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SUBSTRUCTURES IN PLASTICALLY DEFORMED COPPER

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Om Johari and G. Thomas

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Abstract

Transmission electron microscopy techniques were used to study the substructures in single and polycrystals of pure copper deformed under different conditions of static and shock loading. Static loading involved 20% tensile deformation of single and polycrystals at the temperature 77°K, 196°K and 273°K and strain rates of 1 and 10^{-3} per minute. The shock loading was carried out on thin polycrystalline sheets. The pressures ranged from 16 to 200 kbars.

In tensile deformed specimens no systematic correlation between observed substructure and deformation conditions was found. The substructures varied widely in the same specimen from one region to another. It does not seem possible to correlate changes in cell size to various deformation conditions for two reasons: first, the cells are not always observed, and secondly, there is a wide change in cell size in the areas which show cells. From single crystal observations it was found that at least two slip systems should operate for formation of cells and that the dislocation density is much higher in polycrystals than in single crystals.

In shock loaded specimens the cells were uniform and the cell size decreased with increasing pressure until microtwinning occured. The critical cell size is about 0.15 μ. It is shown that the present theories are insufficient to specify the exact quantitative conditions for the onset of twinning.

I. INTRODUCTION

Transmission electron microscopy has been used extensively to study the substructures resulting from plastic deformation in various face-centered and body-centered cubic metals. Reviews of the various dislocation structures which have been observed have been presented by Swann[1] and Keh and Weissmann.[2] In these reviews various characteristics of cell structure and the factors affecting their size are reported. Warrington[3] observed that the cell size depends on the temperature of deformation such that an inverse square root relationship between subgrain size and flow stress was obeyed. Cell structures are not observed in metals of low stacking fault energy. The complex dislocation arrangements existing in cell walls are often called tangles. However, tangles can also be present without a visible cell structure.

A great deal of transmission electron microscopy has been done on the deformation substructure of Cu single crystals. Howie[4] found that during the second stage of deformation, a non-uniform cell structure was obtained. He also found that the dislocation density within the cell was about five times smaller than that in the cell walls. Young[5] and Livingston,[6] carried out some dislocation density calculations using etch pit techniques for copper single crystals. According to Livingston,[6] after sufficient strain the dislocations are distributed in a rough cell structure. It may be mentioned, however, that the cell structure normally resolved by electron microscopy is not resolved by the etch pit method. The recent results of Basinski[7] on dislocation arrangements in copper crystals deformed at low temperatures throw more light on the crystallography of the deformation process.

Nolder and Thomas[8] studied the effects of shock wave propagation in nickel and observed that the cell size decreased with increasing shock pressure. After a critical cell size of ~ 0.15 μ was reached, the mode of deformation changed to twinning. This occurred at and above 350 kbar. It was suggested that the pressure required to initiate twinning is related to the amount of plastic work given prior to twinning. In turn, this energy is related to the stacking fault energy. The twin thickness was observed to be < 5000 Å and was not detected by previous replication methods, e.g., Glass et al.[9] did not find any twins in their investigations on explosively deformed copper single crystal using x-ray diffraction and microscopy. As far as twinning is concerned, Venables[10] has reported that a resolved shear stress of 15 kg/mm^2 (≈ 1.5 kbar) is required for twinning in tensile deformed copper single crystals. For example, in copper single crystals deformed in tension, a deformation of 75% or more is required at 4.2°K. Venables[10] reported that the width of the twins was less than 5000 Å, as was also found in explosively deformed nickel.[8]

No electron microscopy studies have been reported concerning the effect of grain size or rate of deformation on the substructure. It is from this point of view that the present work was undertaken. It was believed that a correlation between cell size and the variables strain rate, grain size, temperature, flow stress and energy of cold work might be found. For this work, two grain sizes--single crystal and polycrystals (70-100 grains/cm), and two strain rates--10^{-3} in./in./min. and 1 in./in./min., and three temperatures--77°K, 196°K and 273°K were selected. It was further believed that a relation between cell size and the onset of twinning in terms of temperature and strain required may be obtained. Following Nolder and

Thomas' [8] work and noting that a more uniform cell structure is obtained by explosive deformation, work was done on explosively deformed copper. It was expected that apart from correlating cell sizes this work would allow further study of strain rate effects and twinning.

II. EXPERIMENTAL PROCEDURES

A. Single Crystals and Polycrystals for Tensile deformation

For all tensile deformed specimens copper of 99.999% purity was used. The copper was supplied by United Mineral & Chemical Corporation of New York, U.S.A. The single crystals were prepared by the moving furnace method using Bridgman's technique. The metal, in the form of a sheet 5" x 5/8" x 1/32", was placed in a graphite mold.. The furnace was maintained under a vacuum of the order of 10^{-4} mm of mercury. The crystals produced were cut to size on a spark cutter so as to give a 2-inch gauge length for the tensile tests on the Instron machine. The width of the specimens was about 0.3". Six crystals were used in this work and Fig. 1 gives their orientation. These crystals, after spark cutting, were annealed in vacuum at cyclic temperatures between 750° - 1000°C for 24 hours. Table I describes the deformation conditions for each crystal.

TABLE I

Crystal No.	Temp. of deformation	Strain rate	Strain
1	196°K	10^{-3} per min.	20%
2	77°K	10^{-3} " "	"
3	273°K	10^{-3} " "	"
4	273°K	1 " "	"
5	196°K	1 " "	"
6	77°K	1 " "	"

For polycrystals a copper strip was rolled and annealed in steps until finally the thickness was about 0.025". This thickness was used to avoid any buckling which might occur during handling of very thin foils. Final annealing gave specimens of grain size 100-150 μ, six of which were used in this work.

B. Deformation

The tensile deformation of polycrystals and single crystals was carried out on an Instron machine. A gauge length of 2 inches was used for all deformations. The deformation was carried out at three temperatures: 273°K, 196°K and 77°K. Two strain rates were used: 1 and 10^{-3} inch per inch per minute. After deformation, the specimens were prepared for transmission electron microscopy, as described later.

Sandia Corporation[11] of New Mexico supplied the explosively deformed copper specimens. These specimens were ACS reagent grade copper. The thickness of the foils was 0.002", and these were deformed at 16, 28, 100, 150, and 200 kbars pressure. The shock loading was carried out in assembly shown in Fig. 2.

C. Preparation of Foils

The tensile deformed specimens were first chemically polished in a solution containing 50% nitric acid, 25% acetic acid, and 25% orthophosphoric acid. When the specimens were thin enough they were electropolished in an electrolyte of 33% nitric acid and 67% methyl alcohol at -30°C. As the explosively deformed foils were only 0.5 μ in diameter, they were mounted on a copper sheet with a circular hole in the center and then similarly polished to avoid the buckling which would otherwise occur. The foils were examined in a Siemens Elmiskop I-b electron microscope operated at 100 kV.

III. RESULTS

A. Tensile Specimens

The transmission electron micrographs of tensile specimens showed various types of dislocation arrangements. Although it was expected that some kind of cell structure would be obtained in all specimens, it was found that such was not always the case. Figure 3 shows some of the micrographs from the polycrystals deformed at different conditions of temperature and strain rate. From areas such as those shown in (a) and (b) for two strain rates at 273°K it was possible to estimate the cell size. Some such areas were observed in specimens deformed at lower temperature also. Figure 3-c and -d are areas showing a very high dislocation density and extremely small cells from specimens deformed at 196°K. The substructures from specimens deformed at 77°K are shown in Fig. 3-e and -f. Apart from the areas shown in these figures other micrographs showed a wide range of structures between the extremes of areas showing cells and areas showing only dislocation tangles. From these observations it is clear that the cell sizes are not uniform and that the dislocation arrangements vary widely from grain to grain in the same specimen. No crystallographic characteristics could be assigned to these dislocation structures.

Figure 4 shows typical micrographs of the effects of grain boundaries. It is seen that although the cells are observed at the areas near the grain boundaries, no obvious dislocation sources were observed at the grain boundary. However, there appears to be a high concentration of dislocations along the boundary.

Widely varying orientations of single crystals were examined to investigate any orientation dependences of cell formation. The electron micrographs from single crystals showed such wide variations in

substructure that no quantitative comparison as to the effect of temperature or strain rate could be made. However, they show clearly that the dislocation density for the same strain for single crystal specimens is much less than that for polycrystalline specimens, e.g., compare Figs. 3 and 5. Cell structures are observed only when two independent slip systems operate, (e.g., Figs. 5a and d). It is also interesting to note here that within the range of temperature and strain rate chosen, no systematic variations were observed in substructure (e.g., compare Figs. 6a, b and c from the same specimen). Crystallographic directions are marked on the micrographs and it is seen that the dislocations tend to lie very close to the active glide planes. The operation of second glide planes (duplex slip) gives structures similar to those reported by Basinski.[7]

Summary of observations on tensile deformed specimens:

Although it is difficult to obtain quantitative correlations between substructure and testing conditions due to the wide range of substructures within each specimen, the following qualitative conclusions are apparent.

(1) Cell formation is favored when two or more independent slip systems operate.

(2) Cell size decreases for increasing strain rate.

(3) Cell size decreases for decreasing temperature of testing.

(4) The dislocation density is a function of grain size, being greater in polycrystals than in single crystals. This effect can be understood if boundary sources are operating during plastic flow.

(5) In the case of single crystals, the dislocations tend to lie very close to the glide planes. This is not usually true for polycrystals.

B. Explosively Deformed Specimens

Transmission electron micrographs of specimens after 16 kbar pressure revealed a structure of densely tangled dislocations primarily arranged in cell walls enclosing relatively small dislocation-free areas. The cell size is quite uniform and the average cell diameter for this specimen was found to be 0.23 μ (range 0.15-0.4 μ) and an example of the structure is shown in Fig. 7a. The 28 kbar specimen showed interesting effects. In the same specimen, areas with twins (Fig. 8) and areas with cell structure (Fig. 7b) are observed. The cell size could not be measured very accurately, but from the few cells which were clear enough to be measured, a cell diameter of 0.12 μ (in the range 0.06 to 0.2 μ) has been found. From the diffraction pattern in Fig. 8b from the twinned region shown in Fig. 8a, it is seen that no extra spots arising from twins are present but, as is shown in Figs. 9a and 10, the twin spots of a (001) orientation actually fall on the matrix spots. Hence, the structure consists of twins--a fact which was also confirmed by dark field work.

The results obtained on specimens after deformation at 100, 150 and 200 kbar are strikingly similar to those reported by Nolder and Thomas.[8] In these specimens, apart from twinned areas, there were regions which contained a very high density of dislocations. The twinning planes in each of the micrographs are {111}, as confirmed by trace analyses and by contrast-reversal using the dark field technique. It was thus possible to identify the spots arising from twinned areas. Figure 11a shows twinning in the 150 kbar specimen. Figure 11b is its diffraction pattern. The foil has an orientation of $(5\bar{1}0)$ and the orientation of the twins T_1 and T_2 are respectively (134) and $(13\bar{4})$. This is consistent with $(1\bar{1}\bar{1})$ are $(1\bar{1}1)$ twinning planes as shown by the trace analysis in Fig. 12 and contrast reversals

by dark field work in Figs. 11-c and -d. From the trace analyses (Fig. 10 and 12), it should be noted that there is a possibility of spots from matrix and twin overlapping. This is shown in Fig. 9 by two expected diffraction patterns after twinning on suitable (111) planes for (001) and (110) orientations.

Figure 13a is the electron micrograph from a 200 kbar specimen. The diffraction pattern shown is similar to that in Fig. 8b. The areas at the twins have been preferentially attacked giving rise to thickness contours. Since the twins are very thin, streaks are produced along <111>, the projection of which in the (001) orientation gives rise to the appearance of streaks along <110> directions (which are also the traces of twin planes (Fig. 10)).

IV. DISCUSSION

A. Cell Size After Tensile Deformation

From the accompanying micrographs it is quite clear that a statistical estimate of cell size cannot easily be made. There are some regions in every specimen which show cell-free substructure. However, from the various micrographs, a statistical analysis of cell size has been made for polycrystals from many micrographs and is presented in Table II. It is interesting to note that the cell sizes lie in the range 0.4-0.65 μ for all the polycrystalline specimens. The results of this work are not in agreement with those of Warrington.[3] Variation in cell size for a given condition is of the order of the total change so that no quantitative correlation between cell size and temperature, grain size, strain rate, energy of cold work, or flow stress could be established. However, if Warrington's[3]

data on temperature dependence of cell size are extrapolated, it is found that for 20% deformation even the lowest temperature experimentally obtainable will give a cell size of 0.4 μ which is 3 times the critical cell size for the onset of twinning (≈0.15 μ). This means that 20% tensile deformation in copper will not produce twinning, even at liquid helium temperature, a fact confirmed by this work and in agreement with Venables'[10] results.

Average Cell Size for Polycrystalline Specimens

Temp. of deformation	Strain rate	Average cell size in μ
273°K	1 per min.	0.59
273°K	10^{-3} " "	0.55
196°K	1 " "	0.45
196°K	10^{-3} " "	0.55
77°K	1 " "	0.41
77°K	10^{-3} " "	0.63

The single crystal micrographs do not show any systematic variations in substructure with change in strain rate or temperature of deformation. The type of structures obtained are similar to those reported, for example, by Basinski.[7] It is seen from Figs. 5-a and -d that dislocations lie on two operating glide planes. Some cross slip has also occurred and the tangles which might lead to formation of cell walls are beginning to appear. In Figs. 5-b and -c there is only one slip system operating and dislocations lie parallel to the one glide plane. These observations may indicate what

happens in those polycrystals in which there is a high dislocation density in tangles. It is quite probable that, e.g., in Fig. 3f, the cell wall is in the plane of the foil, and hence does not give a clear picture of the cell structure. This also explains that the phenomenon of cell formation within the specimen is so random that no quantitative relationships can be derived. Depending on the size of a cell and the orientation in which it is viewed, various substructures could be obtained. It is difficult to pin-point these effects more accurately because the cell walls do not lie along well-defined planes. This is possibly why a great scatter in substructures in polycrystals is obtained, and consequently, the effects of various deformation parameters to substructure cannot be correlated quantitatively without ignoring some of the observed features.

B. Explosively Deformed Copper

From Fig. 7 it is seen that the cell size decreases with increasing pressure, and after a critical cell size is reached the mode of deformation changes to twinning. The relation between critical cell size and the pressure is quite different for Cu and for Ni,[8] e.g., in the case of nickel twinning was observed at a pressure of about 350 kbars and 0.15 μ cell size,[8] while in the case of copper, twinning was observed at a pressure of 28 kbar and nearly 0.15 μ cell size. This large difference in pressures causing twins cannot be due to stacking fault energy only, because the stacking fault energy of Ni[12] is ~80-100 ergs/cm^2 and of Cu is ~60 ergs/cm^2. (Various values of stacking fault energy of copper have been reported and they range from 40[13] to 80[10] ergs per sq. cm. An average value of 60 was therefore selected for this work.) However, there were some differences between the conditions under which experiments were done on nickel[8] and copper.

The specimen size and shape were also different (for nickel-3-inch diameter slabs, 1/4 inch thick; for copper-0.5 inch diameter, 2 mil. thick sheets). The method used to calculate the pressure values in two cases were also different. Thus, no quantitative comparison of the two data can be made without error. However, explosive deformation at higher pressures does give rise to twinning in both these metals after a critical cell size is reached. From the results so far obtained it is not possible to conclude that this critical cell size is constant for various metals, or that this depends on stacking fault energy and other properties of the metal.

At present it is not yet fully known whether the criterion for twinning is a critical shear stress or a critical total plastic work. Venables[10] has given a detailed treatment of the critical shear stress required for twinning. It is generally agreed that the stress must take into consideration a local stress concentration factor. From our experiments we have only two values, the pressure at which twinning was observed and the critical cell size prior to the onset of twinning. Since the mode of deformation changes from slip to twinning after the critical cell size is reached, it should be possible to equate the critical stress expression to give a required cell size, with the critical stress expression for the onset of twinning as given by Venables.[10]

The physical model of the change from slip to twinning is considered to be the following. With increasing plastic deformation the dislocation density increases to some value such that a critical length of pinned dislocation $\bar{\ell}$ is attained (e.g., in cell walls). The stress[14] to bow out such a dislocation is

$$\tau \cong Gb/\bar{\ell} \quad . \qquad (1)$$

If this stress exceeds the stress required to operate a twinning dislocation of length $\bar{\ell}$, then twinning will be the preferred mode of deformation for cases where the critical cell size is reached. At room temperature both in Cu and Ni this is only accomplished by explosive deformation. The critical shear stress expression for twinning due to Venables,[10] is

$$n\tau_T = \frac{S}{b_1} + \frac{Gb_1}{2a_o} \qquad (2)$$

where: a_o is the radius of a source acting as a twin nucleus

n is a stress concentration factor

S is the stacking fault energy of metal

G is the rigidity modulus of metal

b and b_1 are the Burger vectors of unit and Shockley partial dislocations respectively, i.e., $b = \frac{a}{\sqrt{2}}$ and $b_1 = \frac{a}{\sqrt{6}}$

$\bar{\ell}$ is the average distance between dislocations. We assume $\bar{\ell}$ is equal to $2a_o$.

Combining (1) and (2) and taking n = 1 and using the following values:

$$S = 60 \text{ ergs/cm}^2$$

$$G = 4.5 \times 10^{11} \text{ dynes/cm}^2$$

$$b^2 = \frac{a^2}{2}$$

$$b_1^2 = \frac{a^2}{6}$$

and $a = 3.615 \times 10^{-8}$ cm.

One gets $\bar{\ell} = 0.012\ \mu$ and stress equal to 9.5 kbar. This value of $\bar{\ell}$ is nearly $\frac{1}{12}$th the value of the critical cell size. The value of this shear stress is nearly $\frac{1}{3}$rd of the pressure at which the onset of twinning was observed. A similar calculation for Ni shows that the critical value of $\bar{\ell}$ is $0.013\ \mu$ (almost identical to that for copper) and critical shear stress is 13.5 kbar, which is different by a factor of 25 from the observed

value for the onset of twinning.[8] From the calculation of stress values, it appears that the agreement with actual values is not good. However, from this, it cannot be concluded that stress is not the critical factor. In the derivation of the formulae it has been assumed that the length $\bar{\ell}$ of the bowing out dislocation is equal to the length of the leading partial involved in twinning. These assumptions are not exactly true, but $\bar{\ell}$ and $2a_o$ are not expected to be very much different. The pressure values used experimentally do not give the value of the shear stress actually acting for the deformation process. Since the specimen is not subjected to changes in dimensions in the two directions normal to the wave front, some other constraints also come into play. This would mean that the shear stress acting is much different from the applied hydrostatic pressure.

Davis and Jackson[15] have proposed that in the case of shock loading the shear stress acting is about 0.25 of the normal stress. If this is assumed to be so, the results for Cu are in very good agreement with the critical shear stress model[10]. The difference for Ni may be due to the various factors mentioned earlier. This presents another interesting possibility. If the actual value of the critical stress for the onset of twinning were known, a method may be obtained to find a correlation between the applied hydrostatic pressure and acting shear stress.

Nolder and Thomas[8] found an empirical relation between the amount of work done during the deformation and the onset of twinning assuming the model described above. From their results, obtained by combining the expression due to Venables[16] and Kuhlmann-Wilsdorf[14], the criterion for twinning is given by

$$s < \frac{G}{2\,h\bar{\ell}}\,(b^2 - b_1^{\,2}) \qquad (3)$$

where h is an adjustable parameter of the order of unity. For copper this gives a value of $\bar{\ell}$ to be 0.033 μ. This is related to critical cell size by a factor of nearly $\frac{1}{4}$th. This is in agreement with the value obtained by Nolder and Thomas[8]. The corresponding energy values can be obtained by using this value of $\bar{\ell}$ and the expression for plastic work obtained by Wilsdorf's work hardening theory[14] giving $E = \frac{K \cdot G \cdot b^2}{2\bar{\ell}^2}$ where K is a constant nearly equal to 300. The energy value comes out to be 4.5×10^9 ergs/cm^2.

In order to obtain the energy input required for plastic deformation due to shock loading, the pressure strain curve derived from the Hugoniot curve[17] is assumed. From this the value of energy can be obtained following the method outlined previously[8]. The pressure value corresponding to an energy of 4.5×10^9 ergs/cm^2 is 130 kbar. Though this value is nearly 5 times the actual pressure value, in view of the many assumptions involved in the calculation, the apparent dependence of twinning on total plastic work is not proved. In view of the survey of various ideas supporting the critical stress criterion by Reid et al.[18] and the empirical nature of the total plastic work model, it seems that more work must be done to further solve the problem. To this end more experiments are being carried out on explosively deformed Cu-alloys of known stacking fault energies.

Finally, some comments can be made regarding the micrographs. The cell size in Fig. 7a (16 kbar) is very uniform. It appears that, as the strain rate increases, the cell size drops and so the scatter in cell size is less. In all the twinning micrographs, there were only one or two active twinning planes and the other two did not operate at all. It was also seen that the dislocation density was much smaller near the twinned regions as

compared to those where twinning had not occurred. Since the specimen was polycrystalline, it may be that some grains were orientated more favorably for twinning. In the 200 kbar specimen there were fewer areas free from twins than in specimens deformed at lower pressures. The actual width of the twins is smaller as compared to the projected width on the micrographs. The details within the twins were not clearly resolved in all specimens, but from Fig. 13a it is seen that some twins do show fringe contrast and are probably coherent. It also seems that very few twins cross each other; most of them butt against each other where the twins begin to occur on the other plane. Presumably immobile stair rods are formed at the intersection preventing continued propagation of the intersecting twin.

V. CONCLUSIONS

1. The dislocation substructures formed in copper polycrystals varies widely depending on orientation and the previous thermal and mechanical history of the specimen. The same specimen shows so many different structures that no quantitative relationships regarding cell size and temperature, stress, or strain rate could be established.

2. Cells do not form unless at least two independent slip systems operate. For the same conditions the dislocation density is much higher in polycrystals than in single crystals.

3. In explosive deformation, a uniform cell structure is formed unless the pressure is large enough to cause microtwinning.

4. A microtwinning structure has been observed in explosively deformed copper shock loaded to 28 kbar or greater. It was found that twins always occur on one or, at the most, two of the {111} planes after explosive deformation.

5. At present, the correlation between conditions for onset of twinning and critical stress or critical amount of total plastic work cannot be exactly determined because of unknown conditions in the shock wave front. However, a critical cell size of the order of 0.15 μ is attained prior to twinning both in copper and nickel.

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Figure Captions

Fig. 1. Orientation of single crystal specimens.

Fig. 2. Assembly for shock loading (Courtesy Dr. D. Munson and Sandia Corporation) All measurements are in inches.

Fig. 3. Transmission Electron Micrographs from 20%.tensile deformed polycrystals at

(a) 273°K and 1 per min. strain rate
(b) 273°K and 10^{-3} per min. strain rate
(c) 196°K and 1 per min. strain rate
(d) 196°K and 10^{-3} per min. strain rate
(e) 77°K and 1 per min. strain rate
(f) 77°K and 10^{-3} per min. strain rate.

Fig. 4. Transmission electron micrographs showing grain boundaries in

(a) Same specimen as in Fig. 3a
(b) Same specimen as in Fig. 3d.

Fig. 5. Electron micrograph showing dislocation structure in single crystals after 20% deformation at

(a) 273°K and 1 per min. strain rate
(b) 273°K and 10^{-3} per min. strain rate
(c) 196°K and 1 per min. strain rate
(d) 77°K and 10^{-3} per min. strain rate.

Fig. 6. Electron micrographs from specimen 20% deformed at 77°K and 1 per min. strain rate. The dislocation density varies from (a) to (c).

Fig. 7. (a) Electron micrograph from 16 kbar explosively deformed specimens showing uniform cell structure.

(b) Electron micrograph showing dense dislocations in 28 kbar explosively deformed specimens.

Fig. 8. (a) Electron micrograph showing twins as observed in 28 kbar specimen. The twins lie on traces of {111} planes along <110> directions. The foil orientation is (001).

(b) Diffraction pattern from micrograph 8a, shows no extra spots due to twinning.

Fig. 9. (a) Expected diffraction pattern by twinning on (111) planes for (001) orientation.

(b) Expected diffraction pattern by twinning for (110) orientation.

For both cases no extra spots appear and the spots from twinned material overlap those from matrix.

Fig. 10. Stereogram showing movement of poles by twinning on two {111} planes for crystals in (001) orientation.

Fig. 11. (a) Electron micrograph from 150 kbar explosively deformed specimen. The micrograph shows twins and dense dislocations lying away from twins.

(b) Diffraction pattern of micrograph (a). The foil orientation is $(5\bar{1}0)$.

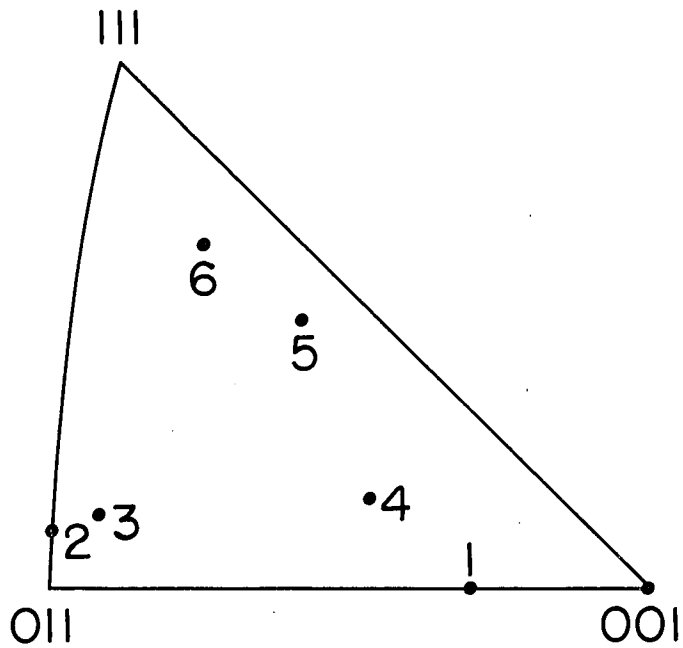
(c) Dark field micrograph from Spot A.

(d) Dark field micrograph from Spot B. Dark field at C gives no change.

Fig. 12. Stereographic analysis of twinning in $(5\bar{1}0)$ orientation.

Fig. 13. (a) Electron micrograph showing twin structure in explosively deformed specimen at 200 kbar pressure. Note the fringe contrast at twin interfaces, the high dislocation density in areas away from twins and preferential etching of twinned material. The foil orientation is (001) and the twins are on {111} planes and appear as traces in <110> directions.

(b) Diffraction pattern corresponding to above figure. (Compare with Fig. 8a).



MU-31429

Fig. 1

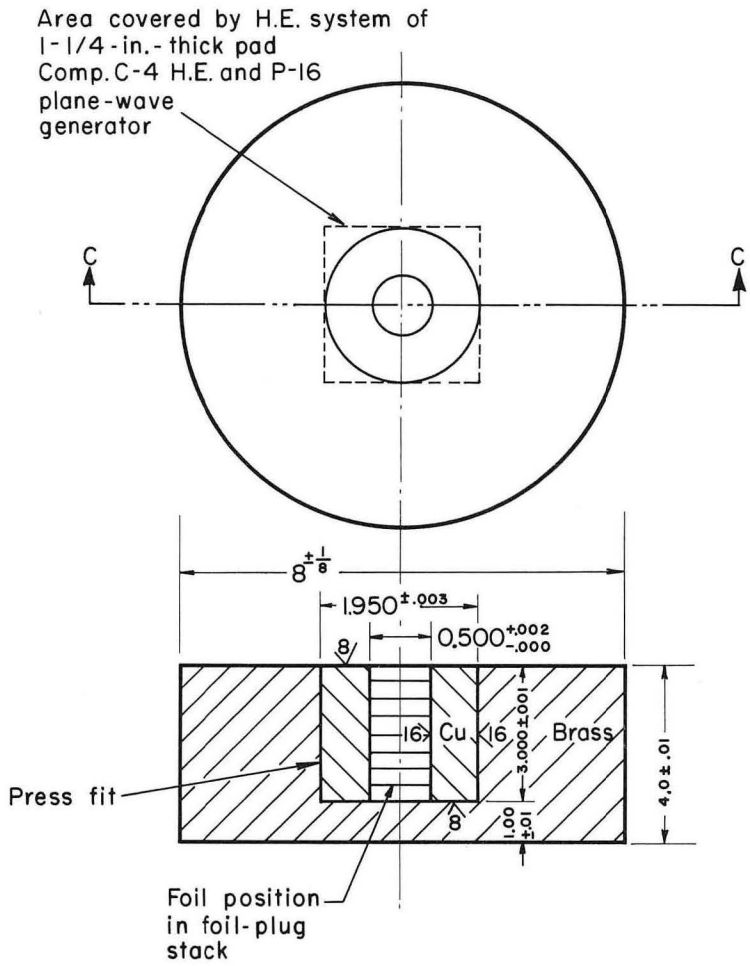
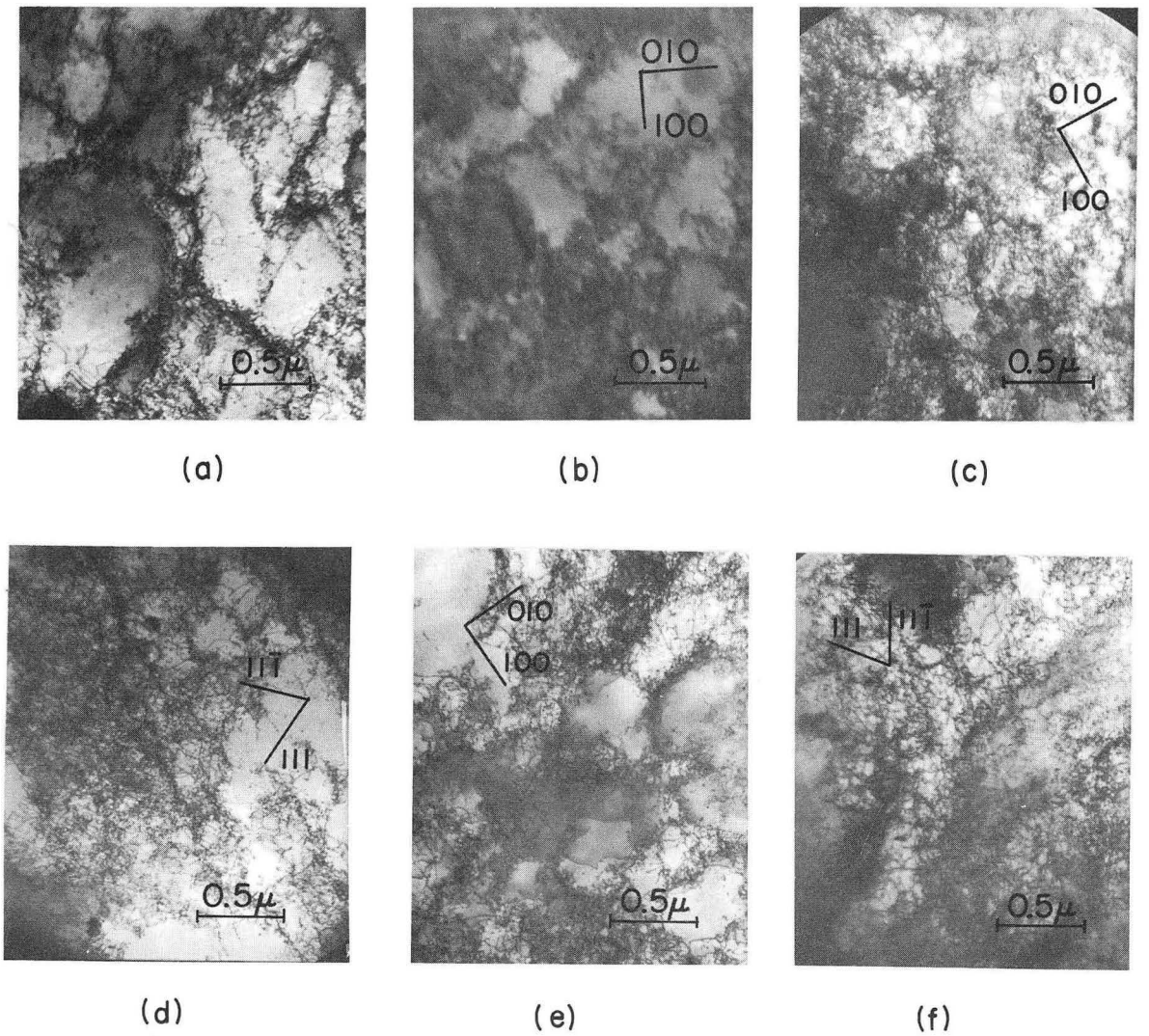
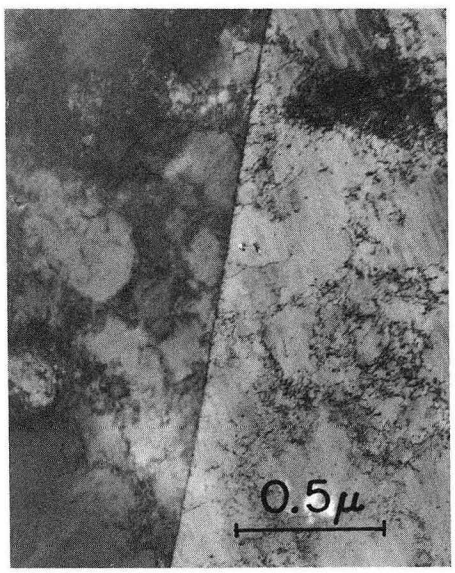

Area covered by H.E. system of
1-1/4-in.-thick pad
Comp. C-4 H.E. and P-16
plane-wave
generator
C
C
8 ± 1/8
1.950 ± .003
0.500 +.002 -.000
8
16
Cu
16
Brass
3.000 ± .001
4.0 ± .01
Press fit
8
1.00 ± .01
Foil position
in foil-plug
stack
MU-31430

Fig. 2

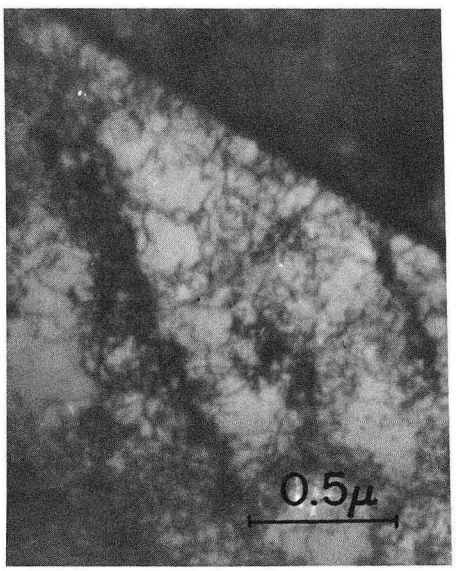

0.5μ
(a)
010
100
0.5μ
(b)
010
100
0.5μ
(c)
111̄
111
0.5μ
(d)
010
100
0.5μ
(e)
111
111̄
0.5μ
(f)

ZN-4034

Fig. 3



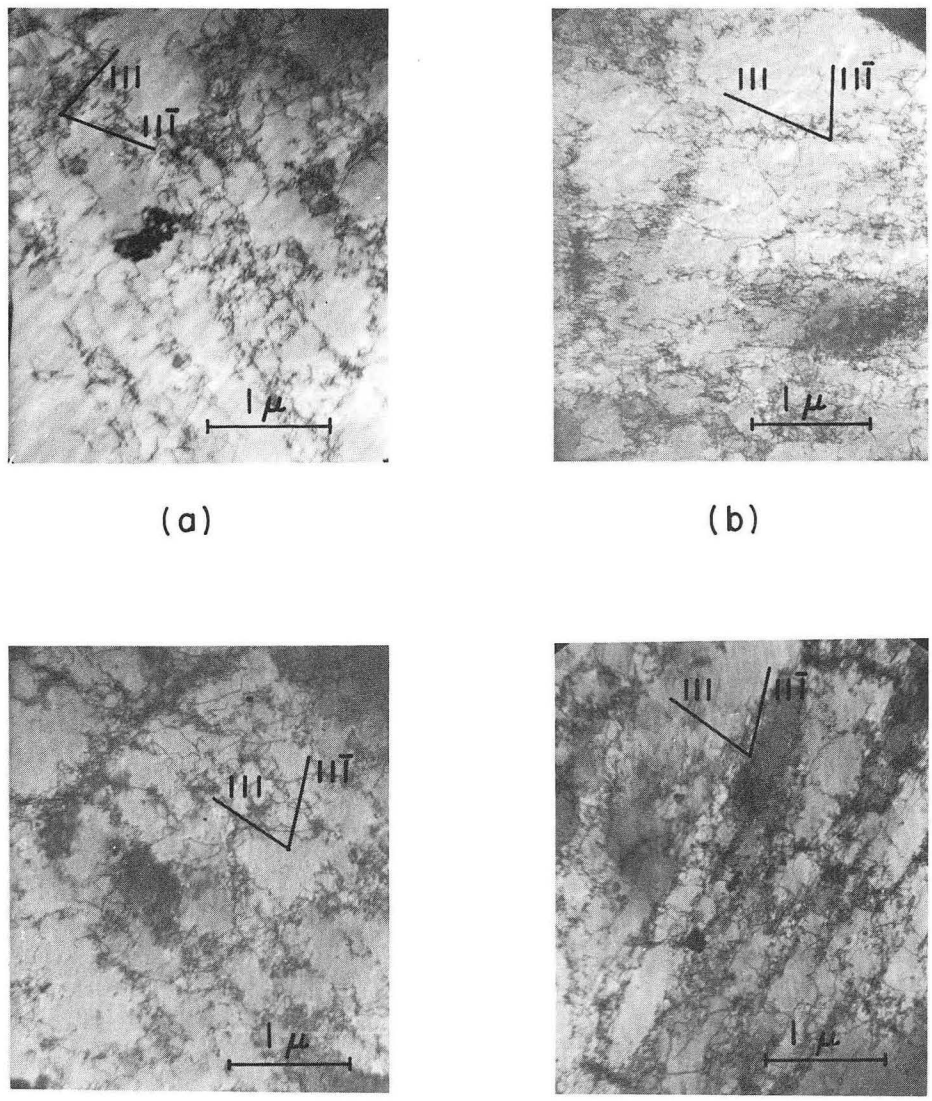
(a)



(b)

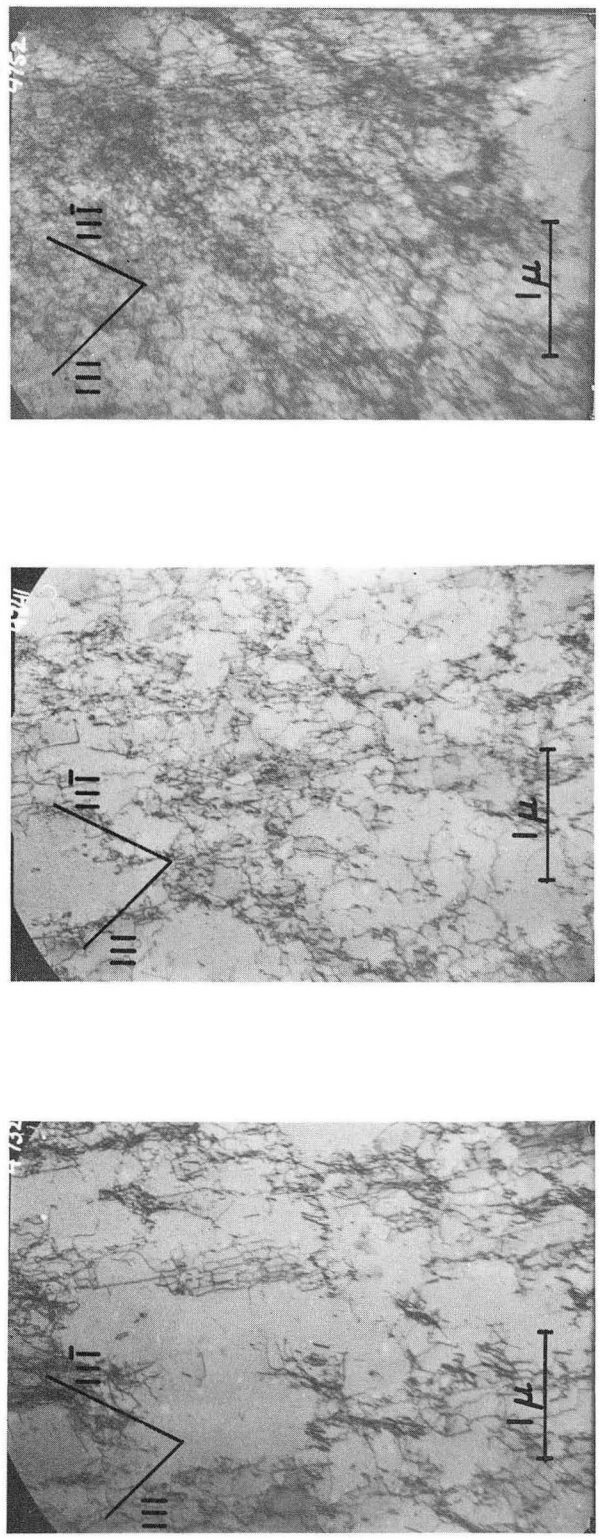
ZN-4035

Fig. 4



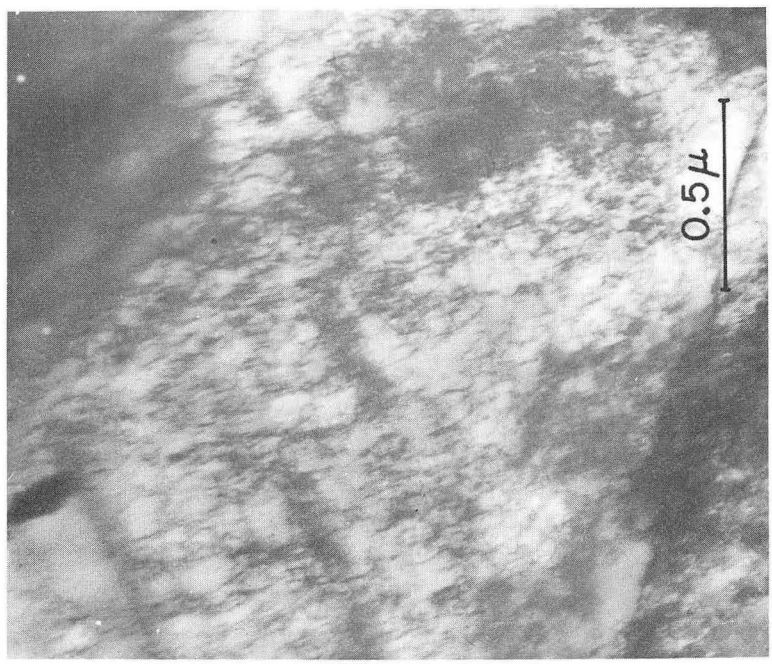
ZN-4036

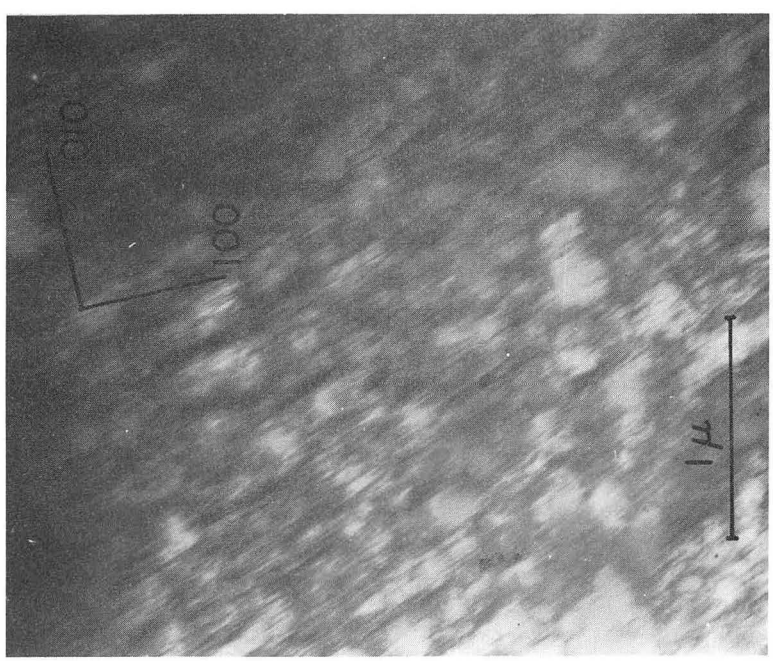
Fig. 5



ZN-3917

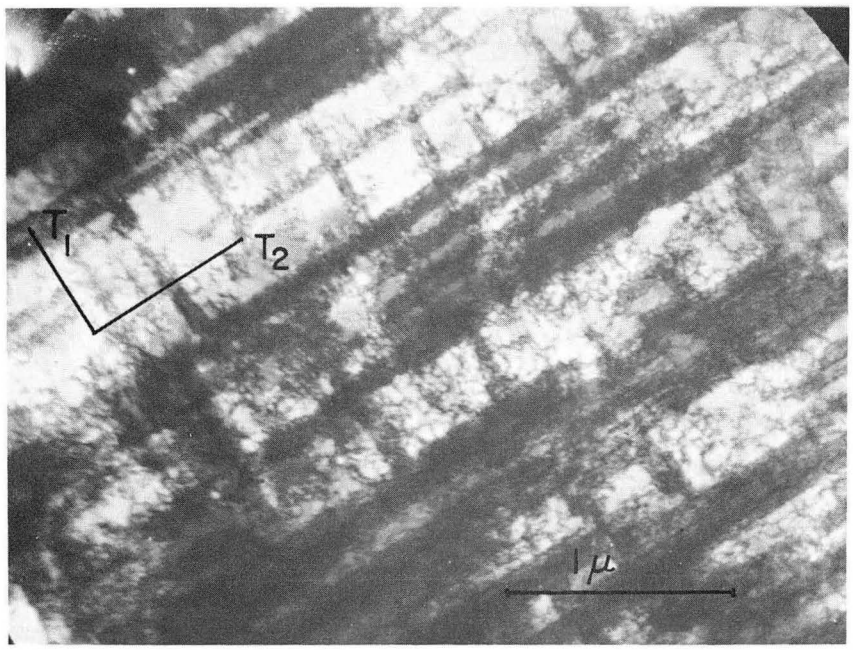
Fig. 6



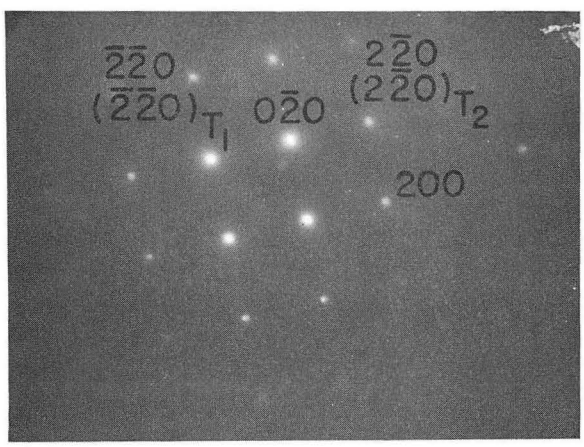


ZN-3919

Fig. 7



(a)



(b)

ZN-3920

Fig. 8

$\dot{6}\bar{6}0$ $(228)_{T_1}$ $(\bar{6}\bar{6}0)_{T_2}$	$\dot{6}\bar{4}0$	$\dot{6}\bar{2}0$	$\dot{6}00$ $(\bar{2}44)_{T_1}$ $(\bar{2}\bar{4}\bar{4})_{T_2}$	$\dot{6}20$	$\dot{6}40$	$\dot{6}60$ $(\bar{6}\bar{6}0)_{T_1}$ $(2\bar{2}\bar{8})_{T_2}$
$\dot{4}\bar{6}0$	$\dot{4}\bar{4}0$ $(\bar{4}\bar{4}0)_{T_2}$	$\dot{4}\bar{2}0$ $(024)_{T_1}$	$\dot{4}00$	$\dot{4}20$ $(0\bar{2}\bar{4})_{T_2}$	$\dot{4}40$ $(\bar{4}40)_{T_1}$	$\dot{4}60$
$\dot{2}\bar{6}0$	$\dot{2}\bar{4}0$ $(204)_{T_1}$	$\dot{2}\bar{2}0$ $(\bar{2}\bar{2}0)_{T_2}$	$\dot{2}00$	$\dot{2}20$ $(\bar{2}20)_{T_1}$	$\dot{2}40$ $(20\bar{4})_{T_2}$	$\dot{2}60$
$\dot{0}\bar{6}0$ $(4\bar{2}4)_{T_1}$ $(\bar{4}\bar{2}4)_{T_2}$	$\dot{0}\bar{4}0$	$\dot{0}\bar{2}0$	$\dot{0}$	$\dot{0}20$	$\dot{0}40$	$\dot{0}60$ $(\bar{4}2\bar{4})_{T_1}$ $(42\bar{4})_{T_2}$
$\dot{2}\bar{6}0$	$\dot{2}\bar{4}0$ $(\bar{2}04)_{T_2}$	$\dot{2}\bar{2}0$ $(2\bar{2}0)_{T_1}$	$\dot{2}00$	$\dot{2}20$ $(220)_{T_2}$	$\dot{2}40$ $(\bar{2}0\bar{4})T_1$	$\dot{2}60$
$\dot{4}\bar{6}0$	$\dot{4}\bar{4}0$ $(4\bar{4}0)_{T_1}$	$\dot{4}\bar{2}0$ $(024)_{T_2}$	$\dot{4}00$	$\dot{4}20$ $(0\bar{2}\bar{4})_{T_1}$	$\dot{4}40$ $(440)_{T_2}$	$\dot{4}60$
$\dot{6}\bar{6}0$ $(6\bar{6}0)_{T_1}$ $(\bar{2}28)_{T_2}$	$\dot{6}\bar{4}0$	$\dot{6}\bar{2}0$	$\dot{6}00$ $(2\bar{4}\bar{4})_{T_1}$ $(244)_{T_2}$	$\dot{6}20$	$\dot{6}40$	$\dot{6}60$ $(\bar{2}\bar{2}\bar{8})_{T_1}$ $(660)_{T_2}$

MU-31440

Fig. 9a

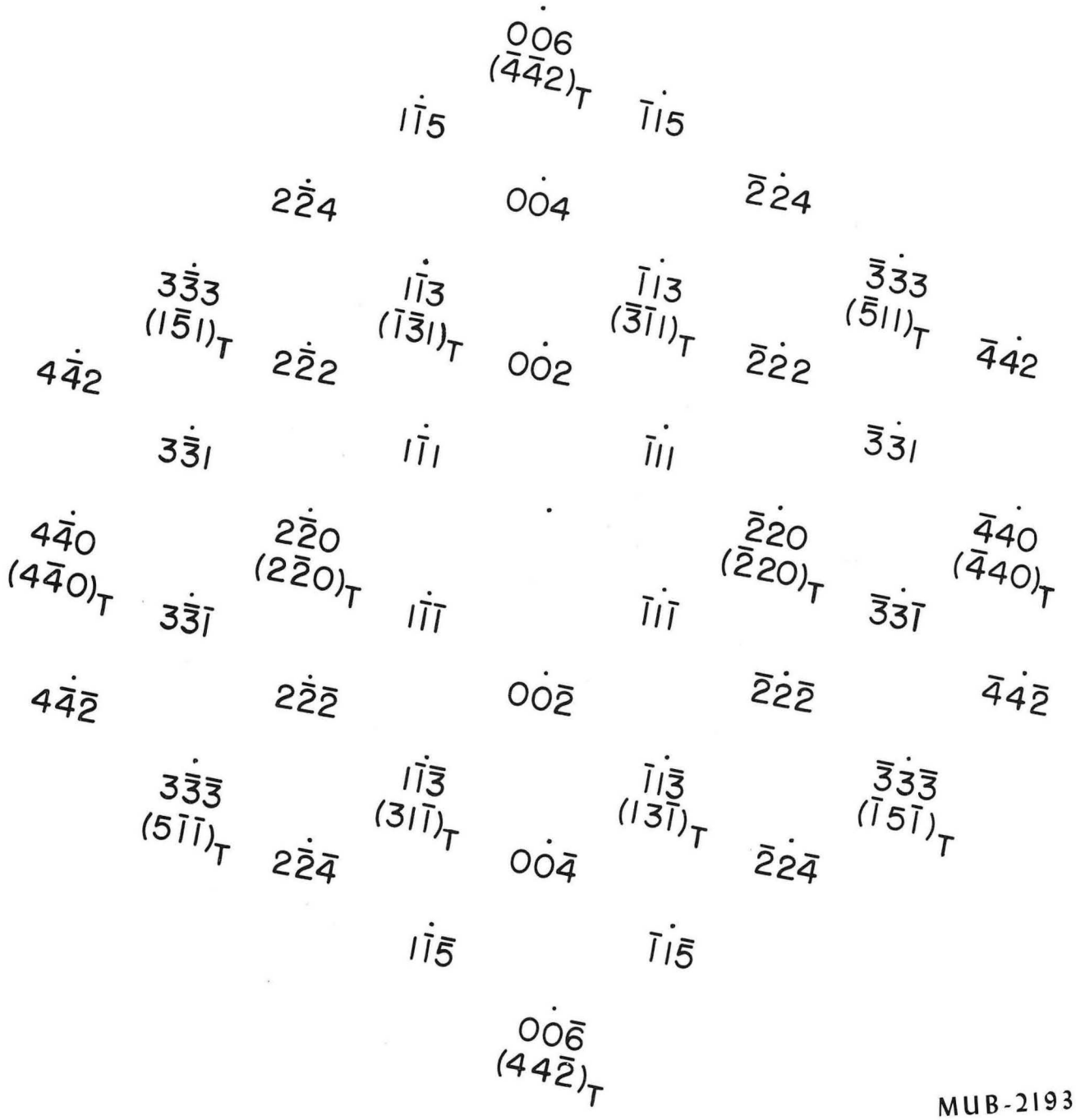

$00\dot{6}$
$(\bar{4}\bar{4}2)_T$
$1\dot{\bar{1}}5$
$\dot{\bar{1}}15$
$2\dot{\bar{2}}4$
$00\dot{4}$
$\dot{\bar{2}}24$
$3\dot{\bar{3}}3$
$(1\bar{5}1)_T$
$1\dot{\bar{1}}3$
$(\bar{1}\bar{3}1)_T$
$\dot{\bar{1}}13$
$(\bar{3}\bar{1}1)_T$
$\dot{\bar{3}}33$
$(\bar{5}11)_T$
$4\dot{\bar{4}}2$
$2\dot{\bar{2}}2$
$00\dot{2}$
$\dot{\bar{2}}22$
$\bar{4}\dot{4}2$
$3\dot{\bar{3}}1$
$1\dot{\bar{1}}1$
$\dot{\bar{1}}11$
$\dot{\bar{3}}31$
$4\dot{\bar{4}}0$
$(4\bar{4}0)_T$
$2\dot{\bar{2}}0$
$(2\bar{2}0)_T$
$\dot{\bar{2}}20$
$(\bar{2}20)_T$
$\bar{4}\dot{4}0$
$(\bar{4}40)_T$
$3\dot{\bar{3}}\bar{1}$
$1\dot{\bar{1}}\bar{1}$
$\dot{\bar{1}}1\bar{1}$
$\dot{\bar{3}}3\bar{1}$
$4\dot{\bar{4}}\bar{2}$
$2\dot{\bar{2}}\bar{2}$
$00\dot{\bar{2}}$
$\dot{\bar{2}}2\bar{2}$
$\bar{4}\dot{4}\bar{2}$
$3\dot{\bar{3}}\bar{3}$
$(5\bar{1}\bar{1})_T$
$1\dot{\bar{1}}\bar{3}$
$(31\bar{1})_T$
$\dot{\bar{1}}1\bar{3}$
$(13\bar{1})_T$
$\dot{\bar{3}}3\bar{3}$
$(\bar{1}5\bar{1})_T$
$2\dot{\bar{2}}\bar{4}$
$00\dot{\bar{4}}$
$\dot{\bar{2}}2\bar{4}$
$1\dot{\bar{1}}\bar{5}$
$\dot{\bar{1}}1\bar{5}$
$00\dot{\bar{6}}$
$(44\bar{2})_T$
MUB-2193

Fig. 9b

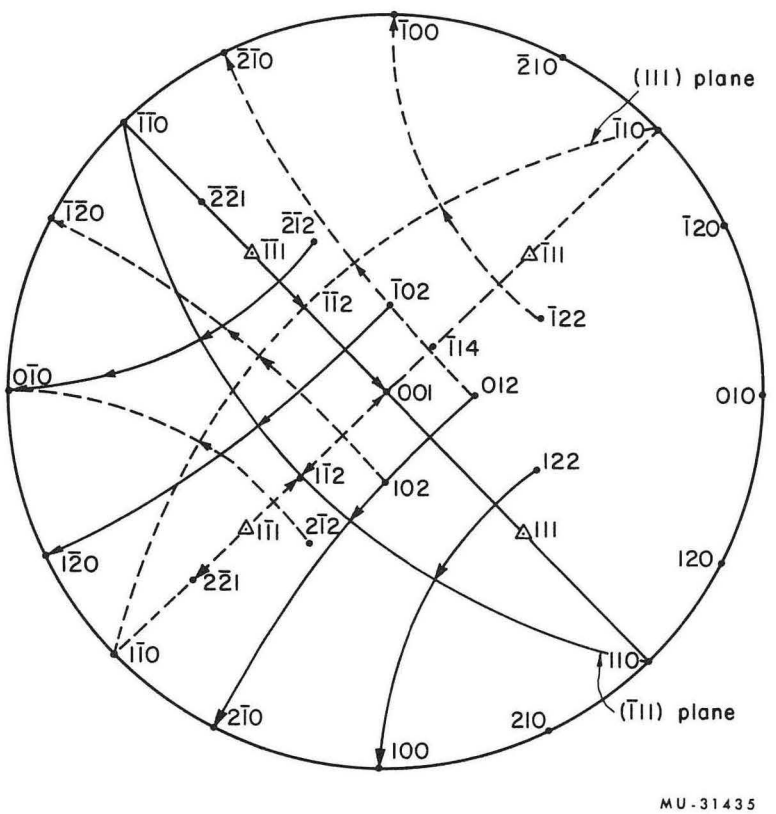
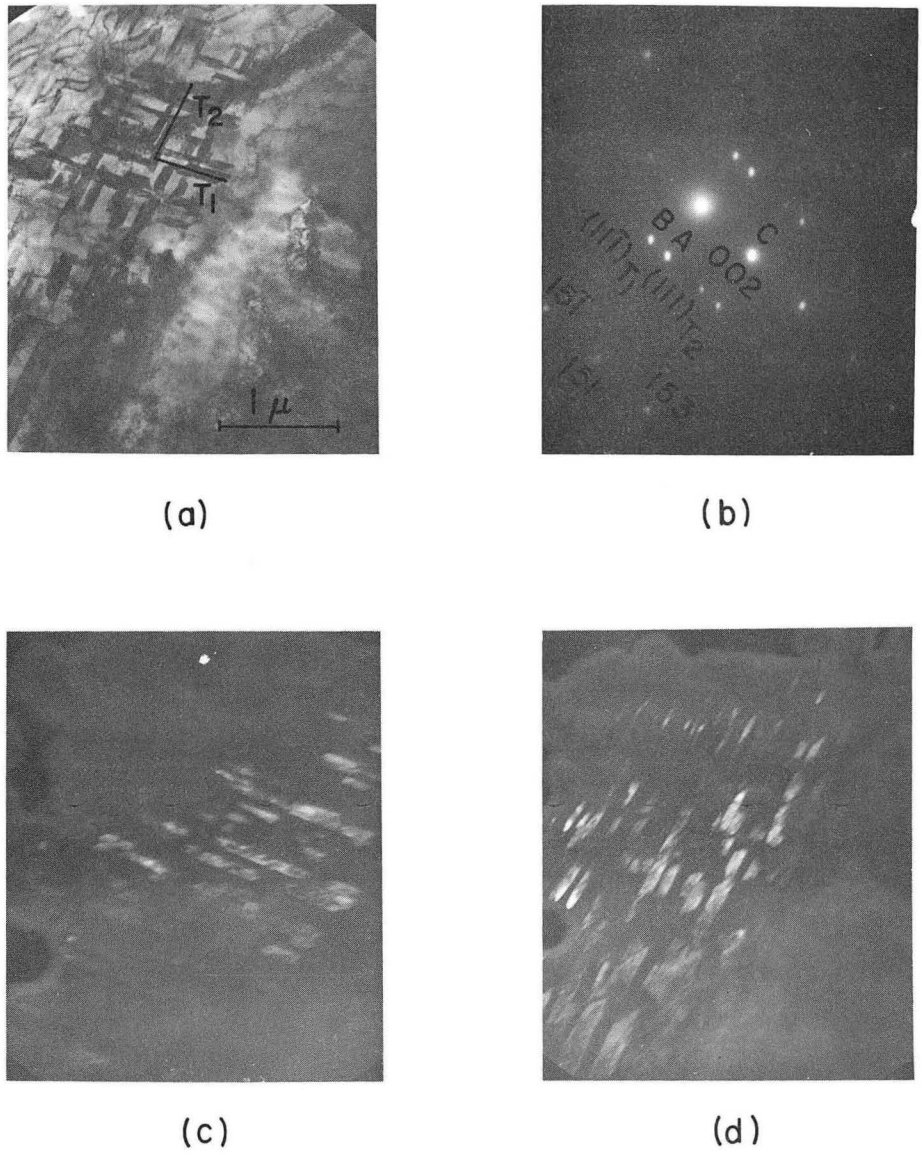

$\bar{1}00$
$\bar{2}\bar{1}0$
$\bar{2}10$
(111) plane
$\bar{1}\bar{1}0$
$\bar{1}10$
$\bar{2}\bar{2}1$
$\bar{1}\bar{2}0$
$\bar{2}\bar{1}2$
$\bar{1}20$
$\bar{1}\bar{1}1$
$\bar{1}11$
$\bar{1}\bar{1}2$
$\bar{1}02$
$\bar{1}22$
$\bar{1}14$
$0\bar{1}0$
001
012
010
122
$1\bar{1}2$
102
$1\bar{1}1$
$2\bar{1}2$
111
$1\bar{2}0$
120
$2\bar{2}1$
$1\bar{1}0$
110
$2\bar{1}0$
210
($\bar{1}$11) plane
100
MU-31435

Fig. 10



ZN-4037

Fig. 11

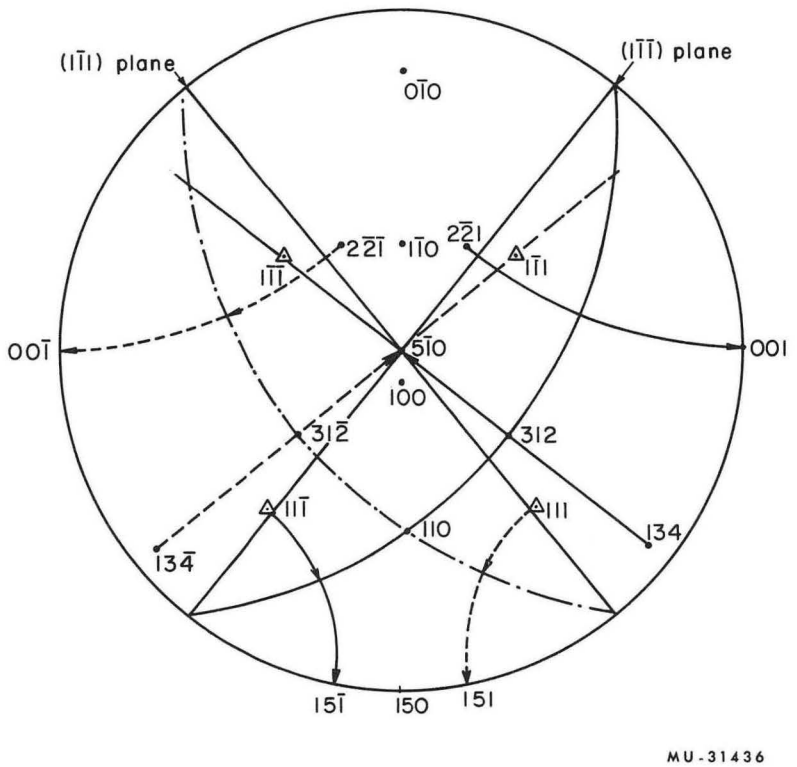
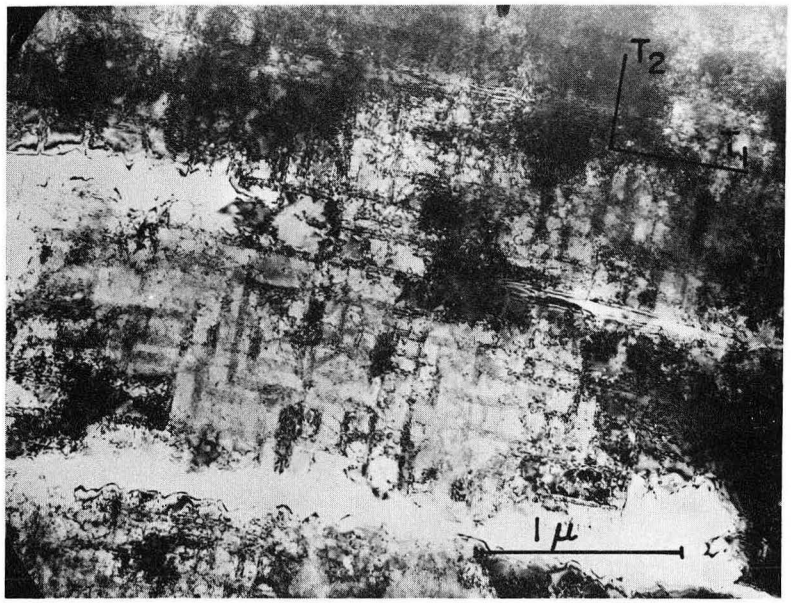
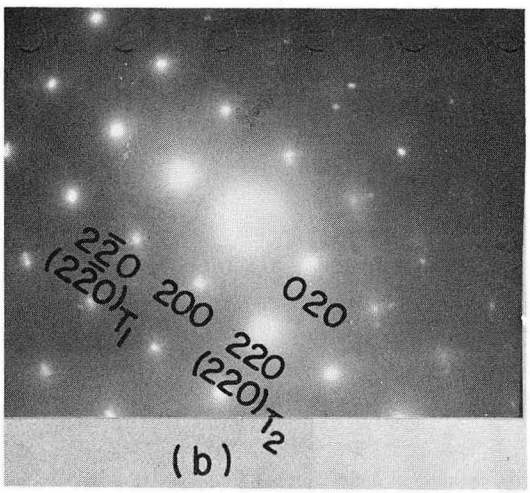

(1$\bar{1}$1) plane
(1$\bar{1}\bar{1}$) plane
0$\bar{1}$0
2$\bar{2}$1
1$\bar{1}$0
2$\bar{2}$1
1$\bar{1}$1
1$\bar{1}$1
00$\bar{1}$
5$\bar{1}$0
001
100
31$\bar{2}$
312
11$\bar{1}$
111
110
134
13$\bar{4}$
15$\bar{1}$
150
151
MU-31436

Fig. 12



(a)



(b)

ZN-3925

Fig. 13

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