to 2400 F, at 50 F intervals, with holding times of one and six hours (Table IV). The resulting microstructures are shown in Figures 12 and 13.

Table IV. HOMOGENIZATION TREATMENT
(Vacuum-Induction Heat)

Specimen	Temperature (F)	Time (hr)	Hardness†
a*	–	–	31 R_c
b	2200	1	87 R_b
c	2250	1	87 R_b
d	2300	1	83 R_b
e	2350	1	84 R_b
f	2400	1	84 R_b
g	2200	6	84 R_b
h	2300	6	86 R_b

*As-forged condition
†Average value - 5 readings

Based on the results of the trial homogenization treatments, it was decided that future treatments would be fixed at 2300 F for 6 hours. It can be seen treatment at 2350 F for 1 hour resulted in excessive grain growth (Figure 12e) whereas at 2400 F for 1 hour treatment a second phase formed at the grain boundaries (Figure 12f). The 2300 F treatment for 6 hours (Figure 13b) provided the best compromise for reducing segregation and controlling grain growth. The relative insensitivity of hardness to homogenization temperature or time (Table IV) indicates that the amount of carbides dissolved is not great enough to affect the strength of the material.

12
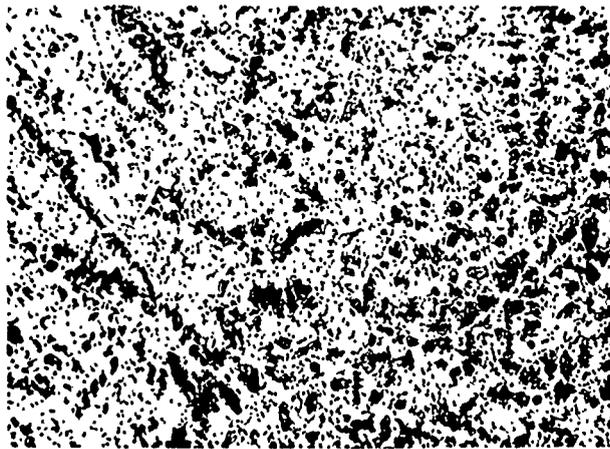
a. As Forged
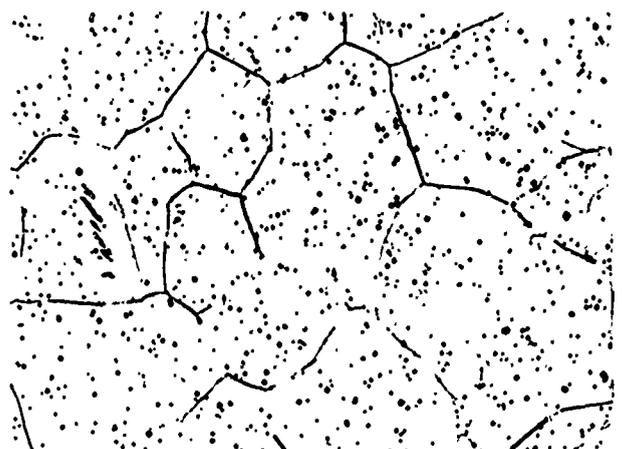
b. 2200 F - 1 hr.
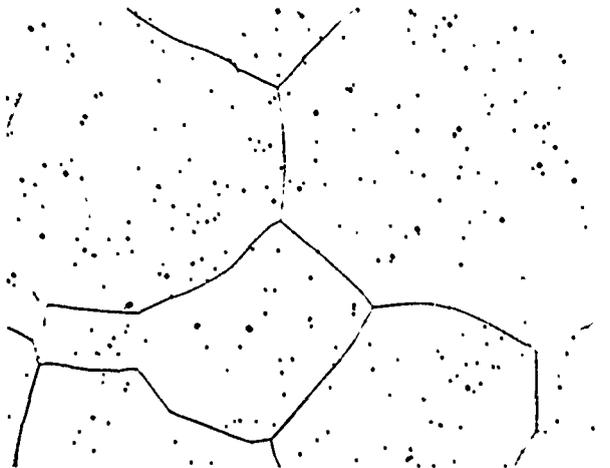
c. 2250 F - 1 hr.
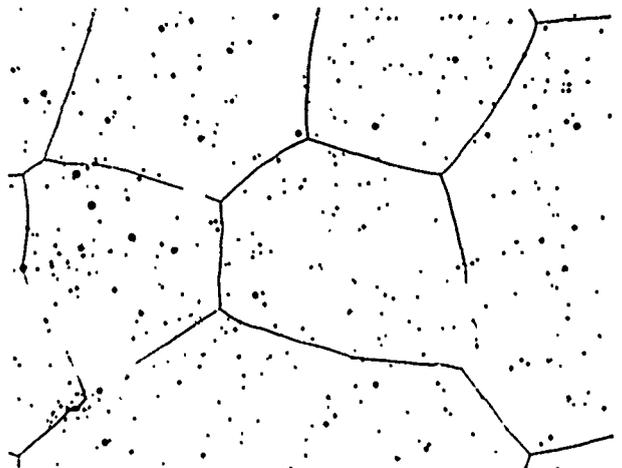
d. 2300 F - 1 hr.

e. 2350 F - 1 hr.
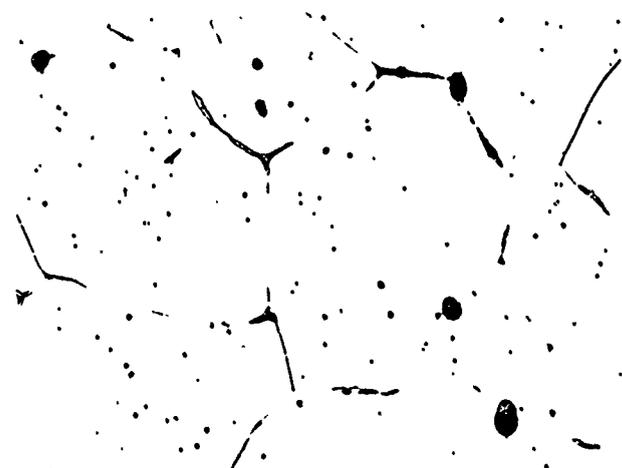
f. 2400 F - 1 hr.

Figure 12. Homogenization Treatment, 1 Hour (Vacuum-Induction Ingot) Mag. 100X.
Etch: 10 ml HNO_3, 30 ml HCl, 3 drops $FeCl_3$.

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Processing of Air-Melted Ingot

A 40-lb air-induction melted ingot (Heat D, Table I) was selected for processing. The chemical analysis in weight percent was:

C	Mn	Si	Cr	Mo	Ni	Al	P	S	H	O	N
0.32	2.07	2.08	8.81	4.01	7.41	0.04	0.003	0.006	0.0006	0.0054	0.0490

The ingot was machined to 3-1/2 inches in diameter by 11-1/2 inches long and subsequently homogenized in vacuum at 2300 F for 6 hours and furnace cooled.

The ingot was hammer forged at 2100 F to a 1-inch square. A short section of this stock was additionally forged to final diameters of 1, 7/8, 5/8, 1/2, and 3/8 inch. Radiographic inspection revealed that the material reduced below 1 inch in diameter had centerline bursting. This behavior was independent of billet history.

Under certain conditions, the inside of a rod is subjected to high tensile stresses when compressed between two parallel platens. Continual compressive forces on the circumference of a bar finally give rise to a cavity along the centerline. This behavior is actually useful when employed in the manufacture of seamless tubing by rotary piercing. Bursting can be avoided by providing support for the bar in a V-shaped anvil while it is being reduced. This technique loads the center of the bar in compression. This method, however, is difficult to apply to small-diameter rods.

a. 2200 F - 6 hr.   b. 2300 F - 6 hr.
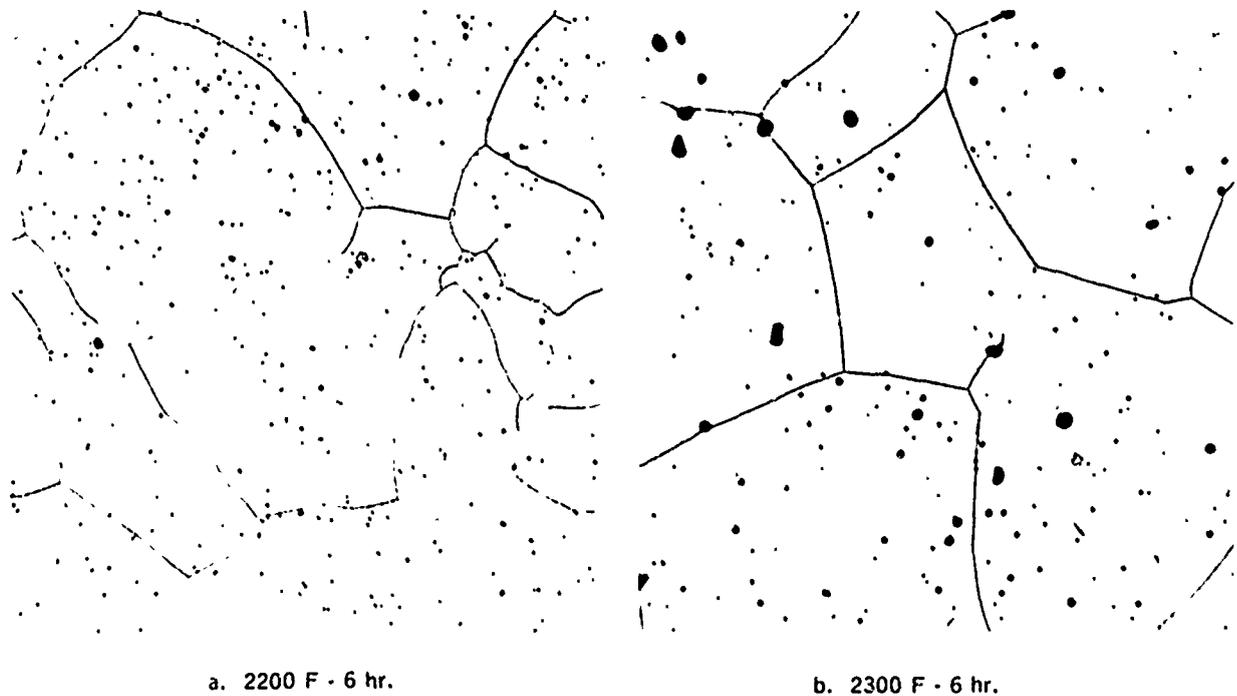
Figure 13. Homogenization Treatment, 6 Hours (Vacuum-Induction Ingot), Mag. 100X.
Etch: 10 ml HNO_3, 30 ml HCl, 3 drops $FeCl_3$.

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A selected quantity of the 1-inch square stock was processed according to the schedule shown in Table V. The stock was rolled at 2100 F to various thicknesses based on maintaining a constant final thickness after warm working for reductions of 0, 20, 40, 60, and 80 percent.

The hot rolling was followed by austenization at 2200 F for one hour, water quenching and subsequent processing at 850 F to a final thickness of approximately 0.100 inch.

Material at 0 percent reduction has been machined into sheet tension specimens (t=0.050"); tension tests have not yet been conducted.

VI. RECOMMENDATIONS AND CONCLUSIONS

Air-Induction Melting

The melting procedure utilizing an argon shield appears to be satisfactory for the melting of the A-2 TRIP steel alloy. Loss of chromium was minimized, and excessive absorption of embrittling gases was prevented. Radiographically sound ingots can be produced.

Slow solidification of the metal in the sand mold resulted in an extremely coarse structure in the ingots. Castings produced in the machined graphite mold resulted in a less coarse structure due to the more rapid solidification. Since this is a highly alloyed material, a slowly solidified structure is less conducive to homogenization and, therefore, less suitable for forging. A rapidly solidified ingot, such as that which would be produced with a metal mold, would be most desirable for this TRIP alloy.

Alloy Composition

Precipitation of carbides and the resulting matrix depletion of carbon and other alloying elements increases the M_s and M_d near the carbides. A method for monitoring martensitic transformation is essential. In this area, the ferromagnetic technique currently offers the most promise; its evaluation will continue.

Pending the development of a reliable technique for martensitic transformation temperature measurements, compositional studies will be curtailed.

Processing of TRIP Steel

Although a homogenization treatment of 2300 F for 6 hours yielded the best structure, other time-temperature conditions should be evaluated. It must also be determined how billet history affects the resulting mechanical and physical properties.

Table V. ROLLING SCHEDULE, HEAT D

R.A (%)	t_f*	t_f†
0	0.119	–
20	0.114	0.115
40	0.173	0.104
60	0.261	0.104
80	0.505	0.101

*Hot rolled at 2100 F to this thickness

†Warm rolled at 850 F to this thickness

The evaluation of properties resulting from warm working through selected deformation schedules should be continued.

AISI 303 STAINLESS STEEL

R_o = ORIGINAL RESISTANCE
R_i = INSTANTANEOUS RESISTANCE
0.252" DIA TENSILE SPECIMEN
NECKING OCCURRED AT $\epsilon = 0.54$

Experimental

Theoretical $\left(\dfrac{R_i}{R_o} = e^{2\epsilon} \right)$

$\dfrac{R_i}{R_o}$

ϵ
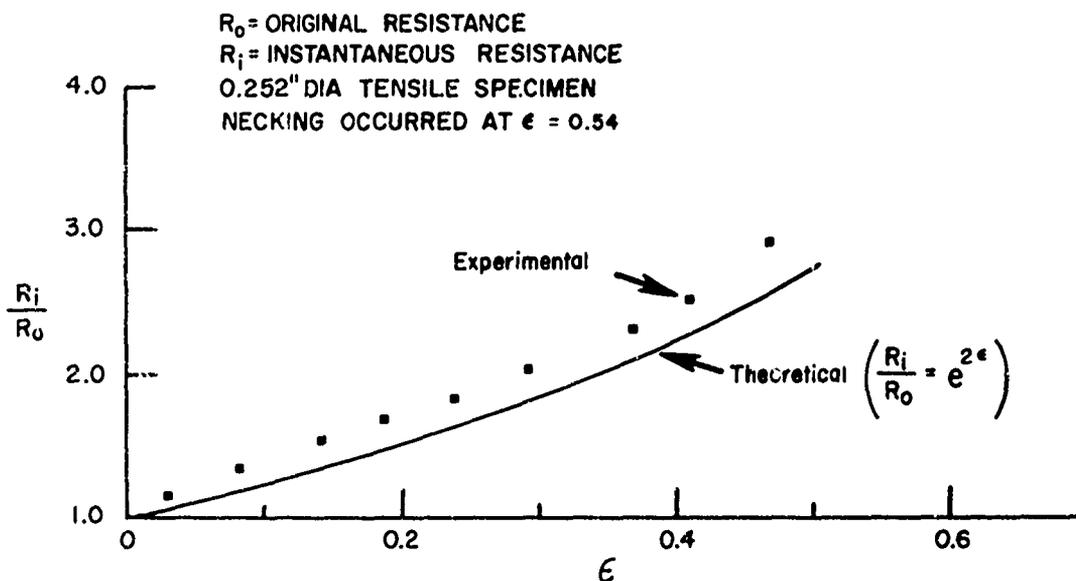
Figure A-1. Resistance Ratio Versus True Longitudinal Strain.

● TRIP PROCESSED
⊕ TRIP PROCESSED ANNEALED 1000 F 1HR

ALLOY A-2 THICKNESS 0.110"
INDENTOR - 10MM DIA BALL
LOAD 2000 KG

CRACKED

MAGNEGAGE READING

INDENTATION TEMPERATURE, °C
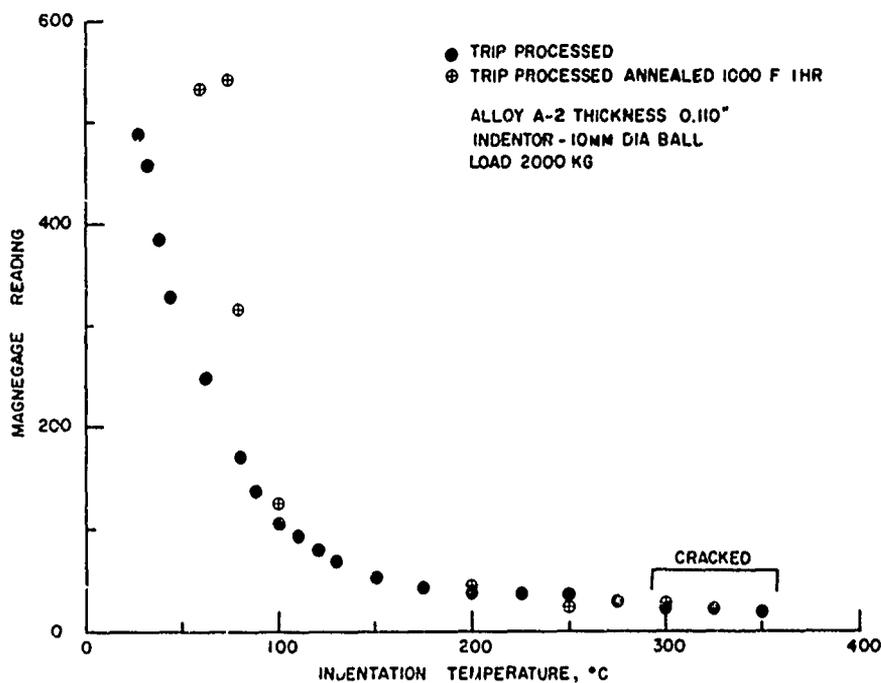
Figure A-2. Magnetic Attraction Versus Indentation Temperature.

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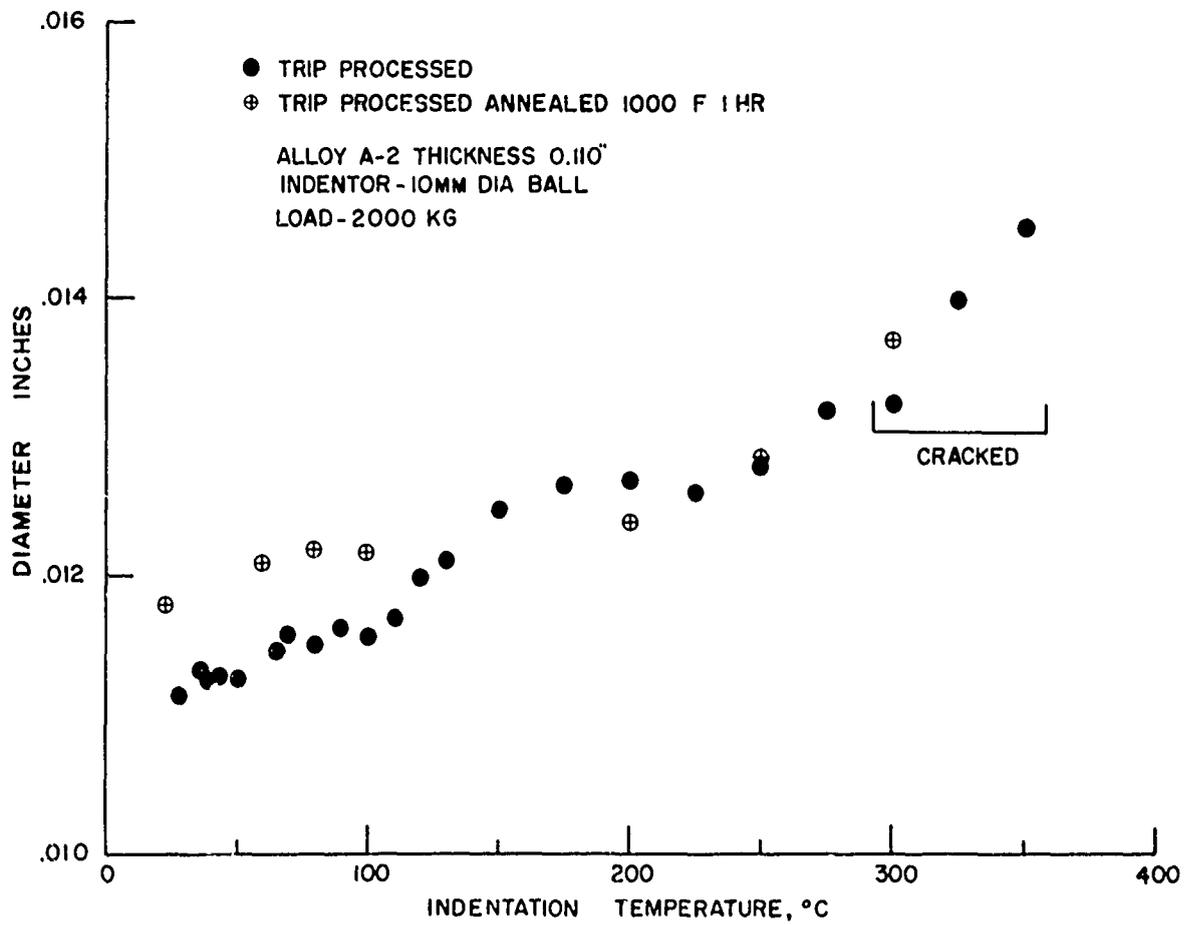
Figure A-3. Indentation Diameter Versus Indentation Temperature.

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