600-4072-2

TECHNICAL PROGRESS REPORT

Strain Hardening and Ductility of Iron:
Axisymmetric vs. Plane Strain Elongation

Submitted to:

U. S. Department of Energy
Materials Science Program
Division of Basic Energy Sciences
Washington, DC 20545

Submitted by:

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May 1979

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- B. G. Langford: Rapid Discovery of the $\langle 110 \rangle$ Axis in a Thin Foil Made from Strongly $\langle 110 \rangle$ Textured BCC Wire.
- C. G. Langford: High Resolution Diffraction Analysis of Heavily Deformed Iron; Proposal for No Cost Use of High Voltage Electron Microscope Facilities at the Argonne National Laboratory (first two pages only).

G. Langford - Strain Hardening and Ductility of Iron: Axisymmetric vs. Plane Strain Elongation

Summary

The strain hardening of iron at high strains in plane strain elongation (strip drawing) is shown to fall increasingly below that of drawn iron wires at true strains above 2. This makes it unnecessary to invoke shear band formation simultaneously as a strengthening mechanism and as a ductility reducing mechanism in the drawn strip. Rather, shear bands may be a weakening mechanism in all contexts. A set of specimens of interstitial-free iron deformed in three of the four main classifications of deformation symmetry (wire, strip, and chips, representing axisymmetric elongation, plane strain elongation, and pure shear) has been prepared in the form of mechanical test specimens and thin foils for high resolution selected area diffraction. Free-compression specimens will be prepared later. A simple technique for rapid discovery of the $\langle 110 \rangle$ axis of foils of strongly textured bcc wire has been worked out. A paper describing the progress of the present work has been submitted to the organizers of the Fifth International Conference on the Strength of Metals and Alloys, to be held at Aachen, West Germany in August, 1979.

I. Introduction

A. There are three main concepts on which the present work on strain hardening and ductility of iron is based:

1. At high strains strain hardening is due to the accumulation of excess dislocations of one sign which is physically equivalent to lattice curvature. Strain hardening vanishes when the iron crystals no longer have to bend or curl to accommodate each other, and is persistent when lattice bending is a necessary part of the deformation process (as in the well-known curly grain structure of drawn body centered cubic and hexagonal close packed wires).
2. The symmetry of the combination of slip systems active for a particular deformation process (wire drawing, rolling or strip drawing, free compression, pure shear, etc.) controls the rate of generation of lattice curvature. Body centered cubic and hexagonal close packed wire have two fold symmetry of their slip systems and form ribbon shaped grains which must interfold (curl) to maintain grain boundary continuity. The $\langle 111 \rangle$ component of face centered cubic wire has three fold symmetry leading to the formation of triangular grains; the $\langle 111 \rangle$ component of compressed bcc metals would be expected to form triangular grains analogously to $\langle 110 \rangle$ textured fcc compression specimens which form curly grains like $\langle 110 \rangle$ textured bcc wire. However, the curvature developed by the shape-change interactions between grains in wires is accentuated by the concurrent reduction in cross sectional dimensions (hence the linear strain hardening of bcc wire) whereas the curvature developed similarly in compression specimens would be reduced by the ongoing increase in cross sectional dimensions.

3. The symmetry of the shape change during deformation affects the microscopic homogeneity of the deformation; non-uniform strain (shear bands) consequently limits the residual macroscopic ductility of heavily deformed metals. Axisymmetric elongation appears to have the least tendency to form shear bands, plane strain elongation may rank second best, while pure shear may be the most prone. Axisymmetric compression is an unranked unknown in this respect. Direct comparisons are difficult because the hydrostatic component of the stress state involved in each deformation mode is quite different.
- B. The scope of the present work is to test these theses by comparing the strain hardening, ductility, and microstructure of iron in plane strain vs. axisymmetric elongation of wire, sheet, and tubes, and to apply the results to the manufacture of tubes.
- C. The following report summarizes the experimental and theoretical progress of the past year within the present project. The main accomplishments in this area are:
1. Discovery that the strain hardening of plane-strain elongated (strip drawn) iron falls below that of persistently hardening wire drawn iron at true strains greater than 2. This reduces the difficulty of accounting for the strain hardening rate of strip vs. wire.
 2. Observation of wide variations in cell size from region to region (each region containing large numbers of cells) in the cellular substructure of strip drawn iron. If these patches of cells have similar orientations, the work of

characterizing the microstructure may be greatly reduced.

3. Preparation of a large collection of thin-foil specimens of the same model material produced by the different deformation modes (wire drawing, strip drawing, and pure shear) for detailed characterization by high resolution selected area diffraction.
 4. Development of a rational technique for correctly orienting strongly textured thin foils in the electron microscope.
 5. Development of the tools and techniques for characterizing the flow stress and mechanical stability of the strain hardening state of specimens deformed in complex ways. The plane strain compression test can now be used to measure the flow stress of a material volume not much larger than that deformed in an ordinary hardness test.
- D. Several important advances were published by others in the current literature. These reinforce the present work as explained in the text.
1. Texture inhomogeneities in rolled strip and sheet of fcc metals have been explained in terms of short-comings of the deformation-zone geometry at low strains and in terms of frictional contact with the work rolls in high strains.
 2. Microstructure-texture correlations have been extended to materials with duplex textures, which are shown to have corresponding duplex microstructures.

3. Chip cutting experiments have shown that the cutting stress is independent of chip size over a wide range, and that periodic variation in the deformation zone geometry of chip cutting can explain some tool-vibration problems because of the resulting periodic variation in cutting force.
4. Shear band formation has received a great deal of theoretical attention, and several major accomplishments were reported. Strain localization has been associated with both yield surface vertices (multiple slip) and deviations from Schmid's Law (sensitivity of flow stress on a given slip system to the normal stress on the shear plane). It has been shown that shear bands can form even during strain hardening if these deviations from Schmid's Law are large enough. A theoretical analysis of such deviations has been published. Inhomogeneous deformation (macroscopic) has been related to textural instabilities.

II. Progress

A. Experimental

1. Axisymmetric Elongation

a. Wire Drawing

All wire drawing has been completed and the dies returned to their source, the U. S. Steel Research Center, Monroeville, Pa.

b. Tensile Testing of Drawn Wires

The tensile specimens made from large diameter wires (true strains up to $\epsilon = 2\ln(d_0/d) = 3$) have been tested using the apparatus shown in Fig. 1. Note that we have designed and constructed a device to enable continuous variation of the crosshead speed of our Instron machine. This is necessary to maintain a uniform strain rate during the necking of the tensile specimen (1). Although a record was made of the size and shape of the neck in each tensile test, analysis of that data has not yet been done (in the manner of Langford and Cohen (2)), although the locus of tensile strengths of these specimens shows the expected linear rate of strain hardening (see Fig. 4 of Appendix A).

c. Characterization of Microstructure

Several thin foils have been prepared for quantitative analysis by conventional bright-field and high-resolution selected-area electron diffraction (SAD) by the method of Langford and Cohen (3). Representative bright-field transmission electron micrographs (TEM) and light optical micrographs (LOM) are shown in Fig. 2. A method for correctly orienting these extremely strongly $\langle 110 \rangle$ textured specimens for quantitative SAD in the electron microscope has been worked out by the author (see Appendix B).

Figure 1. Apparatus for Tensile Testing of Heavily Drawn Wire.

Figure 2. Photomicrographs of Transverse Sections of Drawn Fe-0.17% Ti Wire. Top: Transmission Electron Micrographs; Bottom: Light Micrographs (Fry's Reagent). (a) True Strain = 2.0, (b) True Strain = 4.0.

2. Plane Strain Elongation

a. Strip Drawing

We have completed the strip drawing up to a true strain of $\epsilon = \ln(t_0/t) \approx 4$ for two deformation-zone geometries: $\Delta = 0.5$ and $\Delta = 1.0$. Both these geometries are known to give macroscopically homogeneous deformation (4). We have not been able to draw the strip to higher strains, not because of any material limitations, but simply because the strip is so thin at that stage (about 0.01 in) that it is too difficult to indent accurately for the next drawing pass. Further deformation will have to be carried out by rolling.

b. Plane Strain Compression Tests on Drawn Strip

We have completed evaluation of the monotonic strain hardening of the strip drawn to $\epsilon = 3.5$ by the two deformation zone geometries mentioned above. The results are presented in Fig. 1 of Appendix A and show that there is no difference in strain hardening between the two deformation zone geometries and that the resistance to positive lateral strain (plane strain elongation in the long transverse direction of the drawn strip) is the same as the monotonic strength. This is shown in Fig. 1(c) of Appendix A. (There is some tendency towards greater strain hardening at high plane strain compression test strains there; we don't yet know why.) The plane strain compression test setup and tooling are shown in Fig. 3.

c. Characterization of Microstructure of Drawn Strip

We have prepared thin foils of both edge and transverse cross sections of the drawn strip for quantitative bright field and high resolution SAD as for the wires. Representative TEM and LOM are shown in Fig. 4. Note the nonregularity of these microstructures compared to the drawn wires - not only is there evidence of shear bands, but there is also

(a)

(b)

Figure 3. Tooling for Plane Strain Compression Tests
(a) Overall Setup, (b) Subpresses with
Punches Removed to Show Details.

Figure 4. Photomicrographs of Edge Sections of Drawn Fe-0.17% Ti Strip. Top: Transmission Electron Micrographs; Bottom: Light Micrographs (Fry's Reagent). (a) True Strain = 1.8, (b) True Strain = 2.6.

evidence of greatly differing cell sizes from place to place within the drawn strip. The technique described in Appendix B for orienting thin foils of $\langle 110 \rangle$ textured drawn wires only applies in a general way here: we are hoping that each group of cells that have similar size (within a distinctive region of the thin foil) will also have a similar orientation, and that the foil can be manipulated to bring that orientation into strong diffraction so that all cells within the field will have similar SAD patterns. If this can be done, our work will be greatly simplified (to only slightly harder than for the drawn wires). Several such areas of each foil would have to be characterized and compared with each other to help to explain the micro-structure - shape change - texture interrelationships.

3. Pure Shear

a. Orthogonal Chip Cutting

The apparatus for making machining chips (as a model of pure shear) was completed (Fig. 5) and several good chips made. The available range of strains is quite small ($\epsilon = \ln \ell/\ell_0 = -0.5$ to -1.6), since smaller strains require excessive rake angles on the tool, and larger strains cause fragmentation of the chip. However, the experiment has served much of its intended purpose by providing samples of our model material deformed by pure shear even though somewhat imperfect. When we attempted to make a chip sufficiently thick to be able to make a plane strain compression test specimen with its gauge length parallel to the shear direction of the chip, the higher strain chips deviated substantially from plane strain conditions by becoming semicircular in cross section.

b. Plane Strain Compression Tests on Chips

Plane strain compression test specimens were prepared with the gage length nearly perpendicular to the shear direction of the chip - so

(a)

(b)

Figure 5. Apparatus for Chip Cutting. (a) General View Showing Force Recorder and Dial Indicator Used to Measure Depth of Cut; (b) Detail View Showing Chip Cut From Full Edge Thickness of Specimen.

that the change in length during the plane strain compression test was opposite to the change in length of the specimen during making of the chip. This was the most convenient procedure for initial testing and provided information on the mechanical stability of the chip with respect to reversed deformation. The results are shown in Fig. 2 of Appendix A. Note that the plane strain compression test produces intersecting (double) shear whereas chip cutting (ideally) produces single shear. During the remainder of the contract period we expect to complete testing of specimens prepared at right angles to these to evaluate properly the rate of monotonic strain hardening in chip cutting.

c. Characterization of Microstructure of Chips

Thin foils have been prepared from the chips for quantitative analysis by conventional bright-field and high-resolution SAD. We are using edge sections; representative TEM and LOM are shown in Fig. 6 and indicate that the microstructure of the chip varies between the free surface of the chip and the tool-chip interface. This is to be expected from the results of Walker and Shaw (5), who found that the normal stress on the shear plane strongly influences the flow stress during pure shear tests. We are confining our present investigation to the region where the deformation is most uniform (near the tool-chip interface where the normal stress on the shear plane is highest).

d. Shear Spinning Tooling and Preliminary Experiments

The shear spinning tooling has been prepared and several conical cups spun from scrap material for practice. These and the setup are shown in Fig. 7. The results so far indicate that the shear spinning process is too difficult to control to obtain a model of pure shear - there is too much bending of the blank as evidenced by a contraction of the blank diameter.

Figure 6. Photomicrographs of Edge Sections of Fe-0.17% Ti Chip (Nominally Pure Shear 30° Rake Angle, Longitudinal True Strain $\epsilon = -0.74$ (Shear Strain $\gamma = 1.7$)). Top: Transmission Electron Micrographs; Bottom: Light Micrographs (Fry's Reagent). (a) Near Free Surface of Chip; (b) Near Tool-Chip Interface.

Figure 7. Shear Spinning Tooling Shown After Removal from Milling Machine. Practice Cups in Foreground, Undeformed Blank Mounted on Conical Arbor.

In principle it should be possible to evaluate the mechanical properties of these shear spun cones by extracting plane strain compression test specimens from the cone wall in the desired orientation. Therefore, the technique retains some validity for the evaluation of shear spinning. However, we are dropping further work in this area in favor of the Walker and Shaw technique (5) as described in the accompanying renewal proposal.

4. Progress Expected in the Remainder of the Current Contract Term

a. Wire Drawn Strip and Strip Drawn Wire

We have completed preparation of specimens of wire which were deformed initially by strip drawing to $\epsilon = 3.5$ and of specimens of strip which were deformed initially by wire drawing to the same strain. Since these specimens are now quite small, testing them is a delicate and exacting process; however, we expect to complete sufficient tests to enable an evaluation of their mechanical stability with respect to the change in mode of deformation (plane strain to axisymmetric elongation and vice versa). This experiment is a crucial part of our thesis that tubes can be made with microstructure and properties characteristic of one deformation mode or the other, because we cannot accomplish these competing shape changes using identical starting and finishing dimensions of tubes without changing the strain path part way through the tube drawing process.

b. Plane Strain Compression Tests

Two more crucial experiments for which we expect to obtain at least partial data are reversed plane strain deformation of strip drawn material and monotonic plane strain deformation of chips. The former will enable comparison of the mechanical stability of strip drawn material to wire drawn iron as reported by (1,2), while the latter was already explained in section II.A.3.b above.

c. High Resolution SAD Electron Metallography

We have attempted to do high resolution SAD with the scanning transmission electron microscope (STEM) at Lehigh University with the kind help of Dr. Dave Williams; however, we found it impossible in the time available to orient the foil properly to obtain any $\langle 110 \rangle$ SAD patterns in the strongly textured specimen. This prompted development of the procedure described in Appendix B which we have not yet tried out. We expect to complete one or two preliminary sessions with the high voltage electron microscope at the Argonne National Laboratory (Appendix C) by the end of the current contract term. We will analyze one wire foil and one strip foil; we will probably use Argonne's STEM to make orientation maps from the same specimens so that a direct comparison can be made between a HVEM and a STEM for high resolution SAD.

B. Theoretical

1. Progress in this Project

Our present understanding of the high-strain behavior of iron is described in the paper reproduced in Appendix A. The two most important developments in the work this year are:

- (a) Strain hardening during strip drawing is not the same as during wire drawing at true strains above 2 (which may resolve the dilemma posed by the results of Grewen et al.(6) described in last year's progress report - fcc metals have no difference in orientation between shear bands and the adjacent matrix - hence shear band formation may not be a hardening mechanism in either bcc or fcc metals in plane strain elongation),

- (b) There are large patches of cells of homogeneous appearance in the drawn strip, but there are large differences in cell size between patches - thus the task of characterizing and explaining the inhomogeneous deformation of the drawn strip may be greatly simplified if the rigid-block mechanism proposed by Langford (7) for strip drawn pearlite can be applied here on a scale including large groups of cells (such as these patches).

2. Developments in the Current Literature

a. Texture Inhomogeneities in Rolled Strip

Recent papers by Pospiech and Lücke (8) and Asbeck and Mecking (9) claim that texture inhomogeneities between the surface and central regions of rolled metals arise from inhomogeneous deformation ("forth and back shearing") as a result of unfavorable deformation zone geometry (Backofen's $\Delta \approx 2$ (4)) at low strains and from surface friction against slightly rough rolls at high strains. This work supports our decision to use a constant deformation zone geometry ($\Delta = 0.5$ to 1.0) over the whole range of strip drawing strains up to $\epsilon \approx 4$ and to use a pack rolling technique to obtain higher strains (see next year's renewal proposal).

b. Microstructure-Texture Correlations

In addition to the well-known correlations between the microstructure and texture of homogeneously textured metals at high strains, two recent papers by Noda and Huber (10,11) describe similar correlations in inhomogeneously textured copper and aluminum. Specifically, they noted that the cube oriented regions in rolled copper and an aluminum alloy have a different microstructure than differently oriented regions (10) and that the

$\langle 111 \rangle$ and $\langle 100 \rangle$ components of copper and aluminum wires have different microstructures (11), confirming my observations described in last year's progress report. Similar observations have been made on compressed iron by Inokuti and Doherty (12): compressed bcc iron has the same duplex $\langle 111 \rangle$, $\langle 100 \rangle$ texture as fcc wires. Unfortunately, only light optical microscopy was used to characterize this material (in defense of the authors, it was mainly a study of recrystallization behavior). Therefore, we still plan to do our own high-strain compression tests on our model material - with electron microscopy. These important papers mean that interpretation of deformation microstructures after complex deformations may become somewhat simplified once the rates of strain hardening of the various textural components are independently evaluated - we may need only to identify the microstructure by its appearance rather than having to repeat over and over the tedious characterization in the present project.

Takeuchi (13) has investigated the work hardening of $\langle 111 \rangle$ oriented copper crystals; he interprets the strong work hardening of this orientation in terms of two competitive combinations of the same three slip directions and three slip planes (giving mutually intersecting dislocation motion). Takeuchi's work, combined with the present work on the strain hardening of iron in which two quite different types of dislocation operate in an analogously intersecting manner (i.e., the contributory vs. noncontributory dislocations in the curly-grain structure - see Ref. (3)) may eventually be analyzable by the method of Raghuraman and Arsenault (14), although extremely complex optimization computations are involved.

c. Inhomogeneous Deformation Due to Texture Changes

Witzel (15) has discovered that heavily drawn copper wires undergo mechanical instability when tested in torsion - a zone of twist deformation somewhat like a Luder's band (but without any change in cross

sectional area) propagates along the specimen axis. Witzel relates this strain inhomogeneity to changes in the preferred orientation of the polycrystalline copper. This paper should be kept in mind during discussion of the various mechanisms of shear band formation.

d. Chip Cutting

Two recent papers on chip formation support the experimental methods we have used so far. Williams and Gane (16) found that the cutting stress was independent of depth of cut over a wide range (1 to 60 μ m). This agrees with our own limited results and indicates that there is no need to make chips of any particular thickness, as long as the plane strain condition is met (i.e. the chip does not spread). Sullivan, et al. (17) have found extremely serrated chips during the machining of an austenitic stainless steel. Their chips and cutting forces both exhibited periodic variations; in contrast, our own cutting forces are quite steady, lending credence to the notion that steady force and macroscopically homogeneous deformation go hand in hand in chip cutting and that the inhomogeneous microstructure at the free surface of our chips is only a local phenomenon.

e. Shear Band Formation

Needleman and Rice (18) have analyzed two theoretical models for their effect on strain localization (cavitation (i.e., void growth) and yield-surface vertices) and have found that both models make equivalent predictions. Asaro and Rice (19) and Asaro (20) have done a detailed theoretical analysis of the mechanics of shear band formation and conclude that shear band formation can occur while a material is still strain hardening if deviations from Schmid's Law are present or if the specimen is loaded near a yield surface vertex (i.e., if multiple slip is taking place). Kuhlmann-Wilsdorf (21) has written a theoretical paper analyzing the possible sources

of deviations from Schmid's Law arising from dislocation motion. Finally, the work of Walker and Shaw (5) falls on a totally different intellectual plane and should be mentioned again in the present context: if there is not sufficient normal stress on the macroscopic shear plane, steels exhibit marked strain softening in pure shear above a critical strain which is dependent on the magnitude of the normal stress and on the material properties. These observations taken together and in combination with the present work and that described in the preceding paragraphs indicate that there may be a combination of mechanisms for strain localization, some leading to fracture and some not.

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C. Time Devoted to Project by Principal Investigator

Winter	33.3%	from 2/1/78 through 3/31/78	- 0.67 Man Month
Spring	21%	from 4/1/78 through 6/30/78	- 0.63 Man Month
Summer	50%	from 7/15/78 through 9/30/78	- 1.25 Man Month
Fall	12.5%	from 10/1/78 through 12/31/78	- 0.38 Man Month
Winter	25%	from 1/1/79 through 1/31/79	- 0.25 Man Month
	0%	from 2/1/78 through 7/31/79	- <u>0.00 Man Month</u>

Total Time Spent..... 3.18 Man Months

(18% of the 17.5 month contract period)

D. Financial Statement of Estimated Costs

Drexel University

Contract EY-76-S-02-4072

For the Period 1 February 1978 through 31 July, 1979

1. Estimated Total Project Costs for the Current Period	\$52,370
2. Total Amount Chargeable to DoE (95% of line 1)	49,800
3. Cumulative Support Costs	36,363
4. Estimated Total Cumulative Support Costs	86,163
5. Cumulative Support Ceiling	86,163
6. Anticipated Difference Between Items (4) and (5)	-0-

III. Appendices

- A. A. Razavi and G. Langford: Strain Hardening of Iron: Axisymmetric vs. Plane Strain Elongation, submitted to Fifth International Conf. on the Strength of Metals and Alloys, to be held in Aachen, West Germany, August 1979.
- B. G. Langford: Rapid Discovery of the $\langle 110 \rangle$ Axis in a Thin Foil Made from Strongly $\langle 110 \rangle$ Textured bcc Wire.
- C. G. Langford: High Resolution Diffraction Analysis of Heavily Deformed Iron. Proposal for No Cost Use of High Voltage Electron Microscope at the Argonne National Laboratory (first two pages only).

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Appendix B: Rapid Discovery of the $\langle 110 \rangle$ Axis in a Thin Foil Made from Strongly $\langle 110 \rangle$ Textured BCC Wire.

Introduction

Highly $\langle 110 \rangle$ textured foils have such a large fraction of their crystallites (cells) oriented in the same $[110]$ direction that it is very hard to get any $[110]$ electron diffraction patterns unless the foil is within a few degrees of the wire axis. (See G. Langford and M. Cohen, Met. Trans. 6A vol. 6A, 1975, p. 901 - at a true strain of 5.8 for example, almost all the cells' $[110]$ axes were contained within 3° of the wire axis). Trial and error search for the wire axis can therefore take far too long.

Methods of Rapid Location of the Wire Axis

- A. If a large ($>20\mu\text{m}$) field limiting aperture (FLA) is located in the back focal plane of the objective lens, a ring pattern is obtained. The $[110]$ electron diffraction pattern of a bcc crystal has no (310) or (321) spots; consequently, in a properly oriented $[110]$ textured polycrystal there will be no 5th or 7th rings in the above pattern. Therefore, trial and error tilting of the foil when in diffraction mode can be used if the texture is not too sharp. One simply "plays" with the foil orientations until those rings have minimum relative intensity.
- B. After high wire-drawing strains, the above technique is not productive, since the "window" of the desired foil orientation with respect to the microscope's optical axis is only a couple of degrees wide. Therefore a more systematic approach is necessary. Here, we propose to use a sufficiently small ($3\mu\text{m}$) FLA to obtain single-crystal diffraction patterns. Then, use will be made of the fact that a non $[110]$ diffraction pattern (say, $[hkl]$) will have some common tilt axis relative to the $[110]$ pattern. This common axis of tilt will be a $[110] \times [hkl] = [uvw]$ (where $u = l$, $v = \bar{l}$, and $w = k-h$). The principal tilt axes (those which are uniformly distributed

around the desired $[110]$ axis) are shown in Figure 1. It might be hoped that these $[hkl]$ patterns would have a recognizable geometry which would indicate the relative tilt between $[hkl]$ and $[110]$; however, no such pattern exists, as shown in Table I. Sometimes the distance between the $[uvw]$ rows of spots increases with the tilt angle, and sometimes it decreases. This distance is neither especially large nor especially small. The $[002]$, $[1\bar{1}0]$, and $[1\bar{1}2]$ axes have particularly closely spaced spots in the $[002]$, $[1\bar{1}0]$, and $[1\bar{1}2]$ direction, respectively; since no $[hkl]$ patterns within several degrees of $[110]$ have a second row of such closely spaced spots (Table I), these rows of closely spaced spots should indicate the axis of tilt of the foil relative to the microscope axis. Therefore, the technique to be used is relatively simple: scan the specimen in diffraction mode, using the correct objective lens current to place the first specimen image in the same plane as a small (ca. $5\mu\text{m}$) FLA. Observe the general orientation of rows of closely spaced spots in the SAD patterns, and then slowly tilt the specimen about the axis corresponding to the orientation of those rows. If tilts of 10 to 20° in one direction are unproductive, go back to the original foil orientation and tilt about the same axis in the opposite sense. Tilt by no more than 1 or 2° at a time - the SAD pattern will not give any clues to the angle of tilt until some $[110]$ patterns are finally obtained. When this happens, the circle of bright spots will indicate the remaining tilt angle and direction of tilt as shown in the Langford and Cohen paper. Furthermore, the bright field contrast and intensity become minimal due to strong diffraction from virtually every cell and cell wall, as observed by Langford and Cohen.

Table I. Characteristics of the Polycrystalline Diffraction Pattern of a [110] Textured BCC Specimen when [110] Coincides with the Optical Axis

<u>hkl</u>	<u>$h^2 + k^2 + l^2$</u>	<u>ring # in a rotated bcc electron diffraction pattern</u>
110	2	1
200	4	2
112	6	3
220	8	4
310	10	x
222	12	6
321	14	x
400	16	8
330,411	18	9

missing rings in
a [110] pattern

Figure 1. Principal $[uvw]$ Directions in a $[110]$ Electron Diffraction Pattern.

Table II. $hk\ell$ Diffraction Patterns Containing Characteristic $[uvw]$
Principal Directions Also Lying in a $[110]$ Pattern

$[uvw]$ Principal Direction and Tilt Axis	$hk\ell$ Pattern	angle of tilt from $[110]$ axis about $[uvw]$ axis	ratio of spacing of $[uvw]$ rows to spacing of spots in $[uvw]$ rows	spacing of spots in $[uvw]$ rows relative to spacing of spots in $[110]$ row
002	11 12 0	2.5°	16.3	1.414
	10 13 0	7.4°	8.2	1.414
$\bar{1}\bar{1}4$	38 42 1	3.0°	12.60	3.000
	18 22 1	6.0°	6.33	3.000
	10 14 1	10.1°	3.83	3.000
$\bar{1}\bar{1}2$	10 12 1	6.4°	4.08	1.732
	11 13 1	5.8°	2.84	1.732
	20 22 1	3.3°	9.91	1.732
$2\bar{2}2$	11 12 1	4.3°	10.2	2.449
	10 12 2	9.0°	2.63	2.449
$4\bar{4}2$	11 12 2	7.4°	1.06	4.24
	23 24 2	3.7°	3.70	4.24
$\bar{1}\bar{1}0$	10 10 1	4.0°	14.2	1.000
	10 10 2	8.0°	7.14	1.000
	10 10 3	12.0°	14.5	1.000
	10 10 4	15.8°	7.35	1.000
	20 20 1	2.0°	28.3	1.000
	20 20 2	4.0°	14.2	1.000
	20 20 3	6.0°	28.4	1.000

APPENDIX C: Attached

HVEM-TANDEM ACCELERATOR FACILITIES
RESEARCH PROPOSAL SUMMARY FORM

PROPOSAL NUMBER _____

DATE RECEIVED _____

Name of Principal Investigator: Dr. George Langford

Address: Department of Materials Engineering
Drexel University
Philadelphia, Pa. 19104

Name of Project: High Resolution Electron Diffraction Analysis of
Heavily Deformed Iron

Names and Affiliation of Persons Involved - extent of participation of
each individual:

Dr. George Langford - Drexel University 40%

Mr. Ali Razavi - Drexel University 40%

Dr. Ronald O. Scattergood - Argonne National Laboratory 20%

Scientific Purpose and Objectives of Proposal:

Correlate the microstructure of severely deformed iron to the imposed
external shape change and the resulting mechanical properties.

Significance of Expected Results:

We expect to characterize the microstructure of thin foils of heavily deformed iron prepared in appropriate cross sections by measuring the size and crystallographic orientations of the approximately 0.05 to 0.2 μm diam. crystallites in the cellular dislocation substructure of these specimens. This will enable us to prepare maps of 50 to 100 contiguous cells in a representative area of the thin foil showing (a) the crystallographic preferred orientation (texture), (b) the equivalent density of excess dislocations of one sign, and (c) the detailed arrangements of the cells. We expect to show that these microstructures and their strengthening effects are sensitive to the mode of shape change employed in deforming the specimens. This should lead the way to a unified understanding of the strain hardening of metals at high strains, which is presently nonexistent.

Source of Funding and Duration:

U. S. Department of Energy, Materials Science Program, Division of Basic Energy Sciences, Dr. Stanley Wolf, Program Monitor, February 1, 1978 through July 31, 1979; an application for renewal through July 31, 1980 is being submitted.

Estimate of Facility Time hr/wk/yr:

HVEM 100 hrs., spread over 12 to 25 days within 1 year ACCEL. (none)

Estimate of Manhours Required of Facility Personnel:

150 hrs. (mainly for developing about 2000 negatives, etc.)

Source of Support Personnel and Estimated Manhours:

none hrs. (no in-situ experiments contemplated and we expect to operate the microscope ourselves, although we do need instruction for this particular instrument).

Special Facility Requirements:

Earliest Possible Starting Date - Immediately

Latest Possible Starting Date - June 1, 1979

Other Time Requirements - Need at least one half a working day for each specimen; prefer whole days.

Special Equipment Requirements:

High resolution diffraction - 0.05 μ m selected area.

Specimen Stage Requirements:

1. Double tilt, ambient temperature (alternate: tilt/rotate; may even be preferred).
2. Anticontamination "cold finger".

Special Service Requirements

- 1 Process about 2000 negatives (100 per specimen)
2. Retain and clean selected-area aperture to ensure comparability of results.
3. Careful numbering and cataloging of negatives to prevent (disastrous) mixups.