Lawrence Berkeley National Laboratory
Recent Work

Title

STRUCTURE AND PROPERTIES OF TRIP STEELS PROCESSED BY DEFORMATION AND THERMAL CYCLING.

Permalink

https://escholarship.org/uc/item/7bm47011

Author

Adkins, Howard E.

Publication Date

1973-04-01

eScholarship.org Powered by the California Digital Library
University of California

LBL-1491
c.1

STRUCTURE AND PROPERTIES OF TRIP STEELS PROCESSED BY DEFORMATION AND THERMAL CYCLING

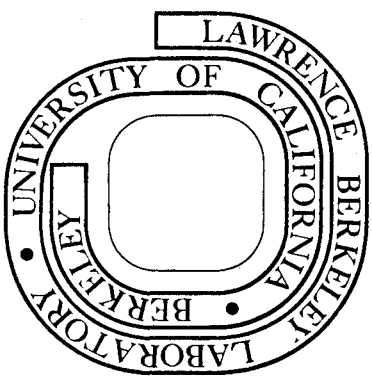
Howard E. Adkins, Jr.
(M.S. Thesis)

April 1973

Prepared for the U. S. Atomic Energy Commission
under Contract W-7405-ENG-48

For Reference

Not to be taken from this room



LBL-1491
c.1

DISCLAIMER

This document was prepared as an account of work sponsored by the United States Government. While this document is believed to contain correct information, neither the United States Government nor any agency thereof, nor the Regents of the University of California, nor any of their employees, makes any warranty, express or implied, or assumes any legal responsibility for the accuracy, completeness, or usefulness of any information, apparatus, product, or process disclosed, or represents that its use would not infringe privately owned rights. Reference herein to any specific commercial product, process, or service by its trade name, trademark, manufacturer, or otherwise, does not necessarily constitute or imply its endorsement, recommendation, or favoring by the United States Government or any agency thereof, or the Regents of the University of California. The views and opinions of authors expressed herein do not necessarily state or reflect those of the United States Government or any agency thereof or the Regents of the University of California.

0 0 0 0 3 9 0 2 7 5 9

STRUCTURE AND PROPERTIES OF TRIP STEELS PROCESSED BY DEFORMATION AND THERMAL CYCLING

Howard E. Adkins, Jr.

Inorganic Materials Research Division, Lawrence Berkeley Laboratory and Department of Materials Science and Engineering, College of Engineering; University of California, Berkeley, California

ABSTRACT

Strengthening of metastable austenitic chromium containing TRIP steels by thermal and thermomechanical processing was investigated. Three processing treatments were studied for their influence on structure and mechanical properties. A considerable increase in strength resulted from the use of small amounts of deformation by rolling at temperatures near the M_s temperature to form "stress induced" martensite. On heating to an elevated temperature, martensite reverted to austenite by a diffusionless reaction. The reversion was accompanied by diffusion based reactions resulting in the formation of austenite of a lower stability. After repeated cycles of the processing treatments, a fine and uniform structure resulted.

Room temperature yield strengths of steels with the fine structure were much higher than that of annealed austenite, but were lower than those of TRIP steels produced by 80% warm work. Selected specimens exibited higher yield strengths in 100°C tests suggesting that the low room temperature yield strengths were due to the stress induced formation of martensite.

I. INTRODUCTION

Investigations of strengthening mechanisms in alloys have stressed the importance of increasing the dislocation density to improve strength.[1,2] In steels, two methods which have been used to increase the dislocation density are thermomechanical processing and the reversion of martensite to austenite by a diffusionless transformation.

High strength metastable austenitic steels of high toughness, known as TRIP (acroynm for TRansformation Induced Plasticity) steels, are one class of steels in which enhancement of strength is achieved by thermomechanical processing.[3] Conventionally, these steels are reduced 70-80 pct in thickness by rolling at temperatures above the M_d in order to produce a high density of dislocations in the austenite. The steels achieve their high toughness and ductility from the deformation induced transformation of austenite to martensite during testing at temperatures below their M_d temperatures.[3-7] One of the major obstacles in the processing of TRIP steels is the large amount of mechanical deformation required to attain the high yield strength.

Thermal reversion of martensite to austenite, as a means of achieving a high dislocation density in austenite, has been the subject of considerable investigation. Several researchers have reported the reversion of martensite in Fe-Ni,[8,9] Fe-Ni-C,[10] and Fe-Ni-V-C[11] alloys. These investigations have shown that reverted austenite (austenite formed by the thermal reversion of martensite) is stronger than annealed austenite, and that significant improvements in properties can sometimes be achieved with multiple cycles of martensite to austenite reversion. The enhanced strength has been attributed to

lattice imperfections[8] and a high concentration of tangled and jogged dislocations[9] in reverted austenite. Hyatt and Krauss[10] suggested that in addition to the diffusionless reversion of martensite to austenite, two diffusion based changes could take place in Fe-Ni-C alloys. These are, recrystallization of reverted austenite, and precipitation of carbides.

Recently, Koppenaal[12] used the reversion of martensite to attain high strength in an Fe-Ni-Mo-C TRIP steel. He reported that thermal cycling of the steel between -196°C and 704°C resulted in mechanical properties comparable to those attained by the use of thermomechanical processing. As pointed out by Koppenaal, thermal cycling eliminates the requirement that large amounts of deformation be used to obtain a high density of dislocations.

The current investigation was initiated with two main objectives. First, Koppenaal's work showed the applicability of martensite reversion in a Fe-Ni-Mo-C TRIP steel, but did not indicate whether TRIP steels of other compositions could be strengthened by thermal cycling. Most TRIP steels have compositions considerably different from the one used by Koppenaal for his investigation. Among these are the corrosion resistant TRIP steels in which chromium, a moderate carbide-former, is an essential alloying element.[13] Also, in these steels, the nickel contents are considerably lower than those in the steels studied by Koppenaal and most other workers. It was of interest to examine whether thermal cycling could be universally employed to strengthen all TRIP steels.

0 0 0 0 3 9 0 2 7 0

Secondly, it was implied in Koppenaal's work that thermal cycling could not be conveniently applied to TRIP steels with M_s temperatures below -196°C. Many of the potentially useful TRIP steels have M_s temperatures below -196°C, and to process them by thermal cycling, process modifications would be needed. Several investigators have shown that in steels of low stability, large amounts of martensite should be formed by small amounts of deformation (low stresses) at temperatures near, but above, the M_s temperature.[14-16] Generally, martensite formed at low stresses is called "stress induced" martensite, and martensite formed by macroscopic plastic strain is called "strain induced". This distinction is some what arbitrary because it is the local state of stress that induces martensite formation in both cases.[16] It was believed that if TRIP steels could be designed to meet the above stability criterion, then they could be strengthened by the process of martensite reversion even if the M_s temperature was below -196°C.

II. EXPERIMENTAL PROCEDURE

A. Alloy Preparation and Initial Processing

The alloys were induction melted in an Argon atmosphere and cast into 20 lb ingots in a copper mold. The ingots were homogenized at 1150°C for three days and allowed to air cool. During homogenization, the ingots were placed in steel tubes packed with cast iron chips in order to reduce surface decarburization.

Following homogenization, the ingots were upset forged at 1100°C into plates 1/2" thick and 2 1/2" wide. These plates were rolled at 450°C into sheets 0.150" thick. The sheets were pickled in an acid bath to remove the surface scale. The pickled sheets were sealed in stainless steel containers, austenitized at 1200°C for one hour and brine quenched. The austenitized sheets were cut into tensile specimen blanks (TSB) with the length parallel to the original rolling direction. The chemical compositions of the steels are given in Table I.

B. Final Processing

The tensile specimen blanks (TSB) were treated by one of three thermo-mechanical processes described below.

1. Thermal cycling with no mechanical deformation (process T-1). This process, used only with steel A, consisted of cooling the TSB to -196°C, air warming to room temperature, and holding for two minutes in a salt bath set at a predetermined temperature for the reversion of martensite to austenite. The TSB was subsequently air cooled. The cycle described above was repeated up to ten times. A schematic for this process is shown in Fig. 1.

2. Thermal cycling with a small amount of mechanical deformation in the first cycle (process TM-1). The first step in this process

0 0 0 0 3 9 0 2 7 1

consisted of rolling the TSB by a small amount (9 to 10 per reduction in thickness) at a cryogenic temperature (-78°C or -196°C). Subsequently, the TSB was treated as in process T-1. A schematic for process T-1 is shown in Fig. 2a for steel A and in Fig. 2b for steel B.

3. Thermal cycling with a small amount of mechanical deformation during each cycle (process TM-2). The TSB was deformed by a small amount (9 to 10 per reduction in thickness) at a cryogenic temperature (-78°C or -196°C), heated for two or three minutes in a salt bath at a predetermined temperature for the reversion of martensite to austenite, and air cooled to room temperature. This cycle was repeated up to seven times. Figures 3a and 3b present process TM-2 schematics for steels A and B respectively.

C. Mechanical Testing

Sheet tensile specimens of one inch gage length were machined from the blanks processed as described in the previous section. A sketch of the specimen is shown in Fig. 4. Tensile tests were carried out in air on an Instron testing machine at room temperature and 100°C; a strain rate of 0.04/min was employed. The yield stress was taken as the upper yield point when a yield point occurred. In the absence of a yield point, the 0.2 pct offset method was employed to calculate the yield stress. Specimen elongation was determined from measurements, made both before and after testing, of the distance between two small indentations on the gage section. A traveling microscope with an accuracy of ±0.004 inch was used for the measurements.

D. Magnetic Measurements

Saturation induction of selected specimens was measured before and during mechanical testing. The observations were converted to volume percentage of martensite with corrections being made for the influence of alloying elements.[17,18] The equipment and technique used have been discussed elsehwere.[19]

E. Dilatometry

Cylinderical specimens approximately 1/4" in diameter and one inch long were machined from sheets of steel processed to form martensite either by cooling or by deformation. The cylinders were heated in a dilatometer at a rate of approximately 50°C per minute. Quartz rods were used to measure the change in length and the change was continuously recorded as a function of time on an X-Y recorder. Temperature was measured by a thermocouple spot welded to the specimen surface.

F. Metallography

Specimens were mechanically ground and polished, and then chemically etched in a solution of 5.0 gm cupric chloride, 100 ml hydrochloric acid, 100 ml methyl alcohol, and 100 ml distilled water. A Carl Zeiss Optical Microscope was used for observation and photography.

G. Fractography

Fracture surfaces of tensile specimens were examined in a JEOLCO JSM-U3 scanning electron microscope operated at 25 kV.

0 0 0 0 5 9 0 2 7 2

III. RESULTS AND DISCUSSION

Steels A and B were alloyed such that their M_s temperatures were below -78°C and -196°C respectively. Though the M_s temperatures were not experimentally determined, steel A specimens cooled to -78°C were found to be austenitic, and specimens cooled to -196°C were found to contain both austenite and martensite by metallographic techniques. Metallographic examination of steel B specimens cooled to 196°C revealed an austenitic structure. Both steels formed large amounts of martensite when deformed by small amounts at -78°C and -196°C. The above structural observations were confirmed by magnetic tests.

Preliminary tests were conducted using the dilatometer to determine if a phase change occurred on heating partially martensitic specimens of steel A, as well as the approximate temperature at which it occurred. Heating rates obtained in the dilatometer were less than those in the salt baths used for heat treating the TSB's. This difference in heating rates could cause a change in the temperature required for the phase change. Therefore, subsequent tests to determine the optimum processing conditions involved hardness measurements and metallographic examinations of specimens heat treated in salt baths.

A marked decrease in volume, indicating a phase change from martensite to austenite, occurred in specimens of steel A heated in the dilatometer. The temperatures for the start and completion of the phase change were estimated as 550°C and 714°C respectively. Metallographic examination at the end of the test revealed that no new grains were nucleated in specimens undergoing the phase transformation. Based on this finding, the phase change was identified as the diffusionless

reversion of martensite to austenite. A phase change occurred at 195°C when the specimens were cooled. At the end of the test, the specimen was weakly magnetic. The phase change at 195°C was identified as the transformation of austenite to martensite. Though the M_s temperature of the steel was initially below -78°C, it was apparent that a depletion of carbon and alloying elements from the matrix by carbide precipitation raised the Ms to 195°C. In repeated tests on the specimen, martensite reversion started at progressively lower temperatures and was completed at a high temperature. Martensite formation during cooling, on the other hand, occurred at progressively higher temperatures in repeated tests. It was also noted that specimens became increasingly magnetic in subsequent tests. These observations indicated that there was a progressive decrease in austenite stability due to increasing amount of carbide precipitation in successive tests. This indication was confirmed by the finding that annealing the specimen at 860°C to attain partial dissolution of carbides lowered the temperature for the start of the austenite to martensite transformation. All the above features of the dilatometry curves are illustrated in Fig. 5 for steel A.

As stated earlier, dilatometric tests were used to determine the approximate reversion temperatures. Subsequent tests were conducted using the salt bath to heat alloy specimens to several temperatures in the neighborhood of the dilatometrically determined reversion temperature with heating times of 2, 3, 5 and 10 minutes. Hardness measurements were taken on each specimen and the variation of hardness with time and temperature indicated that a temperature of 765°C and a time of 2 min were the optimum conditions for reverting the martensite in steel A. Selection

0 0 6 0 3 9 0 2 7 3

of the temperature of 780°C and the time of 3 minutes for steel B was additionally influenced by the fact that specimens with this heat treatment had the smallest amount of martensite, as indicated by magnetic tests.

The post processing mechanical properties of steels A and B are shown in Figs. 6 and 7. The room temperature yield strengths for both steels were less than those reported[6] for similar steels produced by conventional processing. After two cycles, the elongation values of steels A and B were considerably less than those reported for conventionally processed steels. These results were attributed to a number of factors associated with the peculiar behavior of Fe-Cr-Ni-C alloys. These factors which adversly affect the mechanical properties, are discussed below.

As discussed earlier, Hyatt and Krauss pointed out that three processes could operate during the reversion of martensite to austenite. Normally in Fe-Ni-C alloys, the diffusionless reversion results in an austenitic structure with increased dislocation density and enhanced strength. This was also observed by Koppenaal in a Fe-Ni-Mo-C steel.[12] However, in TRIP steels with chromium, carbide precipitation in both austenite and martensite accompanied reversion, and this caused a depletion of austenite stabilizing elements from the matrix. Subsequent reversion of martensite resulted in austenite of low stability which, during subsequent mechanical testing, transformed to martensite at a low stress and caused a low yield strength, a characteristic feature associated with the "stress induced" transformation of austenite to martensite.[16]

The large changes in microstructure during thermomechanical processing were attributed both to diffusionless and diffusion based reactions in austenite. In process TM-2, applied to both steels, the structure changed from a non-uniform mixture of austenite and martensite after the first cycle to a fine and uniform structure after the fourth cycle. These changes are illustrated for steel A in Fig. 8; and for steel B in Fig. 9. During rolling, martensite formed in regions where the local stresses exceeded that required to induce its formation and therefore martensite formation was not uniform. Thus, the microstructure after rolling consisted of isolated martensitic regions in cold worked austenite. On heating to the reversion temperature (765°C for steel A and 780°C for steel B) martensite was tempered. Due to the non-uniformity of cold work, the degree of tempering was expected to vary from one martensitic region to another. In austenite, the dislocation density produced by the cold work could conceivably increase the diffusivity of alloying elements and induce precipitation of alloy carbides.[2] Precipitation caused a decrease in austenite stability, and during subsequent cycles of process TM-2, martensite formation was easier.

At the end of the first cycle, three constituents were observed. One phase was retained austenite (etched light in Fig. 8a). The second phase was believed to be austenite formed by reversion of martensite, but could also be tempered martensite. Both Krauss[9] and Koppenaal[12] have commented on the difficulty in distinguishing between tempered martensite and reverted austenite by optical means. The third constituent of the microstructure was a fine structure appearing in areas that were martensitic before reversion. This constituent was probably a mixture

0 0 3 0 3 9 0 2 7 4

of reverted austenite, tempered (but unreverted) martensite, and very small platelets of martensite that formed in reverted austenite. The amount of this finely formed constituent increased in successive cycles until after the fourth cycle in steel A, it was the only constituent present (Figs. 8e and 10). Magnetic tests indicated that the microstructure at the end of four cycles consisted of approximately 40% martensite in steel A and 80% martensite in steel B (Table 2). Microhardness measurements made on steel A with the Vickers diamond indentor showed that retained austenite present after the first cycle was harder than annealed austenite. The other two structures were approximately equal in hardness and were both considerably harder than retained austenite. Also the hardness of these second and third constituents increased with increasing number of cycles.

In steel B the application of process TM-2 resulted in a microstructure with the three constituents as described above. However, the third constituent appeared to be finer than in steel A. After four cycles there were three phases present in a non-uniform mixed microstructure (Fig. 9c). Even after seven cycles, the original grain boundaries were still apparent (Fig. 9d). These differences in structure between the two steels A and B were probably due to compositional differences, but no experimental evidence is available at present.

Metallographic observations of specimens after the first few cycles of process TM-2 indicated that martensite reversion was more complete in steel B than in steel A, as can be observed from a comparison of Figs. 8c and 8d for steel A with Figs. 9a and 9b for steel B. The reason for this behavior was not known.

The observed increase in strength between the fourth and seventh cycles of process TM-2 for both steels was due mainly to the cold work applied to the specimens as part of the processing. There was little difference in the fracture appearance of tensile specimens of steel B after five cycles (Fig. 11a) and seven cycles (Fig. 11b). In both cases fracture surfaces exhibited features of ductile failure and also those due to cleavage. The ductile fracture features were those typically observed in strain induced martensite.[20]

Figures 12a and 12b show microstructures of steel A after four and five cycles respectively of process T-1. Initially, when the steel was cooled to -196°C, martensite formed at the grain boundaries. This martensite was tempered on heating. In subsequent cycles, martensite plates formed throughout the grains. An increase in yield strength was not observed until the end of the fifth cycle. The microstructure of a specimen at the end of the fifth cycle exhibited a large amount of martensite. The increase in strength was accompanied by a marked decrease in elongation (Fig. 6).

Examination of the fracture of surface of a specimen tested after one cycle of process T-1 indicated intergranular failure. This is shown in Fig. 13. Also seen in Fig. 13 are fracture features indicative of dimpled rupture typically associated with the failure of strain induced martensite.[20]

In several room temperature tension tests, duplicate specimens with the same processing treatment did not exhibit the same elongation and ultimate strength, but did exhibit the same yield strength. A typical example was the observed difference in elongation values (25% vs 54%)

0 0 0 0 3 9 0 2 7 5

for specimens with 2 cycles of process T-1. This difference resulted from a premature failure of the specimen with the lower elongation. The behavior was typical of specimens in which the martensite content reached large values during the test. Even the smallest flaw would cause failure in the martensitic regions.

The processes TM-1 and TM-2 involved the same processing steps in the first cycle. However, specimens were not cold rolled after the first cycle in the TM-1 process. Therefore, the increase in dislocation density in subsequent cycles could result only from the austenite to martensite transformation. As in the case of specimens treated by the TM-2 process, carbide precipitation occurred during reversion and the consequent reduction in the stability of austenite resulted in the formation of increased amounts of martensite on cooling.

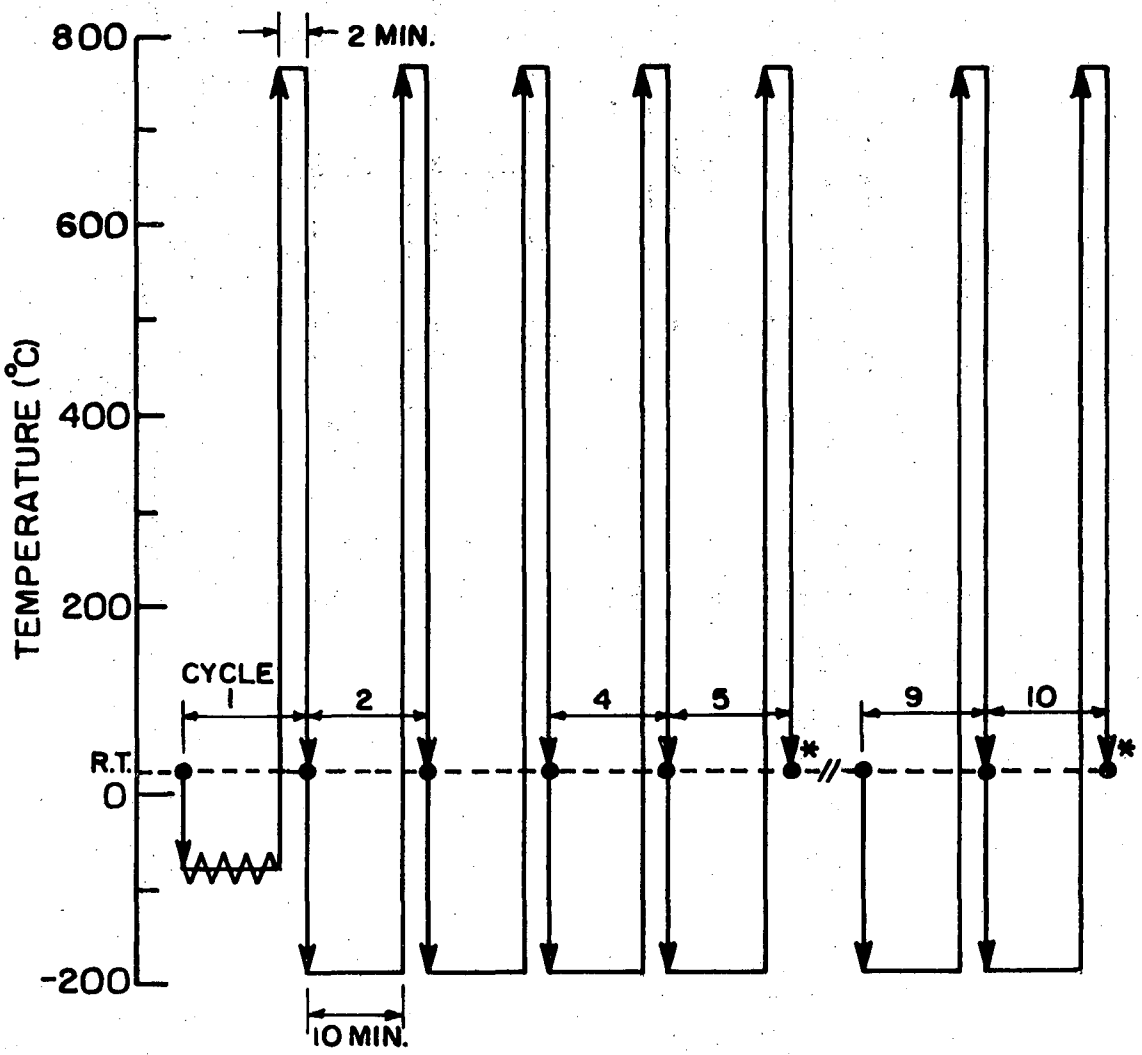
Microstructures of steels treated by the TM-1 process were similar to those that resulted from the TM-2 process (Figs. 14 and 15). However, the finely dispersed mixture of reverted and partially reverted martensite, tempered martensite, and martensite formed on cooling became the predominant micro constituents only after seven cycles. The mechanical properties of steels treated by the TM-1 process were not as good as those of steels treated by the TM-2 process. This difference was attributed to the presence of more austenite in the steels treated by the TM-1 process. This was confirmed by the magnetic test results presented in Table 2. The austenite was very unstable and yielded at a lower stress due to the stress induced formation of martensite.

The fracture surfaces of steel B specimens tested after five and ten cycles of process TM-1 are illustrated in Fig. 16. Features of

intergranular cracking and transgranular cleavage were clearly observed. There was also some indication of ductile failure characteristic of strain induced martensite.

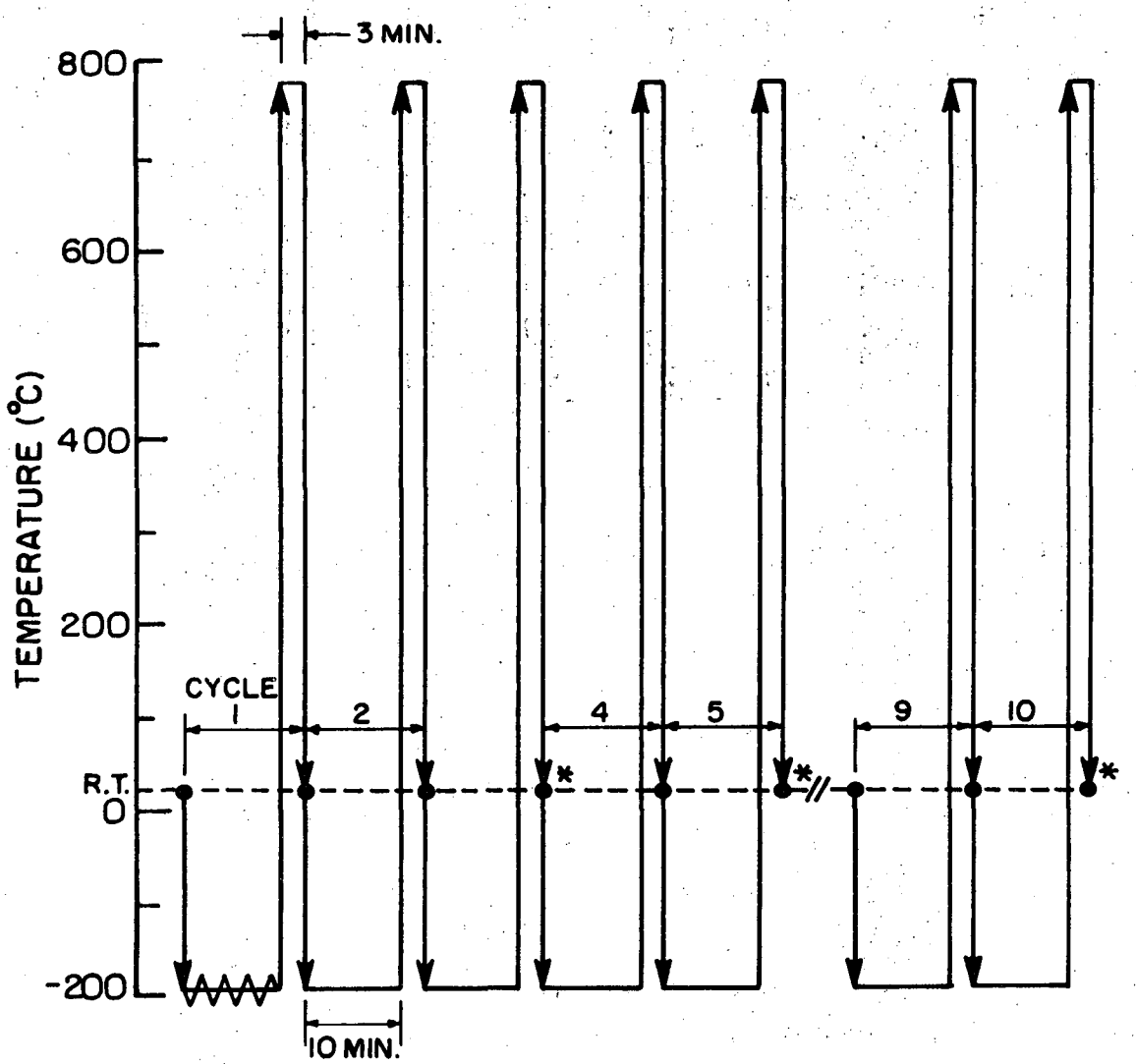
A low room temperature yield strength resulting from the stress induced formation of martensite was a common feature of both steels used in the current investigation. In addition, the room temperature ultimate strength was high and the elongation was low when the stress induced transformation occurred. Selected specimens were tested at 100°C where the austenite was more stable than at room temperature. Higher yield strength values than those observed at room temperature were observed in 100°C tests. Tests results for several specimens are given in Table 3.

0 0 0 0 5 9 0 2 7 8 2



XBL 733-5941

Fig. 2a



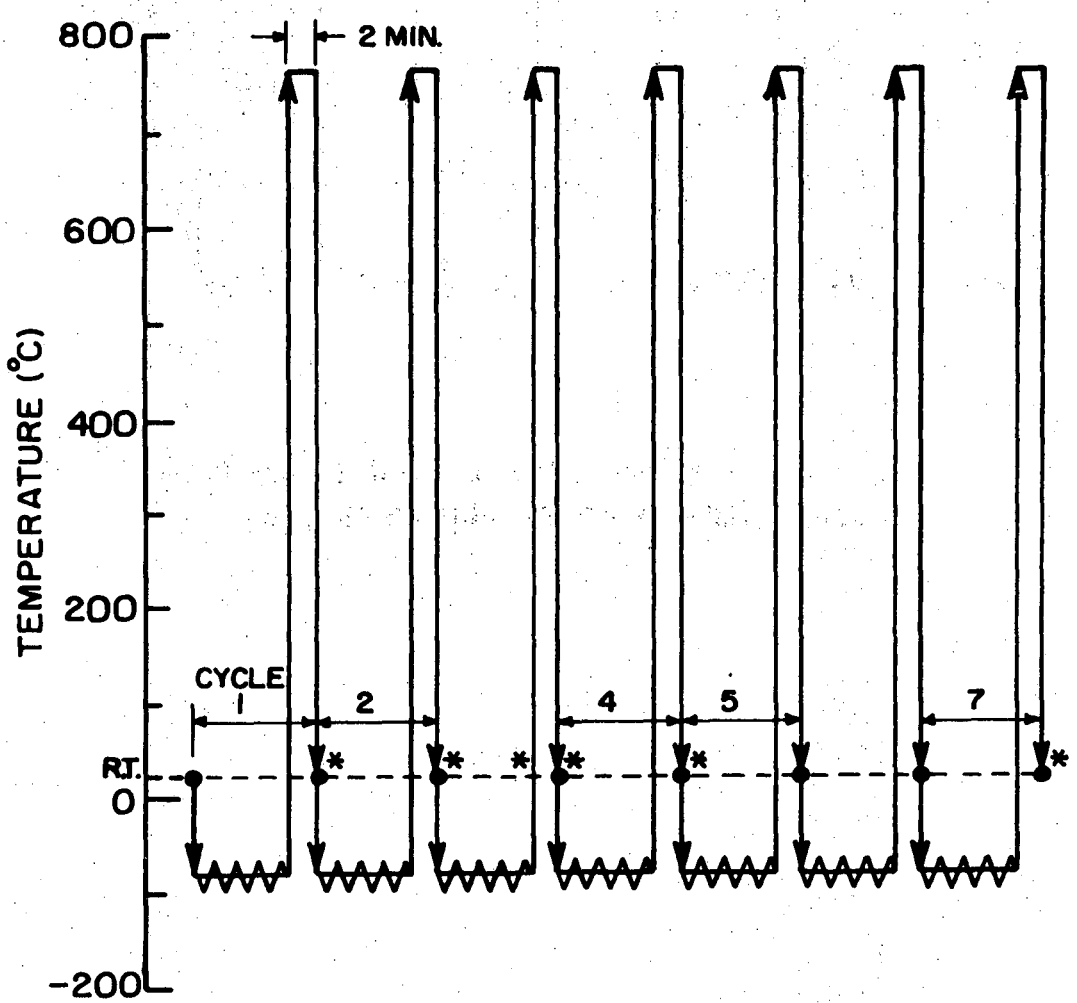
MECHANICAL DEFORMATION OF TENSILE SPECIMEN BLANKS

* MICROSTRUCTURE AT THESE POINTS SHOWN IN FIGURE 15

XBL 733-5940

Fig. 2b

0 0 0 0 3 9 0 2 7 8 3

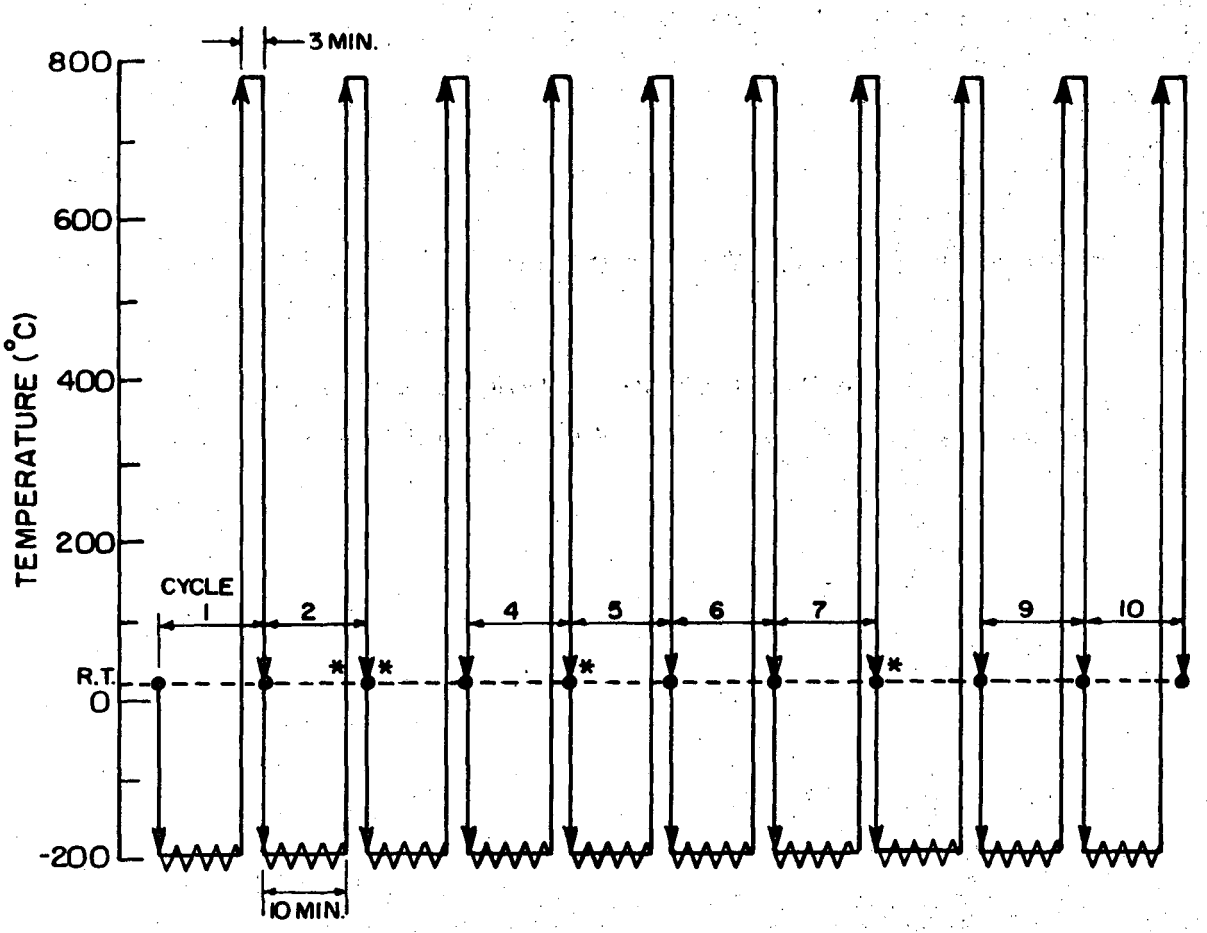


MECHANICAL DEFORMATION OF TENSILE SPECIMEN BLANKS

* MICROSTRUCTURE AT THESE POINTS SHOWN IN FIGURE 8

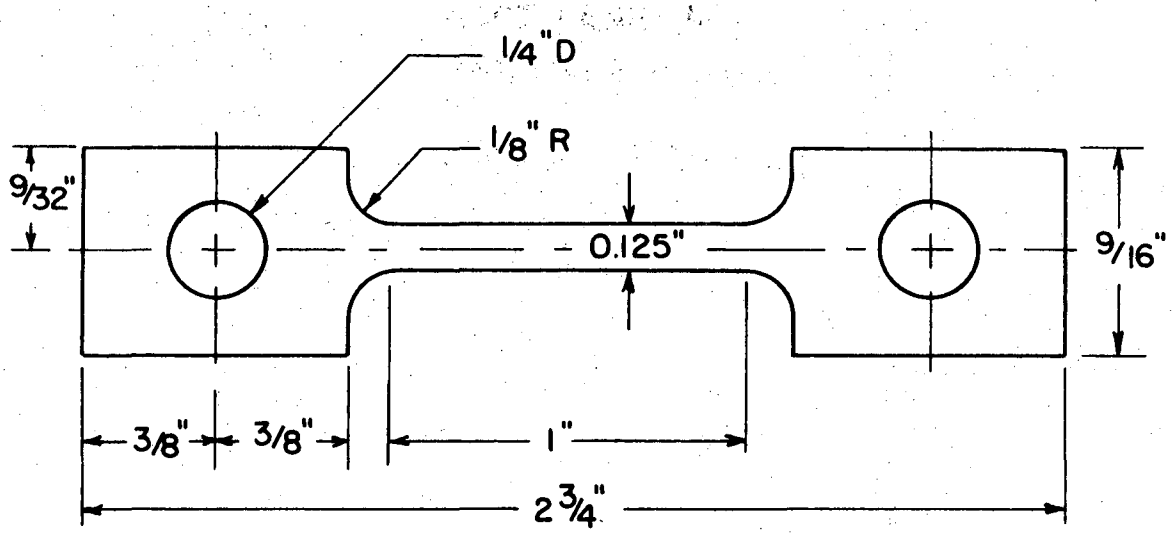
XBL 733-5942

Fig. 3a



XBL 733-5939

Fig. 3b



GAGE LENGTH: 1"
THICKNESS: 0.05"

TENSILE SPECIMEN

XBL 705-911

Fig. 4

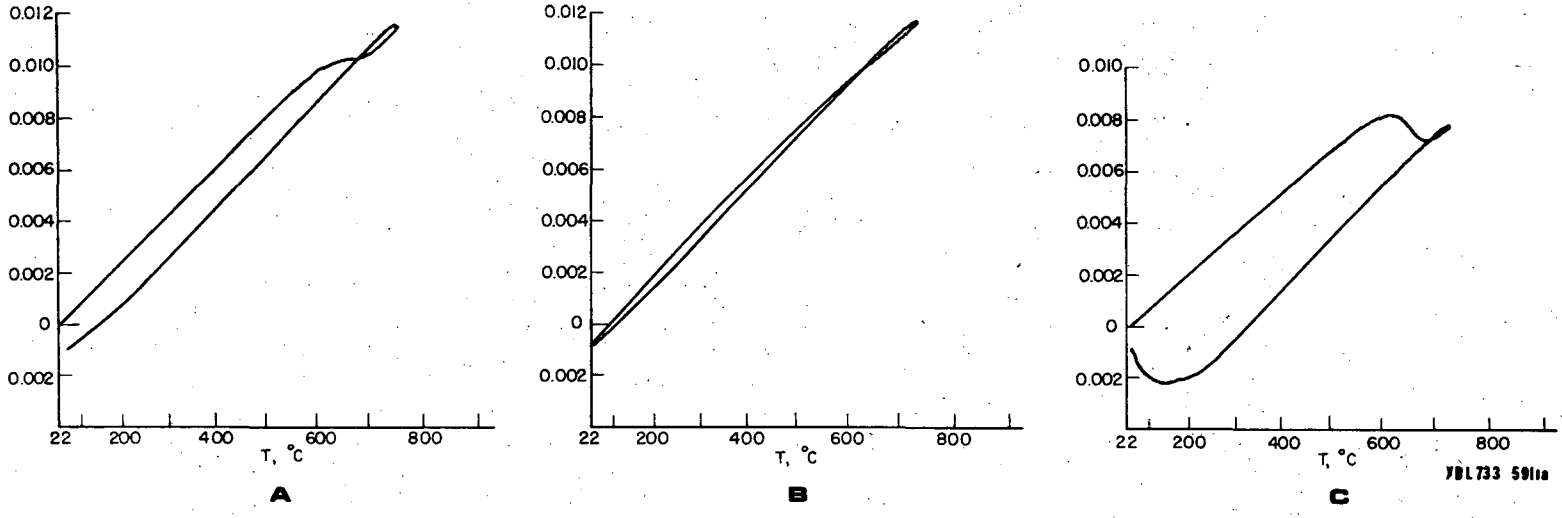
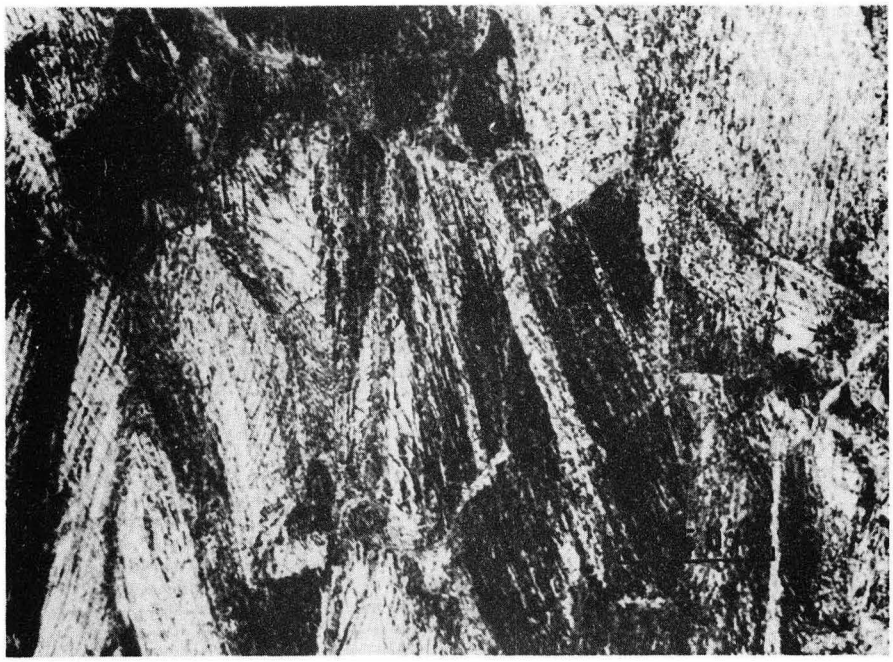
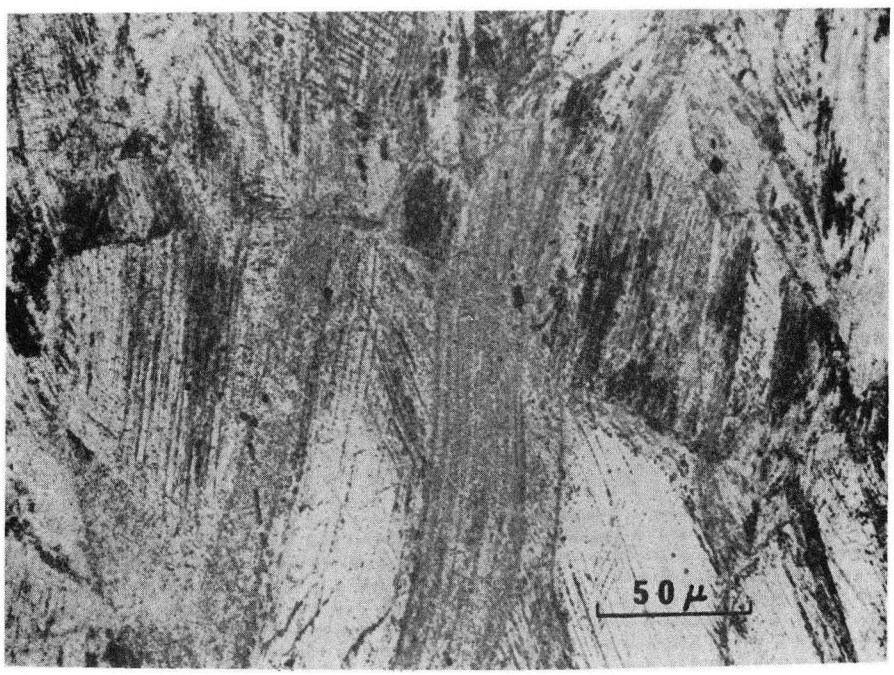

0.012
0.010
0.008
0.006
0.004
0.002
0
0.002
22
200
400
600
800
T, °C
A
B
C
YBL733 591a

Fig. 5

0 0 0 0 3 9 0 4 7 3 6



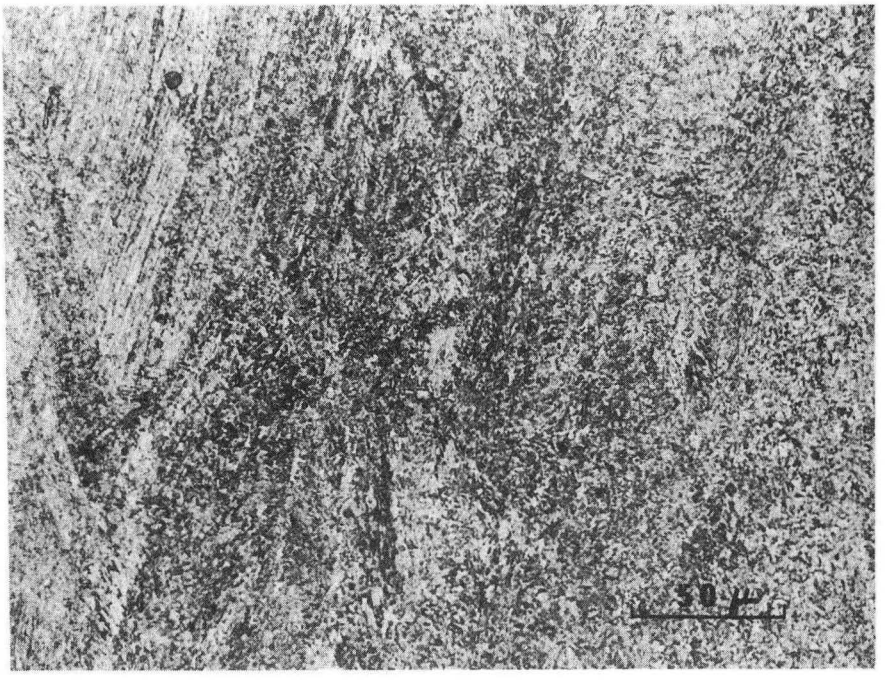
(c)



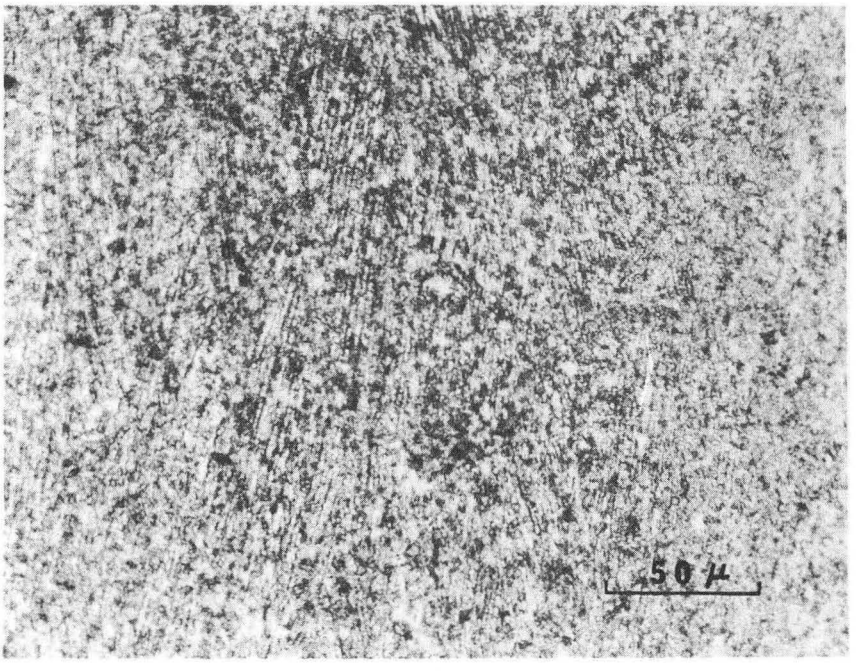
(d)

XBB 733-2174

Fig. 8 continued



(e)

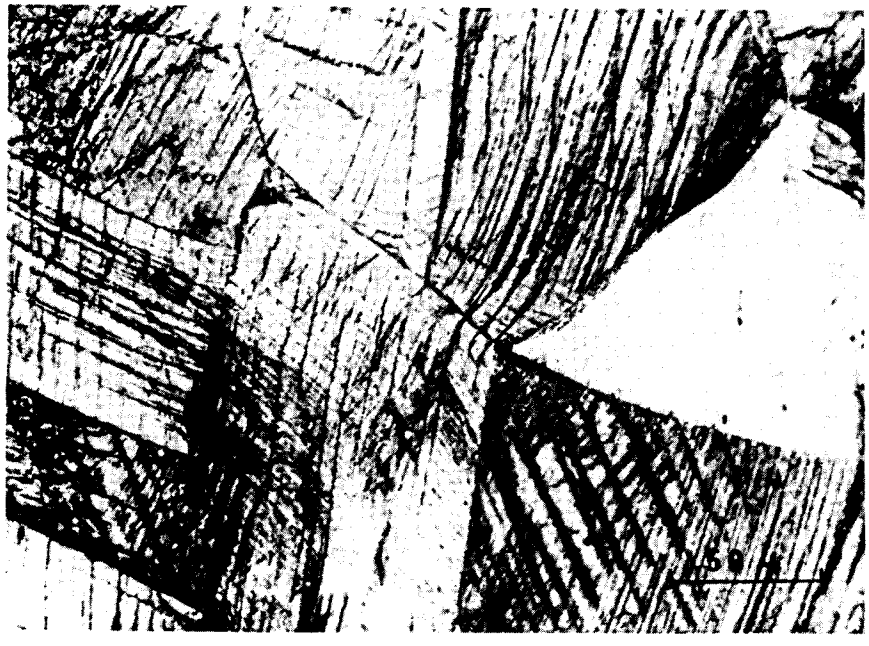


XBB 733-2184

(f)

Fig. 8 continued

(a)

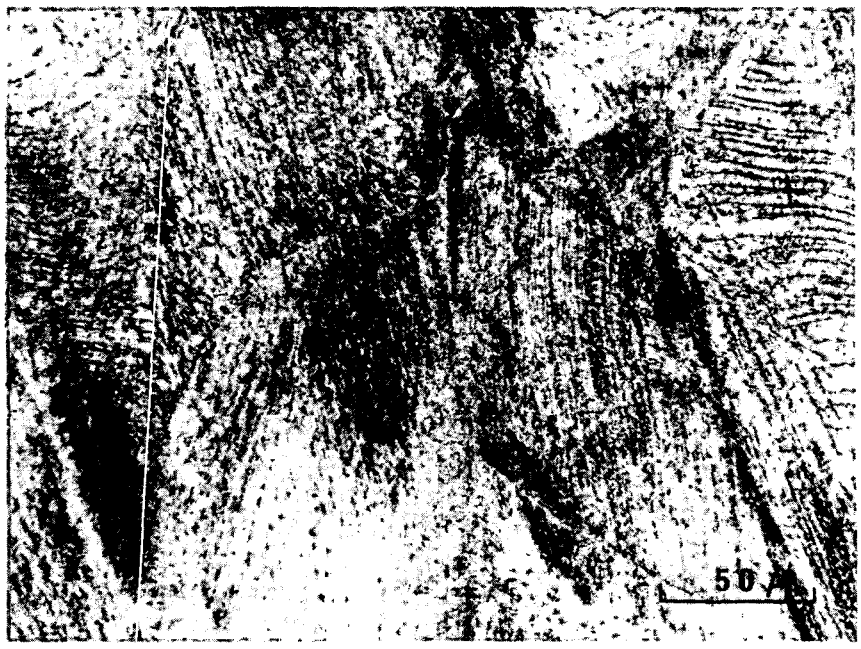


XBB 733-2183

(b)

Fig. 9

(c)



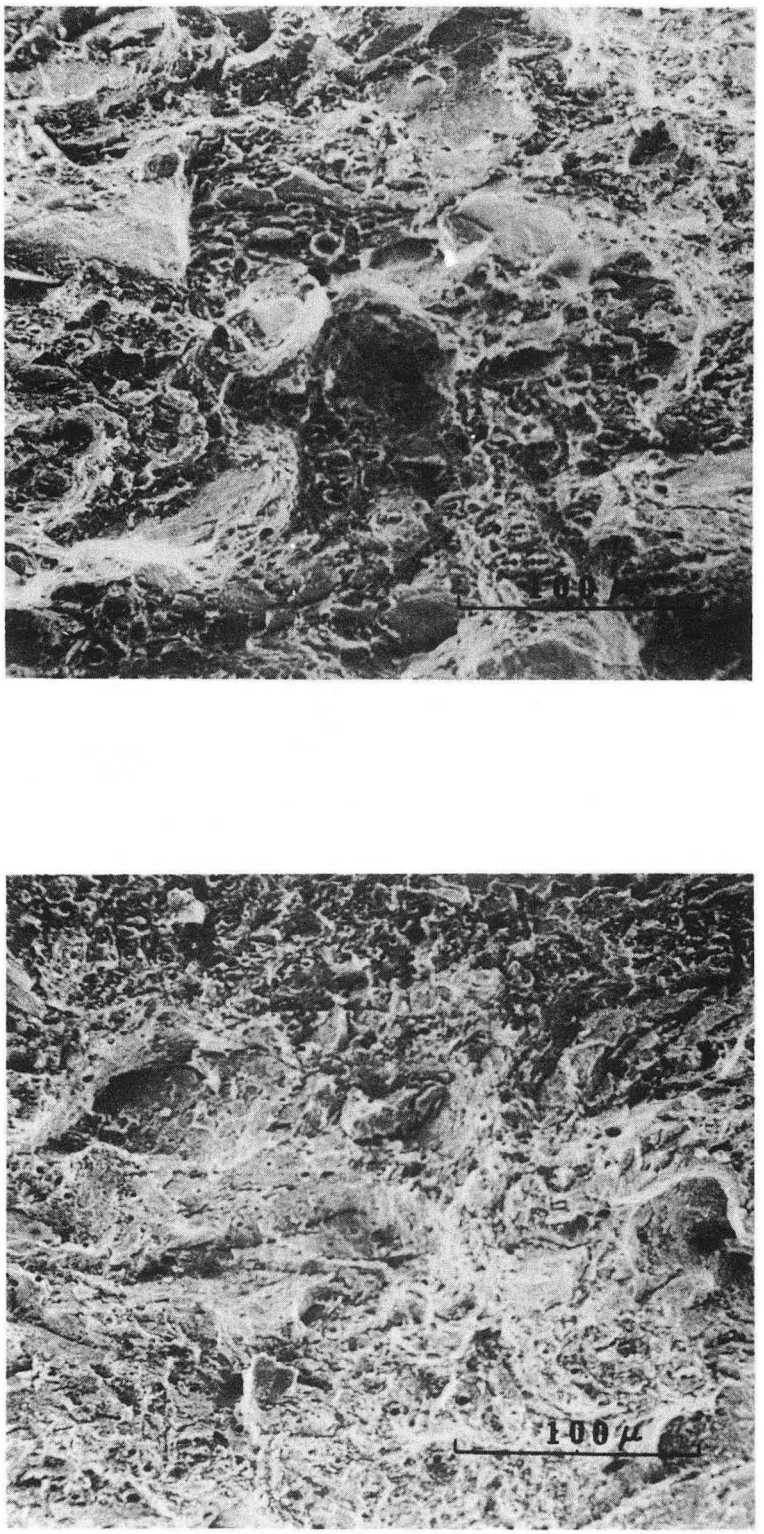
(d)

XBB 733-2182

Fig. 9 continued

XBB 733-2185

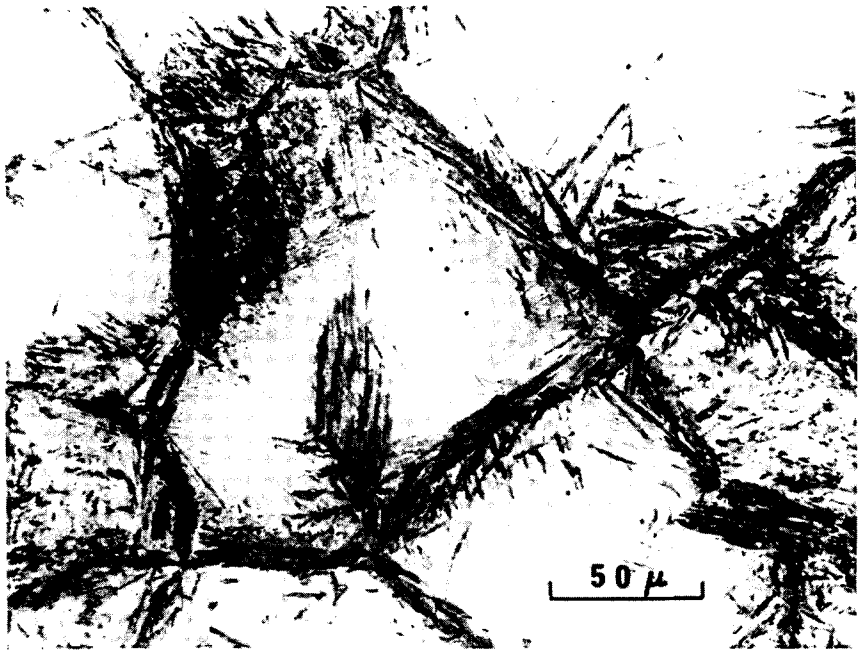
Fig. 10



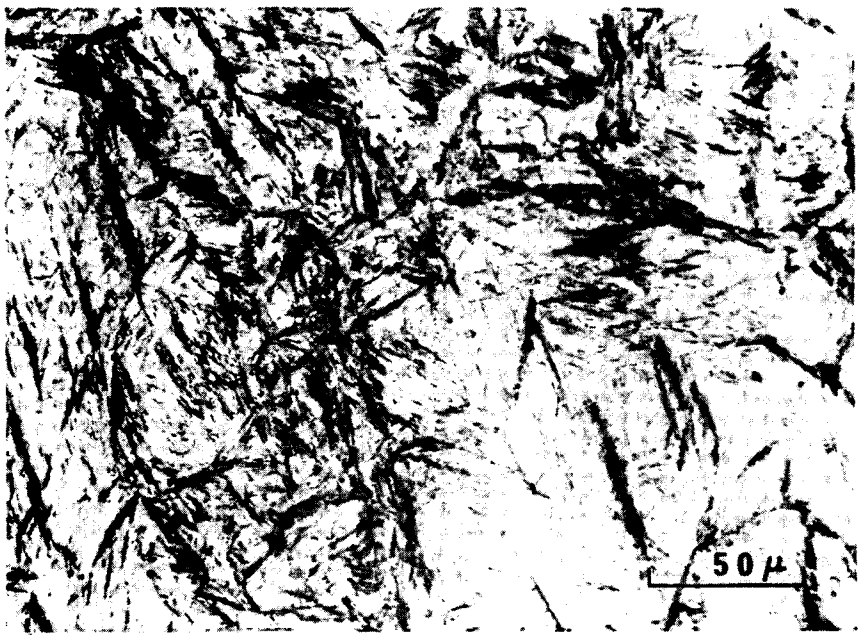
XBB 733-2177

(b)

Fig. 11



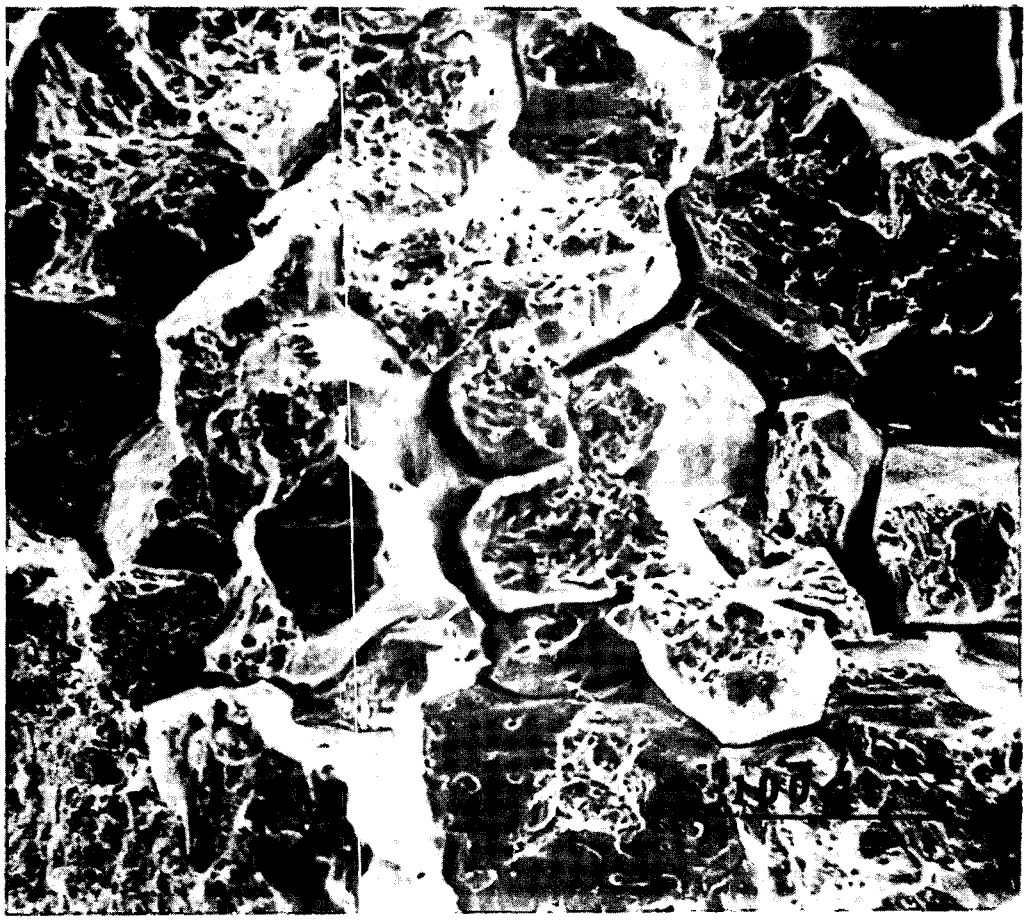
(a)



XBB 733-2178

(b)

Fig. 12



XBB 733-2173

Fig. 13

IV. SUMMARY

1. In chromium containing TRIP steels which were designed to have low austenite stabilities, small amounts of deformations by rolling at temperatures above the Ms temperature resulted in the formation of large amounts of martensite by a stress induced mode. There was a considerable increase in strength over that of annealed austenite.
2. The martensite that formed by small amounts of deformations at temperatures above the Ms reverted, at an elevated temperature, to austenite by a diffusionless transformation. However, diffusion based reactions involving martensite tempering and precipitation of carbides in austenite accompanied the reversion.
3. Three processing treatments were investigated for their influence on structure and mechanical properties. Repeated cycles of these treatments resulted in a fine and uniform structure which was believed to be a mixture of tempered martensite, austenite formed from the reversion of martensite, and martensite formed during cooling from the reversion temperature.
4. The present investigation showed the martensitic reversion could be used in the processing of chromium TRIP steels. However, diffusion based reactions accompanying the diffusionless reversion of martensite to austenite caused a decrease in austenite stability and resulted in a structure with a large fraction of martensite. Steels with this structure exhibited a high strength, but had a low ductility.

In tests at 100°C, higher yield strengths were observed. It was also shown that TRIP steels having a M_s below -196°C could be strengthened by small amounts of deformation to form large amounts of martensite followed by a subsequent reversion of the martensite.

0 0 0 0 3 9 0 2 7 7

V. RECOMMENDATIONS FOR FUTURE WORK

The thermal reversion of martensite to austenite can be used to strengthen chromium containing TRIP steels. However, a limitation in its use is the occurrance of diffusion based reactions that result in a loss of austenite stability. Future work should investigate the possibility of maintaining sufficient austenite stability, so the reduction in stability caused by the diffusion based reactions does not result in the formation of "stress induced" martensite during testing. To achieve this objective, it is necessary to modify the composition of the existing alloys.

Considerations that should be taken into account in the modification are:

1. The austenite stablizing elements that are added should not be strong carbide formers.

2. A reduction in carbon content while causing a loss in austenitic stability will decrease the tendency for carbide precipitation.

3. The maintenance of a high chromium content (>12 pct) is desired to have a high corrosion resistance.[13]

ACKNOWLEDGEMENTS

The author wishes to express his appreciation to Dr. Victor Zackay for helpful discussions and guidance. Appreciation is also extended to Dilip Bhandarkar for helpful discussions, encouragement and his critical review of this manuscript.

This research was performed under the auspices of the U. S. Atomic Energy Commission through the Inorganic Materials Research Division of the Lawrence Berkeley Laboratory.

REFERENCES

1. V. F. Zackay and E. R. Parker, High-Strength Materials, V. F. Zackay, ed., (John Wiley and Sons, Inc., New York, 1965) p. 130.

2. G. Thomas, D. Schmatz and W. Gerberich, ibid, p. 251.

3. V. F. Zackay, E. R. Parker, D. Fahr and R. Busch, ASM Trans, 60, 252 (1967).

4. G. R. Chanani, The Strength and Ductility of Metastable Austenitic Steels as a Function of Composition and Test Temperatures (M. S. Thesis) UCRL-17805, September 1967.

5. W. W. Gerberich, P. L. Hemmings, V. F. Zackay and E. R. Parker, Fracture 1969, Proceedings of the Second International Conference on Fracture, Brighton, 1969, p. 288.

6. D. Fahr, Enhancement of Ductility in High Strength Steels (Ph. D. Thesis) UCRL-19060, September 1969.

7. G. R. Chanani, S. D. Antolovich and W. W. Gerberich, Met. Trans 3, 2661 (1972).

8. G. Krauss, Jr. and M. Cohen, Trans. TMS-AIME 224, 1212 (1962).

9. G. Karuss, Jr., Acta Meta. 11, 499 (1963).

10. B. Hyatt and G. Krauss, Trans ASM 61, 168 (1968).

11. C. Apple and G. Krauss, Met. Trans. 2, 1785 (1971).

12. T. J. Koppenaal, Met. Trans. 3, 1549 (1972).

13. A. J. Baghdasarian, Corrosion Resistance of TRIP Steels (M. S. Thesis) LBL-806, May 1972.

14. T. Angel, J. Iron Steel Inst. 177, 165 (1954).

15. R. H. Richman and G. F. Bolling, Met. Trans. 2, 2451 (1971).

16. D. Bhandarkar, V. F. Zackay and E. R. Parker, Met. Trans. 3, 2619 (1972).

17. R. M. Bozorth, Ferromagnetism (D. Van Nostrano, Co, New York, 1951).

18. K. Hoselitz, Ferromagnetic Properties of Metals and Alloys (Clarendon Press, Oxford, 1952).

19. B. de Miramon, M. S. Thesis, University of California, Berkeley, September 1967.

20. W. W. Gerberich, P. L. Hemmings, V. F. Zackay and E. R. Parker, Fracture 1969, P. L. Pratt, ed. (Chapman and Hall Ltd, London, 1969) p. 288.

0 0 0 0 3 9 0 4 7 9

Table 1. Chemical composition of steels.

Steel	Fe	Cr	Ni	Mn	Mo	C
A	BAL	12.0	7.8	1.9	---	0.28
B	BAL	9.0	8.0	2.0	2.0	0.30

Table 2. Percent martensite in selected specimens before and after tension testing.

Steel	Process	Number Of Cycles	Percent Martensite	
			Before Testing	After Testing
A	T-1	4	18.6	59.3
A	TM-2	2	11.9	67.0
A	TM-2	4	38.9	78.7
B	TM-2	3	55.0	83.1
B	TM-2	4	80.9	89.0

Table 3. Mechanical properties of selected specimens tested at 22°C and 100°C.

Steel	Process	Number Of Cycles	Test Temp. °C	Yield Strength (psi)	Ultimate Tensile Strength (psi)	Elong. (%)
A	TM-2	3	22°C	122,000	219,000	11.9
			100°C	152,000	191,000	11.3
A	TM-2	6	22°C	132,500	208,000	8.6
			100°C	169,700	204,000	7.2
B	TM-1	3	22°C	61,300	154,000	16.8
			100°C	89,000	144,500	13.5

FIGURE CAPTIONS

Fig. 1. Process T-1 schematic for steel A.

Fig. 2. Process TM-1 schematic

a. For steel A

b. For steel B

Fig. 3. Process TM-2 schematic

a. For steel A

b. For steel B

Fig. 4. Flat tensile specimen.

Fig. 5. Dilatometric curves of steel A specimen:

a. After 6% deformation at -78°C.

b. After completion of heating cycle in (a),

c. After cooling to -196°C following the cycle in (b).

d. After cooling to -196°C following the cycle in (c).

e. During heating to 860°C after (d), held for 5 min, and cooled.

Fig. 6. Room temperature tensile mechanical properties of steel A at several stages of processes T-1, TM-1 and TM-2.

Fig. 7. Room temperature tensile mechanical properties of steel B at several stages of processes T-1, TM-1 and TM-2.

Fig. 8. Microstructures of steel A in various stages of process TM-2:

a. After completion of first cycle.

b. After completion of second cycle.

c. After deformation in the third cycle, before heating to the reversion temperature

d. After completion of third cycle.

e. After completion of fourth cycle.

f. After completion of seventh cycle.

Fig. 9. Microstructure of steel B in various stages of process TM-2.

a. After deformation in second cycle, before heating to the reversion temperature.

b. After completion of second cycle.

c. After completion of fourth cycle.

d. After completion of seventh cycle.

Fig. 10. Microstructure of steel A after completion of four cycles of process TM-2.

Fig. 11. Scanning electron fractographs of steel B tested at 22°C

a. After completion of five cycles of process TM-2.

b. After completion of seven cycles of process TM-2.

Fig. 12. Microstructure of steel A.

a. After four cycles of process T-1.

b. After five cycles of process T-1.

Fig. 13. Scanning electron fractograph of steel A tested at 22°C after the first cycle of process T-1.

Fig. 14. Microstructure of steel A.

a. After five cycles of process TM-1.

b. After seven cycles of process TM-1.

c. After ten cycles of process TM-1.

Fig. 15. Microstructure of steel B.

a. After three cycles of process TM-1.

b. After five cycles of process TM-1.

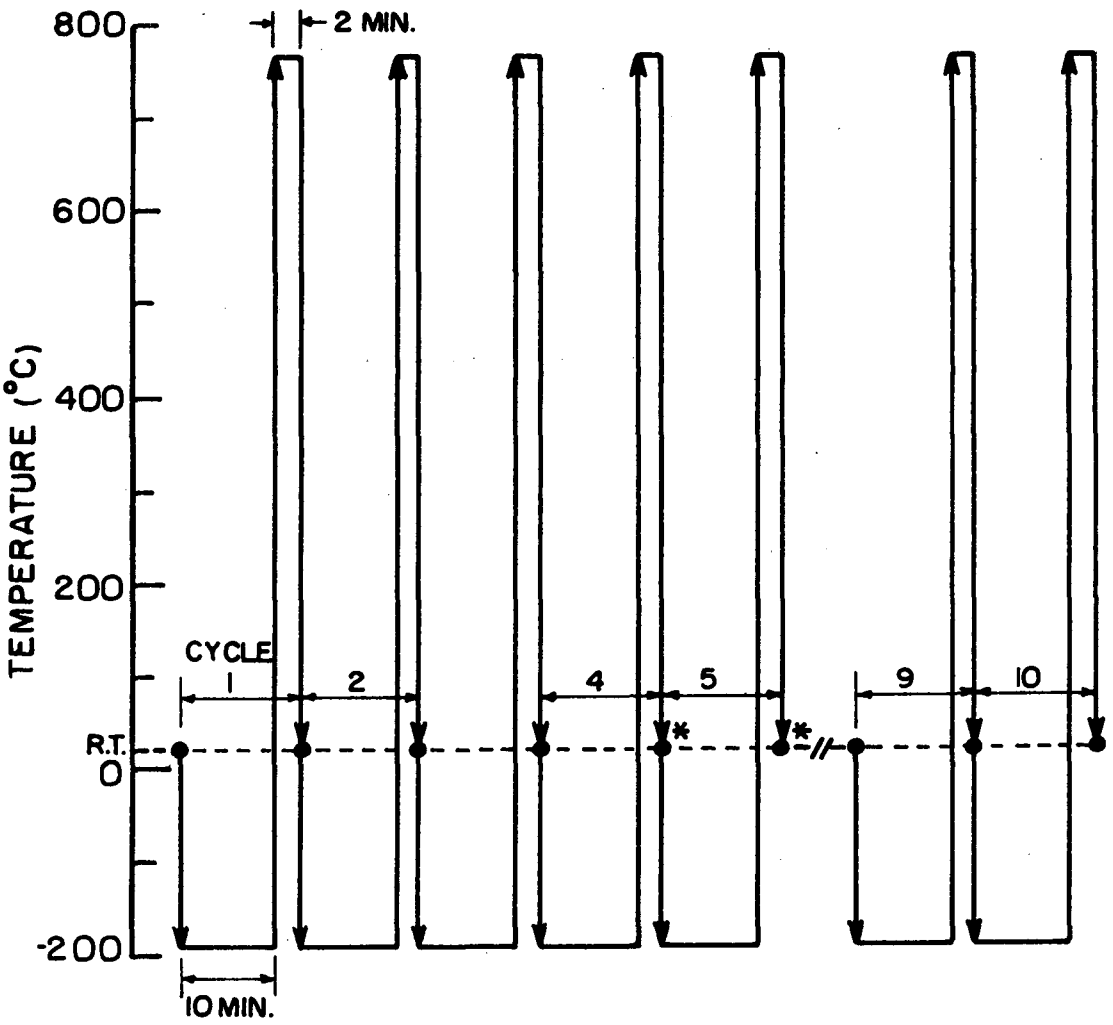
c. After seven cycles of process TM-1.

d. After ten cycles of process TM-1.

Fig. 16. Scanning electron fractographs of steel B tested at 22°C.

a. After five cycles of process TM-1.

b. After ten cycles of processing TM-1.



* MICROSTRUCTURE AT THESE POINTS SHOWN IN FIGURE 12

XBL 733-5943

Fig. 1

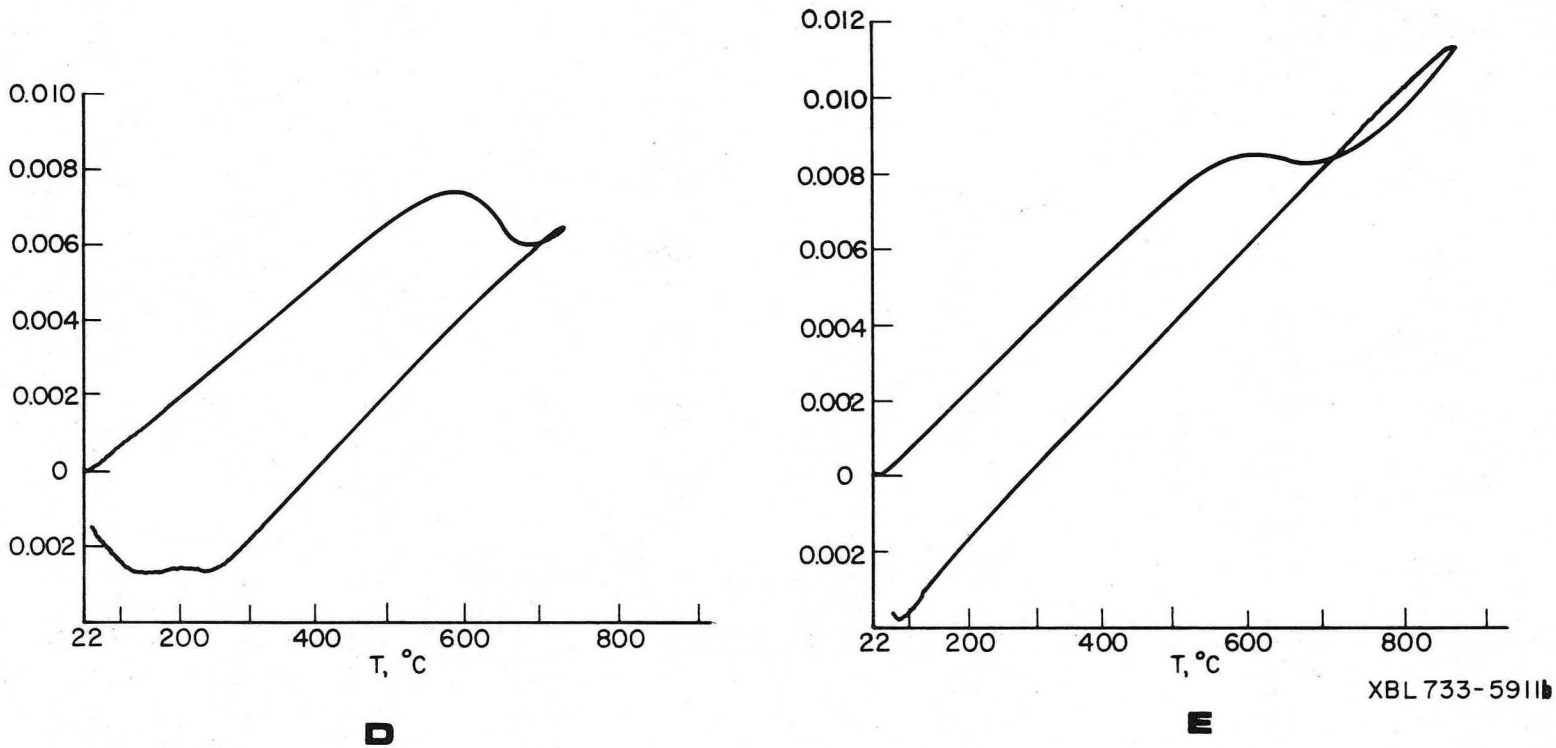


Fig. 5 continued

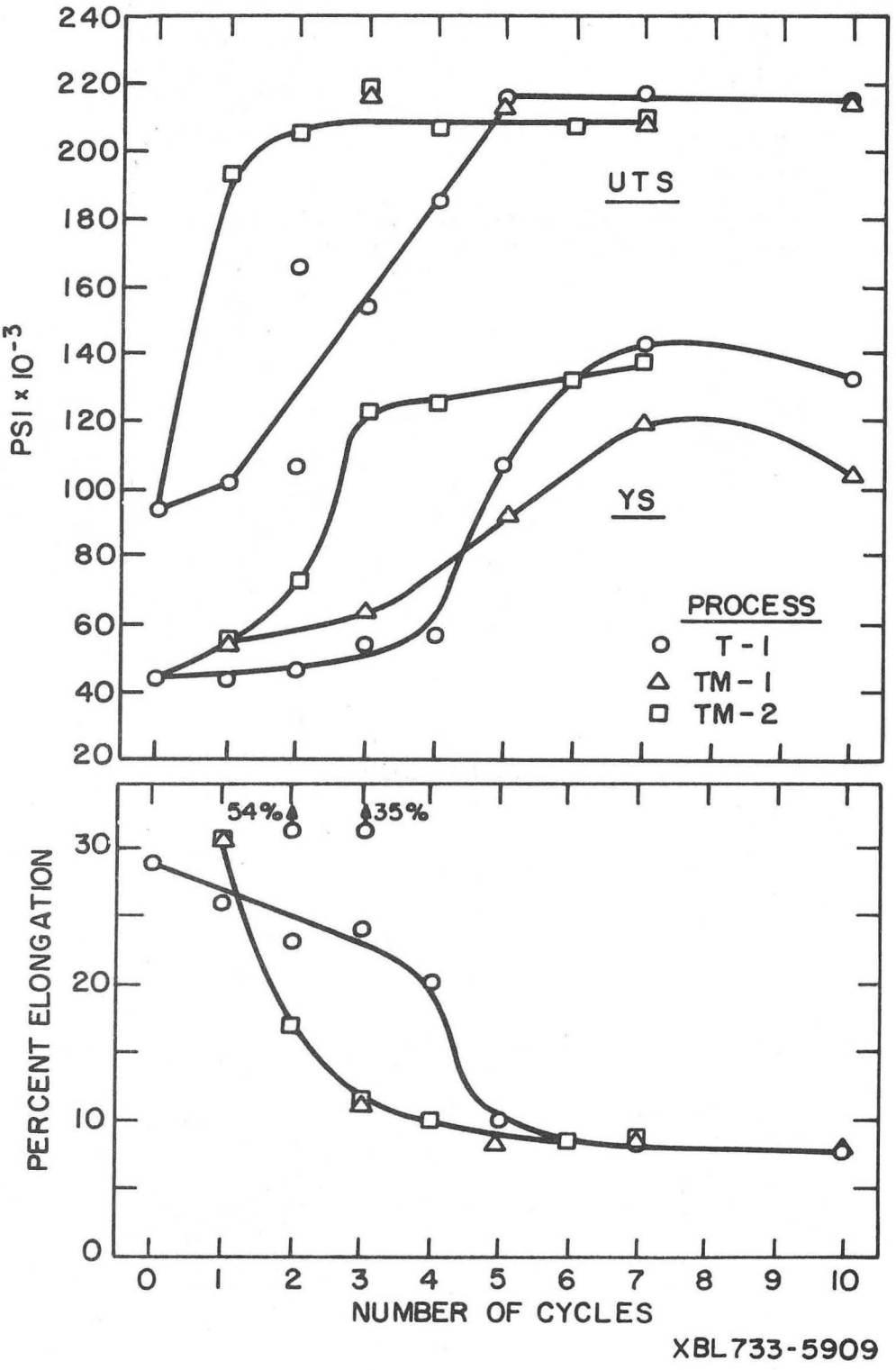

240
220
200
180
160
140
120
100
80
60
40
20
PSI x 10^-3
UTS
YS
PROCESS
T-1
TM-1
TM-2
54%
35%
30
20
10
0
PERCENT ELONGATION
0 1 2 3 4 5 6 7 8 9 10
NUMBER OF CYCLES
XBL733-5909

Fig. 6

0 0 0 0 3 9 0 2 7 8 5

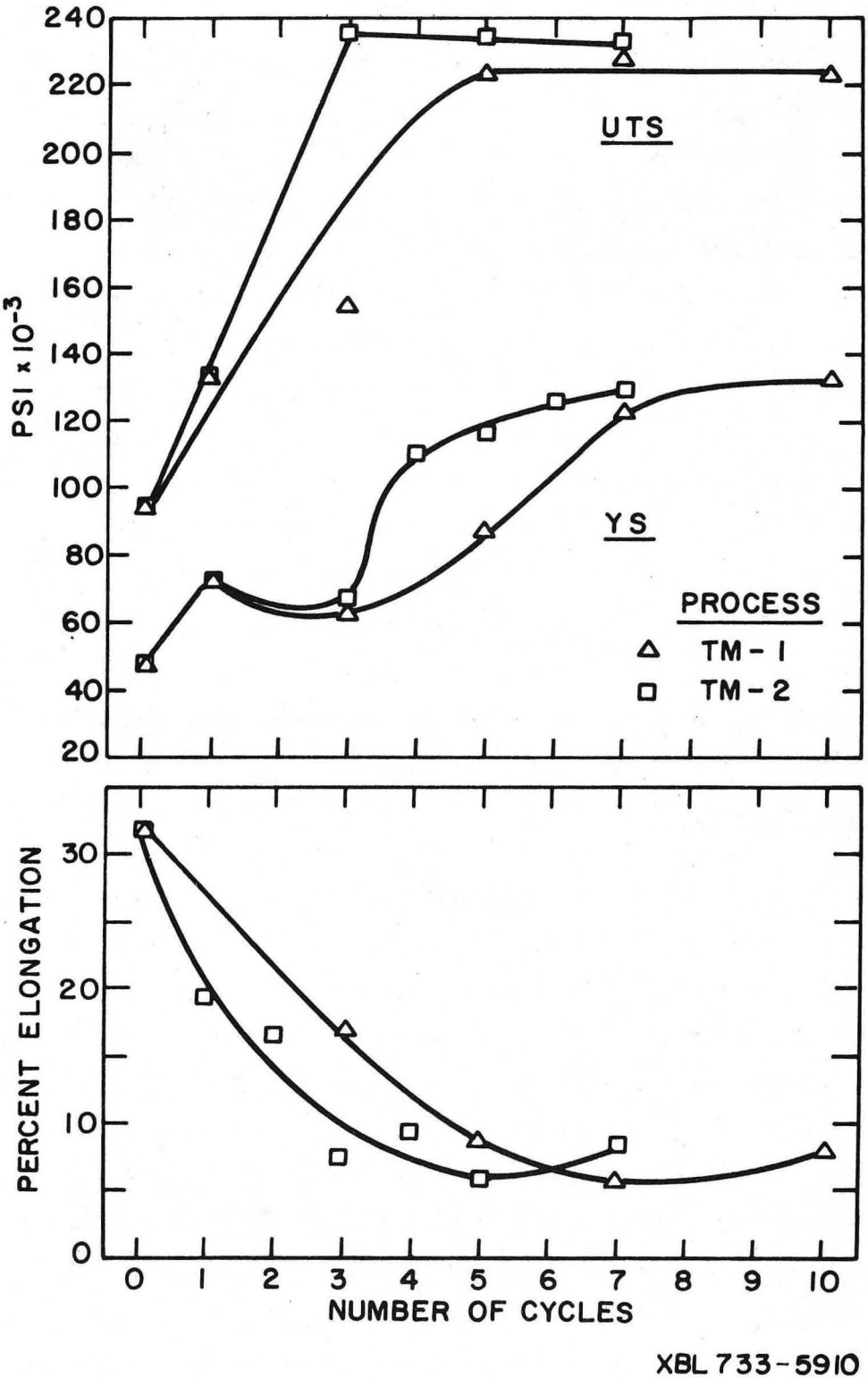
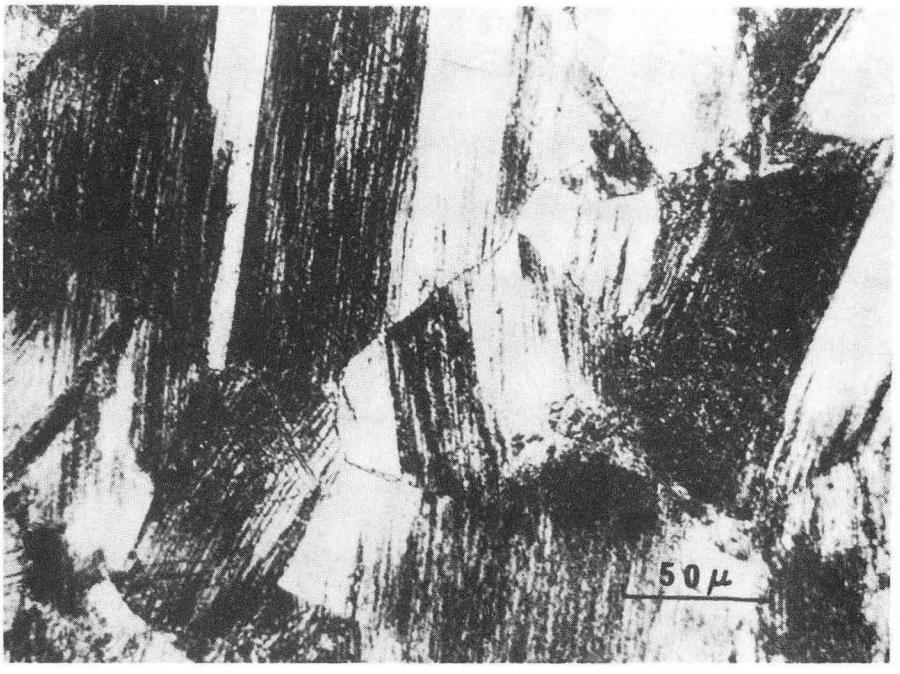
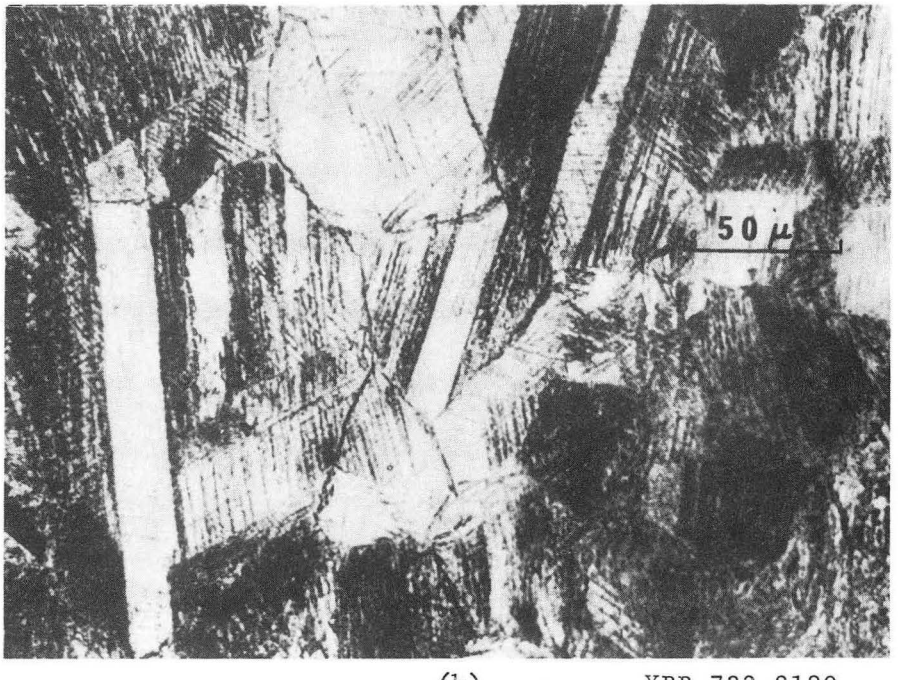


Fig. 7



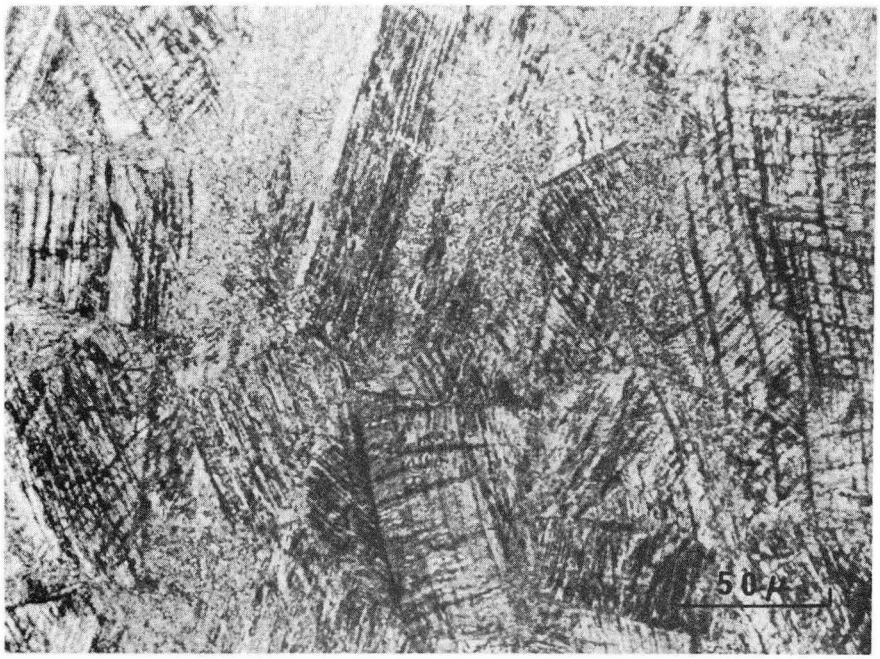
(a)



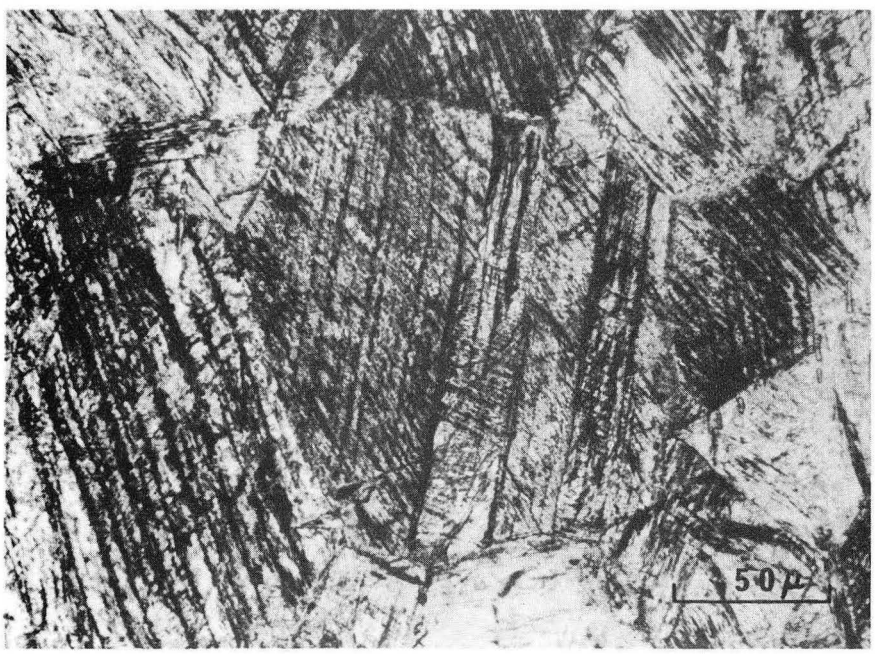
(b) XBB 733-2180

Fig. 8

0 0 0 0 3 9 0 4 7 8 9



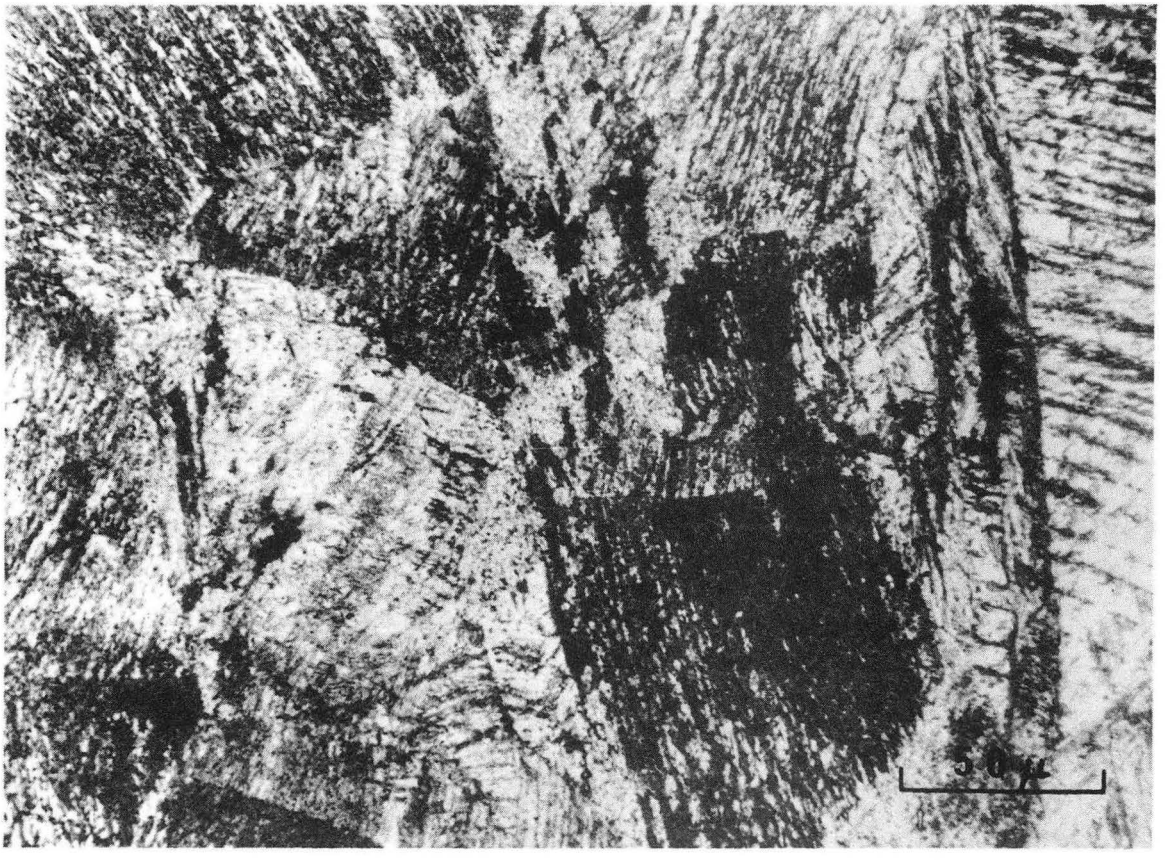
(a)



XBB 733-2179

(b)

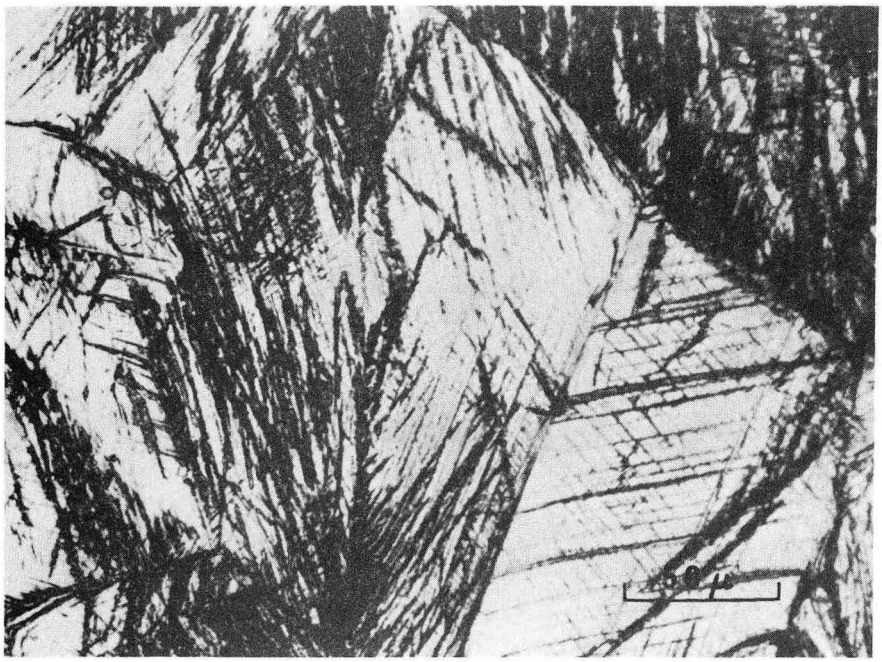
Fig. 14



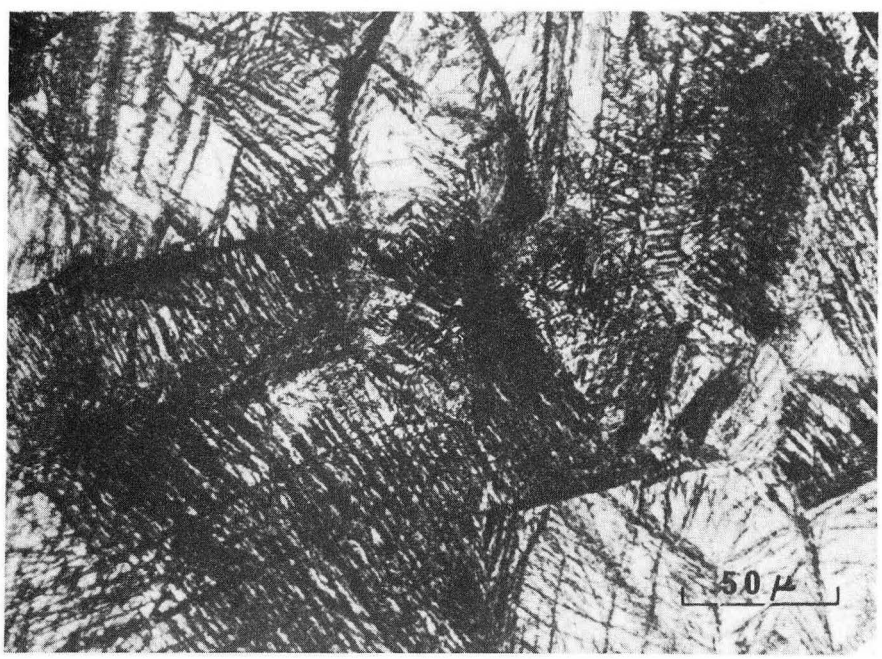
XBB 733-2186

Fig. 14c

0 0 0 0 3 9 0 2 / 9 0



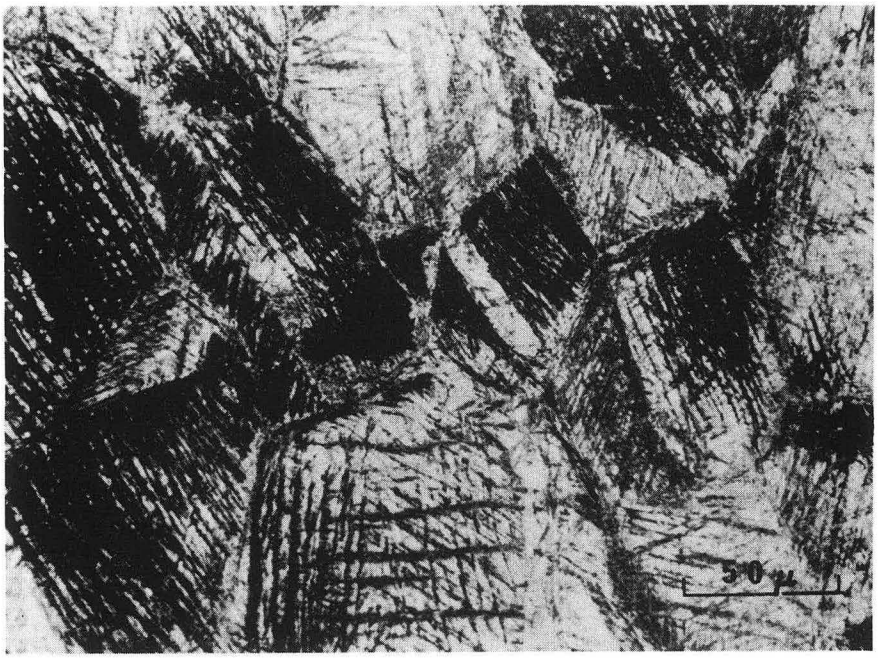
(a)



XBB 733-2181

(b)

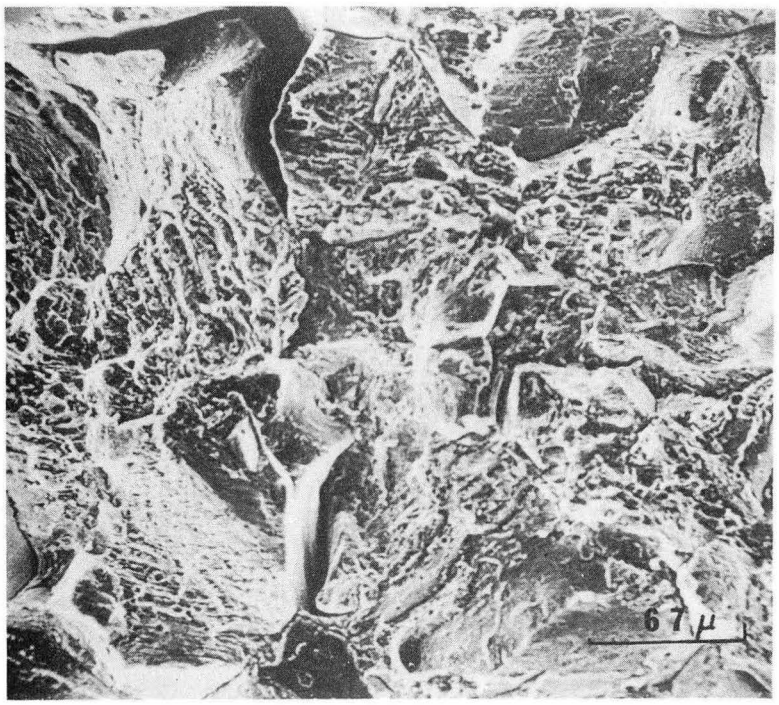
Fig. 15



(c)

(d) XBB 733-2176

Fig. 15 continued



(a)

(b)

XBB 733-2175

Fig. 16

LEGAL NOTICE

This report was prepared as an account of work sponsored by the United States Government. Neither the United States nor the United States Atomic Energy Commission, nor any of their employees, nor any of their contractors, subcontractors, or their employees, makes any warranty, express or implied, or assumes any legal liability or responsibility for the accuracy, completeness or usefulness of any information, apparatus, product or process disclosed, or represents that its use would not infringe privately owned rights.

TECHNICAL INFORMATION DIVISION
LAWRENCE BERKELEY LABORATORY
UNIVERSITY OF CALIFORNIA
BERKELEY, CALIFORNIA 94720