Submitted to Metallurgical
Transactions

UCRL-19692
Preprint

c.2

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August 1970

AEC Contract No. W-7405-eng-48

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STRUCTURE AND MECHANICAL PROPERTIES OF Fe-Ni-Mn-C STEELS

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ABSTRACT

The structure and mechanical properties of tempered martensite and lower bainite were investigated in a series of high purity 0.25% carbon steels with varying amounts of Ni and Mn. The martensites in .25C-5Ni-Fe and .25C-3Mn-Fe alloys were mainly untwinned, while those in .25C-5Ni-7Mn-Fe and .25C-7Mn-Fe alloys were heavily twinned. Manganese appears to promote carbide precipitation along the lath boundaries in tempered martensite. At equivalent yield and ultimate tensile strength levels, the tempered martensite of lower manganese steels showed better impact toughness than the tempered martensite of higher manganese steels. The impact toughness (compared at similar strength levels) of untwinned tempered martensite of 0.25% carbon steel with Widmanstatten precipitation of carbide was higher than that of lower bainite, which showed unidirectional carbides.

The reasons for the difference in impact toughness between the alloys, and also between the structures are rationalized in terms of internal twinning, grain boundary precipitation and carbide morphology together with other microstructural features.

I. INTRODUCTION

In recent years it has been argued as to whether the bainitic or martensitic structure has better toughness when compared at the same strength level. Holloman et al.[1] have shown that an AIS13135 steel at a similar nominal tensile strength the toughness of bainite was lower than that of tempered martensite. In a 0.2%C, Mn, Ni, Cr, Mo steel, Herres and Lorig[2] reported that the notched-bar impact toughness of bainite was inferior to that of tempered martensite. Similar advantages of tempered martensite have been established by Griffiths et al.,[3] Sachs et al.[4] and also Bailey.[5] On the other hand, opposite results have been shown by others.[6-12] These latter workers found the toughness of bainitic structures to be superior to the conventional quenched and tempered ones. It seems that in certain cases tempered martensite may possess better toughness than bainite of similar strength levels and in some cases the reverse may be true. It has long been suspected that this controversy is due to the difference in microstructure.

It has been suggested[13,14] that the presence of internal twinning in martensite may reduce its ductility. In a recent investigation[15] it was found that at equivalent strength levels the toughness of lower bainite was higher than that of the heavily twinned tempered martensite and it was suggested that untwinned martensites possess better toughness than that of bainites. Unfortunately fully bainitic structures could not be obtained in the low carbon alloys[15] so that dislocated martensite and bainite could not be compared. Therefore, the present authors have carried out a similar investigation to determine whether dislocated martensite (i.e. without internal twinning), is tougher than bainite or not. Furthermore, compositional factors affecting the substructure of martensite (i.e. dislocated vs. twinned)[16,17] were also investigated.

It is known that carbon lowers the M_s temperature drastically and promotes twinning. Most alloying elements in steel also lower the M_s temperature. Manganese has the strongest effect followed by nickel and chromium, while cobalt raises M_s, although cobalt does not seem to reduce twinning.[15] Therefore, by adjusting the composition, it is possible to vary the M_s temperature, the yield strength, and the extent of internal twinning. Except for a discussion by Boniszewski[18] who stated in Fe/0.2C/(2-7Mn) steels that increased manganese increased the amount of internally twinned martensite, little is known generally about the effect of manganese on martensite substructure. Thus this investigation on steels with varying manganese content will add to the existing data of the effect of carbon and nickel on the extent of internal twinning in martensite.

It has been reported that manganese additions improved the toughness of very low carbon steel (less than 0.037%C)[19] and also some commercial steels.[20-23] It is interesting to note the structure of all these steels consists of ferrite and pearlite and none of them contained more than 2 wt. pct of manganese. Some investigators have found that manganese was deleterious to the toughness of 18% Ni maraging steel.[24,25] However, the effect of manganese on the toughness of Fe-C martensitic steels is not understood. Thus, by using a series of high purity 0.25% carbon alloys with varying manganese contents, an attempt has been made in the present work to investigate whether manganese is injurious to the toughness of tempered martensitic steels or not, and if so, why.

II. EXPERIMENTAL PROCEDURE

The compositions of the alloys under investigation are given in Table I along with their M_s temperatures. The steels will be referred to by their

heat numbers as given in Table I.

The alloys were provided by Republic Steel Corporation and were obtained as 70 lb. air induction heats, subsequently rolled to sheet and plate. Specimens from these were austenitized at 1650°F for one and one-half hours in an argon atmosphere, and water quenched. Then they were immediately immersed in liquid nitrogen and refrigerated for 20 hours to ensure complete transformation to martensite. Tempering was done in a neutral salt bath. All the specimens were double tempered for 2 plus 2 hours (with intermediate quenching and refrigeration) at different temperatures. This completely eliminated the retained austenite. Bainite was obtained by quenching the specimens directly into the salt bath followed by isothermal transformation at the required temperature. Carbon analysis before and after heat treatment showed no significant decarburization. Nevertheless, after heat treatment, tensile specimens were carefully ground from both sides in order to eliminate any possible effects of surface decarburization from austenitizing. Flood cooling was employed to minimize specimen heating during grinding.

Tensile specimens of 0.060 in. thick were tested on an Instron machine at a cross head speed of 4×10^{-2} in/min. Standard size Charpy V-notch specimens were used for measuring the impact toughness. All tests were conducted at room temperature.

Thin foils for transmission electron microscopy were prepared from 0.060 in. thick heat treated specimens, first mechanical grinding, then chemical thinning and finally electropolishing in chromic-acetic electrolyte using the window technique. Foils were examined in a

Siemens Elmiskop IA microscope, operated at 100 kV. Two stage plastic/carbon replicas of the freshly broken fracture surface of Charpy specimens were prepared and examined in the electron microscope at 60 kV.

III. RESULTS

A. Mechanical Properties

Figures 1 and 2 show the variation of mechanical properties versus tempering temperature. Steels with 5% nickel, i.e. 804, 805 and 806 (Fig. 1) show similar tempering characteristics. The yield and ultimate tensile strengths drop and elongation and Charpy V-notch impact energy increase with tempering temperature, as expected. By comparing steel 804 (0% Mn), 805 (1.9% Mn), and 806 (3.8% Mn), it can be seen that the addition of first ~2% Mn is very effective in increasing tempering resistance. The Charpy V-notch toughness of 805 and 806 show an embrittlement when tempered around 600°F. This is the so-called 500°F embrittlement which generally occurs through the range of 400 ~ 700°F. It is interesting that this does not exist in steel 804 and the degree of embrittlement increases from steel 804 to steel 805, to steel 806, in that order. This increase in degree of 500°F embrittlement of tempered martensite with increasing manganese content is very similar to that observed by Payson[26] in carbon steels, where increasing carbon also enhances 500°F embrittlement.

Figure 2 shows the tempering behavior of steel 807, 808 and 809. Again the yield and ultimate tensile strengths drop and elongations generally increase with tempering temperature. Similarly to steel 806, steel 807 and 808 show embrittlement when the tempering temperature is 600°F. Steel 807 possesses high toughness when the strength is low. However,

the CVN impact energy for steel 808 and 809 are so low, that the toughness never exceeds 20 ft-lbs through the whole range of strength levels. The mechanical properties will be correlated with structure later.

In order to show the variation of toughness with manganese content at similar strength levels curves are drawn in Fig. 3 for steel 804 to 809. All the points are interpolated from the tempering curve. It can be seen that the toughness of 5Ni-.25C steels, i.e. 804, 805 and 806, increases slightly up to addition of ~2% manganese but drops at higher manganese contents. Also, the toughness of .25% carbon martensitic steels (807, 808 and 809) drops with increasing manganese from 3 to ~7%.

The Charpy V-notched toughness obtained at various yield strengths for different heats are shown in Fig. 4. The ultimate tensile strength-toughness varies in like manner. Again, steels 804, 805 and 807 show that toughness decreases with increasing strength, whilst steels with higher manganese, i.e. 806, 808 and 809 are very brittle and the toughness values are low even at low strength levels. At equivalent strengths the marked difference in toughness between lower and higher manganese steel indicates that increased manganese content above ~2% in 5Ni-.25C steel or ~3% in plain .25C steel is deleterious to the toughness of these tempered martensitic steels.

The mechanical properties of as-transformed bainite are relatively poorer than that of tempered martensite of the same steel. The yield strength drops and the toughness increases with increasing isothermal transformation temperature. To compare the toughness of martensite with bainite, it is important to ensure that a fully bainitic structure

is obtained. The attainment of fully bainitic structures in steels 804, 805 and 807 was verified by examining the microstructure of bainite after tempering at low temperatures. Since the carbide morphology in bainite and in tempered martensites is different, the presence of any martensite in as-transformed bainite could be detected by examining foils in the electron microscope. Further confirmation was obtained from the mechanical properties, which do not change much by tempering, except slight lowering of strength, whereas mixed structures of bainite and martensite would have shown different mechanical properties after tempering. It was not possible to obtain 100% bainite in steels 806, 808 and 809 so these alloys are excluded from this discussion.

The comparison of toughness between lower bainite and tempered martensite of steels 804, 805 and 807 is shown in Fig. 5 by plotting CVN impact value against strength for the two structures. It can be seen that at similar strength levels, the toughness values of the bainite of these steels are inferior compared to those of tempered martensite. The strength levels of bainite are limited due to the restriction on the isothermal transformation, i.e. higher isothermal transformation temperature gives upper bainite which is not desired.

Summarizing the above results at similar yield and ultimate tensile strength levels, the tempered martensites in steels of lower manganese content (i.e. 804, 805 and 807) show better toughness when compared to steels of higher manganese content(i.e. 806, 808 and 809). The tempered martensite of steels 804, 805 and 807 has better toughness than the bainite of the same steels.

B. Structure of Martensite

Both electron and x-ray diffraction observations showed no evidence of epsilon martensite in any of the alloys investigated. All the martensitic structures described in this paper were identified to be body centered alpha martensite. The microstructure of untempered martensites in steels 804, 805 and 807 are typically dislocated lath type. Some microtwins similar to those reported recently[27,28] were found in laths of steel 805. Appreciable twinning (mainly plate martensite) was observed in steel 806 and 809 whilst 953 was fully twinned. Examples of the microstructures are shown in Figs. 6 and 7. This variation in structure correlates with % Mn - increasing manganese content increases the amount of twinning. This variation also correlates with Ms temperature which decreases with increasing % Mn. These results are contrary to what is stated by Christian,[29] who indicated that martensite in high-alloy steels containing manganese is not internally twinned.

C. Structure of Tempered Martensite

The microstructures of martensite after various tempering treatments have also been examined. The martensite of steel 804 tempered at 400°F showed a mixture of epsilon carbide and cementite. These can only be distinguished by habit plane analysis since single crystal diffraction patterns do not provide unique solutions. It is known that the epsilon carbide has $\{100\}_\alpha$ habit and grows in $\langle 100 \rangle_\alpha$ directions, whereas cementite forms with {110} habit (Fig. 8a,b). Upon increasing the tempering temperature to 600°F, the precipitates are all found to be cementite with {110} habit. Spherodized carbides of 200Å to 500Å diameter are found on tempering at 850°F (fig. 8c). Some dislocation polygonization (recovery) occurs at this temperature. Figure 8d shows the structure of martensite of the same steel tempered at 1000°F for 2 + 2 hours. The carbides agglomerate on the martensite lath boundaries.

It is also known that in the presence of twins the cementite precipitates preferentially on the twin boundaries with {112} habit.[30] Figure 9 shows the structure of martensite of steel 805 tempered at 400°F, which shows a packet of internally twinned martensite laths with carbides, precipitated on the twin-matrix interface. The selected area diffraction pattern shows the matrix to be in approximately $[\bar{1}\bar{3}1]$ and the twins to be approximately $[10\bar{1}]$. Twinning on $(\bar{1}21)$ plane will bring $[\bar{1}\bar{3}1]$ matrix into an orientation which is about 5 to 6 degrees from $[10\bar{1}]$. This is similar to the observation of Kelly and Nutting on twinned plates in high carbon steels.[31] Dark field analysis distinguishes between the twins and the carbides and it can be seen that

these carbides precipitate on the twin boundaries. Figures 10(a), (b) show 850°F and 1000°F tempered martensite in steel 807. Both show Widmanstatten precipitation of cementite, which becomes coarser at higher temperature. By comparing 8(c) wtih 10(a) it is seen that when tempered at 850°F the carbides in steel 804 are well spherodized while those in steel 807 are still platelike. The morphological difference between the carbides of steel 804 and 807 is even clearer at 1000°F. It is evident that steel 807 possesses stronger tempering resistance than steel 804. This agrees with the variation of strength with temperature as shown in Fig. 1 and Fig. 2. The compositions of steel 804 and 807 are .26C-5Ni-Fe and .25C-3Mn-Fe respectively. Thus, as an alloying element in steel, manganese provides greater tempering resistance than nickel.

Unlike the lower manganese steels (i.e. steels 804, 805 and 807), tempering of higher manganese steels (806, 808 and 809) produces long precipitates along the lath boundaries and the carbide distribution is rather non-uniform. Also, as described before, since manganese promotes internal twinning, the cementite morphology is now {112} in the twinned crystals rather than {110} as in laths. As a typical example for the higher manganese steels, Figure 11 shows the structure of steel 808 tempered at 400°F for 2+2 hours which contains continuous carbide precipitates along the twin boundaries. Widmanstatten {110} cementite in the twin free regions is also observed. Aside from the twin boundary precipitates, some areas of 400°F tempered martensite of steel 809 show continuous lath boundary precipitates (Fig. 12). On tempering at higher temperatures, the long precipitates along the boundaries became coarser

and broke up into segments.

Summarizing the results, in lower manganese steels, epsilon carbide precipitates below 400°F, and is subsequently replaced by cementite at higher tempering temperatures. The manganese containing steel has stronger tempering resistance than a nickel steel. In higher manganese steels, the cementite precipitates preferentially along the twin and lath boundaries suggesting that manganese may promote lath boundary precipitation.

D. Structure of Bainite

It has long been known that there are two mainly different bainitic morphologies, namely upper and lower bainite.[32]

Figures 13(a) and (b) show the typical structure of lower bainite of steel 804 transformed at 720°F and of steel 807 transformed at 710°F for 24 hours respectively. The carbides are typically precipitated at an inclined angle of 55 ~ 65° to the long direction of the bainite ferrite grain.[32,33] The bainitic ferrite laths are found in packets separated by small misorientations as checked by selected area diffraction analysis. The bainite laths are very similar to low carbon martensite laths except for the carbide morphology inside the ferrite laths. The carbides are identified to be cementite and the trace analysis indicates that the habit plane of the carbides is {112} similar to the observation made by Baker et al.[34] No internal twinning was observed in bainitic ferrite, and the reason why cementite precipitates on only {112} planes still needs further investigation. It is found that generally the strength decreases and the toughness increases with increasing transformation temperature. The changes in microstructure are similar to those on

tempering martensite, viz. the carbide particles become coarser and dislocation recovery occurs.

As mentioned before, fully bainitic structures were not obtained in steels 806, 808 and 809 by isothermal transformation at various temperatures for 24 hours.

E. Electron Fractographic Observations

The tempered martensites of lower manganese steels possess relatively high toughness and show dimple rupture, e.g. Figure 14 from steel 804 which is typical of the fractography at 400°F, 800°F and 1000°F.

The fracture surfaces of tempered martensite and bainite of similar strength levels were also examined. Figure 15(a) represents the fracture mode in martensite of steel 805, tempered at 800°F. The main feature is that the fracture surface is composed largely of well-defined dimples of various sizes. In contrast, the fracture mode in bainite of the same steel transformed at 610°F, shows a mixture of dimples and cleavage facets, which is a low energy fracture (Fig. 15b). This agrees with the toughness data which gave CVN values of 48 and 13.5 ft-lbs for the tempered martensitic and bainitic structures respectively. The mixed structure of cleavage facets and dimples in bainite indicates that when a cleavage crack is arrested, some deformation and ductile fracture occur until a new crack is initiated as pointed out earlier.[35] Fracture mode comparison of the tempered martensite with bainite of similar strength level was also made for steel 807. Figure 16(a) shows the fracture surface of 800°F tempered martensite of steel 807 where small equiaxed dimples are observed. The fracture surface of bainite of the same steel transformed at 710°F [Fig. 16(b)] indicates the herringbone structure, which is

typical of cleavage.[36] There are dark irregular bands at some boundaries where the carbon replicating film has collapsed, suggesting very large topographical differences or intergranular cracks at these boundaries as observed by Low.[37] The CVN toughness values of these tempered martensite and bainite are 43 and 26 ft-lbs respectively.

From the above comparison, there is good correlation between the fracture modes and the impact data. Fractographs of steels 806, 808, and 809 show predominantly intergranular fracture. Some small dimples were occasionally seen. A typical example is shown in Fig. 17, which is the fracture surface of steel 809 double tempered at 400°F for 2+2 hours. The observations also match the low impact toughness values of the higher manganese steels.

No microcracks in any of the high manganese steels were observed. Thus the low impact toughness of tempered martensite of high manganese steel could not be due to the presence of microcracks, as was found in high carbon steels.[38]

IV. DISCUSSION

A. Toughness of Tempered Martensite and Bainite

Based on similar strength levels, toughness was compared either between tempered martensite of varying manganese content or between different structures of the same composition such as tempered martensite and bainite.

1. Comparison of Toughness Between Bainite and Tempered Martensite of Lower Manganese Steels with a Little Twinning

The results have shown that tempered martensite of lower manganese steel possesses better toughness than bainite. Assuming other things being equal, the difference in toughness will be attributed to the grain size and orientation of ferrite, and carbide morphology.

Both the size and orientation of the grains are important in controlling the toughness. It is known[37] that the smaller the grain size the greater is the energy absorbed during fracture and the more difficult is the process of crack propagation. In the present case, various sizes of laths were found in both bainite and martensite, but in many cases it seems that in the same alloy at similar strengths the bainite laths are narrower than the martensite laths. However, the contribution to toughness by narrow laths of bainite is superseded by other factors, as discussed below.

As mentioned before, the fracture mode of the bainite of the steels under investigation is predominantly cleavage. It has been shown from the microstructural observations that the bainite consists mainly of laths, which are separated from each other by low angle boundaries. It is thought that the cleavage cracks can propagate easily in these laths. If the bainite consisted of randomly oriented grains the change in orientation at individual grain boundaries should impede the propagation of cleavage cracks by causing localized deformation near the boundary. However, in the structure of parallel laths, less energy is required for the crack to propagate, because of the small orientation difference across small angle boundaries. Thus, the cleavage crack can cross from one ferrite

grain to another without much deviation in direction until the crack meets large angle boundaries, e.g. at prior austenite grain boundaries. This could be one of the factors responsible for the low toughness of bainite.

Also, the morphology of the carbide precipitation affects the toughness. In the twin free region of martensite, carbides precipitate in {110} Widmanstatten patterns, while in the bainitic structure, relatively few, coarse, widely spaced carbides precipitate unidirectionally inside the ferrite grain. It can be seen that the chance of the crack being blocked by Widmanstatten precipitates of carbide is greater than that of the bainitic carbides. This may be another factor which is responsible for the difference in toughness between the two structures.

The above argument can be demonstrated by comparing Fig. 13(b) with Fig. 10(a). These show the structure of bainite and tempered martensite of steel 807 at equivalent yield strength level of about 130 ksi. The bainitic structure [Fig. 13(b)] shows not only a high dislocation density but also a few unidirectional carbides, while the tempered martensitic structures [Fig. 10(a)] of the same strength level as bainite show numerous uniformly distributed carbides on {110} planes. Similar comparison can also be made for bainite and tempered martensite for steel 805.

2. <u>Comparison of Toughness Between Tempered Martensite of High and Low Manganese Steels</u>

The mechanical test results have shown that the toughness of martensite of lower manganese steel is higher than that of higher manganese steel. Again assuming other things being equal, the difference

in toughness may partially be attributed to microstructural differences such as internal twinning and partially to grain boundary precipitation.

The role of internal twinning in lowering the toughness of tempered steels can be explained as follows. The micrographs have shown that cementite precipitates preferentially all along the twin boundaries. Thus the carbides are very long and continuous as shown in Fig. 11. These long precipitates of carbides could act as a strong barrier for the movement of dislocations and so induce high stress concentrations, which lead to crack nucleation. Evidence of cracks induced by twins in metals and alloys have been observed before.[39] Also, in the presence of internal twinning, the martensite is more likely to be deformed by mechanical twinning rather than slip as reported by Krauss et al.[40] and Bevis et al.[41] It is known that the plastic deformation is restricted when the deformation mode is mechanical twinning which subsequently results in low toughness. Therefore, the presence of internal twinning and its effect on the carbide precipitation in the martensite plate together with deformation by mechaical twinning can give rise to the low energy fracture.

The presence of precipitation or segregation at the grain boundary is another major reason for lowering the toughness and causing intergranular failure. Grain boundary precipitates help in nucleation of cracks by acting as a strong barrier to dislocation motion and to the transfer of plastic strain energy from grain to grain. The grain boundary carbides seem to be preferentially formed as the manganese content increased.

B. Twinning

The martensite of steel can be described by its transformation substructure, i.e. twinned martensite and dislocated martensite. The M_s-M_f temperature range has been generally agreed to be one of the main parameters in affecting twinning by most researchers,[14-17] and has been the subject of a recent review.[42]

It is observed in the present work that the martensites of .25C-3Mn-Fe and .25C-5Ni-Fe alloys are mainly untwinned, but the martensite of .25C-7Mn-Fe and .25C-5Ni-7Mn-Fe alloys are almost fully twinned, showing that internal twinning in martensite is promoted by manganese. Thus the extent of internal twinning in martensite depends partially on M_s temperature is further confirmed in pure Fe-Ni-Mn-C and Fe-Mn-C alloys as it is in Fe-C[31] and Fe-Ni-C[16] but not in cobalt bearing alloys.[15]

Fundamentally, the deformation mode of bcc materials, i.e. whether they deform by twinning or slip depends on the critical resolved shear stress (CRSS) for slip or twinning at the transformation temperature. At the transformation temperature if the CRSS for twinning is lower than that for slip then the structure should be twinned.

It is known that deformation mode and CRSS for slip in body centered materials is strongly temperature dependent. Thus at low temperatures, the CRSS for slip may be so high that twinning is the only mode of deformation. Generally the lower the transformation temperature, the greater is the tendency for martensite to twin. Also, martensite start (M_s) to martensite finish (M_f) covers a range of about 200°C.[43] Thus in many cases mixed structures of slipped and twinned martensites are observed. In addition, however, CRSS for slip (and twinning) is composition dependent. Therefore, the strength of the martensite at the temperature of the transformation is the most important factor in determining whether the martensite will twin or slip. This explains the effect of cobalt which, although increasing M_s does not appear to decrease twinning.[15]

V. CONCLUSIONS

1. The substructure of martensite in a 0.25% carbon steel depends on manganese content, increasing Mn increases the amount of twinning.
2. At similar yield and ultimate tensile strength levels, the tempered martensite of low manganese steels (which is mainly untwinned) shows better toughness than the tempered martensite of higher manganese steel (which is twinned).
3. It appears that manganese promotes carbide precipitation along the lath boundaries of tempered martensite. This coupled with the effect of manganese on twinning may be responsible for the low toughness of high manganese steels.
4. At equivalent strength levels, the toughness of untwinned tempered martensite of 0.25% carbon steel with {110} Widmanstatten precipitation

of carbides is higher than that of lower bainite, which shows unidirectional carbides on {112}.

5. In the lower bainite, the strength decreases and toughness increases with increasing isothermal transformation temperature. This follows the usual strength-toughness relationship.

6. No transformation twins are observed in lower bainitic ferrite.

ACKNOWLEDGEMENTS

The authors wish to thank the Republic Steel Corporation for supplying the alloys and to the staff of its research center for helpful discussions. This work was done under the auspices of the United States Atomic Energy Commission through the Lawrence Radiation Laboratory in Berkeley.

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TABLE I

Chemical Composition of the Alloys* (Given in wt. %)
With Their Ms Temperature

Heat No.	C	Ni	Mn	P	S	Ms Temp. °F
804	.26	5.0	---	.004	< .005	603
805	.26	4.9	1.9	.004	< .005	520
806	.26	4.85	3.8	.005	< .005	385
807	.25	---	3.0	.004	< .005	601
808	.25	---	4.9	.007	< .005	454
809	.24	---	6.85	.005	< .005	367
953**	.26	5.1	7.5	ND	ND	148

* [illegible]
** Steel 953 contains 0.02% Si and 0.02% Al.

ND: Not Determined

Figure Captions

Fig. 1. The effect of tempering temperature on strength and toughness of steel 804, (no manganese) 805 (1.9% Mn) and 806 (3.8% Mn).

Fig. 2. The effect of tempering temperature on strength and toughness of steel 807 (3% Mn) 808 (4.9% Mn) and 809 (6.85% Mn).

Fig. 3. Variation of Charpy V-notch impact energy with manganese content for steel 804 through 809 at similar yield strengths.

Fig. 4. Comparison of Charpy V-notch impact energy of various steels at different yield strength levels.

Fig. 5. Comparison of CVN toughness and yield strength between tempered martensites and bainites in the same steels.

Fig. 6. Mainly dislocated lath martensite of steel 805 which is typical of zero and lower manganese steels, 804 and 807. Some fine internal twins can be seen in some of the narrow laths (see also fig. 9).

Fig. 7. Twinned martensite of steel 809 which is typical of higher manganese bearing steels (806, 808, 809).

Fig. 8. Structure of martensite of steel 804 double tempered for 2 + 2 hours at (a) 400°F, showing fine precipitates of epsilon carbides, arrow marks indicate cementite of {110} habit has also begun to form; (b) another region with higher magnification, showing cementite only, all three families of {110} cementite are in contrast, foil orientation <111>. (c) 850°F, showing spherodized carbides of 200 ~ 500Å diameter in the matrix; dislocation recovery has started. (d) 1000°F, showing dislocation polygonization in recovered structure, and coarse carbides on the martensite lath boundaries.

Fig. 9. Structure of martensite of steel 805 tempered at 400°F for

2 + 2 hours. (a) shows a packet of internally twinned martensite laths with carbides precipitated on the twin-matrix interface, (b) dark field of $(0\bar{2}0)$ twin spot reverses the contrast of the twins, (c) dark field of carbide spot A reverses contrast at the twins, showing that cementite precipitates on the twin boundaries, (d) dark field of $(\bar{1}0\bar{1})$ matrix spot reverses contrast for the whole packet of martensite laths, (e) selected area diffraction pattern and indexing showing that the orientation of the matrix and twin are approximately $[\bar{1}3\bar{1}]$ and $[10\bar{1}]$, respectively. Twin plane is $(\bar{1}21)$.

Fig. 10. Martensite of steel 807 tempered for 2 + 2 hours at (a) 850°F, showing fine Fe_3C precipitates with {110} habit, (b) 1000°F, some of the cementite still retain mainly platelet shape, but some parts have begun to spherodize.

Fig. 11. Martensite of steel 808 tempered at 400°F for 2 + 2 hours, showing continuous carbide precipitates along the twin boundary, Widmanstatten {110} cementite plates are also observed in the twin free regions.

Fig. 12. 400°F tempered martensite of steel 809, showing continuous elongated carbide precipitation along the lath boundaries.

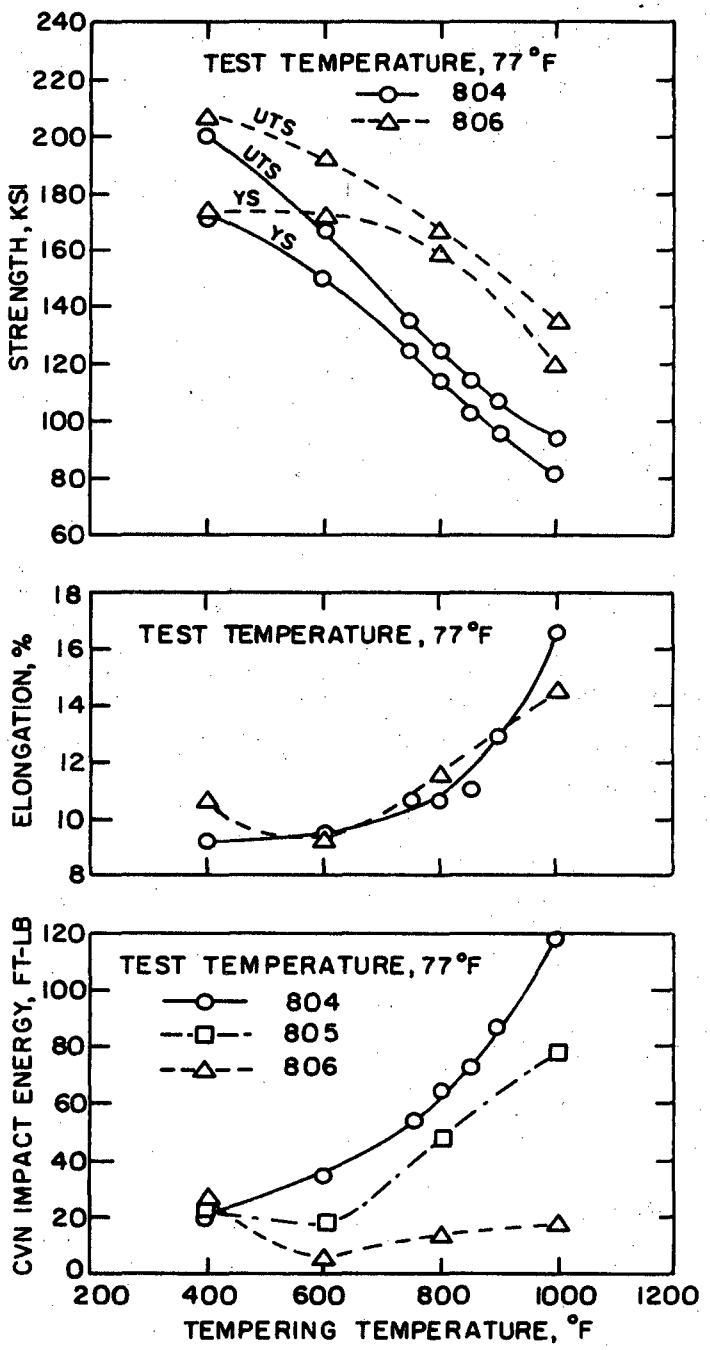
Fig. 13. Structure of isothermally transformed lower bainite in (a) steel 804 after 24 hours at 720°F, (b) steel 807 after 24 hours at 710°F, both show side by side nucleated ferrite laths and unidirectional carbides.

Fig. 14. Electron fractograph of Charpy V-notch specimen of steel 804 tempered for 2 + 2 hours at 1000°F, showing elongated, high energy dimples. [Compare to fig. 8(d).]

Fig. 15. Electron fractograph of tempered martensite and bainite of similar strength level of steel 805 (a) 800°F tempered martensite, showing various sizes of well-defined dimples, (b) lower bainite obtained by isothermal transformation at 610°F for 24 hours, showing mixture of dimples and cleavage facets.

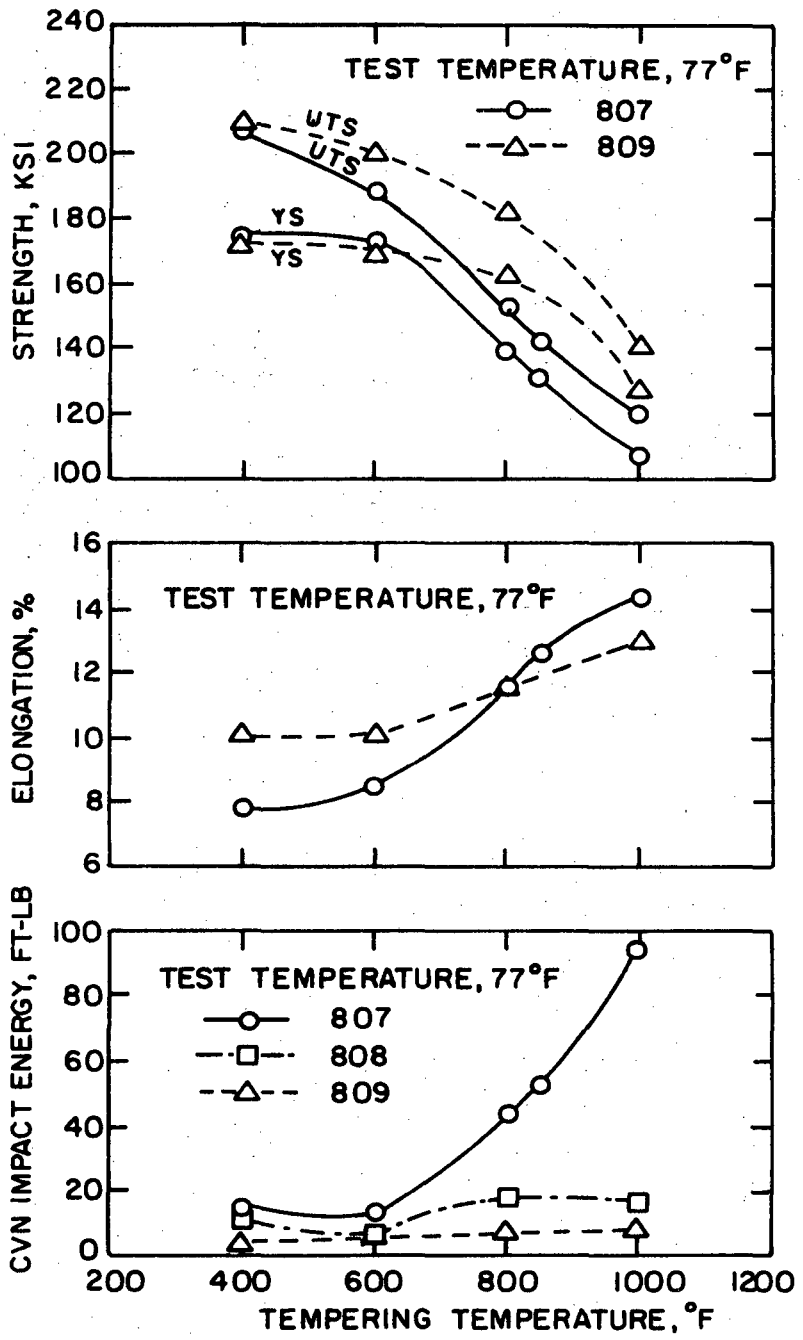
Fig. 16. Electron fractograph of tempered martensite and bainite of similar strength levels of steel 807. (a) 800°F tempered martensite, showing equiaxed dimples, [compare to fig. 10(a)]. (b) lower bainite formed at 710°F after transforming for 24 hours, showing herringbone structure, which is typical of cleavage. [compare to fig. 13(b)].

Fig. 17. Electron fractograph of steel 809 double tempered at 400°F for 2 + 2 hours, showing intergranular fracture, (compare to fig. 12).



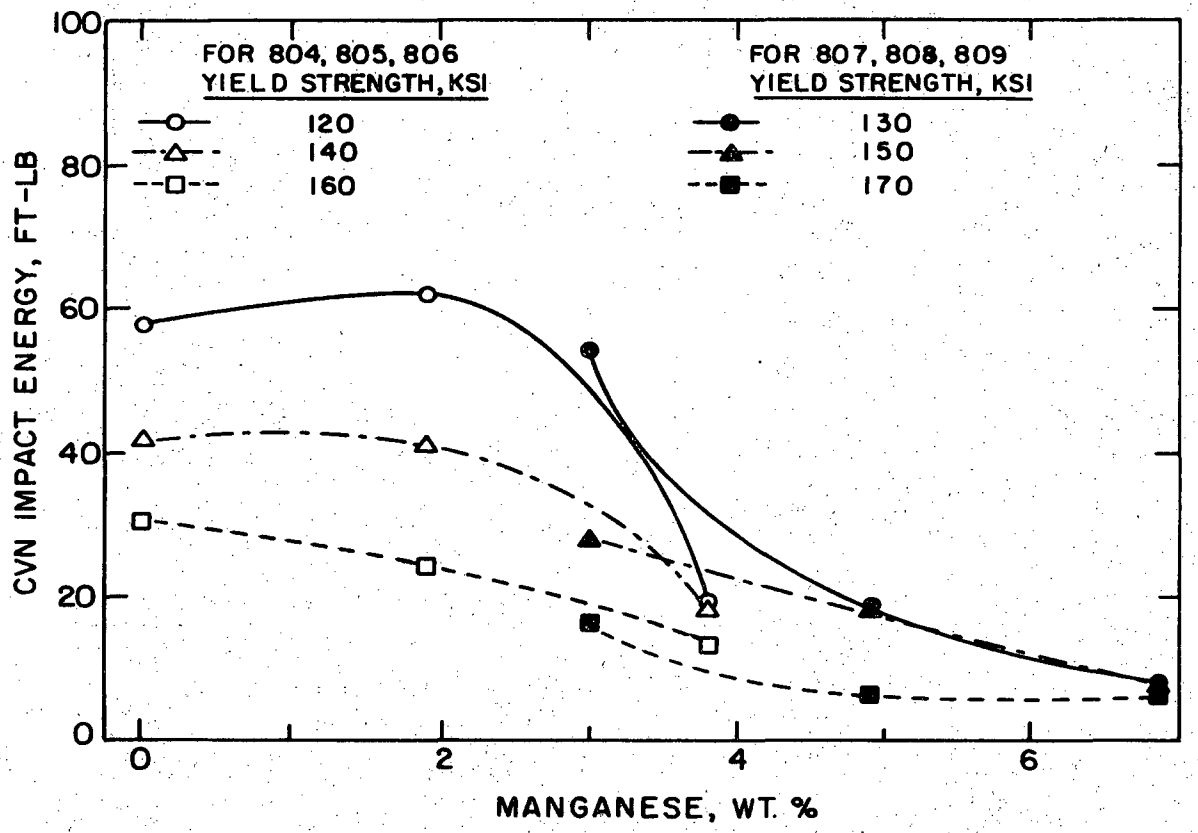
XBL 707-1635

Fig. 1.



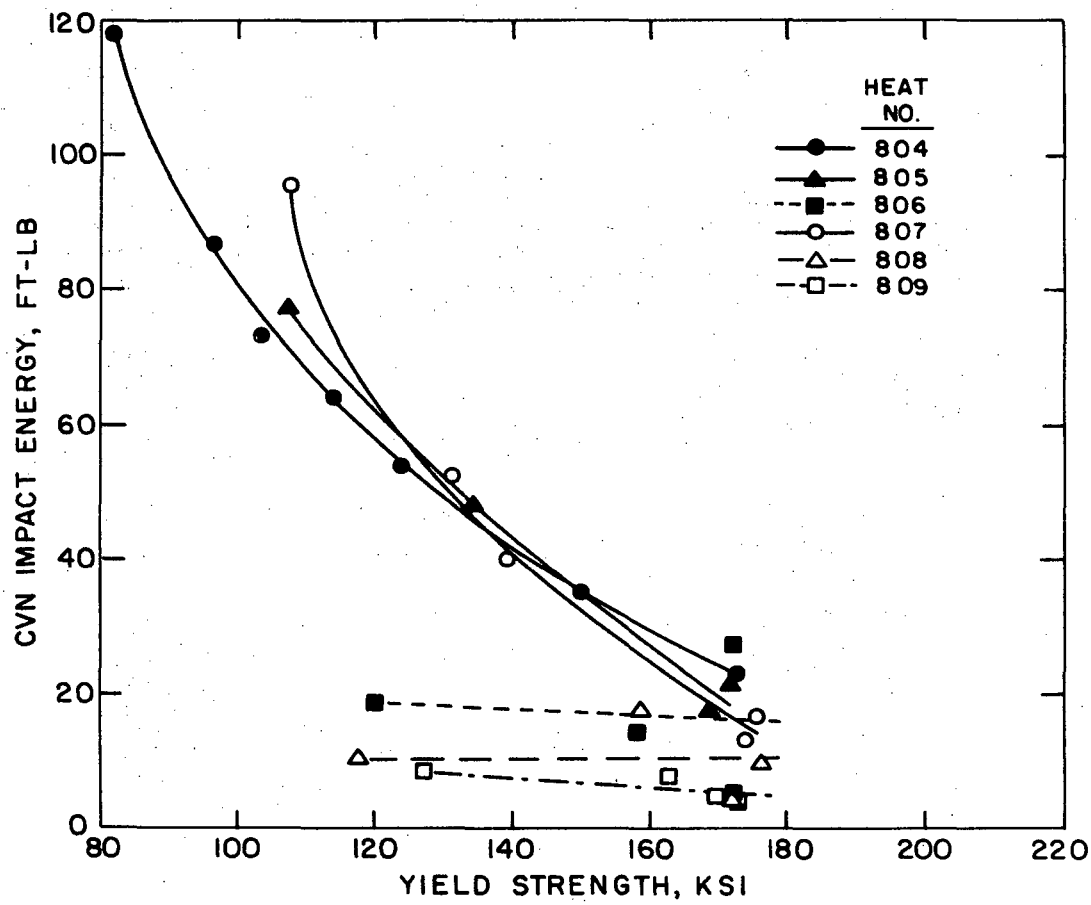
XBL 707-1636

Fig. 2.



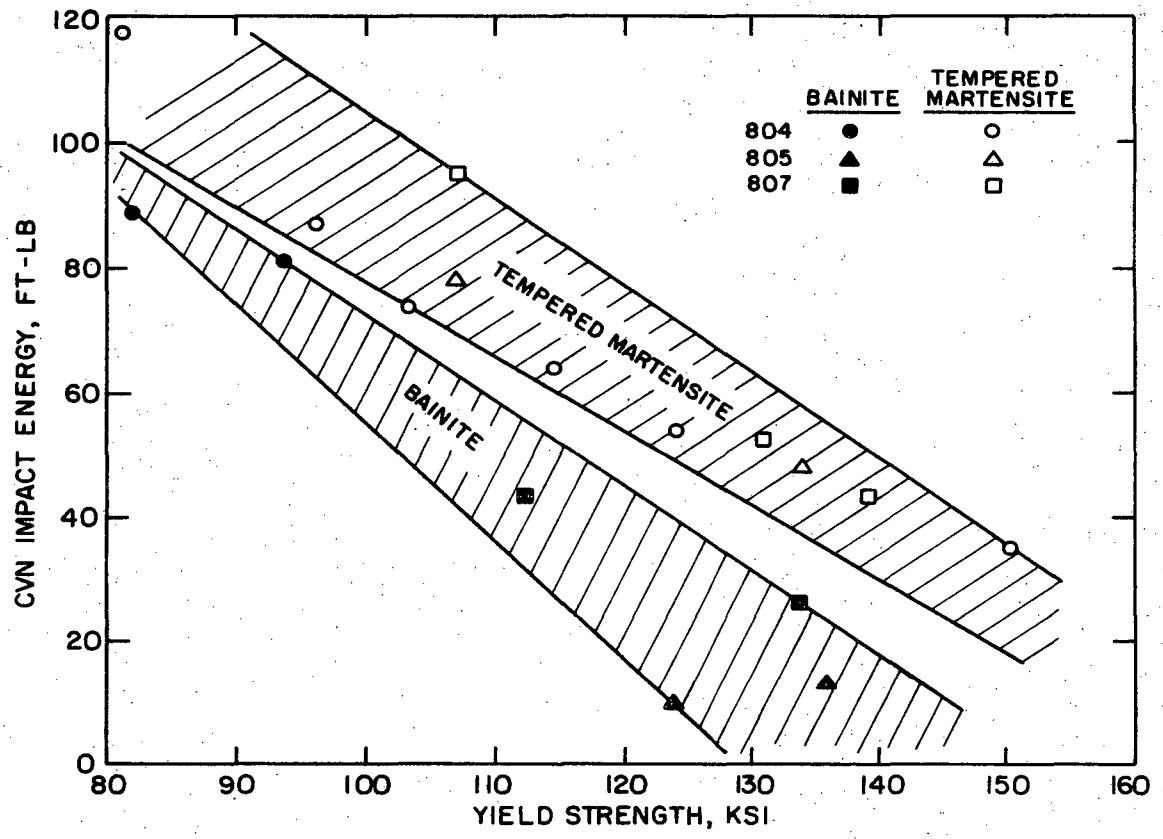
XBL 708-1707

Fig. 3.



XBL 698-1313

Fig. 4.



XBL 707-1634

Fig. 5.

Fig. 6.

Fig. 7.

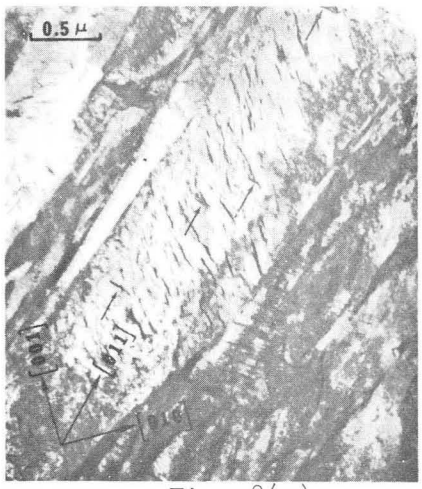


Fig. 8(a).

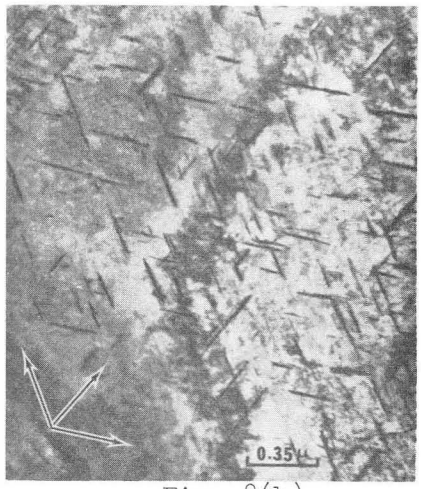
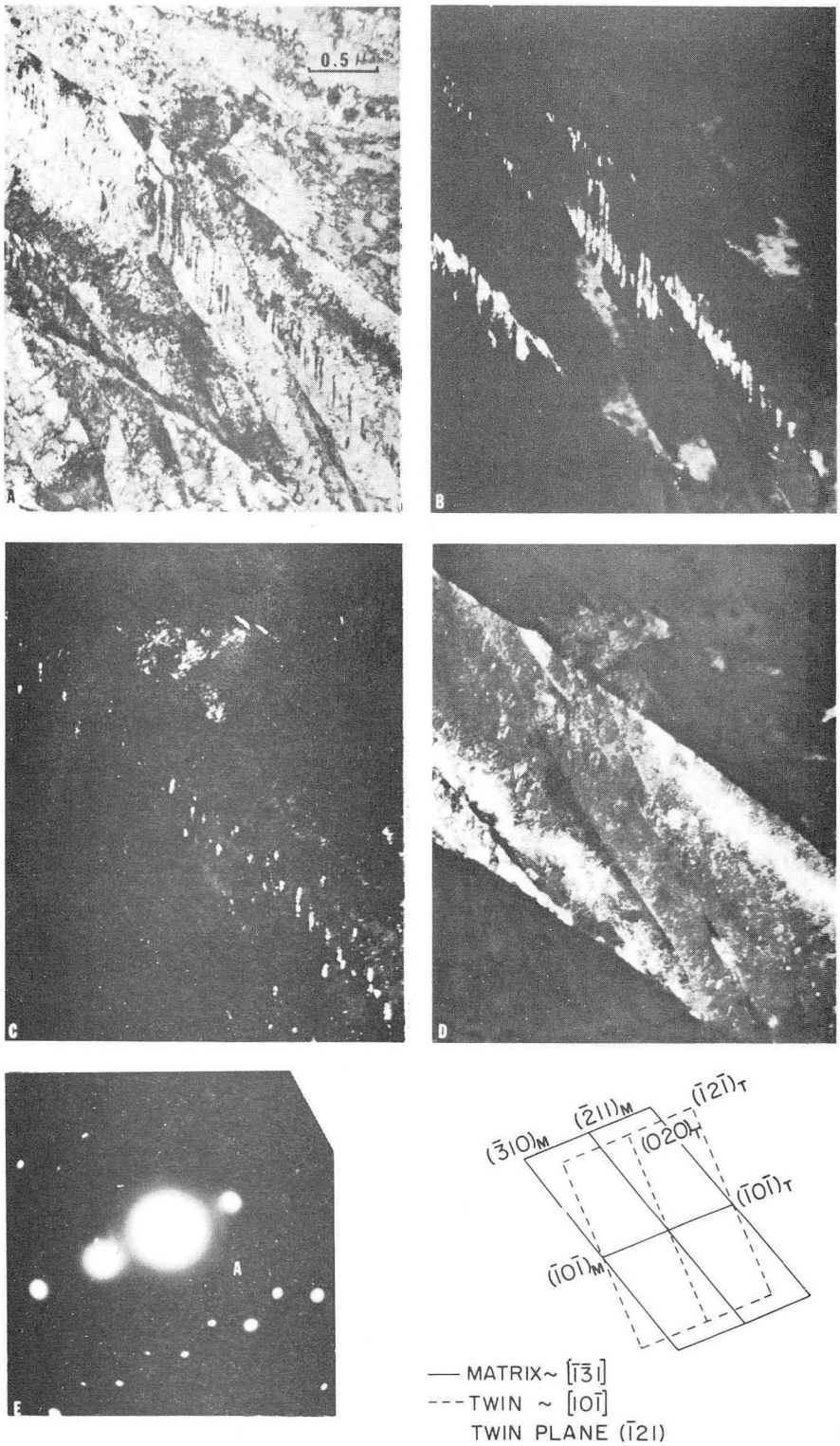


Fig. 8(b).

Fig. 8(c).

Fig. 8(d).

XBB 708-3537



XBB 708-3534

Fig. 9.

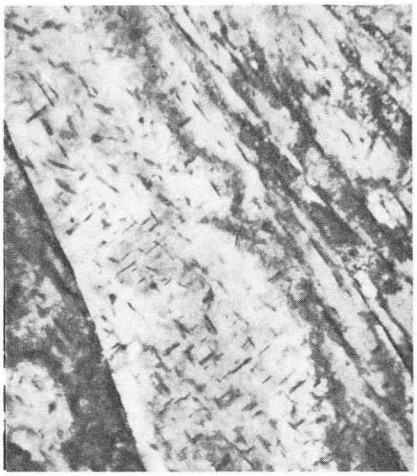


Fig. 10 (a).

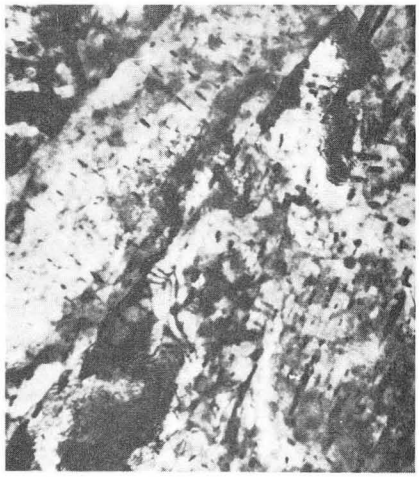


Fig. 10 (b).

Fig. 11.

Fig. 12.

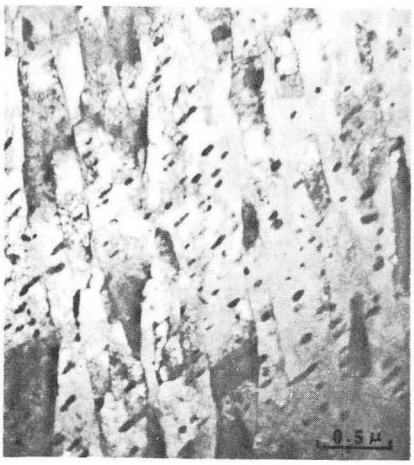


Fig. 13 (a).

Fig. 13 (b).

XBB 708-3536

Fig. 14.

Fig. 17.

Fig. 15 (a).

Fig. 15 (b).

Fig. 16 (a).

Fig. 16 (b).

XBB 708-3535

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