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## STRUCTURE AND MECHANICAL PROPERTIES OF Fe-Ni-Co-C STEELS

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## ABSTRACT

The structure and properties of tempered martensite and bainite were investigated in a series of steels with varying C, Ni, and Co contents. At similar  $M_s$  temperatures, the martensite in the 0.24% carbon steels exhibited very small amounts of twinning compared to that of the 0.4% carbon steels. At similar yield and ultimate tensile strengths, the mainly untwinned martensite showed considerably higher toughness than the heavily twinned ones.

The structure of lower bainite consists of laths or plates of dislocated ferrite with internal carbides but no internal twins. The strength and toughness of isothermally transformed lower bainite was found to decrease with increasing transformation temperature. This was associated with increased coarsening and grain boundary precipitation of the carbides at higher transformation temperatures. At similar strength levels the toughness of lower bainite was found to be superior to that of the heavily twinned martensite, but inferior to that of mainly untwinned martensite. Thus, the role of microtwins in lowering the toughness has been demonstrated in two ways. Firstly, by comparison between the heavily twinned and mainly untwinned martensite, and secondly, by comparing the martensite with bainite. It follows that in cases where twinned martensites are obtained, isothermal aging may be more appropriate than quenching and tempering if high strength, tough steels are desired.

## INTRODUCTION

In recent years considerable interest has grown in the properties of bainitic structures in steels. High strength bainitic steels obtained both by isothermal transformation (1) and continuous cooling (2,3) have been successfully developed. Nevertheless, considerable doubt exists as to whether at equivalent strength levels the bainitic structure is tougher than the tempered martensitic or not. Davenport et al. (4) have shown that in a plain carbon steel (0.74%C) the toughness of bainite obtained by isothermal transformation at 580°F is higher than that of tempered martensite. Similar advantages of bainitic structures have been reported by Waterhouse (5), Klinger et al. (6), and also Hehemann et al. (7). The work on 9 Ni - 4 Co - 0.45C high strength steel has also established the superior toughness of bainite over tempered martensite at equivalent strength levels (1). On the other hand, other researchers (8-10) found the toughness of bainitic structures to be poorer than that of conventional quenched and tempered ones. However, in the latter cases either low carbon ( $\sim 0.2\%$ C) steels (8) or very high isothermal transformation temperatures (above 700°F) were employed (9,10). Thus it appears that in certain cases tempered martensite may possess better toughness than bainite of similar strength levels and in certain cases it may not. In the present work an attempt has been made to investigate whether such variations exist or not, and if so, why.

There have been suggestions (11,12) that the presence of internal twins in martensite may reduce its ductility. So one of the aims of the present work was to see whether micro-twins have any influence on notch toughness or not. It has also been argued (12-14) that the extent of internal twinning in martensitic steels may partially depend on  $M_s$  temperature; the lower

the  $M_s$  temperature, the greater is the tendency for martensite to twin rather than slip. Since carbon and nickel are known to lower  $M_s$  temperature and cobalt raises it, by adjusting Ni and Co contents it is possible to attain the same  $M_s$  temperatures in steels of two different carbon levels. This was necessary to assess the role of individual alloying elements on the occurrence of internal twinning in martensite apart from their influence on  $M_s$  temperature. The composition of the alloys designed on the above basis are given in Table 1 along with their  $M_s$  temperatures. The steels will be referred to by their heat numbers as given in Table 1.

### EXPERIMENTAL

The specimens were austenitized at 1600°F for one hour in an argon atmosphere and then either water quenched (for 0.24%C steels), or oil quenched (for 0.4%C steels). Then they were immediately refrigerated in liquid nitrogen to ensure complete transformation to martensite. To obtain the bainitic structures the specimens were directly quenched into the salt bath at the required temperature and then isothermally transformed. After heat treatment, all the specimens were carefully ground from both sides using excessive coolant. The carbon analysis before and after the heat treatment showed no significant decarburization.

The tensile tests were done on 0.060 in. thick sheet material. Fatigue precracked single edge notch specimens (15) were used for measuring fracture toughness. These specimens were tested at -196°C to obtain plane-strain conditions in the 0.06 inch specimens, but since plane-strain conditions were not obtained in the low carbon series, the parameter K rather than  $K_{IC}$  has been used to denote the apparent fracture toughness values. The room temperature fracture toughness values were obtained on crack-line loaded samples of the type suggested by Mostovoy, et al. (16). The half height (H) of the specimen was 1.2 inches, the width (W) 2.0 inches, the cracklength (a) 0.5 inches, and the thickness (t) 0.5 inches. All tensile tests were done using an Instron machine at a cross head speed of  $4 \times 10^{-2}$  in./min.

The heat treated 0.060 in. thick specimens were first mechanically ground, then chemically thinned and finally electropolished in chromic-acetic acid solution. Thin foils were examined in a Siemens Elemiskop I microscope at 100 kV. Two stage plastic-carbon replicas of the fractured surface of notched specimens were prepared in the usual way.

## RESULTS

### Mechanical Properties of Tempered Martensite and Bainite:

The mechanical properties of tempered martensitic structures are tabulated in Table 2 and those of bainitic structures in Table 3. The values given are the average of three or more tests. Since it is important to compare the toughness of various structures at the same yield and ultimate tensile strength levels, each steel was tempered at various tempering temperatures for the same time. This treatment gave a range of strength levels, so that the toughness could be compared at any desired strength level.

The low carbon steels 141, 142, and 145 showed similar tempering behavior, as typically illustrated in Fig. 1. The high carbon steels 522, 523, and 524 also showed similar tempering characteristics and is represented by Fig. 2. With increasing tempering temperature the yield strengths of the low carbon steels remain more or less constant in the temperature range shown. The elongation increases continuously but the fracture toughness first decreases up to 800°F and then increases on increasing the tempering temperature. This behavior is different from that observed by Pascover and Matas (17) in CVN room temperature impact tests of a steel similar to 141. Their results show a small increase in the toughness up to 800°F after which it rises sharply. Thus, there is an appreciable embrittlement around 800°F when the fracture toughness tests were carried out at -196°C. Similar observations have been made by Payson (18) who observed appreciable embrittlement at -100°F, but little at 80°F in similar low alloy steels. At the same tempering temperatures the strengths of steels 522 and 523 were nearly the same, but those of 524 were always higher. Steel 523 contained higher nickel and 524 higher cobalt than 522. It seems that whilst nickel has

little influence on the tempering resistance of martensite, cobalt greatly increases the tempering resistance, as has been observed before (19).

The fracture toughness values obtained at various yield and ultimate tensile strength levels are shown in Figs. 3 and 4, respectively. The fracture toughness values in the embrittled condition are poor and have not been included in these. There is some scatter in the toughness values obtained for 0.24% carbon steels because the toughness changes on tempering whereas the yield strength is nearly constant. However, all values lie within the band shown. It is seen that at similar yield and ultimate tensile strength levels the toughness of steels 141, 142, and 145 are higher than those of 522, 523, and 524. Also, the steels 145 and 524 have lower toughness than steels 142 and 522 respectively, the former having 7% cobalt and the latter only 4%. This shows that other factors remaining the same, the addition of excessive cobalt lowers toughness at the same strength level.

The bainitic structures in steels 522 and 524 show a lowering of yield and ultimate tensile strengths, and notch toughness, with increasing isothermal transformation temperature. The mechanical properties of bainite are best at the lowest possible transformation temperature above  $M_s$ . The yield strengths obtained in the as-transformed bainite of steels 141 and 145 are poor compared to the yield strengths of the tempered martensite of the same steel. On tempering, the yield strength was found to increase, but the ultimate tensile strength decreased slightly. The transmission electron microscopic observations showed the presence of martensite in these as transformed bainite samples, which means that the austenite did not transform completely to bainite even after isothermal holding for 100 hours. Such mixed structures give poor mechanical properties as is widely recognized.

It was found that the increase in yield strength and fracture toughness on tempering was greater for greater amount of martensite present. Tempering of bainite in steels 522 and 524 showed no increase in yield strength and toughness, but rather a small drop in strength level. This result showed that the transformation to bainite was complete in these two steels. The structural observations also verified this.

Table 4 compares the fracture toughness of bainite and tempered martensite at equal strength levels. The interpolated values for fracture toughness of tempered martensite at the same strength level as that of bainite were taken from the tempering curves. It is seen that the fracture toughness of tempered bainite of steels 141 and 145 is much inferior compared to that of tempered martensite. Although 100 percent bainite could not be obtained in these two steels, still it can be seen that as the amount of bainite increased from steel 145 to 141, and by increasing transformation time in 141 the toughness decreased markedly. The fully bainitic steels 522 and 524, have better notch toughness properties than is obtained in these steels after tempering martensite, both at similar strength levels. Although the toughness of bainite at  $-196^{\circ}\text{C}$  is not so much greater than that of tempered martensite in steels 522 and 524, the room temperature toughness values show a large difference which confirms the results of Pascover and Matas (1).

In summary, at similar yield and ultimate tensile strength levels, for tempered martensitic structures the 0.24% carbon steels show better toughness than the 0.4% carbon steels. On the other hand, the toughness values of bainite in steels 522 and 524 are better than that of the tempered martensite. Increasing the cobalt content of the steels from 4 to 7 percent reduces their toughness.

### Structure of Martensites:

The morphology of martensite in the low carbon steels 141, 142, and 145 is similar and consists mostly of dislocated martensite laths as seen in Fig. 5. Isolated examples of internal twins were also seen in very few plates. Mostly the laths were about 0.25 micron wide and were separated by low angle boundaries whereas laths about 1 micron width were found to be twin related. Unlike the 0.24% carbon steels many examples of completely internally twinned martensite plates were found in the 0.4% carbon steels 522, 523, and 524, as seen in Fig. 6. Very often the fine internal twins gave rise to streaks in the  $\langle 112 \rangle$  direction in the diffraction patterns as seen in Fig. 7. In addition to the twin spots there are other extra spots due to double diffraction.

By examining a number of foils a relative comparison of the twin density was made. It was found that the volume fraction of the twins in all the three 0.4% carbon steels was much higher than that in the 0.24% carbon series, irrespective of their  $M_s$  temperatures. The steels 142 and 524 have nearly the same  $M_s$  temperatures, but the extent of internal twinning is much less in 142 compared to 524 (Fig. 8). This means that even by controlling the quantity of substitutional elements such as nickel and cobalt to obtain the same  $M_s$  temperatures in steels of differing carbon contents, the extent of internal twinning can still be widely different. Thus,  $M_s$  temperature alone does not control whether martensite is twinned or not.

### Structure of Tempered Martensite:

The microstructures of martensite tempered at various temperatures were examined. Structures typical of low and high carbon alloys are given in Fig. 9 and 10, respectively. Tempering of 0.24% carbon steels at 400°F resulted in precipitation of mostly cementite. Since the martensite in

these steels show very little twinning, the cementite precipitated predominantly in a Widmanstätten pattern with  $\{110\}$  habit. Precipitation along lath boundaries was not detected until tempering was done at  $1000^{\circ}\text{F}$ . This precipitation did not cause intergranular failure as shown by Fig. 14(c).

The martensite of the 0.4% carbon steels tempered at  $400^{\circ}\text{F}$  showed mainly  $\epsilon$ -carbide with a small amount of  $\text{Fe}_3\text{C}$ . Although these  $\epsilon$ -carbide precipitates did not give diffraction patterns satisfactory for unique identification, the trace analysis of the precipitates showed a  $\{100\}_{\alpha}$  habit. It is known (20,21) that  $\epsilon$ -carbide has  $\{100\}_{\alpha}$  habit and also grows in  $\langle 100 \rangle_{\alpha}$  directions. The amount of  $\epsilon$  decreased with increasing tempering temperature; at  $800^{\circ}\text{F}$  only cementite was detected.

In twinned martensites, cementite precipitated preferentially on the  $\{112\}$  twin boundaries, as showing in Fig. 10(c). There are also precipitates along the grain boundaries. Similar observations have been made by Baker et al. (22), but they did not observe carbides at the martensite plate boundaries in the presence of twins. Frequently, the twins are not very clear due to the presence of carbides but by proper tilting they can be brought to better contrast [Fig. 10(d)]. This shows that it is not easy to remove the twinned structure by tempering, unless recrystallization occurs ( $>1100^{\circ}\text{F}$ ).

Structure of Bainite:

As discussed before, the strength and toughness of bainite deteriorated with increasing isothermal transformation temperature, even in the lower bainite range as was found by Pascover and Matas in 9Ni-4Co-0.4C steels (1). Hence, the structures of lower bainite with increasing isothermal transformation temperature were examined as shown in Fig. 11. The structure of lower bainite is known to consist of acicular ferrite with carbides at 55-65° to the long direction of ferrite (23,24). The carbide particles were found to become coarser with increasing isothermal transformation temperature. As expected, the yield strength consequently decreases. Also on transforming at 600°F, the carbides start precipitating at the plate boundaries as can be seen in Fig. 11(b). The lower bainite formed at 550°F in steel 524 is similar to that in steel 522.

At higher transformation temperatures, in the low carbon steels, mostly bainite laths were observed instead of plates. These laths were found to be parallel with only slight misorientations between them as seen in Fig. 12. This structure is similar to that of dislocated lath martensite. Similar examples were also found in bainites of high carbon steels. The number of carbide particles is small in these low carbon bainites, compared to that in the high carbon ones, which is expected. It is also interesting to note that there is no precipitation at the lath boundaries in 0.24% carbon bainite although they were formed above 600°F, whilst the 0.4% carbon bainite showed lath boundary precipitation. In no case was internal twinning observed in the bainitic ferrite. The precipitation of cementite on the twin boundaries does not remove internal twins in martensite even on tempering at 1100°F as has been seen before. Thus, if there were internal twins in the bainitic

ferrite they could not have been removed by aging for 4 hours at 550°F. Figure 13 shows an incompletely transformed bainite where internal twinning is seen in martensite, but not in bainite. Even at the lowest transformation temperature (400°F), no twinning was found in bainite. This is in disagreement with what has been discussed by Shackleton and Kelly (24).

As has been mentioned before the bainite reaction in steels 141, 142, and 145 did not go to completion even after isothermal holding for 100 hours. When examined the structures showed some areas of martensite. Sometimes it becomes very difficult to distinguish between martensite and bainite at least in alloys where the martensite is not twinned. They may be distinguished by observation of the carbide morphology after tempering.

## DISCUSSION

### Toughness of Tempered Martensite and Bainite:

In many high strength steels the so-called 500°F embrittlement is observed when tempering in the 400-700°F range. This has been associated with the re-resolution of  $\epsilon$ -carbide and simultaneous precipitation of cementite (25). In the present investigation embrittlement was observed in almost all cases around 800°F. In the 0.24% carbon steels little  $\epsilon$ -carbide was observed at 400°F, and at 600°F the carbide was almost all  $Fe_3C$ . Hence, the observed embrittlement cannot be explained by the carbide transition mechanism. However, it is very difficult to draw any conclusions as to the cause of embrittlement from the observed microstructural changes.

The toughness difference between the martensite and bainite can now be related to the observed structural differences and the following factors appear to be important.

1. Twinning: The mechanical test results have established that at equivalent strength levels the toughness of lower bainite structure is not always superior to that of tempered martensite. Only in the 0.4% carbon steels did the mainly twinned tempered martensite show poor toughness compared to the bainitic structures. The martensite of 0.24% carbon steels which showed a small amount of twinning possessed a very high toughness, compared to both the twinned martensite and dislocated bainite.

However, it remains to be seen whether twinning directly or indirectly is really responsible for the loss in toughness. The problem is complicated in comparing untempered martensites because of solid solution strengthening due to carbon; furthermore, in bainite and martensite the carbide morphology is usually different. However, in spite of these complications, a correlation between twinning and embrittlement is apparent. Previous work indicates

that the embrittlement may be associated with the restriction of slip due to the presence of transformation twinning. For example, Krauss and Pitsch (26) observed mechanical twins intersecting the transformation twins in martensite plates that were deformed after transformation. Also Richman (27) found mechanical twinning to be effectively the only plastic deformation mode in high carbon martensite.

## 2. Size and Distribution of Ferrite and Carbide:

The ferrite grain size is very important in controlling the toughness (28). The smaller the ferrite grain size the greater is the energy absorbed during cleavage. In the present case the ferrite lath size in the martensite and bainite were nearly the same. Hence, their contribution to toughness can be taken to be the same in both cases.

In the 0.24% carbon steels the martensite consists of a mixture of laths, that are separated by low angle boundaries and multidirectional plates, whereas the bainite in the same steel shows predominantly laths. So the fracture path may be easier in these laths compared to multidirectionally arranged plates, because in the latter case the fracture path has to change direction frequently in going from one grain to the other, and so will absorb more energy. This can contribute to the higher toughness of the martensite in low carbon steels compared to bainites. The presence of internal twins in the multidirectional plates of 0.4% carbon martensite over-rides this effect and lowers the toughness.

Unless precipitate particles can deform with the matrix, carbides can act as stress raisers and crack nucleators. However, it is observed that provided there is little grain boundary precipitation, the toughness is better for the Widmanstätten {110} cementite habits, such as is obtained on tempering the 0.2% carbon steels.

At the same carbon level the volume fraction of carbide in the bainitic and tempered martensitic structures do not appear to be greatly different. However, in the lower bainite the volume fraction of carbides appears to be increasing with increasing isothermal transformation temperature which may be partly responsible for the loss in toughness.

### 3. Grain Boundary Carbides:

The grain boundary carbides were prominent in martensite only on tempering at the higher temperatures. These grain boundary carbides in the 600°F bainitic 0.4% carbon steels may be somewhat responsible for their lower toughness compared to the 500°F bainitic steels. In the tempered martensite the fine discontinuous carbides at the lath or plate boundaries do not appear to be very effective in promoting intergranular failure, (Fig. 14). Since both high and low carbon tempered martensite showed such discontinuous grain boundary precipitation, its weakening effect can be more or less the same. These discontinuous boundary precipitates do not reduce toughness very much unless the precipitates are continuous all along the boundary, as has been found in some precipitation hardening aluminum alloys (29).

### Effect of Composition on Internal Twinning in Martensite:

It has been shown that at similar  $M_s$  temperatures the martensite in a high carbon steel (524) shows much more internal twinning than that in a lower carbon steel (142). Thus, it appears that increasing carbon promotes internal twinning in the martensite, in addition to lowering the  $M_s$  and  $M_f$  temperatures. This is in contrast to the suggestions by Kelly and Nutting (30) that the carbon content of the steel has only an indirect effect on the morphology of martensite through its control of  $M_s$  and  $M_f$  temperatures.

In simple binary alloys of Fe-Ni or Fe-C decreasing  $M_s$  temperature gives increasing internal twinning, but at the same  $M_s$  temperature all of them will not have the same extent of twinning (e.g., Fig. 8). The situation is further complicated when several other alloying elements such as Co, Cr are also present, in which case their influence on the occurrence of microtwins in the martensite cannot be predicted from their simple effect on  $M_s$  temperature. Whether a martensite plate will be twinned or slipped will depend on two factors. Firstly, the critical resolved shear stress necessary to cause slip or twinning and secondly, the stress available from the transformation. At a particular  $M_s$  temperature if the CRSS for twinning is less than that for slip the martensite will be twinned rather than slipped as has been discussed by Johari and Thomas (13). The composition may influence the CRSS for slip and for twinning differently. Thus, although the addition of cobalt raises the  $M_s$  temperature, cobalt does not reduce twinning unless the  $M_s$  temperature is such that the CRSS for slip is lower than that for twinning. Cobalt is thus not useful in this respect for high carbon steels. The stress generated during the transformation depends on the crystallography of the martensite formation, which is again composition dependent. This stress may be proportional to the driving force of the reaction. The differences  $(T_0 - M_s)$ ; where  $T_0$  is the equilibrium temperature  $[\frac{1}{2}(A_s + M_s)]$  or more appropriately the area of the hysteresis loop of the martensitic transformation can be a measure of the driving force. A larger hysteresis gap will mean a greater driving force and consequently a greater chance for the martensite plate to be twinned. This idea is similar to that discussed recently by Pascover and Radcliffe (31).

Thus, the influence of the various alloying elements on twinning in martensite can be understood by their effect on the area of the hysteresis loop together with their influence on CRSS for slip and twinning.

Design of Alloys:

One aim in alloy design is to achieve high toughness at a given strength level. The addition of carbon is helpful in achieving high strength, but in martensitic steels it also increase % twinning which severely impairs toughness. In the case of heavily twinned martensite, the bainitic structure should be more desirable for high toughness, hence isothermal transformation is more promising for high carbon steels than for low carbon steels. In the latter, twin free tempered martensite produces good mechanical properties. Without resorting to thermomechanical treatments such as ausforming, which is helpful in decreasing twinning in martensite (32), it seems that future research needs to be directed towards finding suitable alloying additions which can prevent twinning in martensite. The addition of cobalt has been shown not to be very effective in this respect, even though it raises  $M_s$ .

CONCLUSIONS

1. Although it was difficult to obtain exactly the same strength levels in the low and high carbon steels, at similar yield and ultimate tensile strength levels the higher toughness of tempered martensite in the 0.24% carbon steels compared to the 0.4% carbon steels is related to the greater extent of internal twinning in the high carbon steels.
2. On increasing the isothermal transformation temperature, the strength and toughness of the bainite decreases. This is mainly due to carbide particle coarsening and carbide precipitation at the grain boundaries.
3. At equivalent strength levels the toughness of lower bainite transformed at lower temperatures is higher than that of the heavily twinned tempered martensite.
4. In the absence of internal twinning the toughness of tempered martensite is higher than that of lower bainite. This is due mostly to the differences in morphology of the ferrite and carbides in the two structures.
5. The mechanical properties of the incompletely transformed bainite are very poor because of the presence of virgin martensite.
6. The addition of cobalt >4% decreases the toughness.
7. The  $M_s$  temperature of the alloy is not a sufficient indication of the extent of internal twinning in martensite. Although Co raises  $M_s$ , it does not decrease the % twinning. It is suggested that the driving force for transformation, as given by the area of the hysteresis loop may give a better indication of the occurrence of twinning in martensite.
8. The ferrite in lower bainite shows no internal twinning at all. The proportion of laths in the lower bainite increase with increasing isothermal transformation temperature.

9. Isothermal transformation is a preferred heat treatment to yield a steel with superior high strength-toughness properties than quenching and tempering when the composition is such as to give mainly twinned martensite (e.g., high carbon steels).

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FIGURE CAPTIONS

- Fig. 1 - The effect of tempering temperatures on strength and toughness of steel 142, which is typical of the 0.24%C series.
- Fig. 2 - The effect of tempering temperature on strength and toughness of steel 523, which is typical of 0.4%C series.
- Fig. 3 - Comparison of fracture toughness (K) of various steels at different yield strength levels.
- Fig. 4 - Comparison of fracture toughness (K) of various steels at different ultimate tensile strength levels.
- Fig. 5 - (a) Dislocated lath martensite in steel 145 which is also typical of steels 141 and 142. The laths occur in bundles with slight misorientations between laths in each bundle, (b) selected area diffraction of encircled area in (a); (c) and (d) dark-field of spots A and B, respectively.
- Fig. 6 - Twinned martensite in steel 522, showing various plate sizes most of which are twinned.
- Fig. 7 - Selected area diffraction from a twinned martensite plate (see Fig. 6) showing streaks in  $\langle 112 \rangle$  direction, twin spots and double diffraction spots.
- Fig. 8 - (a) Martensite of steel 142 with no twinning, (b) Martensite in steel 524 showing twinning. The  $M_s$  temperature in both is the same.
- Fig. 9 - Martensite of 0.24% carbon steels tempered for 4 hours at (a) 400°F (steel 141), (b) 1000°F (5Ni-0.5Co steel) showing the Widmanstatten morphology of cementite.
- Fig 10 - Martensite of steel 522 tempered for 4 hours at (a) 400°F, (b) 600°F, (c) 1100°F, (d) same as (c) but twins in another area are brought to better contrast by tilting.

- Fig. 11 - Structure of isothermally transformed lower bainite in (a) steel 522 after 24 hours at 500°F, (b) steel 522 after 24 hours at 600°F, the grain boundary carbide is pointed out by the arrow mark.
- Fig. 12 - Structure of lower bainite in steel 141 after isothermally transforming for 24 hours at 600°F.
- Fig. 13 - Mixed structure of twinned martensite and twinfree bainite in steel 523 isothermally transformed at 400°F for 140 hours.
- Fig. 14 - Electron fractographs of notched specimens fractured at -196°C, (a) Quasi cleavage failure of tempered martensite in steel 522 tempered for 4 hours at 800°F with some dimples, (b) Equiaxed dimples in the lower bainite of steel 522 transformed for 24 hours at 500°F, (c) Dimpled rupture of martensite in steel 142 tempered for 4 hours at 400°F, which is also typical of 1000°F tempering.

TABLE 1

Chemical Composition of the Alloys With Their  $M_s$  Temperatures

Heat No.	C	Mn	Si	Cr	Mo	V	Ni	Co	$M_s$	Temp. °F
141	.24	.33	.086	.48	.49	.083	8.4	3.9	590	
142	.24	.31	.053	.44	.48	.086	12.5	3.8	498	
145	.24	.31	.094	.44	.44	.075	11.3	7.5	584	
522	.43	.40	.14	.52	.54	.087	8.6	3.97	436	
523	.39	.38	.10	.50	.54	.087	11.95	4.0	286	
524	.40	.40	.15	0.52	0.57	0.09	8.3	7.2	486	

TABLE 2

Mechanical Properties of Tempered Martensitic Structures\*

Heat No.	Tempering Temperature °F	Y.S. x10 <sup>3</sup> psi (0.2% off set)	UTS x10 <sup>3</sup> psi	% Elongation	K, KSI-IN <sup>1/2</sup> At -196°C
141	400	176.7	207	9.8	135.0
	600	175.5	195.6	9.0	118.7
	800	172.2	183	10.2	97.8
	1000	175	182	11.9	115.4
142	400	173	203	9.4	158.2
	600	175.4	194.6	10.3	139.7
	800	173.5	186.3	11.9	126.5
	1000	175.2	180.7	12.7	139
145	400	172.0	204.0	9.5	139.3
	600	172.5	194.9	9.6	123.7
	800	174.3	187	10.8	47.5
	1000	176.2	181.9	12.1	101.4
522	400	226.0	266.4	8.5	40.1
	600	216.7	243.2	7.3	46.4
	800	192.5	224.3	9.3	38.1
	1000	191.3	215.5	10.0	64.8
	1100	194.3	211.8	9.8	74.6
523	400	229	266	8.1	43.5
	600	208.6	242.2	8.9	43.9
	800	194	230	10.4	36.6
	1000	190	214	11.4	63.8
	1100	185.4	211.3	11.8	68.2
524	400	251	288	6.8	35.6
	600	237.5	267	6.8	33.9
	800	214.5	240.2	7.6	29.3
	1000	200.1	224.3	8.1	39.2
	1200	184.3	195.1	8.9	

\* Double Tempered at each temperature for 2 + 2 hours.

TABLE 3

## Mechanical Properties of Bainitic Structures

Heat No.	Isothermal Transformation Temperature	Transformation Time in hours	Tempering Temperature °F	Y.S. x10 <sup>3</sup> psi (0.2% off set)	U.T.S. x10 <sup>3</sup> psi	Elongation %	Kat -196°C Ksi - in <sup>1/2</sup>
141	620	24	-	146.4	188.2	11.4	54.6
	620	24	1000	173.6	186	12.9	56.8
	620	100	-	145	176	10.5	45.8
	620	100	1000	158.5	176.5	12.0	46.1
145	620	24	-	135.2	200.3	12.3	54.1
	620	24	1000	176.8	191.9	13.6	73.1
522	500	24	-	188	232.3	10.6	50.5
	620	28	-	169.0	212.0	11.3	32.6
	500	24	1000	183.0	198.5	12.1	49.2
524	550	24	-	205.4	237.8	7.9	35.7
	550	24	1000	202.0	219.5	8.7	32.6
	600	24	-	181.5	229.0	6.9	34.1

\* Double tempered at each temperature for 2 + 2 hours

TABLE 4

Comparative Toughness Values of Tempered Martensitic and Bainitic Structures at Similar Strength Levels

Heat No.	Strength Level Ksi	K <sub>Ic</sub> at 25°C Ksi-in <sup>1/2</sup>		K at -196°C Ksi-in <sup>1/2</sup>	
		Tempered Martensite	Bainite	Tempered Martensite	Bainite
141	Y.S. ~ 173	-	-	102	56.8
	UTS ~ 186	-	-	98	56.8
145	Y.S. ~ 177	-	-	111	73.1
	UTS ~ 192	-	-	93	73.1
522	Y.S. ~ 190	-	-	44	50.5
	UTS ~ 230	66.5	90.2	41.5	50.5
524	Y.S. ~ 205	54.2	82	33.5	35.7
	UTS ~ 238			30	35.7

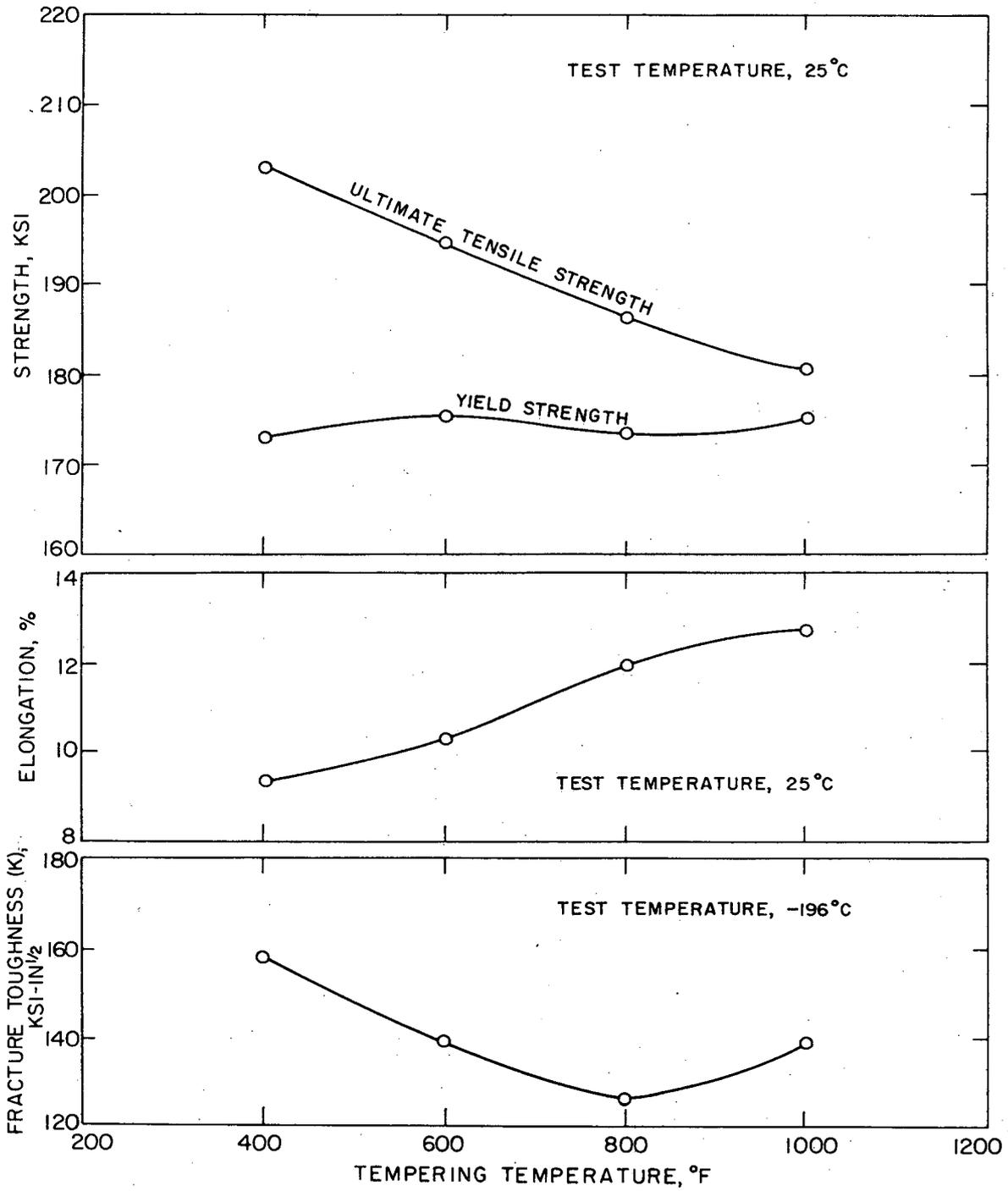


Fig. 1

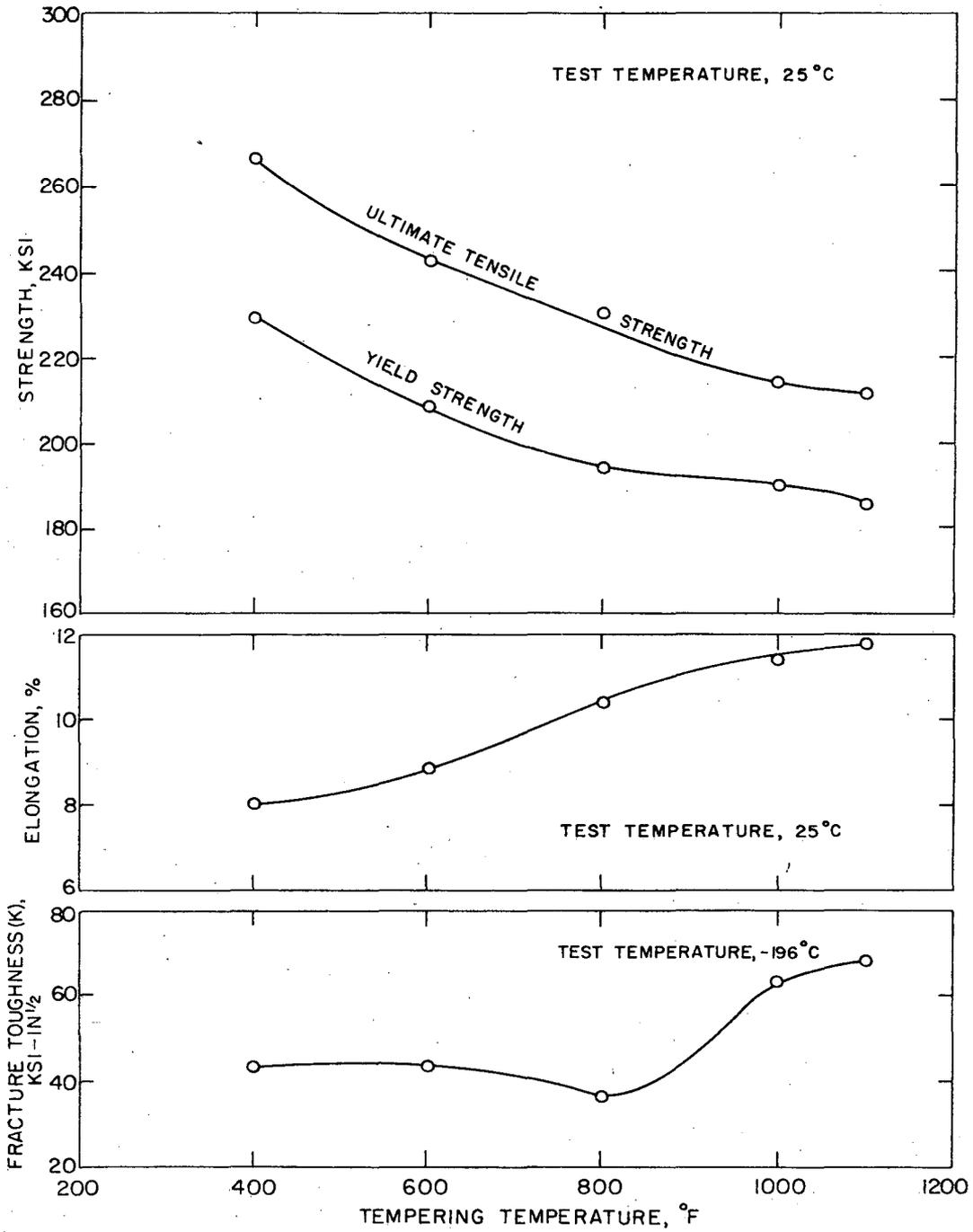


Fig. 2

XBL 6811-6187

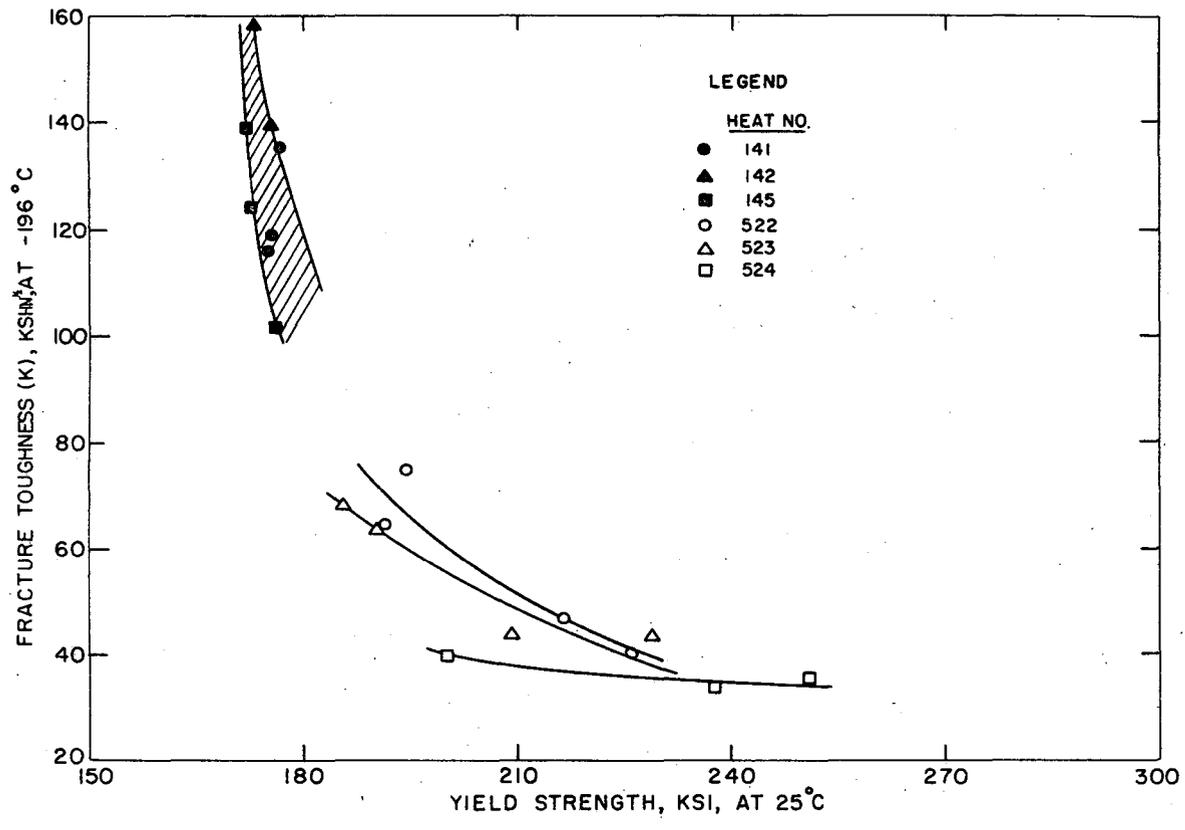
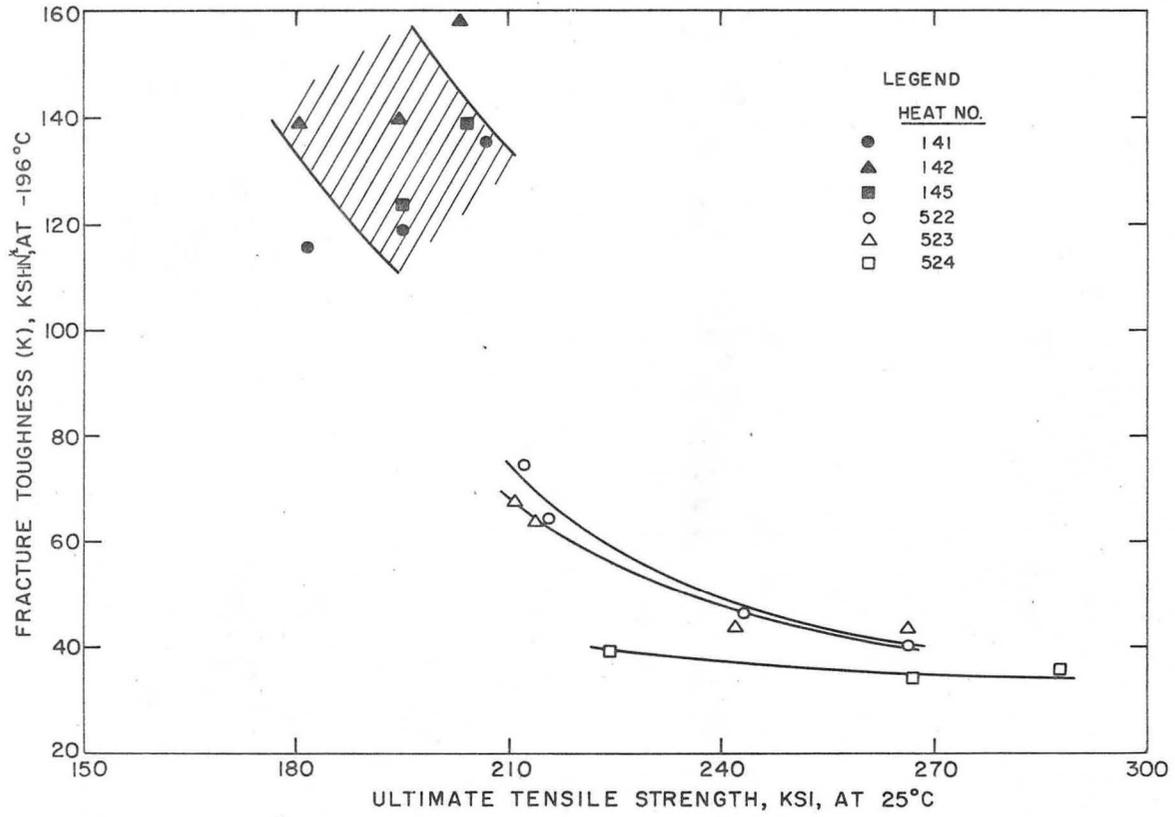


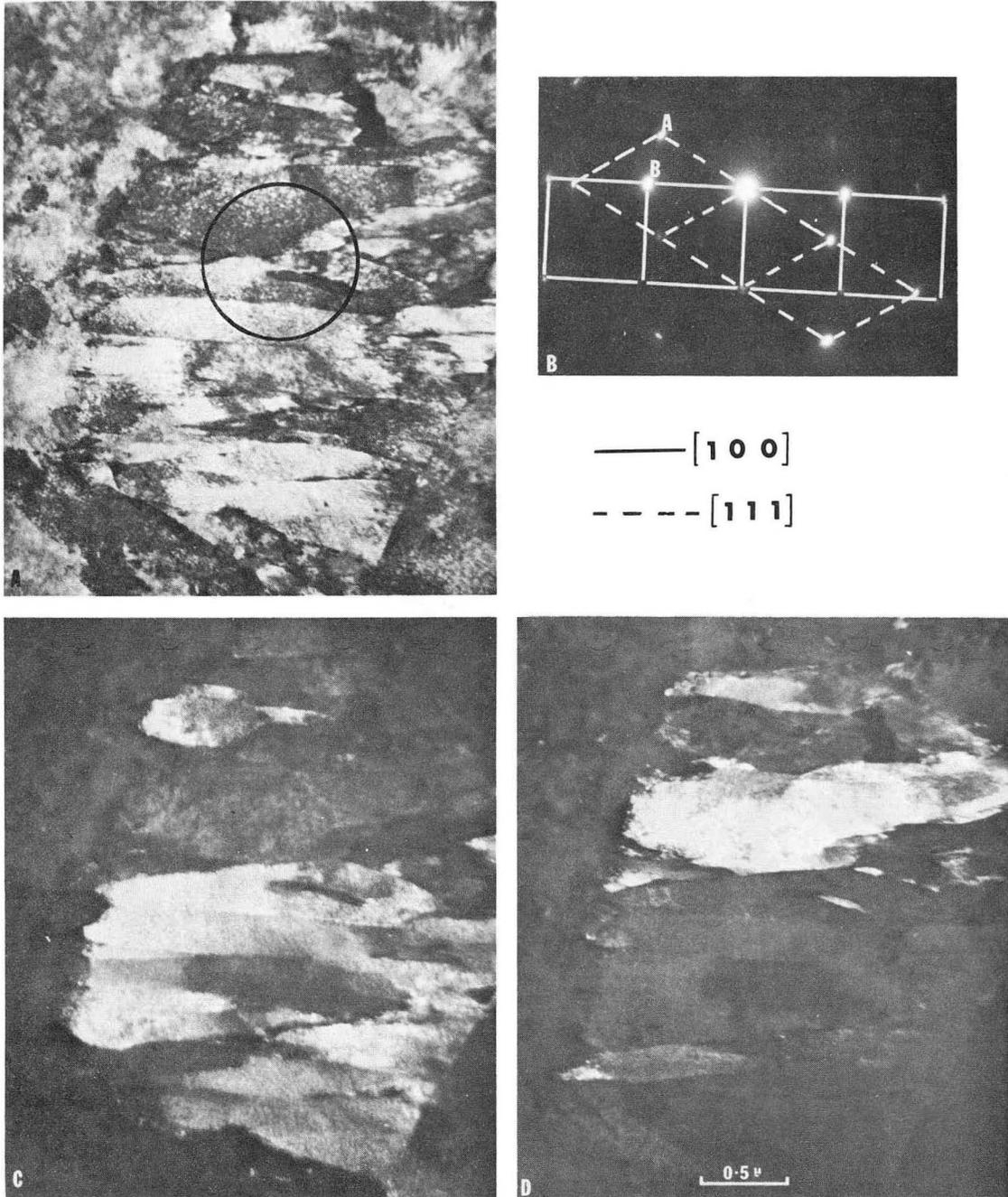
Fig. 3

XBL 6811-6181



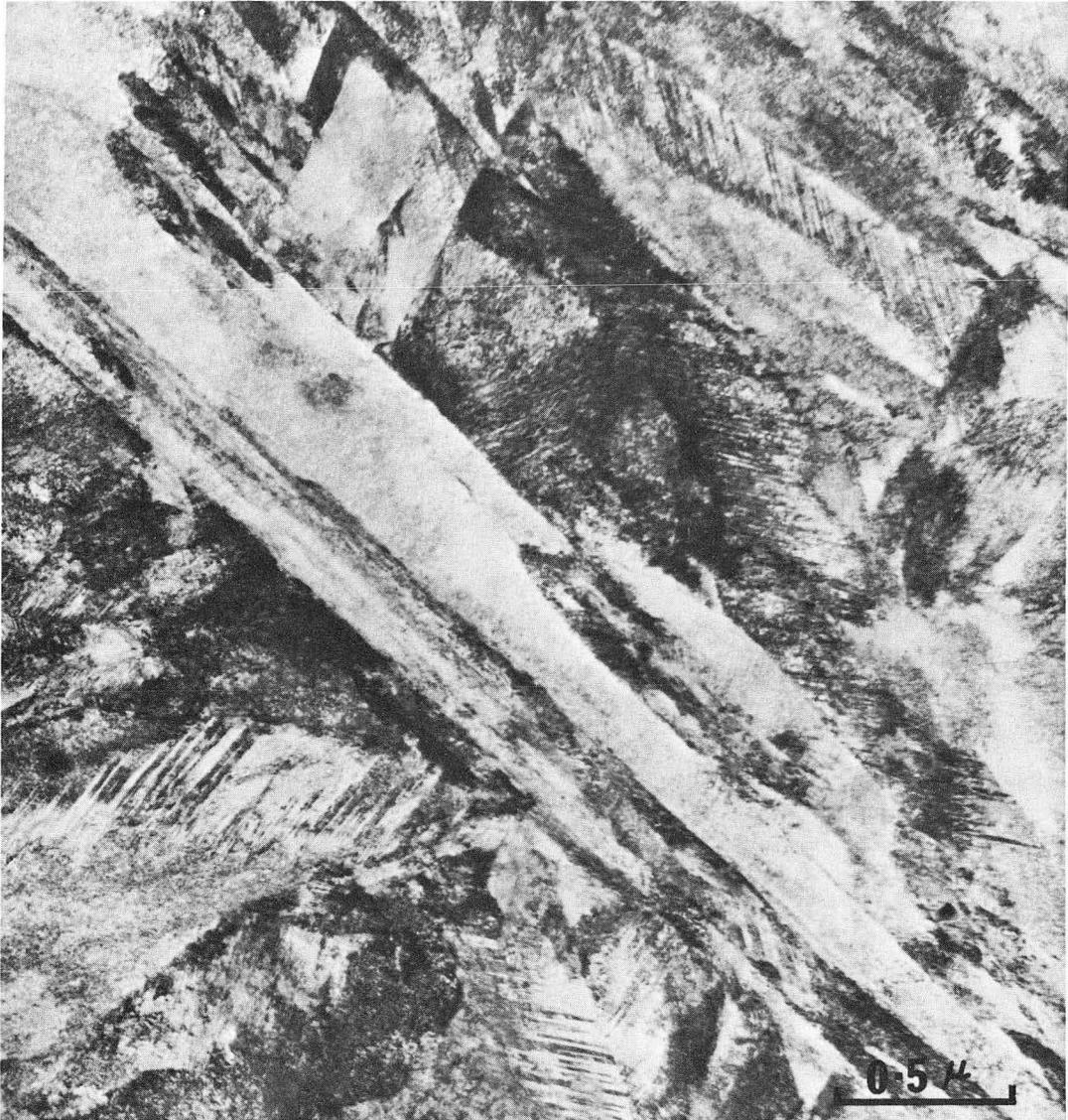
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Fig. 4



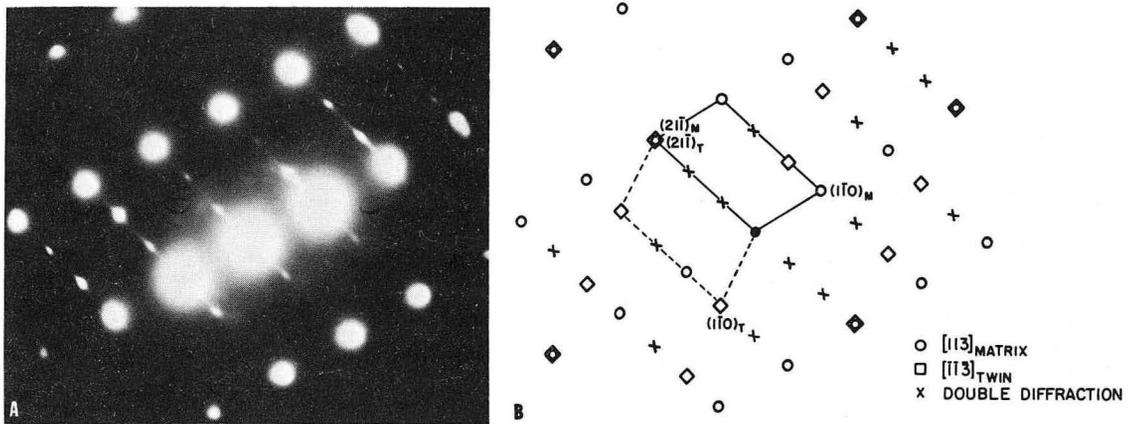
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Fig. 5



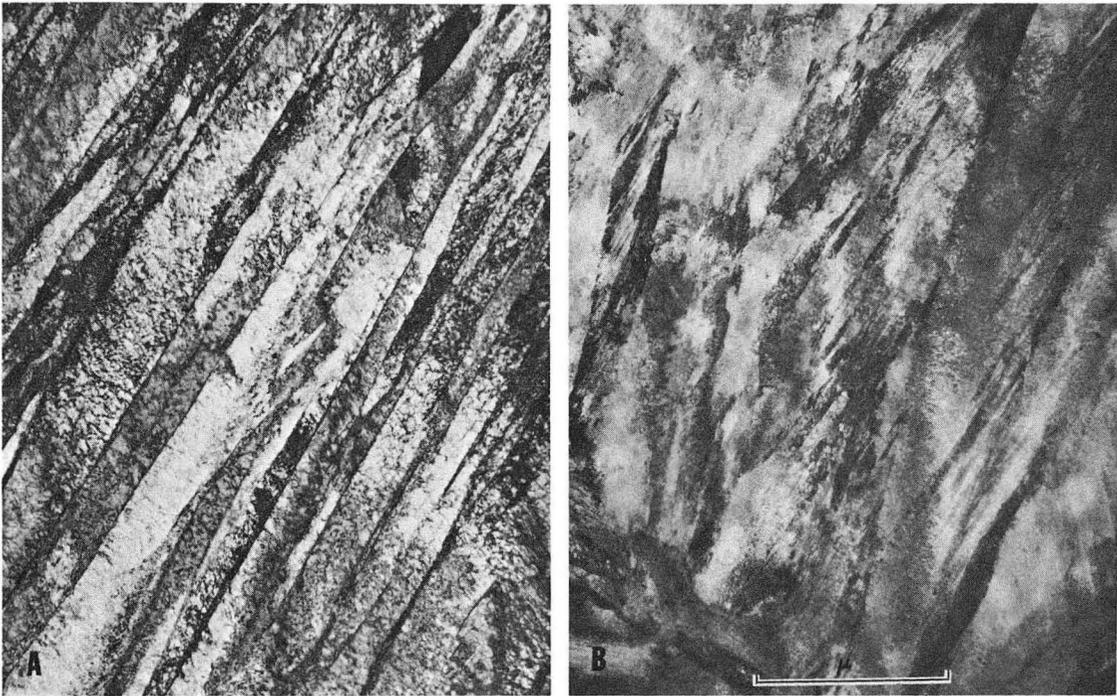
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Fig. 6



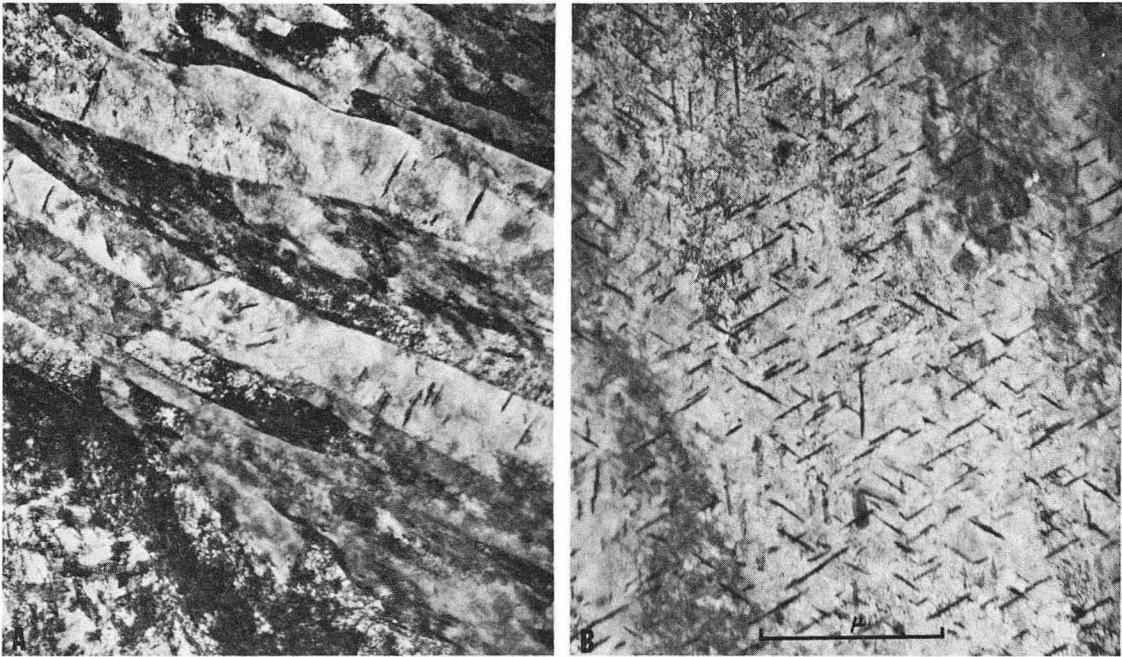
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Fig. 7



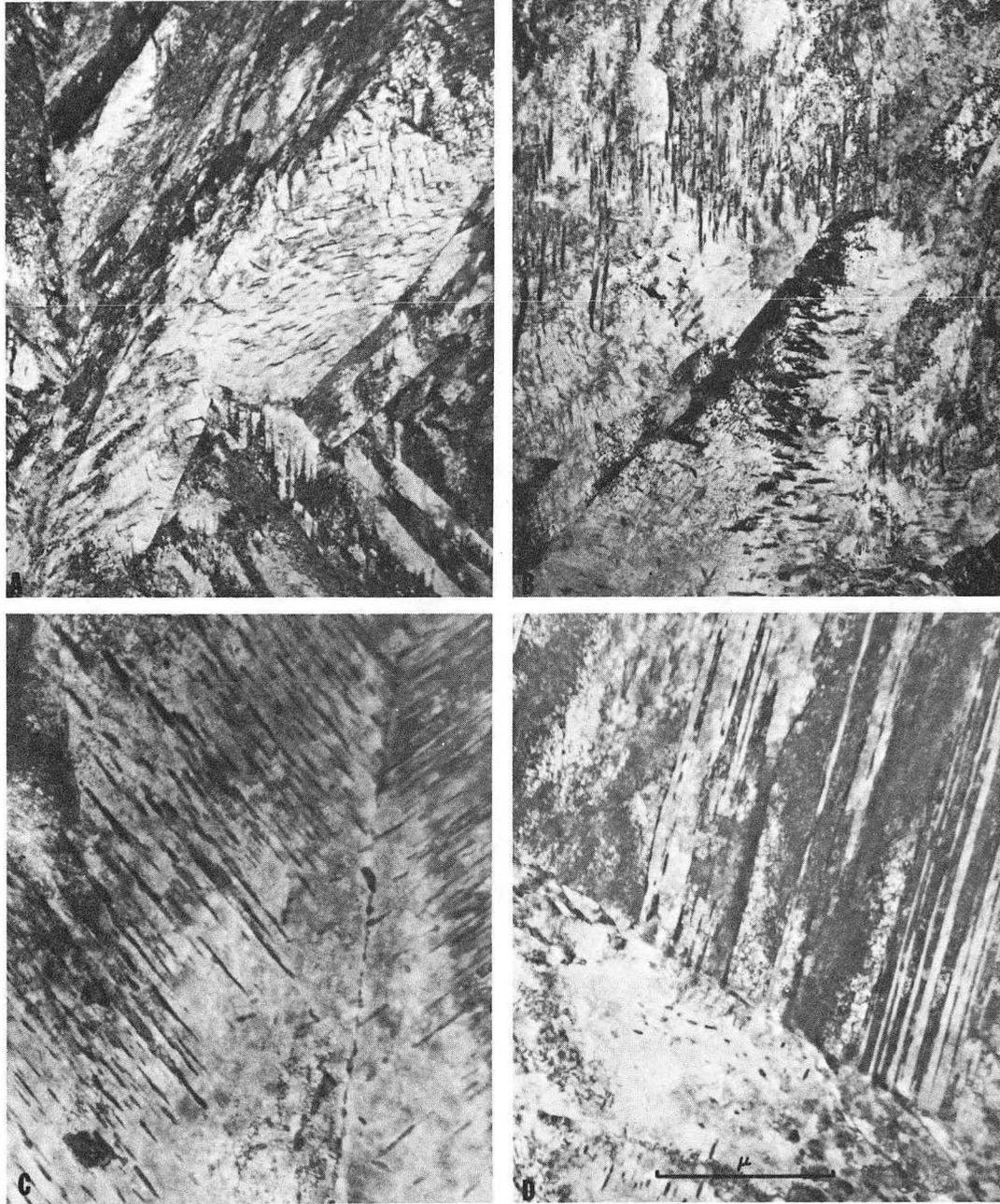
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Fig. 8



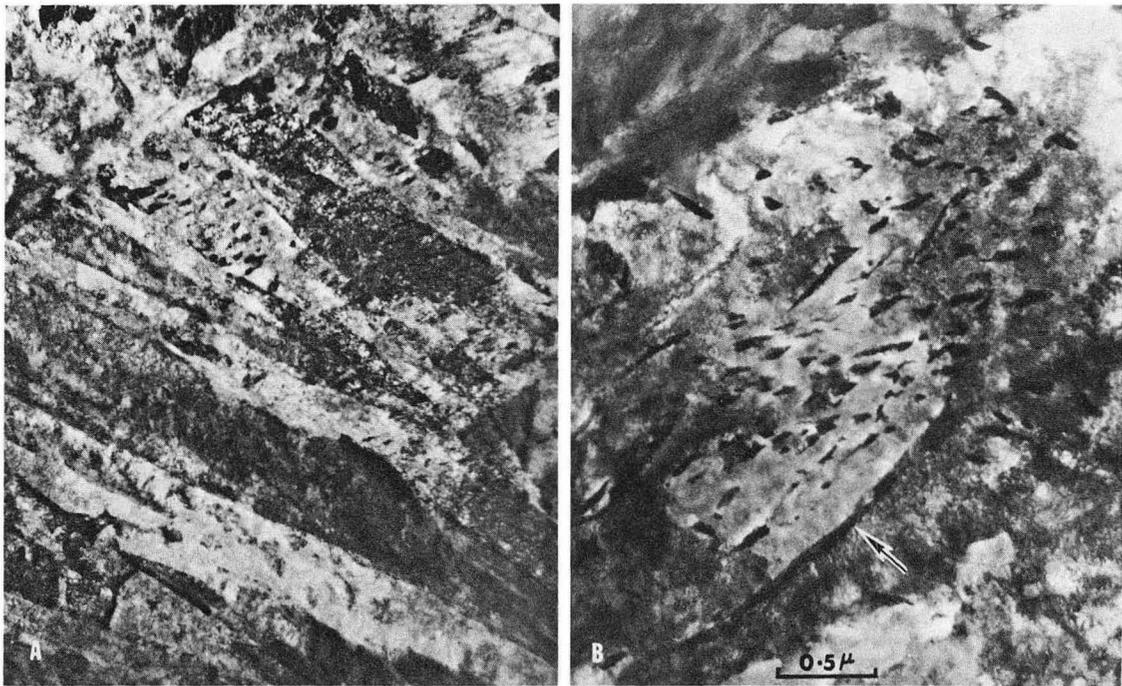
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Fig. 9



XBB 692-1538

Fig. 10



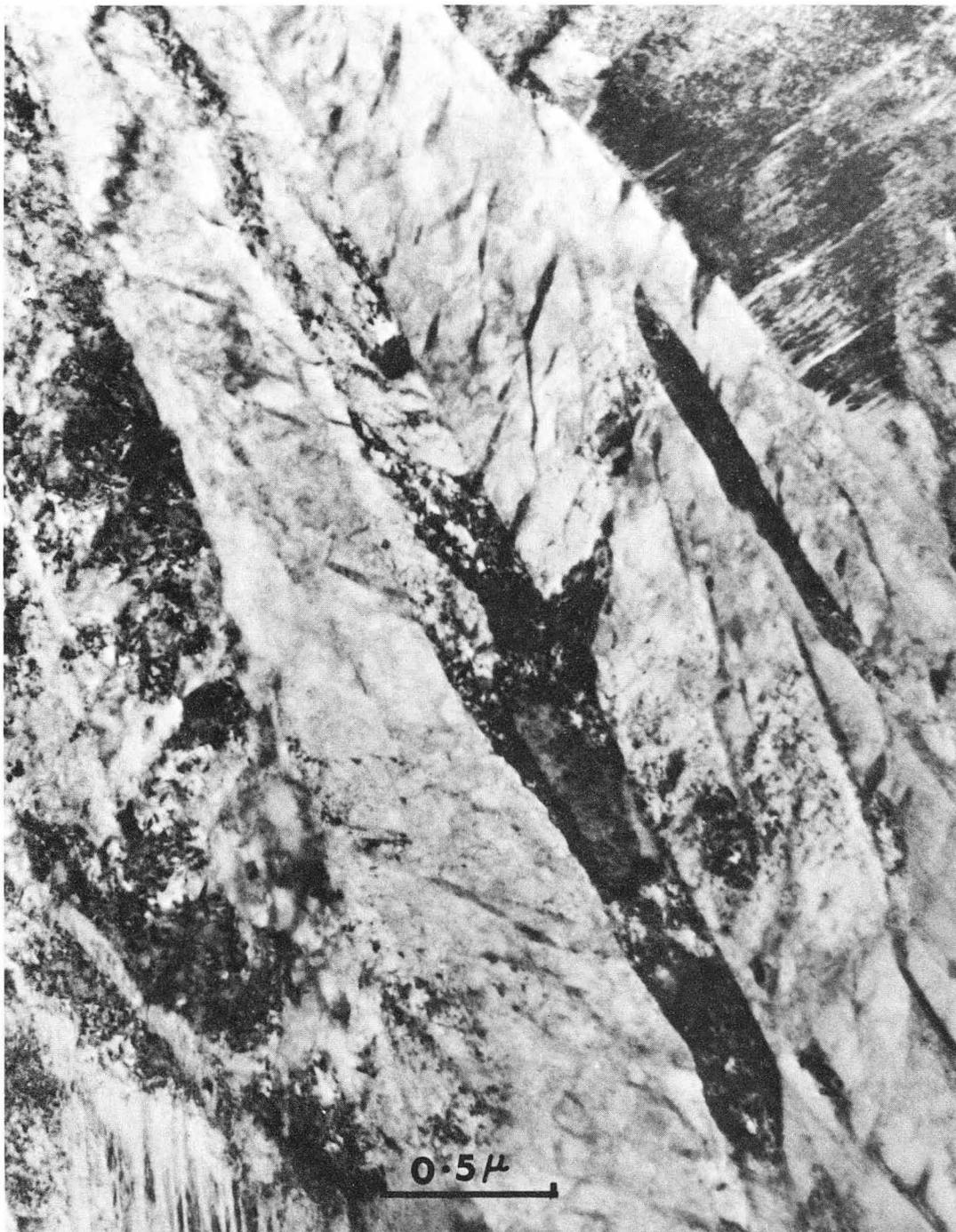
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Fig. 11



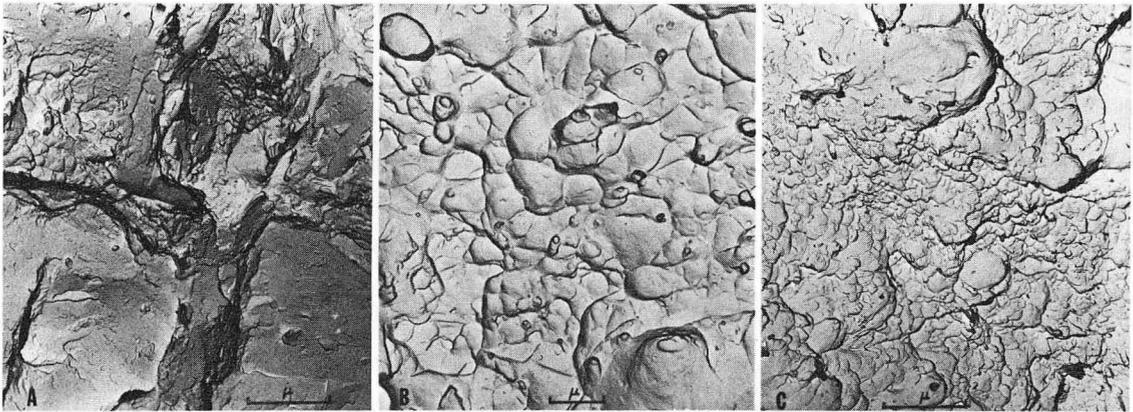
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Fig. 12



XBB 692-1543

Fig. 13



XBB 693-1974

Fig. 14

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