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The development of ductility in 0.1 pct. carbon steels for possible deep-drawing applications.

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THE DEVELOPMENT OF DUCTILITY IN 0.1 pct. CARBON STEELS FOR POSSIBLE DEEP-DRAWING APPLICATIONS

A Thesis
Submitted to the Faculty of Graduate Studies through the
Department of Engineering Materials in Partial Fulfilment
of the Requirements for the Degree of
Master of Applied Science at the
University of Windsor

by

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ABSTRACT

Notched tension tests and "R and N" tests were carried out on twelve grades of rimmed steel ranging in carbon content from 0.002 pct. to 0.165 pct. in the process-annealed and normalized conditions. Process-annealing was carried out at 670°C for three hours and normalizing at 925°C for 1/2 hour.

It was found that notch elongations were very sensitive to carbide morphology and distribution in sub-critically annealed low carbon steels. High elongation values were associated with a uniform distribution of fine carbides. Coarser carbides, segregated at the grain boundaries, are associated with low notch elongation values. Normalizing resulted in a partially pearlitic structure and slight grain coarsening which generally lowered the yield strengths.

Similar trends were noted in the notch elongation values and minimum coefficient of anisotropy values in process-annealed materials, as both parameters are closely linked with the finishing and coiling temperatures. Process-annealed materials in general showed high planar anisotropies and work hardening coefficients, dependent on grain size. Normalizing resulted in a reduction in the planar anisotropy and an increased work hardening coefficient, the latter being associated with a slight grain coarsening.

Normalizing of the two 0.1 pct. carbon grades resulted in significant improvement in the planar anisotropy, minimum coefficient of anisotropy and work hardening coefficient. These two grades,

in the normalized condition were judged to have superior deep-drawing properties to the 0.028%C steel in the process-annealed condition.

Indirect evidence indicated that steels which had been open-coil annealed developed different textures on normalizing than steels which had been previously batch annealed.

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Worthy of special mention is my wife Antoinette, who typed the thesis.

CONTENTS

Page

ABSTRACT iii

ACKNOWLEDGEMENTS v

CONTENTS vi

FIGURES viii

TABLES x

I INTRODUCTION 1

A. Production of Steel for Deep-Drawing 1

B. Laboratory Tests Commonly Used to Predict Deep-Drawing Behavior 2

C. Criteria for Prediction of Drawability 5

D. Preferred Orientation 8

II EXPERIMENTAL PROCEDURE 10

A. Material 10

B. Mechanical Testing 10

1. Specimen Preparation 11

2. Tension Test 12

3. R and N Test 12

C. Metallographic Examination 13

1. Grain Size Determination 14

2. Carbide Morphology 14

III RESULTS AND DISCUSSION 15

A. Tension Tests 15

1. Preliminary Tension Tests 15

Page

2. 0.002%C and 0.0095%C Steels 19
3. 0.028%C and 0.045%C Steels 22
4. 0.055%C, 0.06%C, 0.065%C and 0.075%C Steels 25
5. 0.095%C and 0.098%C Steels 27
6. 0.13%C and 0.165%C Steels 29

B. R and N Results 31
1. 0.002%C and 0.0095%C Steels 33
2. 0.028%C and 0.045%C Steels 34
3. 0.055%C, 0.06%C, 0.065%C and 0.075%C Steels 37
4. 0.095%C and 0.098%C Steels 39
5. 0.13%C and 0.165%C Steels 41

IV SUMMARY AND CONCLUSIONS 43

V RECOMMENDATIONS 45

REFERENCES 46

APPENDIX A MILL DATA OF MATERIAL USED 48

APPENDIX B FURNACE COOLING CURVES FOR HEAT TREATMENTS 62

APPENDIX C COMPUTER PROGRAMS AND THEORY FOR R AND N TESTS 64

APPENDIX D NOTCHED TENSION TEST RESULTS AND GRAPHS 73

APPENDIX E R AND N TEST DATA AND GRAPHS 82

APPENDIX F GRAIN SIZE TABULATION AND GRAPH 97

APPENDIX G PHOTOMICROGRAPHS 100

LIST OF FIGURES

Page

Figure

B-1 Typical furnace cooling curve for process-annealed specimens 63

B-2 Typical furnace cooling curve for normalized specimens 63

D-1 Notched tension test - Y.S., U.T.S. and pct. elong. versus carbon content for normalized specimens 74

D-2 Notched tension test - Y.S., U.T.S. and pct. elong. versus carbon content for normalized specimens 75

D-3 Notched tension test - "Z" versus carbon content, process-annealed specimens 76

D-4 Notched tension test - "Z" versus carbon content, normalized specimens 77

D-5 Specimen dimension for notched tensile test 78

E-1 R & N test - R versus carbon content, process-annealed specimens 83

E-2 R & N test - R versus carbon content, normalized specimens 84

E-3 R & N test - R_{min} versus carbon content, as-received, process-annealed, normalized specimens 85

E-4 R & N test - R_n versus carbon content, as-received, process-annealed, normalized specimens 86

E-5 R & N test - N versus carbon content, process-annealed specimens 87

E-6 R & N test - N versus carbon content, normalized specimens 88

E-7 R & N test - N_n versus carbon content, as-received, process-annealed, normalized specimens 89

Figure		Page
E-8	R & N test - N_n versus $R_{min.}$ - as-received, process-annealed, normalized specimens	90
F-1	Grain size versus carbon content, as-received, process-annealed, normalized specimens	98
G-1	Process-annealed 0.002%C - equiaxed fine grained structure	101
G-2	Normalized 0.002%C - abnormal grain growth	101
G-3	Process-annealed 0.045%C - massive carbides	102
G-4	Normalized 0.045%C - pearlite colony enveloped by cementite	102
G-5	Process-annealed 0.065%C - massive carbides	103
G-6	Process-annealed 0.075%C - finely dispersed spheroidized carbides	103
G-7	Normalized 0.075%C - coarse pearlite and banded cementite	104
G-8	Process-annealed 0.095%C - even carbide distribution	104
G-9	Process-annealed 0.0983%C - 70-30% carbide distribution between grain boundary and interior of grains	105
G-10	Normalized 0.095%C - pearlite colonies with fine interlammeller spacing	105
G-11	Normalized 0.0983%C - small pearlite colonies with coarse interlammeller spacing	106
G-12	Process-annealed 0.165%C - massive carbides and carbide bands	106
G-13	Normalized 0.165%C - evenly distributed pearlite with cementite bands	107

TABLES

Page

A-1. Carbon Analysis 49

A-2. Summary of Mill Data 50

D-1. Tension Test Results 79

E-1. R and N Test Results 91

F-1. Grain Size 99

I INTRODUCTION

The object of this investigation was to develop in a 0.1% carbon rimmed steel deep-drawing properties equivalent to more dilute alloys (approximately 0.03%C), which are the grades commonly used for this application.

A. Production of Steel for Deep-Drawing

At present adequate deep-drawing properties of steel with 0.1% carbon are only found in fully-killed steels. However, the production of fully-killed steels involves high consumption rates resulting from scrapping of the feeder head, the amount of scarfing to be carried out on the slabs and rejects because of surface defects of cold reduced sheets (1). Additional difficulties are encountered in production as a close control of proper conditions for aluminum nitride precipitation is necessary to ensure an oriented grain structure.

Rimmed steels for deep-drawing applications usually have a carbon content of less than 0.04%. Though special grades are produced with very low carbon levels (0.01% or less) by employing a decarburizing atmosphere during the process-anneal cycle, in general the carbon content of the flat-rolled product is essentially the same as the original ingot. To obtain an ingot with about 0.04%C necessitates pouring a melt containing about 0.06%C, with the rimming action reducing the effective carbon level to about 0.04%. The casting of a sound ingot of this composition involves difficulties, that disappear at higher carbon levels; hence considerable economic gains could be realized if adequate formability could be imported to a 0.1%C rimmed steel.

B. Laboratory Tests Commonly Used to Predict Deep-Drawing Behavior

The development of laboratory tests that adequately predict the behavior of steel sheet during forming operations has been a formidable task. A committee was set up in Britain in 1935, originally under the auspices of the Institution of Automobile Engineers, whose job it was to devise a test to suitably predict the behavior of sheet in the press shop (2). As of 1962 this committee was still sitting, its task unaccomplished.

The concept of drawability is a little nebulous and it is more convenient to view deep-drawing operations as a combination of two basic forming operations: die-drawing and stretching. "Die-drawing is deformation by compression affecting that part of the metal which slides below the blank holder and fills the matrix; stretching is deformation by elongation in many directions for the part which is outside the blank holder and which will be shaped by the nose of the punch" (1). Most industrial operations are a combination of these two types of deformation and are referred to as "mixed drawn operations".

The properties required of a material differ appreciably with respect to these two basic types of deformation. For die-drawing a minimum ductility obtainable in any given material (3). For example, 18/8 stainless steel has a tensile ductility in excess of the best mild steel but performs poorly in pure deep-drawing operations. Titanium, on the other hand, does not possess outstanding tensile ductility but performs the best of any commercially available material in deep-drawing (2). Of greater importance in die-drawing is the orientation of ductility in the plane of the sheet such that planar deformation can take place without an appreciable reduction in thickness. Conversely, stretch forming requires maximum uniform elongation of unsupported metal without necking. Uniform elongation is governed by the rate of work hardening.

Laboratory tests commonly used for the assessment of deep-drawing properties are as follows:

1) Standard tension test
2) Hardness test
3) Cupping tests
4) R & N test (modified tension test)

The conventional tension test yields limited data applicable to deep-drawing. As previously mentioned, a minimum tensile ductility must be exceeded for die-drawing. This of course, is measured in the tension test. One other directly applicable parameter is the yield to ultimate ratio, which is an indication of a material's ability to stretch.

Hardness measurements, in themselves, are not a good criterion as similar deep-drawing behavior is obtainable in materials of widely differing hardnesses. For example, with favourable sheet orientation, deep-drawing of high strength low alloy steels becomes a practical possibility (3). However they are valuable as a monitor within materials of similar grade.

A number of cupping tests have come into being over the years in an attempt to predict formability. These tests are of a simulating nature and differ in principle depending upon their use as a criterion for die-drawing or for stretching (1). Tests for die-drawing include: Flat-bottomed Cup, Swift test, Cup with Hemispherical Bottom, whereas tests for stretching include the Erichsen and Bulge tests. The Fukui test was introduced to simulate both die-drawing and stretching. These tests have proven to be most useful in predicting behavior of a material in the press shop but suffer from the disadvantages that they are slow to carry out and are influenced by such variables as lubrication, surface condition of sheets and tools, speed of testing, test piece thickness, etc.

A modified tension test, in which elongations are measured in two perpendicular directions and which is terminated at approximately 20% elongation, yields two parameters: the work hardening coefficient N, and coefficient of anisotropy R. Lankford et al (4) were the first investigators to relate these parameters to deep-drawing of low carbon steel. The coefficient of anisotropy (R) is useful as a criterion for die-drawing whereas the work hardening index (N) is used primarily as an indicator of the stretching characteristics of a material. For most industrial applications both R and N are of importance as processes are in general a mixture of die-drawing and stretching.

C. Criteria for Prediction of Drawability

To clarify the terminology to be used in the following discussion the definition of a few terms is in order:

Planar anisotropy - variation of the R value with direction in the plane of the sheet

Normal anisotropy - an average R value defined by the equation:

$$R_n = 1/4\ (R_{0^o} + 2\ R_{45^o} + R_{90^o})$$

Normal work hardening coefficient - an average N value defined by the equation:

$$N_n = 1/4\ (N_{0^o} + 2\ N_{45^o} + N_{90^o})$$

Lilet and Wybo (5) have shown that an excellent correlation exists between R and deep-drawing tests involving circumferential compressive strain, while a less significant correlation exists between N and stretch-forming tests. They have also shown that a knowledge of the mean values of R and N, (i.e. R_n and N_n), are often insufficient to adequately assess formability. They emphasize the importance of the minimum R value and variation of R within the plane of the sheet (planar anisotropy). The R_{min} value is of importance because, with perfect tooling, a symmetrical pressing should yield at the die radius in the direction in which the sheet offers the least resistance to thinning. It has also been shown that a high planar anisotropy results in "earing". Wacquez (1) points out that the high plastic anisotropy of aluminum-killed steels, (as expressed by R_n), is offset by a low R_{min} value at 45 degrees, which often results in considerable earing.

With respect to the correlation of R and N values to actual forming operations, Lankford et al (4) came to the conclusion, in predicting the behavior of a material in an unsymmetrical fender draw, that one must take into account both R_n and N_n. Further, the yield to ultimate ratio could be used instead of N_n, but a lower degree of correlation with press behavior was obtained.

An extensive investigation by a special committee of the ASTM (6) was concerned with the correlation of R and N to three different pressing operations:

1. Production of a truck door panel - judged to be predominantly a stretching operation.
2. Production of a blower housing - judged to be predominantly die-drawing.
3. Production of an instrument panel - a combination of die-drawing and stretching.

In the first instance a strong dependence on N and very little dependence on R was found. Good performances were obtained at all levels of R for steels with N values above 0.220 and the only correlation with the coefficient of anisotropy was with R_{min} occurring at 45 degrees to the rolling direction. The production of the instrument panel showed a strong dependence on both R and N. Good performance was obtained if either variable was high and poor performance was only obtained if both values were low. The blower housing production showed a clear dependence on R with little N dependence.

Concerning absolute values of the two parameters, to ensure adequate press performance, Lloyd (7) points out that rimmed steels destined for deep-drawing are marketed with R_n values as low as 0.4 but in his opinion steels should have the following R values to ensure drawability:

$R > 1.7$ - extra deep-drawing quality

$1.7 > R > 1.2$ - ordinary drawing quality

Lankford et al (4) suggest that best results in unsymmetrical draws are obtained with material exhibiting N values in excess of 0.240 and R_t values greater than 1.50, where R_t is defined as the R value in the rolling direction of the sheet.

It appears evident from a survey of recently published test results that optimum properties for a rimmed steel which is to undergo an industrial deep-drawing operation, (combination of varying degrees of die-drawing and stretching), are the following:

1. High normal coefficient, R_n
2. High minimum coefficient, R_{min}
3. Low planar anisotropy
4. High normal work hardening coefficient, N_n.

D. Preferred Orientation

Anisotropy in steel sheet has different causes as exhibited by different mechanical properties. The difference in the longitudinal and transverse ductility of steel sheet is for the most part due to mechanical fibering (8). Another form of anisotropy resulting in directionality in the yield behavior of low carbon sheet is caused by residual stresses introduced by temper rolling. While these two forms of anisotropy influence drawability to some degree, of greater importance is the anisotropy resulting from preferred crystallographic orientation, or texture. A material which has a strong rolling texture will exhibit different yield strengths in different directions. Burns and Heyer (9) have studied the texture development in low carbon rimmed steels in the cold rolled, process-annealed and normalized conditions. They have shown that in the case of a rimmed steel process-annealed at 1300°F for 32 hours, after a 60% cold reduction, the prominent texture is $(\bar{1}11)$ [110] with some evidence of a (100) [001] preferred orientation. Normalizing at 1750°F resulted in a complete disappearance of the (110) [001] texture while a small amount of $(\bar{1}11)$ [110] was retained and a low degree of (100) [001] was developed. They also correlated the planar anisotropy to the various preferred orientations as follows:

(100) [001] - high R in the rolling direction (greater than 2)
- low R in transverse and 45 degree directions (less than 0.5)

(110) [001] texture - low R at 0° and 45° with high R at 90°

($\bar{1}$11) [110] texture - R greater than unity increasing steadily
from 0° to 90°.

As preferred orientations were not determined in this investigation, texture development can only be inferred from the relative values of the coefficient of anisotropy.

This investigation employed both the R and N test and the notched tension test. It was felt that a correlation of notched tension data with R and N data would be valuable, as the determination of R and N values is time consuming and a notched tension test might prove to be of greater use in production control.

A total of twelve rimmed steels ranging in carbon content from 0.002% to 0.165% were tested in the process-annealed and normalized conditions. Some testing was carried out on as-received material but it was decided to sub-critically anneal all grades to eliminate any strain-aging effects due to differences in the interval that elapsed between the final temper roll and testing.

II EXPERIMENTAL

A. Material

The material used in this investigation was plain carbon steel, 0.0359" thick, supplied in 2' x 4' sheets by Dominion Foundries and Steel, Limited. There was a total of twelve rimmed grades ranging in carbon content from 0.002 to 0.165%. The complete mill data for this material are found in Appendix A.

The carbon concentrations reported were supplied by Dominion Foundries. An independent analysis of four grades in the as-received, process-annealed and normalized conditions was conducted by Chicago Spectro Services Laboratory Inc. and appear in Table A-1.

B. Mechanical Testing

Two types of tension test were employed. The standard tension test, on both notched and unnotched specimens, was a test to failure and measured elongations solely in the direction of the principal tensile stress. A second type of tension test, in which elongations are measured in two perpendicular directions, permits calculation of the coefficient of anisotropy (R) and the work hardening coefficient (N). These tests are terminated at approximately 20% elongation and will subsequently be referred to as the R and N tests. See Appendix C for the applicable equations.

1. Specimen Preparation

Specimen preparation was similar for all tests. The 2' x 4' sheets were sheared to form test pieces 1/2" wide and 6" long. Specimens were cut from the sheet in the rolling direction, transversely, and at 45° to the rolling direction. All machining operations were carried out prior to heat treatment. On unnotched specimens a gage length was shaped by grinding on a Tensil-Grind Model 2066 grinder, whereas notched specimens had the notch milled before the gage length was ground. Two notch geometries were employed in an attempt to suitably measure elongation variations between the different grades of low carbon sheet. At the outset a notch with a 3/1000" radius and 45° angle was used and subsequently a notch of 1/1000" radius and 60° angle was employed. Fig. D-5 depicts specimen dimensions for the latter geometry - which proved to give more satisfactory results.

Prior to heat treatment the specimens were thoroughly degreased with xylene. A tube furnace was constructed which permitted the entire heat treatment process to be carried out under an argon atmosphere. The actual specimen tube was 2 1/2" diameter, type 316 stainless steel, which protruded 2' past the hot zone of the furnace at one end. This portion of the tube was wrapped with copper cooling coils in an attempt to simulate "air cooling" conditions in an inert atmosphere. Normalized specimens were held at 925°C for 30 minutes and subsequently moved to the water cooled end of the furnace tube. Process-annealing was carried out at 670°C for 3 hours with a similar cooling cycle. Appendix B contains typical cooling curves for both process-annealed and normalized specimens. These curves were obtained by fixing a thermocouple to the interior of a batch of specimens. It is apparent from these curves that the temperature fell below 300°C in less than 3 minutes for the process-annealed samples and in less than 9 minutes for the normalized specimens.

2. Tension Test

An Instron, Model TT-P, tensile tester was used with extensometers of 1/2" and 1" gage lengths. In the preliminary tension tests on both notched and unnotched specimens, the strain rate was varied from 0.05"/min. to 2"/min. The R and N tests were carried out at a strain rate of 0.2"/min.

3. R and N Test

The R and N test procedure and method of calculation are those employed by Grumbach and Pomey (10), the main difference being the specimen width which was 1/4" in our work as opposed to 1/2" in theirs. The test was terminated at 20% elongation, as measured over a 1" gage length, to avoid straining the specimen beyond a value associated with the true ultimate strength of the material. To facilitate elongation measurements in a direction perpendicular to the principal tensile stress, (i.e. across the width of a specimen), the central portion of the gage length was marked off in 5 mm intervals by fine parallel lines scribed across the specimen using a steel jig specially constructed for this purpose. The specimen width at these points was measured before and after the test using the low power optics of a Leitz microhardness tester.

As the calculation of R and N values from test measurements is a laborious task, two computer programs were written for this purpose and appear in Appendix C. The results of these tests were based on four identical test pieces from each of the three sheet directions, (rolling direction, transverse and 45° to the rolling direction).

An individual R value was calculated for each specimen based on the initial width, final width and total elongation, (in the longitudinal direction) measured over a 1" gage length. An average value for the four identical specimens was then calculated.

The work hardening coefficient (N) calculation involved application of Linear Regression to the data as a whole from tests of the four identical specimens. In this way the accuracy of the calculation was increased over that obtainable by application of Least Mean Squares to data of a single specimen and subsequently averaging the values of the four identical specimens.

C. Metallographic Examination

For the purpose of metallographic examination, sections from each of the four identical R and N test specimens were mounted - each carbon concentration, rolling direction and heat treatment separately.

1. Grain Size Determination

The grain size of each of the mounted samples was determined by the "Intercept Method" proposed by Hilliard (9) to an accuracy corresponding to a standard deviation of 0.3. This method involves counting the number of grain boundary intersections on a circle which is superimposed on the magnified image of a metallographic specimen. The specimen image was projected on the ground glass screen of a Leitz Metallograph at a magnification of either 320X or 160X. The number of grain boundary intersections on the circumference of a superimposed 10 cm diameter circle was then counted. In each case an attempt was made to choose a representative area within a sample and the average of the four identical specimens was taken.

2. Carbide Morphology

An examination of carbide morphology and distribution using 1% nital as etchant was carried out with a 180X oil immersion objective to give a magnification of X1600. A very light etch proved most satisfactory.

III RESULTS AND DISCUSSION

Although the sections of a thesis entitled "results" and "discussion" are commonly separate (thereby avoiding the possible confusion of fact and conjecture) it was felt that the nature of this investigation necessitated the combination of these two sections into one for the sake of clarity.

A. Tension Test

Appendix D contains the conventional tension test results in tabulated and graphical form. Notched and unnotched specimens in the as-received, process-annealed and normalized conditions were tested and each result is an average of at least three specimens.

1. Preliminary Tension Tests

Preliminary tension tests carried out on three grades of steel, (0.002%C, 0.055%C and 0.095%C), were concerned with attempting to establish suitable test conditions whereby significant differences in elongations between the three grades and three heat treatment conditions could be detected.

Tests Nos. 1, 2 and 3 (refer to Table D-1) were carried out on as-received material of the above carbon concentrations. The specimens were unnotched and the strain rate was varied from 0.05"/min. to 0.5"/min. As expected, a slight increase in both the yield strength and ultimate tensile strength was observed as the speed of testing was increased, and with the exception of the 0.002%C material a slight lowering in elongation was noted. As the difference in elongations between the three grades was more pronounced with a strain rate of 0.2"/min. it was decided to carry out subsequent testing of unnotched specimens at this rate.

These tests show the effect of strain rate on the yield to ultimate ratio. The yield to ultimate ratio is commonly used as an acceptance test for formability. It has been recently pointed out (12) that the strain rate dependence of this parameter is one reason for not using it as the sole criterion.

To determine the effect of heat-treatment on unnotched tensile properties Tests Nos. 4 and 5 were carried out on process-annealed and normalized material respectively. In the case of the process-annealed material a 12 percentage points difference in elongation between the 0.055%C and 0.002%C grades was obtained, and a 5% difference between the 0.095%C and 0.055%C materials existed. This was not the case for the normalized specimens as only a 3% difference between the 0.055%C and 0.002%C grades was obtained, while the 0.095 and 0.055%C grades showed identical elongations. In hopes of obtaining an elongation difference between the 0.055%C and 0.095%C materials in the normalized condition it was decided to employ a notched tension test.

Test No. 6 was carried out on as-received steel at a strain rate of 0.05"/min. with specimens notched to a radius of 0.003" ("dull" notch). Comparison of the results of Test No. 1 and Test No. 6 shows that the effect of the notch is to increase both the yield and ultimate strengths approximately 20% while the elongations drop from 40% to less than 10%. The difference in elongations between the three grades of as-received material is not significantly increased.

Tests Nos. 7 and 8 were carried out on identically notched samples as No. 5 but at a strain rate of 0.2"/min. Test No. 7 was on process-annealed material whereas Test No. 8 was on normalized specimens.

Comparing the results of Test No. 4 and No. 7 it appears that the notch increases both the yield and ultimate tensile strength in the 0.095%C grade. The ultimate tensile strength of both the 0.055 and 0.002%C grades is raised about 5000 psi whereas the yield strength is little effected in these two grades by the introduction of a notch.

A similar comparison for normalized material of these three grades (Tests 5 and 8) indicates that the effect of the notch is to raise both the yield and ultimate tensile strengths about 5000 psi in all cases. Once again, the elongation dependence on carbon concentration is rather disappointing - less than 1% difference - but in this case the process-annealed samples of 0.055%C and 0.002%C are identical.

The notch design was changed for subsequent tests to a 0.001" radius and 60° angle which had been found by previous investigators (13) to be the geometry that renders maximum notch sensitivity in high strength sheet.

Test No. 9 shows the results of tests carried out on normalized material of the same three carbon concentrations as previously tested, but with a sharpened notch. Comparing Tests 8 and 9 one sees that the sharper notch has had little effect, other than to further decrease the elongations by approximately 3 percentage points in all cases.

It should be noted that Tests Nos. 1 to 9, inclusive, employed an extensometer with a 1" gage length. It was felt that the notch sensitivity might be further increased by a reduction of the gage length. All later tests were conducter with a 1/2" extensometer.

Tests Nos. 10 and 11 were carried out on all twelve grades using a "sharp" notch, 0.2"/min. strain rate and 1/2" gage length extensometer. Test No. 10 was on process-annealed material whereas No. 11 was on normalized specimens.

Figures 1 and 2 of Appendix D show graphically the results of Tests 10 and 11, where ultimate tensile strength, yield strength and elongation versus carbon content are presented. Figures 3 and 4 show a "toughness" parameter (Z) derived from these tests. This parameter is:

$$Z = (UTS - YTS) \times Elong.$$

UTS - ultimate tensile strength
YTS - yield strength
Elong. - notched elongation

To our knowledge this parameter has never been used as a criterion in assessing the formability of sheet steel, but calculations from tension test results of Lankford et al (4) lead to an interesting correlation with actual press performance. Listed in the following table are the tensile test results for six fully killed low carbon steels. Three of these steels performed satisfactory in a car fender draw whereas three did not.

Lot	Rockwell B	Olsen	YS(KSI)	UTS(KSI)	Elong.%	Z=(UTS-YS)* ($X10^{-5}$) Elong.
Satisfactory						
A	36	444	22.7	44.6	41.5	9.10
D	30	473	17.5	40.4	39.0	8.92
E	40	460	20.8	43.3	43.5	9.80
Unsatisfactory						
B	28	445	21.7	41.5	39.0	7.73
C	27	448	19.7	40.4	40.5	8.38
F	34	450	20.3	41.5	45.0	9.37

Note - all the above data are for specimens from the rolling direction of the sheet.

It appears from these test results that satisfactory press performance is, with two exceptions, indicative of a Z value in excess of 9. It should be emphasized that these results are for unnotched specimens.

2. 0.002%C and 0.0095%C Steels

Comparing notched tensile test results for 0.002% steel in the process-annealed condition, to those of the normalized condition (Figs. D-1 and D-2), one sees that the normalizing has resulted in a drop of 11,000 psi in the yield strength, while the ultimate tensile strength and elongation has been comparatively unaffected. A similar trend is also noted in the case of the 0.0095%C material but here the yield strength was decreased 17,000 psi by normalizing and the ultimate lowered by 8000 psi. These changes are for the most part believed to be caused by excessive grain growth which took place in both grades of steel during the normalizing treatment. Figure G-1 shows the microstructure of the 0.002%C steel in the process-annealed condition and G-2 shows the same steel after normalizing. The process-annealed grain structure is fine and equiaxed whereas the non-uniformity and extreme coarseness of the normalized material made a grain size determination of these two grades impossible. Figures D-3 and D-4 show that in both the 0.002%C and 0.0095%C grades the Z parameter increased by 100% as a result of normalizing.

The results of tensile tests of a 0.005%C alloy by Morrison (11) show that a grain size increase from 8 to 110 microns in this alloy results in a decrease in lower yield strength from 40,000 psi to 19,000 psi. Although not specified in Morrison's publication, it is assumed that all tensile specimens were from the rolling direction of the sheet.

It appears that the 0.002%C and 0.0095%C alloys respond to heat treatment, after cold rolling, in a similar way to hot-worked ingot iron that has also been finished at a temperature below A_3 (15). In the present investigation, the sub-critical anneal, (670°C for 3 hours), did not result in an abnormally large grain size whereas the normalizing treatment did. In the case of ingot iron which had been hot-rolled and subsequently cold rolled, Kenyon observed complete recrystallization at temperatures below A_3 was not possible. Abnormal grain growth would be expected at temperatures above A_3, but not below, if the structure contained the equivalent of 10% deformation prior to being annealed. At 10% deformation the recrystallization temperature is well above 705°C and within the critical range of deformation which would result in abnormal grain growth at 950°C (16). This phenomenon results from the fact that a certain amount of impurities must be present for complete recrystallization (17).

Another possible explanation for the exaggerated grain growth in these alloys could be related to the number of available austenite (γ) nucleation sites upon transformation to the high temperature phase, as the ferrite grain size is generally a reflection of the austenite grain size. The rate of nucleation of γ grains is dependent on the amount of interfacial area between ferrite and cementite (18). These dilute alloys have comparatively few carbides, and hence fewer nucleation sites for the high temperature phase, which would result in a coarse grained austenite.

3. 0.028%C and 0.045%C Steels

Consider the results of the process-annealed specimens of 0.028%C and 0.045%C (Fig. D-1). A distinct drop in notch elongation was found in comparison to the two lower carbon grades. An examination of the microstructures of these two steels in both the as-received and process-annealed conditions revealed the following: a uniform grain size was present from the surface to the centre of the samples. The carbide was of the massive blocky type, fairly evenly distributed between the grain boundaries and the interior of the grains in the case of the 0.028%C material, whereas the 0.045%C sample showed a higher percentage of carbide at the grain boundaries. Figure G-3 shows the microstructure of the 0.045%C material. A slight difference was noted between the as-received and process-annealed specimens. The as-received material contained a small proportion of finely distributed spheroidized carbides which, for the most part, were coarsened by sub-critical annealing. Metallographic examination indicated that these carbides would account for no more than 10% of the carbide present in both cases. It is believed that the massive carbides are a result of the hot-working operation, in particular a high coiling temperature. The coiling temperature of these two steels was relatively high (1260°F and 1280°F). It has been shown (19) that transformation of austenite at 1300°F in low carbon steel results in massive, irregularly shaped cementite particles. In effect this indicates that part of the cooling cycle falls in the three phase ($\alpha+\gamma+$carbide) region on the T-T-T diagram (11). These particles once formed are not substantially refined by subsequent cold-rolling and sub-critical annealing.

The normalizing treatment noticeably increased the notch elongation, about 5%, and lowered the yield strength 8000 psi in both cases. Again it is evident from Figs. D-3 and D-4 that the Z parameter was doubled by the normalizing treatment.

From Fig. F-1, where grain size is plotted as a function of carbon content, it is evident that only in the case of the 0.0028%C grade could the lowering of the yield strength be in part the result of an increased grain size by normalizing, as the 0.045%C material showed essentially the same grain size after normalizing. The normalizing of the 0.028%C material resulted in a grain size increase from ASTM 9 to ASTM 8.

The microstructures of the 0.028 and 0.045%C materials after normalizing were similar. Both grades showed colonies of coarse pearlite which was often enveloped by carbide. The remainder of the carbides in both cases was mostly blocky when situated within the grains or formed bands along the grain boundaries, often extending the length of a grain boundary. In the case of the 0.045%C material these bands were noticeably refined by normalizing (i.e. decreased in thickness). A small amount of finely dispersed spheroidized carbide was also present.

Pearlite formation in low carbon steel can take place if the cooling rate through the ($\alpha+\gamma$) region is sufficiently slow that the last γ grains to transform are sufficiently enriched in carbon (16). This effect has been demonstrated in the case of 0.006%C steel by Rickett (19). A number of specimens were quickly cooled from the austenizing temperature to 1350°F and subsequently isothermally transformed at lower temperatures. In all cases the resulting structures consisted of massive carbides in a ferrite matrix with no indication of pearlite. A second series of specimens (in Rickett's work) was slowly cooled to 1350°F and then isothermally transformed at various lower temperatures. The slower cooling afforded an opportunity for formation of a larger amount of proeutectoid ferrite, and consequently an enrichment of the remaining austenite in carbon. Again, specimens annealed at 1300°F transformed to massive carbides whereas specimens transformed in the range 1200 to 1000°F were pearlitic. The pearlite was finer at lower transformation temperatures as is the case in higher carbon steels.

Figure C-4 indicates that the pearlite in the 0.045%C steel was formed at a high temperature, as it is very coarse or nodular in appearance.

The restriction of pearlite colonies to locations near the surface of the 0.028%C material has two possible explanations. First, a carbon inversion would be expected to lead to this type of a structure, though this is highly unlikely in a rimmed steel. More possibly, the cooling rate near the surface was sufficiently fast to supress the formation of an entirely massive carbide structure.

4. 0.055%C, 0.06%C, 0.065%C and 0.075%C Steels

Consider the process-annealed material in the range of carbon contents 0.055% to 0.075%C. Concerning the ultimate tensile strengths (Fig. D-1), it appears that the 0.06%C exhibits a slightly higher relative value while the 0.075%C is slightly lower than expected. The yield strengths tend towards a minimum in this region with the exception of the 0.06% material which appears about 7000 psi higher than normal. Over this region minima in notched elongations appear at the 0.055% and 0.065%C compositions while the 0.06% and 0.075% grades are substantially higher.

The abrupt increase in yield strength of the 0.06%C grade over that of the 0.055% material is probably, and for the most part, a grain size effect. The 0.06%C material has a grain size of ASTM 9.5 compared to ASTM 7.5 for the 0.055%C material. As previously mentioned, the effect of grain size on yield strength has been investigated by Morrison (14).

The "sea-saw" type behavior of the notched elongation values within this region is believed to be a result of carbide morphology and distribution. The 0.055%C and 0.065%C grades were similar in that the majority of the carbides were present primarily at the grain boundaries in the form of "mosaic" blocks, though in the 0.065%C material 10 to 20% of the carbides were present within the grains but often in close proximity to a massive carbide at the grain boundary (see Figs. G-5 and G-6). The principal difference between the morphology of these two compositions compared to the 0.028%C or 0.045%C grades is in the degree of continuity of the massive carbide along the length of the grain boundary: the lower grades in general show a more continuous mass or band. The degree of "banding" was higher in the 0.065%C material than in the 0.055%C grade, which could in part be the reason for the lower elongation of this grade. The difference in the minimum values of elongation between the 0.055% and 0.065%C steels is not readily explainable solely by differences in carbide morphology.

The 0.06%C and 0.075%C grades, showing elongation maxima, exhibited an entirely different carbide morphology and distribution. The 0.06%C grade showed coarse spheroidized carbides distributed approximately 60%-40% between the grain boundary and interior of the grains. On the other hand, the 0.075%C material contained finely dispersed spheroidized carbides equally distributed between the grains and grain boundaries. This type of structure is generally considered to result in optimum deep-drawing properties and results from maintaining the finishing temperature above A_3 coupled with a reasonably low coiling temperature.

The normalized samples of these four grades did not show as highly dissimilar structures as the process-annealed samples. In general the structures were predominantly coarse pearlite colonies with the remainder of the carbides present as "massive bands" at the ferrite grain boundaries. In general, as the carbon content increased, so did the size and number of pearlite patches. Figure G-7 shows the microstructure of the 0.075%C material. It is interesting to note that a significant decrease in yield strength resulted from normalizing the steels which had previously exhibited a fairly uniform carbide distribution (0.06%C, 0.075%C). This decrease in yield was approximately three times that observed for the two grades which showed carbide segregation at the grain boundaries (0.055%C and 0.065%C).

Both Z curves (Figs. D-3 and D-4) show a minimum within this range of carbon concentration, though the process-annealed material appears to be reaching a plateau at 0.065%C and 0.075%C.

5. 0.095%C and 0.0983%C Steels

Consider the 0.095 and 0.0983%C grades, the two grades which are of particular interest in this investigation.

It is evident from Fig. D-1 that the yield and ultimate strengths of these two grades are approximately the same in the process-annealed condition, though there is about a 2% drop in elongation in going from the 0.095% to 0.0983% material.

From Fig. F-1 it is seen that the 0.0983%C steel has a slightly larger grain size, (ASTM 9.0), compared to ASTM 9.6 for the 0.095%C material. This could account for part of the decrease in notch elongation, but of more importance is the carbide distribution of the process-annealed samples. The 0.095%C grade showed medium sized spherical carbides equally distributed between the grain boundaries and interior of the grains (Fig. G-8), whereas the 0.0983%C contained coarser carbides with approximately a 70-30% distribution between the grain boundaries and interior of the grains. Qualitatively speaking, the mean ferrite path of the randomly distributed alloy (0.095%C) would be shorter and one would expect an increased yield strength (20).

Normalizing of these two alloys resulted in a distinct lowering of the yield strength without appreciably affecting the ultimate tensile strength. The notch elongation results show that normalizing has no effect on the 0.0983%C steel while an appreciable drop (about 4%) is evident in the 0.095%C grade. The microstructures of the normalized samples were similar, exhibiting approximately 90% of the carbides present in the form of pearlite with the remaining 10% as massive cementite. The difference between the two grades was that in general the size of the pearlite colonies was smaller, and the interlamellar spacing larger, in the 0.0983%C material. Figures G-10 and G-11 show typical microstructures of the 0.095% and 0.0983%C steels respectively.

Examination of the mill data for the 0.095%C and 0.0983%C steels shows the following:

% Carbon	Finishing Temp. (°F)	Coiling Temp. (°F)
0.095	1610	1160
0.0983	1650	1250

Of particular interest as far as the form and distribution of carbide is concerned is the coiling temperature. The comparatively high coiling temperature of the 0.0983%C steel would be expected to lead to a coarser, more non-uniform carbide distribution (19). A high coiling temperature permits carbide precipitation at high temperatures resulting in sufficient time for a subsequent increase in size and decrease in number by diffusion. Carbide coarsening may also be the result of the long annealing time (30 hours), which the 0.0983%C grade received after cold reduction.

The interlamellar spacing of the pearlite in the normalized specimens is most likely an indication of cooling rate. The 0.0983%C steel had, in general, a coarser spacing which would indicate a slightly slower cooling rate below the A_3 temperature. Also, the pearlite patches in the 0.0983%C were often enveloped with cementite and more carbide "bands" appeared in this grade.

6. 0.13%C and 0.165%C Steels

The results of the notched tension tests of the two highest carbon alloys, 0.13% and 0.165%, in the process-annealed condition again emphasize the effect of grain size and carbide distribution on the notched elongation and yield strength. The 0.13%C material was extremely fine grained (ASTM 10) and a uniform distribution of carbides was evident, though the carbides were massive. No continuous bands of carbides were present. These two factors result in a comparatively high yield strength and elongation. The 0.165%C material was coarser grained (ASTM 8.5), while the microstructure showed extremely massive spheroidal carbides with a 70-30% distribution between the grain boundaries and interior of the grains with about 30% of the total carbides present as wide continuous bands at the grain boundaries. Consequently the yield strength and elongation are significantly lower than for the 0.13% material. Figure G-12 shows the general appearance of the massive carbides and bands in the process-annealed 0.165%C material.

Normalizing equalized the yield strengths of the two grades at 35,000 psi, which represents a decrease of about 12,000 psi for the 0.13%C grade. The elongation of the 0.13% material dropped 1% whereas the 0.165%C remained unchanged.

Figure G-13 shows the structure of the 0.165% normalized material. The 0.13%C grade was similar except that the degree of carbide banding was less and a higher percentage of large pearlite colonies with finer interlamellar spacing were present. Considerable grain coarsening occurred in the 0.13%C steel as a result of normalizing (ASTM 10 to 8) which was probably the cause of the greater lowering of the yield strength over that observed in the 0.165%C grade.

Again it is most likely that the carbide morphology and distribution in process-annealed specimens is a reflection of the coiling temperature, as the 0.13%C steel was coiled at 1180°F which is 30°F lower than the coiling temperature of the 0.165%C grade.

Figure E-4 shows that the Z parameter remained relatively constant in the range 0.06 to 0.013%C, for process-annealed grades, and at a significantly higher value than the 0.028% to 0.055%C range. It appears that this parameter is sensitive to carbide distribution if the carbides are spheroidized, as grades showing a low value generally had a pronounced segregation of carbides at the grain boundaries.

B. R and N Results

Appendix E contains the results of the R and N tests in tabulated and graphical form.

With respect to the accuracy of the R results it is evident from the tabulated values that there is an appreciable spread in the individual R values within a group of identical specimens. This variation is believed to be due to the method by which the gage length of the samples was machined. The Tensil-Grind Model 2066 grinder does not produce a uniform specimen width over the entire gage length. A 10% variation is not uncommon. In the calculation of the R value an average width was used which would be expected to lead to a random error. Not shown in the tabulation of the N values is the correlation coefficient, which varies from 0.98 to 0.99. The small deviation of this parameter from 1.0 indicates that there is a greater than 99% probability that the data are represented by the true stress equation (21) (see Appendix C).

It should be noted that the small elastic strain component was neglected in the calculation of the coefficient of anisotropy, which would lead to systematically lower R values, but would not affect the comparative values between the different grades of steel.

As previously mentioned it appears evident from a survey of recently published test results that optimum properties for a rimmed steel which is to undergo an industrial deep-drawing operation, (combination of varying degrees of die-drawing and stretching), are the following:

1. High normal coefficient, R_n
2. High minimum coefficient, R_{min}
3. Low planar anisotropy
4. High normal work hardening coefficient, N_n.

With the above criteria in mind the data of the R and N test are presented graphically in the following manner:

1. Figures F-1, E-2, E-5, E-6 present the individual R values for each specimen direction, (0°, 45° and 90° to the rolling direction), versus carbon content in the process-annealed and normalized conditions. From these graphs the degree of planar anisotropy is evident as well as the variation of the work hardening coefficient with direction in the plane of the sheet.
2. Figure E-3 presents the R_{min} value as a function of carbon content and heat treatment.
3. Figures E-4 and E-7 present, respectively, the normal coefficients R_n and N_n as a function of carbon concentration and heat treatment.
4. Figure E-8 pictorially presents the R_{min} - N_n criteria first introduced by Lilet (22). The four zones shown in this diagram represent the following:

Zone 1 - sheet suitable for most complex drawing operations.

Zone 2 - steels suitable for operations in which die-drawing predominates.

Zone 3 - steels suitable for operations in which stretching predominates.

Zone 4 - steels of inferior quality.

1. 0.002%C and 0.0095%C Steels

The two lowest carbon alloys, 0.002% and 0.0095%C alloys show quite a high planar anisotropy in the process-annealed condition with a minimum value at 45 degress to the rolling direction, Figure E-1. Figure E-5 shows that the N variation with respect to direction in the plane of the sheet is also quite pronounced. The effect of normalizing on the properties of these two steels was to cause a pronounced drop in the transverse R values in both cases, and a large drop in the R_0o value only in the 0.002%C grade. The R_0o and R_{min} values of the 0.0095%C grade were raised slightly by normalizing whereas the normal coefficient R_n remained constant. In the 0.002%C grade normalizing has no effect on R_{min} but resulted in a slight decrease in R_n.

It appears from the limited decrease in planar anisotropy of these two grades by normalizing that a larger amount of (110) [001] texture was retained after normalizing than Burns and Heyer (9) observed. The relative positions of the R values in the rolling direction of the two grades in the normalized condition would lead one to believe that a higher degree of (100) [001] was developed in the 0.0095%C alloy than in the 0.002%C alloy. The most prominent effect of normalizing on the properties of these two grades was in the values of the work hardening coefficient. The variation of N with direction was reduced and Figure E-7 shows that N_n was increased by about 0.04 in both cases. Figure E-8 shows that both grades in the process-annealed condition fell in Zone 3, (steel suitable for operations in which stretching predominates), and remained in this zone after normalizing, though the increased N_n values would indicate superior stretching properties could be expected. As already mentioned, both these steels exhibited excessive grain growth during normalizing (Fig. G-2). Morrison (14) has studied the effect of grain size on the work hardening coefficient and has derived an empirical relationship of the form:

$$N_a = \frac{5}{10+d} - 1/2$$

d = grain diameter in millimeters

N_a = work hardening coefficient

This relationship is independent of carbon content if the steel exhibits "single n behavior", i.e. a constant n value over the elongation range 0% to 20%. It is believed that all our specimens exhibited this behavior as shown by the 99% confidence level implied by the correlation coefficient. Therefore, the increased N_n values of the 0.002%C and 0.095%C materials are most likely due to grain coarsening during normalizing.

2. 0.028%C and 0.045%C Steels

Consider the 0.028%C and 0.045%C grades. In the process-annealed condition, the 0.028%C material exhibited the highest planar anisotropy of the twelve grades investigated, while that of the 0.045%C material was somewhat less (Fig. E-1). The normal coefficients (R_n) for the two grades (Fig. E-4) were identical, whereas the R_{min} was slightly higher in the 0.045%C grade. The N_n value of the 0.045%C steel was appreciably higher. This higher N_n of the 0.045%C material could be due to the larger grain size of this grade (ASTM 7.5) as opposed to ASTM 9.0 for the 0.028%C grade. The as-received material of these two grades showed identical grain sizes. The larger grain size of the 0.045%C grade after process-annealing at this laboratory could be a result of the degree of temper rolling. An examination of the mill data for these two grades shows that the 0.045%C steel received a 1.2/1.6% reduction as opposed to a .8/1.07% temper roll for the 0.028%C grade. Based on the R_{min}--N_n criterion of Lilet, both these steels fell into Zone 3 in the process-annealed condition, though the higher N_n value of the 0.045%C material should result in superior behavior in a forming operation in which stretching predominates.

These two grades responded quite differently to normalizing. Figure E-2 shows that the planar anisotropy of both steels was reduced and in the case of the 0.028%C material a dramatic increase in the R_{45^o} value is evident. This value increased from 0.90 in the process-annealed condition to 1.30 in the normalized condition, exceeding the R_{0^o} value in the normalized condition. This means that the minimum R value is no longer at 45 degrees to the rolling direction but now occurs in the rolling direction. Figure E-3 shows that normalizing substantially increased the R_{min} value in the 0.028%C material and had no effect on the 0.045%C material. Figure E-4 shows that the R_n of the 0.028%C material was raised by normalizing while the R_n of the 0.045%C grade was slightly lowered. The R behavior as a result of normalizing would imply that a portion of the $(\bar{1}11)$ [110] texture was retained in both grades with a higher degree of (100) [001] developed by normalizing the 0.028%C grade.

As far as the work hardening coefficient is concerned, Figs. E-5, E-6 and E-7 show that a substantial increase in N was achieved by normalizing the 0.028%C grade, whereas the increase was less in the 0.045% material but essentially no variation in N with sheet direction was noted in this case (Fig. E-6). Figure E-8 shows that the 0.045%C material still remains in Zone 3 but with an increased N value which is not offset by a lowering of R_{min} . The 0.028%C material, after normalizing, has properties falling within Zone 1, which means this steel is suitable for the most complex deep-drawing operations. Concerning the location of the R_{min} value with respect to direction in the plane of the sheet, observed in the 0.028%C grade, Lilet and Wybo (5) have shown that two steels having identical R_n and R_{min} values, but different planar anisotropies and location of R minimas, react differently in an asymmetrical pressing operation. The steel with the lower planar anisotropy coupled with an R value progressively increasing from 0 degrees to 90 degrees, relative to the rolling direction, exhibits better deep-drawing performance.

It is evident from the plot of grain size versus carbon content (Fig. E-1) that the increased N_n values are not entirely due to grain coarsening by normalizing, as the 0.045%C grade showed slightly smaller grain size in the normalized condition though it is most likely a contributing factor in the substantial N_n increase noted in the 0.028%C material.

3. 0.055%C, 0.06%C, 0.065%C and 0.075%C Steels

Consider the grades within the carbon range 0.055 to 0.075%C inclusive. In the process-annealed conditions all grades appear to have roughly the same planar anisotropy with the exception of the 0.065%C grade. This steel shows a slightly larger $R_{max} - R_{min}$ value. From Fig. E-3 it is seen that the R_{min} values are equivalent in all cases, again except the 0.065%C grade which is substantially lower. The normal coefficients R_n of the 0.06 and 0.075%C steels are slightly higher than the 0.055 and 0.065%C grades respectively. The work hardening coefficient N_n is a minimum at 0.06%C with the 0.065 and 0.06%C grades showing substantially lower values than the 0.055 and 0.075%C grades. Figure E-8 shows that the three of these four grades which are tested in the as-received condition all fell in Zone 4 corresponding to steel of inferior quality for deep-drawing. After process-annealing in this laboratory all four grades fell in Zone 3. Although it is doubtful that the annealing texture, as indicated by the R values, is related to carbide morphology it is interesting to note that the two grades (0.06 and 0.075%C) which had a uniform distribution of spheroidized carbides between the grain boundaries and interior of the grains showed slightly higher R_n and R_{min} values than the two grades showing carbide segregation at the grain boundaries (0.055%C and 0.065%C). (See photomicrographs G-5 and G-6). The reason for the very low R_{min} value and comparatively high planar anisotropy of the 0.065%C material is not immediately apparent from the hot-mill data as the coiling temperatures of all grades were essentially constant. There is, however, a difference in the cold-mill practice. The 0.065%C material received a 45% reduction during cold rolling as opposed to about 35% for the other grades in this range. Also this grade was open-coil annealed whereas the remaining three grades were batch annealed. All things being equal, one would expect a more pronounced texture development, as evidenced by the high planar anisotropy, the greater the degree of cold work. The low N_n coefficient for the 0.06%C material is believed to be a result of the comparatively fine grain size, (ASTM 9.5). The low N_n value of the 0.065%C material is not readily explainable as this steel showed identical grain size to the 0.075%C material, which had an appreciably higher N_n.

These four steels in the middle of the range investigated reacted quite differently to the normalizing treatment. The planar anisotropy was radically reduced in all cases except the 0.065%C material as shown in Figs. E-1 and E-2. It is also evident that the minimum R value, in all cases but the 0.065%C steel, occurred in the rolling direction after normalizing. (The 0.065%C grade exhibited a minimum R at 45 degrees in the process-annealed condition, as did the three other alloys in this range). A slight increase in

R_{min} was evident in the 0.055%, 0.06% and 0.075%C grades, while the 0.065%C grade showed a more substantial increase. The R_n values, on the other hand, showed a reverse behavior with all three grades, 0.065%C exempted, showing substantially lower values. The 0.065%C grade R_n remained constant. The normal work hardening coefficient showed nominal improvement by normalizing of the 0.055%C and 0.065%C grades, whereas the 0.075%C grade showed a slightly larger improvement.

The 0.06%C steels' N_n value was substantially raised from 0.230 to 0.290. Figure E-8 shows that all four grades remained in Zone 3 after normalizing. The substantial increase in R_{min} achieved by normalizing the 0.065%C material would undoubtedly make this grade more amenable to die-drawing. The location of the R_{min} value of the 0.055%, 0.06% and 0.075%C materials indicates that in these three grades a sizable proportion of the $(\bar{1}11)$ [110] preferred orientation remained after normalizing. The anomalous behavior of the 0.065%C alloy is believed to be associated with a (110) [001] texture which Burns and Heyer found to be completely removed by normalizing. It is interesting to note that the three grades which were open-coil annealed (0.002%, 0.0095% and 0.065%C), all showed evidence of a high degree of retention of the (110) [001] texture after normalizing. The relative decrease in N_n for the four medium carbon grades are directly proportional to the grain coarsening resulting from normalizing. Refer to Figs. E-7 and F-1.

4. 0.095%C and 0.0983%C Steels

The R values of the 0.095 and 0.0983% grades in the process-annealed condition, show that the 0.095%C material has a higher planar anisotropy, and slightly higher R_{min} and R_n values. These higher R values of the 0.095%C material are believed to be the result of the lower coiling temperature employed in the production of this grade. The effect of coiling temperature on the carbide morphology was described previously (refer to photomicrographs G-8 and G-9). The lower N_n value of the 0.095%C grade, in the process-annealed condition, is a reflection of the finer grain size of this grade. Both steels, in the as-received condition, were in Zone 4 of the Lilet plot and after process-annealing, the 0.0983% grade was elevated to Zone 3, while the 0.095%C grade remained in Zone 4.

Comparison of graphs E-1 and E-2 for these two grades shows that normalizing has effectively nullified the planar anisotropy in both cases. Figure E-3 indicates substantial gains have been realized in the R_{min} values of both grades. The R_{min} increase is offset in the 0.095%C grade by a proportionate decrease in the normal coefficient R_n, while the 0.098%C grade shows little R_n decrease. Figure E-7 shows an appreciable N_n increase in both grades due to normalizing and Fig. E-8 shows that both grades have deep-drawing properties lying within Zone 3. It is also evident from Fig. E-8 that both normalized grades have properties definitely superior to process-annealed 0.028% grade. In fact, according to this criterion, both the 0.095% and 0.0983%C grades should outperform the process-annealed 0.028%C material in an operation where stretching predominates, and the 0.095%C grade should perform better than the process-annealed 0.028%C grade in a deep-drawing test where either die-drawing or stretching predominates. The grouping of the R values about 1.0 for these two higher carbon grades would imply that the texture changes observed by Burns and Heyer (9) for normalized rimmed steel took place in these two grades, i.e. the (110) [001] texture was completely removed while a small amount of $(\bar{1}11)$ [110] preferred orientation remained and a low degree of (100) [001] texture developed. Again, the increased N_n values observed in normalizing were most likely due to grain coarsening.

5. 0.13%C and 0.165%C Steels

The two highest carbon grades, 0.13% and 0.165%C, in the process-annealed condition show slightly different R behavior. The planar anisotropy of the 0.13%C grade is higher but the 0.165%C material showed distinctly lower R_{min} and R_n values. The lower R_{min} and R_n of the 0.165%C grade was most likely due to the high coiling temperature employed. The microstructural differences between the two grades described earlier are again believed to be the result of coiling temperature differences. The N_n value was lower for the 0.13% grade, which again is indicated by the finer grain size of this material. Both steels fell into Zone 3 in the process-annealed condition. The as-received 0.165%C material was situated in Zone 4. The normalizing treatment of these two grades has, as usual, decreased the planar anisotropy and altered the relative positions of the maximum and minimum R values with respect to direction in the plane of the sheet. The 0.13%C steel shows a R_{min} in the rolling direction, slightly lower than the process-annealed value. The 0.165%C material in the normalized condition shows a minimum R in the transverse direction, appreciably improved over the process-annealed value. These observations imply a retention of the (100) [001] texture in the 0.13%C grade with a $(\bar{1}11)$ [110] retention in the 0.165%C grade superimposed on a more pronounced (100) [001] texture. The R_n value was substantially increased in the 0.13%C grade by normalizing and slightly lowered in the 0.165%C grade. The N_n value was greatly increased by normalizing the 0.13%C material and slightly lowered in the 0.165%C grade. The 0.165%C grade is the only material in which normalizing resulted in a decreased N_n value. The raising of N_n in the 0.13%C steel is again a result of grain coarsening whereas the slight decreased N_n in the 0.165%C is not readily explainable on the basis of grain size, as both the process-annealed and normalized grades showed identical grain sizes. Figure E-8 shows that both steels also report to Zone 3.

IV SUMMARY AND CONCLUSIONS

The results of the notched tension tests indicate that notch elongation is very sensitive to carbide morphology in process-annealed materials. High notch elongation values are an indication of spheroidized carbides equally distributed between the grain boundaries and interior of the grains. The carbide morphology and distribution are closely linked with the finishing and coiling temperatures. A high finishing temperature coupled with a low coiling temperature generally results in a uniform distribution of spheroidized carbides (0.0095%, 0.002%, 0.075%, 0.095% and 0.13%C grades). A low finishing temperature and high coiling temperature leads to a segregation of carbides at the grain boundaries and severe coarsening. This type of microstructure is associated with low notch elongation values (0.055%, 0.045%, 0.065%, 0.0983%, 0.165%C). Normalizing of all grades resulted in a partially pearlitic structure and slight grain coarsening in most grades. The carbides not present in pearlite were generally of the massive type. The grain coarsening is believed to result in a general lowering of the yield strengths.

As the finishing temperature has a pronounced effect on the coefficient of anisotropy, it is not surprising that the general behavior of the minimum R values closely resembles trends shown in the notched elongation curves for the process-annealed steels. In general, the process-annealed materials showed a high degree of planar anisotropy. It is general practice to control the work-hardening coefficient by the degree of temper-rolling. As specimens were sub-critically annealed at this laboratory prior to testing, the N values were now dependent on the resultant grain size, which in some cases was associated with the extent of the temper-roll. The effect of normalizing was to decrease the planar anisotropy in most grades and increase the work hardening coefficient. The N increase results directly from slight grain coarsening.

The two carbon concentrations of particular importance in this investigation, 0.095%C and 0.0983%C, showed dramatic improvements in deep-drawing properties as a result of normalizing. Both grades in the normalized condition showed superior R and N values to the 0.028%C grade in the process-annealed condition. The planar anisotropy is less, the R_{min} is equal to that of the 0.028%C grade in one case and higher in the other. Although there is somewhat lower R_n values in the higher carbon grades, it has been demonstrated that this parameter alone is not a good criterion for industrial deep-drawing applications.

The "Z" parameter, though useful as a general indicator of formability, is not felt to be sensitive enough to adequately discriminate between good and poor steels of nearly the same carbon concentration for deep-drawing.

V RECOMMENDATIONS FOR FURTHER STUDY

The normalizing treatment applied to the two 0.1%C steels resulted in significant improvements in parameters commonly indicative of deep-drawing behavior. This particular heat treatment of 1/2 hour at 925°C followed by a simulated air cool did not result in all the carbides being present in the form of pearlite. Massive carbides were present in most cases. It is recommended that a study be initiated to find the optimum normalizing conditions.

Second, texture determinations were not undertaken in the present investigation, but possible preferred orientations resulting from normalizing were inferred from the R values. It was hypothesized that a different preferred orientation was developed from normalizing steels which had been previously open-coil annealed. This texture determination could be of more than academic interest as a higher degree of planar anisotropy is maintained in this case.

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APPENDIX A

MILL DATA FOR MATERIAL SUPPLIED BY DOMINION FOUNDRIES AND STEEL LTD.

TABLE A-1. CARBON ANALYSIS

Grade	Heat Treatment	Wt. Pct. Carbon
0.095	As-received	0.08
	Process-annealed	0.07
	Normalized	0.08
0.098	As-received	0.09
	Process-annealed	0.09
	Normalized	0.09
0.028	As-received	0.03
	Process-annealed	0.03
	Normalized	0.03
0.055	As-received	0.06
	Process-annealed	0.06
	Normalized	0.06

TABLE A-2. SUMMARY OF MILL DATA

Carbon Content wt. pct.	Finish	Coiling	Anneal	
	Temp.(°F)	Temp.(°F)	Time(hrs.)	Temp.(°F)
0.165	1635	1250	30	1300
0.13	1600	1180	--	--
0.0983	1650	1250	35	1300
0.095	1610	1160	6	1270
0.075	---- NOT AVAILABLE ----			
0.065	1590	1150	3	1320
0.06	1610	1150	30	1300
0.055	1570	1130	4	1250
0.045	1630	1280	4	1300
0.028	1600	1260	30	1300
0.0095	1550	1130	2	1300
0.002	1600	1145	2	1300

GENERAL INFORMATION

Grade	TC150
Mill Number	27326B
Order Number	7495-12
Serial	74200
Heat	46847

MELT SHOP PRACTICE

Turndown Temperature	2940°F
Cooling Practice	Nil
Ladle Additions	1200* 80% FeMn, 450* Coke
Mould Additions	Nil

HOT MILL PRACTICE

Hot Band Gauge	.098
Finish Temperature	16350°F
Coil Temperature	1250°F

COLD MILL PRACTICE

Cold Roll Gauge	.0344
Anneal Cycle	Lot 639 - Batch #1 - 1300°F - 30 hrs.
Surface Finish	Code #2R
Temper Mill Elongation	.75/1.0%
Temper Mill Profile T	35-45 micro inches
B	35-45 micro inches

TEST RESULTS

		C	Mn	P	S	Cu	Ni	Cr	Si	Al	Sn	N
Analysis	Ladle	.155	.43	.004	.029	.04	.03	.02	.006	--	.007	--
	Check	.165	.29	.003	.033	.04	.03	.01	TR	.004	--	.0045

Micros	Cleanliness	D3.8	Alumina
	Grain Size	7.1 equiaxed	Fine Carbides

PHYSICAL TESTS

Gauge	.0348
Yield Point	36,000
Tensile Strength	52,600
% Elongation	33.5
Rockwell	51B

GENERAL INFORMATION

Grade	TC160
Mill Number	10208B
Order Number	BC4679-5
Serial	95189
Heat	53208

MELT SHOP PRACTICE

Turndown Temperature	2900°F
Cooling Practice	2000* Lime
Ladle Additions	1200* 8090 FeMn - 8* Al - 315* Coke
Mould Additions	45 oz. Al

HOT MILL PRACTICE

Hot Band Gauge	.125
Finish Temperature	1600°F
Coil Temperature	1180°F

COLD MILL PRACTICE

Cold Roll Gauge	.0359
Anneal Cycle	Batch #1
Surface Finish	Code 2R
Temper Mill Elongation	.55-.75%
Temper Mill Profile T	40-50 Micro Inches
B	40-50 Micro Inches
Temper Mill Date	March 26/68 - 8 a.m.

TEST RESULTS

		C	Mn	P	S	Cu	Ni	Cr	Si	Al	Sn	N
Analysis	Ladle	.15	.32	.004	.014	.025	.025	.01	.004	--	.001	--
	Check	.13	.22	.003	.013	.03	.01	.01	.01	.004	--	.0027

Micros	Cleanliness	13 .15% B3.4 Fine-minute Fe_3C
	Grain Size	8.1 eq.

PHYSICAL TESTS

Gauge	.0368
Yield Point	32,000 P.S.I.
Tensile Strength	50,400 P.S.I.
% Elongation	35.0%
Rockwell	54B

GENERAL INFORMATION

Grade	BC021E
Mill Number	28411B
Order Number	7802-3
Serial	73580
Heat	45845

MELT SHOP PRACTICE

Turndown Temperature	2980°F
Cooling Practice	Nil
Ladle Additions	1000* 80% FeMn, 30* Al
Mould Additions	211 oz. Al

HOT MILL PRACTICE

Hot Band Gauge	.100
Finish Temperature	1650°F
Coil Temperature	1250°F

COLD MILL PRACTICE

Cold Roll Gauge	.0359
Anneal Cycle	Lot 1099 - Batch 2C - 1300°F - 35 hrs.
Surface Finish	Code #2R
Temper Mill Elongation	1.45/1.6%
Temper Mill Profile T	55-65 Micro Inches
B	60-70 Micro Inches

TEST RESULTS

		C	Mn	P	S	Cu	Ni	Cr	Si	Al	Sn	N
Analysis	Ladle	.10	.28	.005	.014	.025	.02	.01	.005	--	.002	--
	Check	.098	.28	.003	.015	.025	.015	.009	.004	.008	--	.0032

Micros	Cleanliness	Al, 4	Fine Iron Carbide
	Grain Size	8.4 Equiaxed	

PHYSICAL TESTS

Gauge	.0372
Yield Point	32,000
Tensile Strength	44,900
% Elongation	38.5
Rockwell	48B

GENERAL INFORMATION

Grade	BC021
Mill Number	15072B
Order Number	8165-10
Serial	31840
Heat	44174

MELT SHOP PRACTICE

Turndown Temperature	3010°F
Cooling Practice	2000* Limestone
Ladle Additions	80% FeMn - 1250*, 42* Al
Mould Additions	230 oz. Al wire

HOT MILL PRACTICE

Hot Band Gauge	.100
Finish Temperature	1610°F
Coil Temperature	1160°F

COLD MILL PRACTICE

Cold Roll Gauge	.0359
Anneal Cycle	Lot 172 - Batch 2C - 1260/1270°F - 6 hrs.
Surface Finish	Code #2R
Temper Mill Elongation	1.19%
Temper Mill Profile T	75-85 micro inches
B	80-90 micro inches

TEST RESULTS

		C	Mn	P	S	Cu	Ni	Cr	Si	Al	Sn	N
Analysis	Ladle	.08	.36	.004	.018	.03	.02	.02	.004	--	.003	--
	Check	.095	.37	.003	.021	.055	.03	.015	.015	.015	---	.004

Micros	Cleanliness	A2, 4 - numerous fine carbides
	Grain Size	7.6 equiaxed

PHYSICAL TESTS

Gauge	.0364
Yield Point	36,300
Tensile Strength	50,700 Y.P.E.
% Elongation	33.5
Rockwell	45, 47, 53B

GENERAL INFORMATION

Grade	TC0210
Mill Number	11984B
Order Number	6260-9
Serial	14034
Heat	34736

MELT SHOP PRACTICE

Turndown Temperature	2900°F
Cooling Practice	Nil
Ladle Additions	1600# 80% FeMn, 44# Al
Mould Additions	186 oz. Al

HOT MILL PRACTICE

Hot Band Gauge	.080
Finish Temperature	1590°F
Coil Temperature	1150°F

COLD MILL PRACTICE

Cold Roll Gauge	.0359
Anneal Cycle	Lot 9274 - Open Coil - 1320 - 3 hrs.
Surface Finish	Code #2R
Temper Mill Elongation	1.1%
Temper Mill Profile T	65-75 Micro Inches
B	60-70 Micro Inches

TEST RESULTS

		C	Mn	P	S	Cu	Ni	Cr	Si	Al	Sn	N
Analysis	Ladle	.07	.32	.006	.016	.07	.02	.02	.003	--	.008	--
	Check	.065	.30	.005	.018	.06	.02	.01	TR	.017	--	.0031

Micros	Cleanliness	A2, 6 D1, 8 Medium Carbides
	Grain Size	7 - 9 equiaxed

PHYSICAL TESTS

Gauge	.0383
Yield Point	28,600
Tensile Strength	43,000
% Elongation	37.0
Rockwell	39B

GENERAL INFORMATION

Grade	TC021
Mill Number	14075B
Order Number	5205-2
Serial	4602
Heat	56408

MELT SHOP PRACTICE

Turndown Temperature	2910°F
Cooling Practice	8000# Limestone
Ladle Additions	1300# - 80% FeMn - 11#Al
Mould Additions	268 oz. Al

HOT MILL PRACTICE

Hot Band Gauge	.125
Finish Temperature	1610°F
Coil Temperature	1150°F

COLD MILL PRACTICE

Cold Roll Gauge	.0359	
Anneal Cycle	Batch #1	30 hrs. and 1300°F
Surface Finish	Code 2R	
Temper Mill Elongation	.6 -.75%	
Temper Mill Profile T	40-50 Micro Inches	
B	40-50 Micro Inches	
Temper Mill Date	April 8/68 - 3 a.m.	

TEST RESULTS

		C	Mn	P	S	Cu	Ni	Cr	Si	Al	Sn	N
Analysis	Ladle	.075	.31	.005	.018	.03	.02	.01	.001	--	.001	--
	Check	.06	.34	.005	.017	.027	.027	NIL	NIL	.01	--	.0025

Micros	Cleanliness	.125% 6 B2, 4 Fine Minute Fe_3C
	Grain Size	9.3 eq.

PHYSICAL TESTS

Gauge	.0352
Yield Point	26,700 P.S.I.
Tensile Strength	49,000 P.S.I.
% Elongation	37.0%
Rockwell	47B

GENERAL INFORMATION

Grade	BC031E
Mill Number	15317B
Order Number	6124-15B
Serial	84037
Heat	33473

MELT SHOP PRACTICE

Turndown Temperature	3030°F
Cooling Practice	8000# Limestone
Ladle Additions	700# 80% FeMn, 40# Al wire
Mould Additions	220 oz. Al wire

HOT MILL PRACTICE

Hot Band Gauge	.111
Finish Temperature	1570°F
Coil Temperature	1130°F

COLD MILL PRACTICE

Cold Roll Gauge	.0350
Anneal Cycle	Lot 12 - Batch 2C - 1250°F - 4 hrs.
Surface Finish	Code #1R
Temper Mill Elongation	.5/.6%
Temper Mill Profile T	50-60 Micro inches
B	50-60 Micro inches

TEST RESULTS

		C	Mn	P	S	Cu	Ni	Cr	Si	Al	Sn	N
Analysis	Ladle	.08	.36	.006	.016	.06	.025	.02	--	--	.006	--
	Check	.055	.34	.001	.004	.055	.04	.025	.034	.012	--	--

Micros	Cleanliness	- A2, 4 Numerous course carbides
	Grain Size	- 6.5 equiaxed

PHYSICAL TESTS

Gauge	.0372
Yield Point	31,300
Tensile Strength	45,600 Y.P.E.
% Elongation	37.0
Rockwell	40, 41, 44B

GENERAL INFORMATION

Grade	TU0210
Mill Number	34241B
Order Number	7674-3
Serial	84308
Heat	47680

MELT SHOP PRACTICE

Turndown Temperature	2975°F
Cooling Practice	8000# Limestone
Ladle Additions	300# 80% FeMn, 7# Al, 500# Lime, 400# Low Carb. Mn.
Mould Additions	322 oz. Al

HOT MILL PRACTICE

Hot Band Gauge	.100
Finish Temperature	1630°F
Coil Temperature	1280°F

COLD MILL PRACTICE

Cold Roll Gauge	.0359
Anneal Cycle	Lot 1166, Batch 2C - 1300°F - 4 hrs.
Surface Finish	Code #1R
Temper Mill Elongation	1.2/1.6%
Temper Mill Profile T	75-85 micro inches
B	85-95 micro inches

TEST RESULTS

		C	Mn	P	S	Cu	Ni	Cr	Si	Al	Sn	N
Analysis	Ladle	.06	.35	.005	.015	.02	.015	.01	.005	--	.001	--
	Check	.045	.31	.008	.015	.02	.01	.01	TR	TR	--	.0037

Micros	Cleanliness	A2, 4 Medium-Coarse Carbides
	Grain Size	7.4 equiaxed

PHYSICAL TESTS

Gauge	.0365
Yield Point	26,350
Tensile Strength	43,000
% Elongation	37.0
Rockwell	39B

GENERAL INFORMATION

Grade	TU0210
Mill Number	73572H
Order Number	3948-1
Serial	81439
Heat	44855

MELT SHOP PRACTICE

Turndown Temperature	2960°F
Cooling Practice	1000# Limestone
Ladle Additions	500# 80% FeMn, 24# Al Bar, 500# Mn
Mould Additions	394 oz. Al

HOT MILL PRACTICE

Hot Band Gauge	.100
Finish Temperature	1600°F
Coil Temperature	1260°F

COLD MILL PRACTICE

Cold Roll Gauge	.033
Anneal Cycle	Lot 1107 - Batch 2C - 1300°F - 30 hrs.
Surface Finish	Code #2R
Temper Mill Elongation	.8/1.0%
Temper Mill Profile T	38-42 Micro Inches
B	40-45 Micro Inches

TEST RESULTS

		C	Mn	P	S	Cu	Ni	Cr	Si	Al	Sn	N
Analysis	Ladle	.065	.30	.003	.014	.05	.02	.01	.004	--	.007	--
	Check	.028	.34	.006	.015	.03	.02	TR	.01	TR	--	.0041

Micros	Cleanliness	A1, 4 D3, 4
	Grain Size	7.5 Equiaxed Coarse Carbides

PHYSICAL TESTS

Gauge	.0322
Yield Point	32,200
Tensile Strength	44,000 Y.P.E.
% Elongation	37.5
Rockwell	44B

GENERAL INFORMATION

Grade	TC0210
Mill Number	28929B
Order Number	7298-1
Serial	79X530
Heat	45643

MELT SHOP PRACTICE

Turndown Temperature	2980°F
Cooling Practice	2000# Pellets Added
Ladle Additions	1100# 80% FeMn, 19# Al
Mould Additions	478 oz. Al

HOT MILL PRACTICE

Hot Band Gauge	.100
Finish Temperature	1550°F
Coil Temperature	1130°F

COLD MILL PRACTICE

Cold Roll Gauge	.0359
Anneal Cycle	Lot 9683 - Open Coil Decarb - 1300°F - 2 hrs.
Surface Finish	Code #1R
Temper Mill Elongation	.8/.9%
Temper Mill Profile T	75-85 Micro Inches
B	75-85 Micro Inches

TEST RESULTS

		C	Mn	P	S	Cu	Ni	Cr	Si	Al	Sn	N
Analysis	Ladle	.065	.28	.004	.011	.02	.02	.01	.007	--	.002	--
	Check	.0095	.33	.003	.011	.023	.014	.009	.003	.001	--	.0036

Micros	Cleanliness	A1, 4 B2, 6
	Grain Size	8.0 Equiaxed

PHYSICAL TESTS

Gauge	.0366
Yield Point	32,000
Tensile Strength	44,900
% Elongation	38.5
Rockwell	44B

GENERAL INFORMATION

Grade	TC021
Mill Number	6990B142
Order Number	1585-15
Serial	22312
Heat	43198

MELT SHOP PRACTICE

Turndown Temperature	2940°F
Cooling Practice	Nil
Ladle Additions	1000# 80% FeMn, 18# Al
Mould Additions	364 oz. Al wire

HOT MILL PRACTICE

Hot Band Gauge	.098
Finish Temperature	1600°F
Coil Temperature	1145°F

COLD MILL PRACTICE

Cold Roll Gauge	.0359
Anneal Cycle	Lot 9086 - Open Coil Decarb - 1300 - 2 hrs.
Surface Finish	Code #1R
Temper Mill Elongation	.9%
Temper Mill Profile T	60-70 Micro Inches
B	65-75 Micro Inches

TEST RESULTS

		C	Mn	P	S	Cu	Ni	Cr	Si	Al	Sn	N
Analysis	Ladle	.07	.32	.004	.016	.05	.03	.015	.003	--	.004	--
	Check	.002	.31	.003	.012	.06	.03	.01	.008	.014	--	--

Micros	Cleanliness	- A2, 4 - no carbides
	Grain Size	- 6.7 equiaxed

PHYSICAL TESTS

Gauge	.0363
Yield Point	28,800
Tensile Strength	44,100 Y.P.E.
% Elongation	38.0
Rockwell	49-53B

APPENDIX B

FURNACE COOLING CURVES FOR HEAT-TREATED SPECIMENS

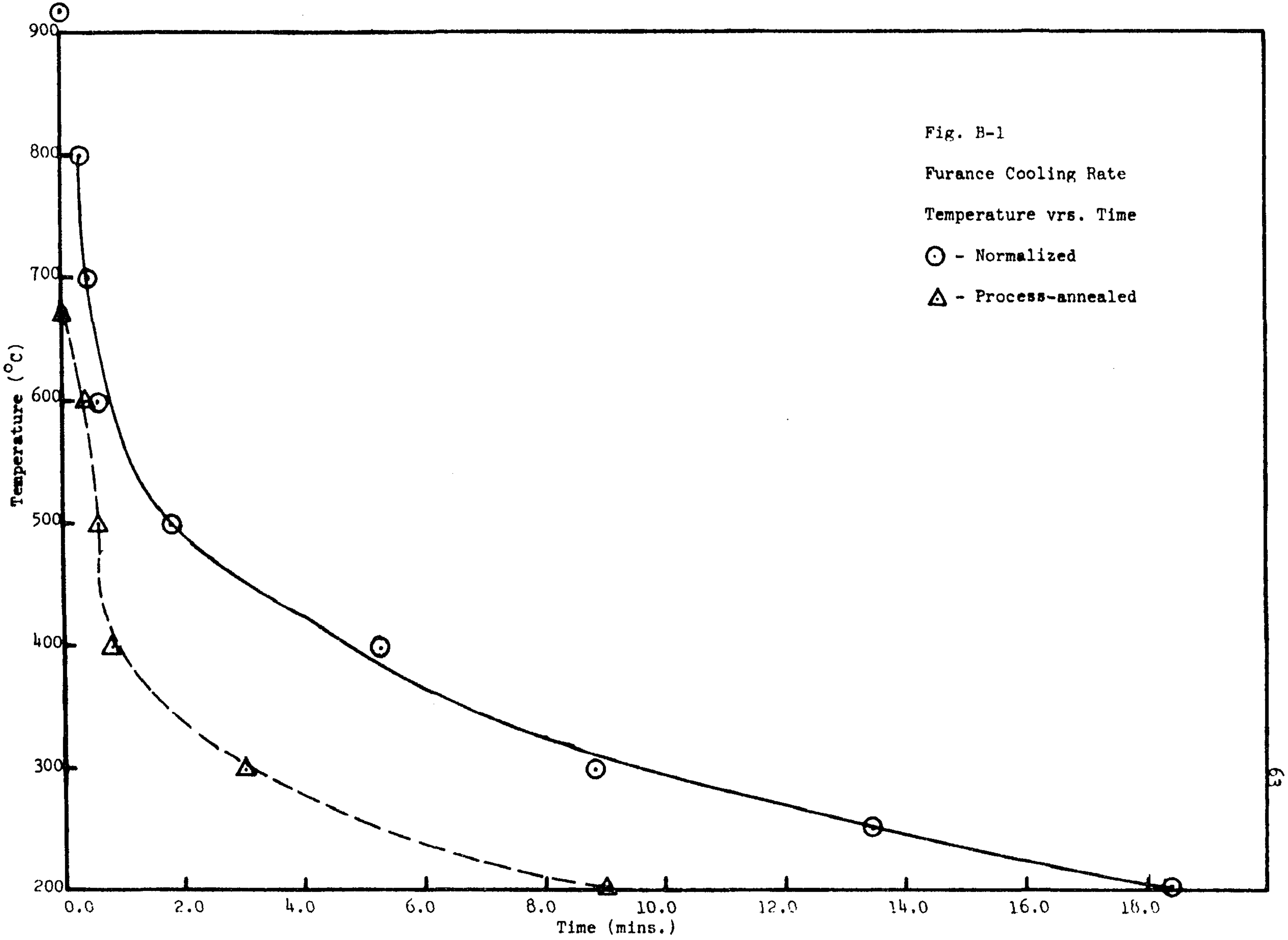


Fig. B-1

Furance Cooling Rate

Temperature vrs. Time

APPENDIX C

EQUATIONS AND COMPUTER PROGRAMS FOR CALCULATION OF COEFFICIENT OF ANISOTROPY AND WORK HARDENING COEFFICIENT

DEFINITION AND FORMULAE

A. Coefficient of Anisotropy

The coefficient of anisotropy is equal to the ratio of the rational deformations in the two perpendicular directions to principal stress, i.e. in the width (w) and thickness (e):

$$(1) \quad R = \frac{\varepsilon_w}{\varepsilon_e} = \frac{\mathrm{Log}\ (w_o/w)}{\mathrm{Log}\ (e_o/e)}$$

- subscript "o" represents the initial state

Log - Naperian logarithm

log - decimal logarithm

- assuming the volume of metal remains constant during the test equation (1) reduces to the following:

$$(2) \quad R = \frac{\log\ (w_o/w)}{\log\ (lw/l_o\, w_o)}$$

l = specimen gage length

w = specimen width

$$(3) \quad R_n = 1/4\ (R_{0^\circ} + 2\ R_{45^\circ} + R_{90^\circ})$$

R_n = normal coefficient of anisotropy

B. Work Hardening Coefficient

The work hardening coefficient is defined as the exponent of the empirical formula $\sigma = k\ \varepsilon^n$ which accounts for the shape of the tensile curves of mild steels and some other materials.

- by definition:

(4) $\sigma = \frac{P}{S} = \frac{\text{load}}{\text{true section}}$ (true stress)

(5) $\varepsilon = \text{Log}\ \frac{l}{l_o}$ (rational elongation)

- taking the logarithms of each side of equation (4) gives:

(6) $\log \sigma = \log\ (\frac{P}{S})$

- assume the volume of metal remains constant during the test equation (6) becomes:

(7) $\log \sigma = \log P - \text{Log}\ S_o + \log \frac{l}{l_o}$

- also

(8) $\log \varepsilon = \log \text{Log}\ (l/l_o)$

S_o = original cross-sectional area

l = gage length at any time

l_o = original gage length

- consider the equation:

$\sigma = k\ \varepsilon^n$

taking logarithms one obtains:

$\log \sigma = n \log \varepsilon + \log k$

Therefore if one plots log σ versus log ε the slope of the curve will be the work hardening coefficient (N). For this investigation log σ was calculated using equation (7) and log ε was calculated by equation (8).

$$N_n = 1/4\ (N_{0^o} + 2\ N_{45^o} + N_{90^o})$$

N_n - normal work hardening coefficient

COMPUTER PROGRAM FOR COEFFICIENT OF ANISOTROPY (R)

```
ZZJOB 5                                   REG HAMILTON
ZZFORX5
C     R-COEFFICIENT OF ANISOTROPY
C     N-NUMBER OF SPECIMENS IN THE TEST
C     M-NUMBER OF X-SECTIONAL MEASUREMENTS PER SPECIMEN
C     MN-TOTAL NUMBER OF TESTS
      DIMENSION OW(100), FW(100), R(100)
      PUNCH 89
   89 FORMAT (13H$REG HAMILTON)
      PUNCH 82
   82 FORMAT (23HR-VALUE DOFASCO PROJECT,/)
      READ 3,NN
      KK=0
    7 READ 1,IT,N,M
      KK=KK+1
      PUNCH 85,IT
   85 FORMAT (12HTEST NUMBER ,13,/)
      K1=M
      K2=N
      K=K1*K2
      M=K1
      N=K2
      READ 2, (OW(I),I=1,K)
      READ 2, (FW(I),I=1,K)
      I=1
      ICNT=0
      DO 20 J=1,N
      SUM1=0.0
      SUM2=0.0
      I 1=M


      T=I 1
      M=I 1
   10 SUM1=SUM1+OW(I)
      SUM2=SUM2+FE(I)
      1CNT=ICNT+1
      IF (ICNT-M)    11,12,11
   11 I=I+1
      GO TO 10
   12 AVI=SUM1/T
      AV=SUM2/T
      I=I+1
      ICNT=0
   20 R(J)=LOGF(AVI/AV)/LOGF ( (1.2*AV)/AVI)
      PUNCH 84
   84 FORMAT (19HINDIVIDUAL R VALUES)
      PUNCH 2, (R(J),J=1,N)
      L1=N
      S=L1
      N=L1
      SUM=0.0
      DO 30 I=1,N
```

```
30 SUM=SUM+R(I)
   AVR=SUM/S
   SUM=0.0
   DO 35 I=1,N
   DIFF=(ABSF (R(I)-AVR))*2.0
35 SUM=SUM+DIFF
   J1=M
   T=J1 - 1
   M=J1
   S=SQRTF (SUM/T)


   PUNCH 86
86 FORMAT (26HRESULTS OF TEST AS A WHOLE,/)
   PUNCH 87,AVR
87 FORMAT (10HAVERAGE R=,F6.2)
   PUNCH 88,S
   IF (NN-KK) 36,37,36
36 GO TO 7
37 CONTINUE
88 FORMAT (19HSTANDRAD DEVIATION=,F6.2)
 1 FORMAT (3X,14,2I3)
 2 FORMAT (4X,5F10.4)
 3 FORMAT (4X,I3)
   END
```

COMPUTER PROGRAM FOR WORK HARDENING COEFFICIENT (N)

```
ZZJOB 5                                        REG HAMILTON
ZZFORX5
C      PROGRAM FOR N VALUES-DOFASCO PROJECT
C      N-NUMBER OF MEASUREMENTS FOR SAMPLE (15)
C      M-NUMBER OF IDENTICAL SPECIMENS IN A GIVEN TEST
C      IT-TEST NUMBER
C      NN-TOTAL NUMBER OF TESTS
C      L(I)-MATRIX FOR PREDETERMINED ELONGATIONS
C      P(I)-LOADS CORRESPONDING TO ABOVE ELONGATIONS
C      W-AVERAGE INITIAL SPECIMEN WIDTH
       DIMENSION E(60),YR(60),SL(30),P(30)
       DIMENSION X(60),Y(60),XX(60),YY(60)
       PUNCH 83
    83 FORMAT (13H$REG HAMILTON)
       PUNCH 82
    82 FORMAT (33HWORK HARDENING COEFF DOFASCO PROJ,/)
       READ 1,N,M,NN
       READ 2,(SL(I),I=1,N)
       JJ=1
    31 K=1
       J=1
       READ 13,IT,ST,W
       PUNCH 84,IT
    84 FORMAT (12HTEST NUMBER-,13,/)
       K1=M
       K2=N
       KK=K1*K2
       M=K1
       N=K2
    20 READ 2,( P(I),I=1,N)


    24 XX(K)=(LOGF (SL(J)))/2.302585
    21 YY(K)=(LOGF (50.0*P(J))-LOGF(W*ST)+LOGF(SL(J)))/2.302585
    22 IF (J-N) 23,25,23
    23 J=J+1
       K=K+1
       GO TO 24
    25 IF (K-KK) 26,28,26
    26 K=K+1
       J=1
       GO TO 20
    28 SSX=0.0
       SSY=0.0
       SXY=0.0
       SUX=0.0
       SUY=0.0
       SSEI=0.0
       SDX=0.0
       DO 60 I=1,KK
```

```
   SUX=SUX+XX(I)
60 SUY=SUY+YY(I)
   I1=KK
   T=I1
   KK=I1
   TX=SUX/T
   TY=SUY/T
   DO 65 I=1,KK
   X(I)=XX(I)-TX
65 Y(I)=YY(I)-TY
   DO 70 I=1,KK
   SSX=SSX+X(I)*X(I)
   SSY=SSY+Y(I)*Y(I)


   SXY=SXY+X(I)*Y(I)
70 SDX=SDX+X(I)
   B=SXY/SSX
   A=TY-B*TX
   PUNCH 3
   PUNCH 4,A,B
   DO 75 I=1,KK
   YR(I)=A+B*XX(I)
   E(I)=ABSF(YY(I)-YR(I))
75 SSEI=SSEI+E(I)*E(I)
   SY=SQRTF((SSEI)/(T-2.0))
   SB=SY/SQRTF(SSX)
   TX=ABSF(TX)
   SA=SQRTF(SY**2.0*(1.0/T+(TX**2.0)/SSX))
   R=SXY/SQRTF(SSX*SSY)
   PUNCH 5
   PUNCH 6, SY
   PUNCH 7
   PUNCH 8,SB
   PUNCH 9
   PUNCH 10,SA
   PUNCH 11
   PUNCH 12,R
   IF (JJ-NN) 29,30,29
29 JJ=JJ+1
   GO TO 31
30 CONTINUE
 1 FORMAT (4X,I3,I3,I3)
 2 FORMAT (4X,6F10.4)
 3 FORMAT (23HREGRESSION COEFFICIENTS,//)
 4 FORMAT (2HA=,E16.8,5X,2HB=,E16.8,//)


 5 FORMAT (30HSTANDARD DEVIATION OF THE MEAN,/)
 6 FORMAT (4HSYB=,E16.8,//)
 7 FORMAT (31HSTANDARD DEVIATION OF THE SLOPE,/)
 8 FORMAT (3HSB=,E16.8,//)
 9 FORMAT (35HSTANDARD DEVIATION OF THE INTERCEPT,/)
```

```
10 FORMAT (3HSA=,E16.8,//)
11 FORMAT (23HCORRELATION COEFFICIENT,/)
12 FORMAT (2HR=,E16.8,///)
13 FORMAT (3X,14,2F10.4)
   END
```

APPENDIX D

RESULTS OF STANDARD TENSION TESTS

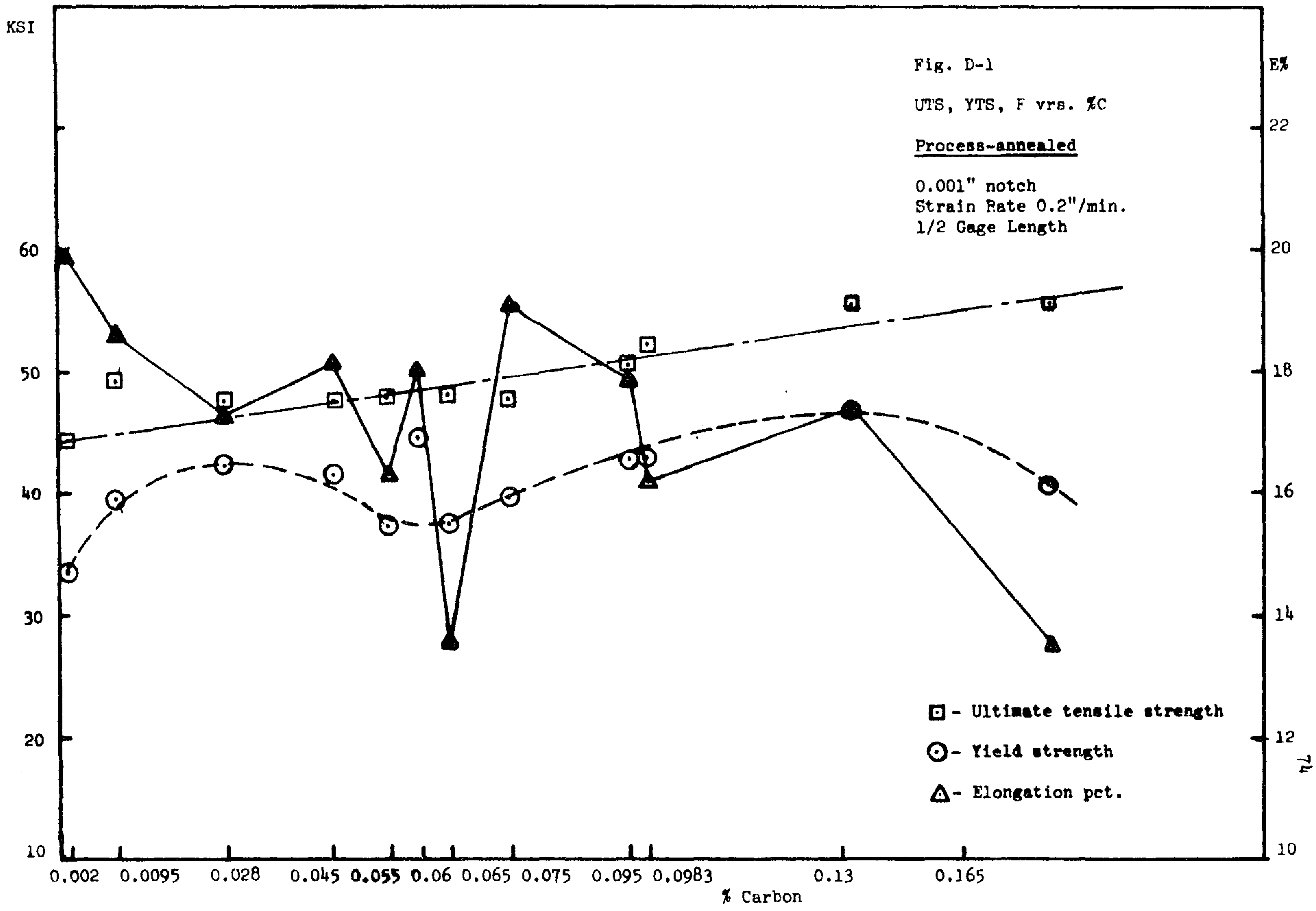


Fig. D-1

UTS, YTS, F vrs. %C

<u>Process-annealed</u>

0.001" notch
Strain Rate 0.2"/min.
1/2 Gage Length

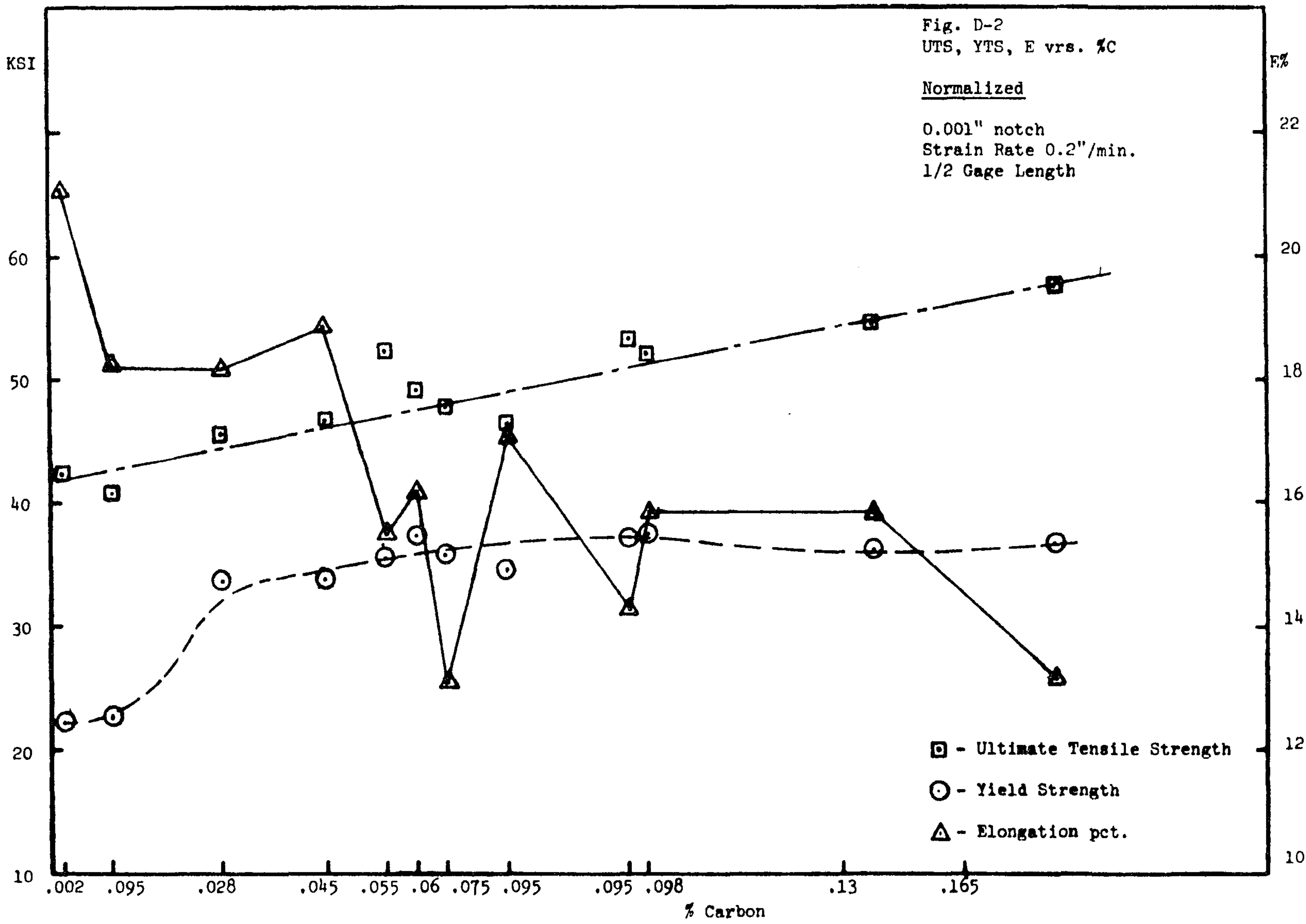


Fig. D-2
UTS, YTS, E vrs. %C

Normalized

0.001" notch
Strain Rate 0.2"/min.
1/2 Gage Length

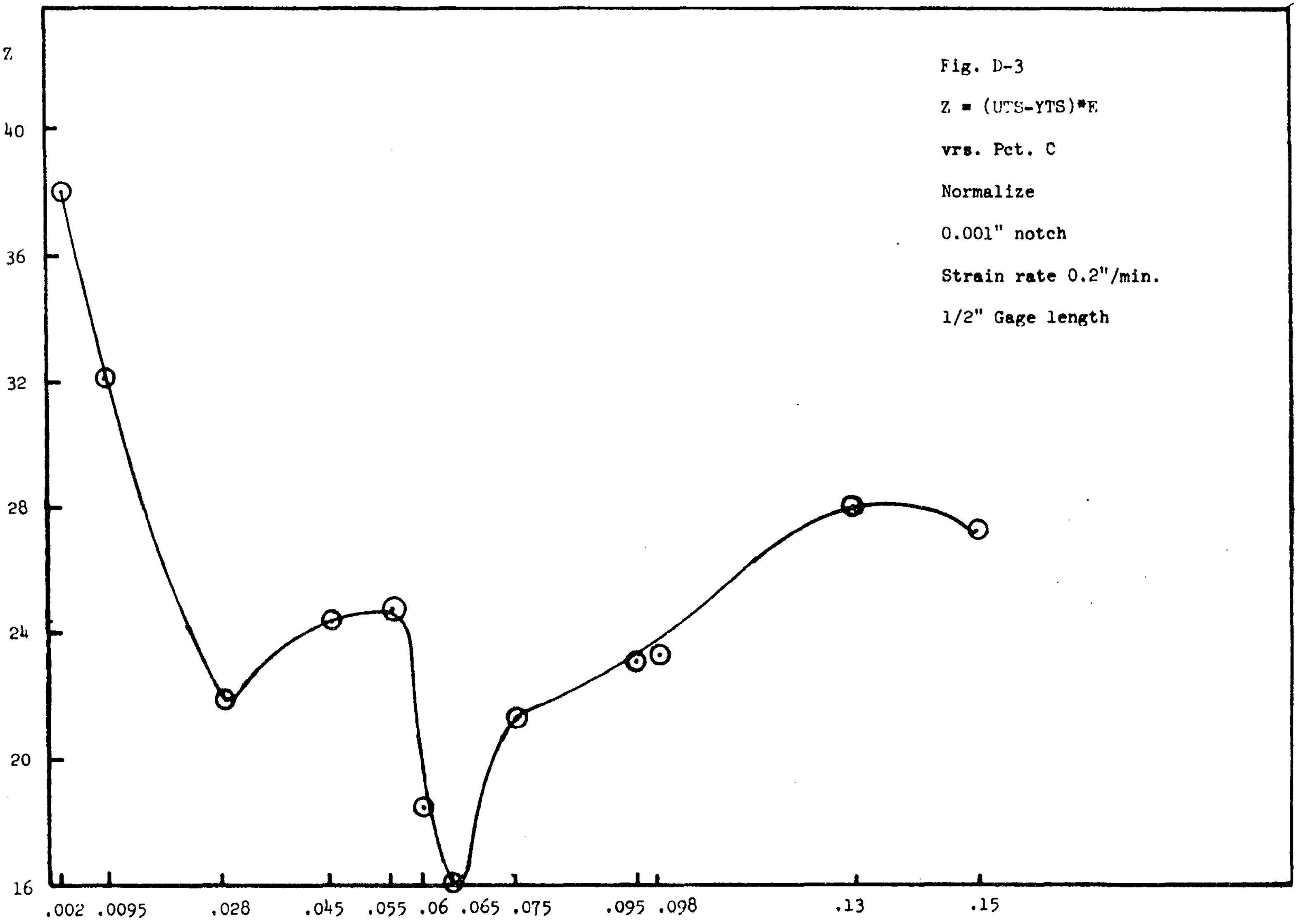


Fig. D-3

Z = (UTS-YTS)*E

vrs. Pct. C

Normalize

0.001" notch

Strain rate 0.2"/min.

1/2" Gage length

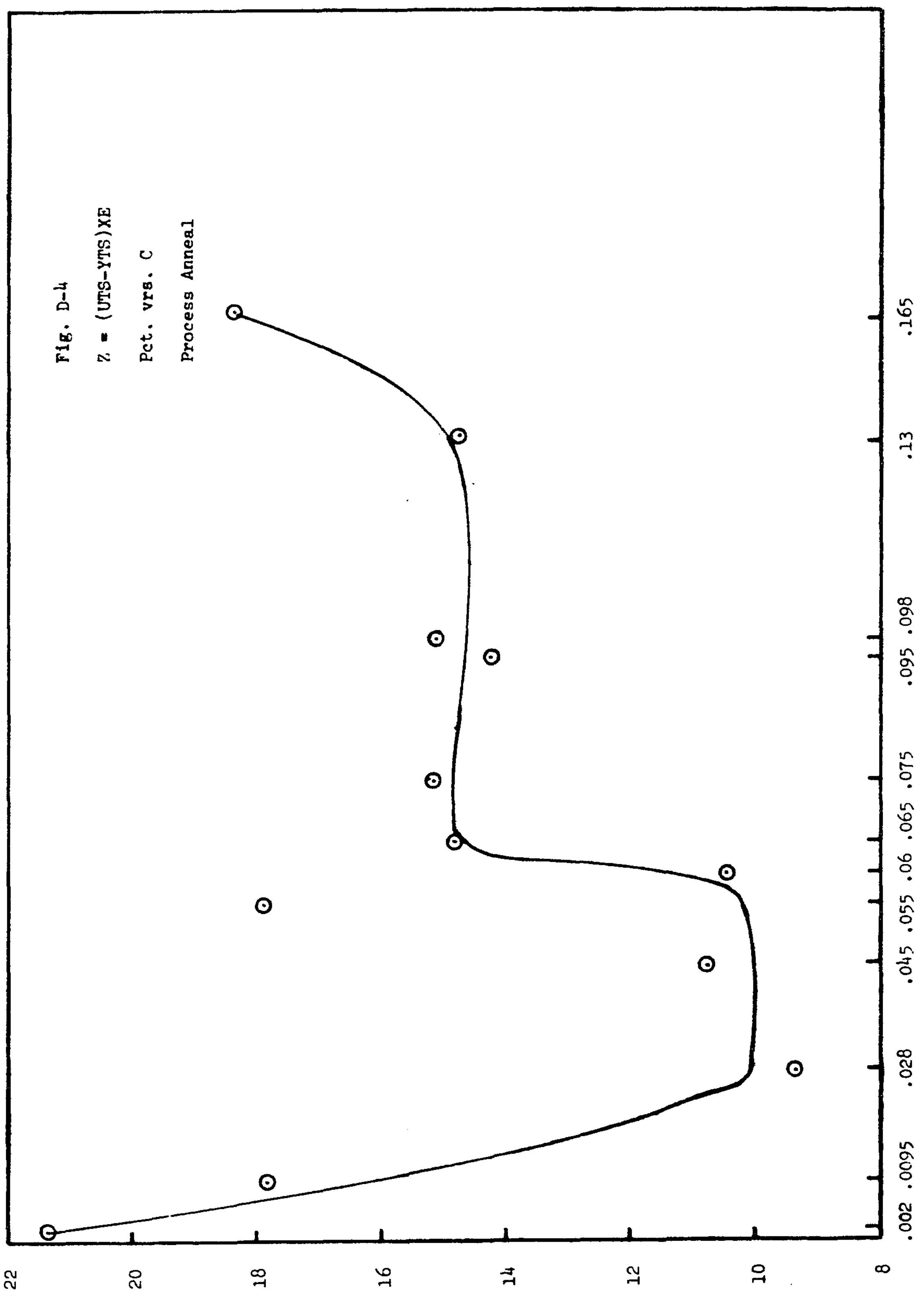


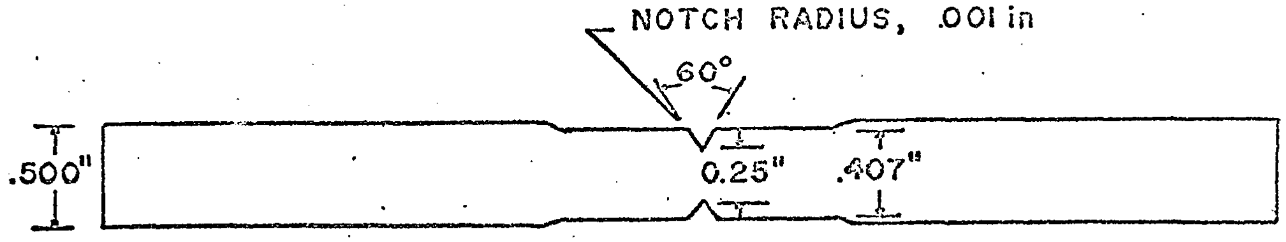
Fig. D-4

Z = (UTS-YTS)XE

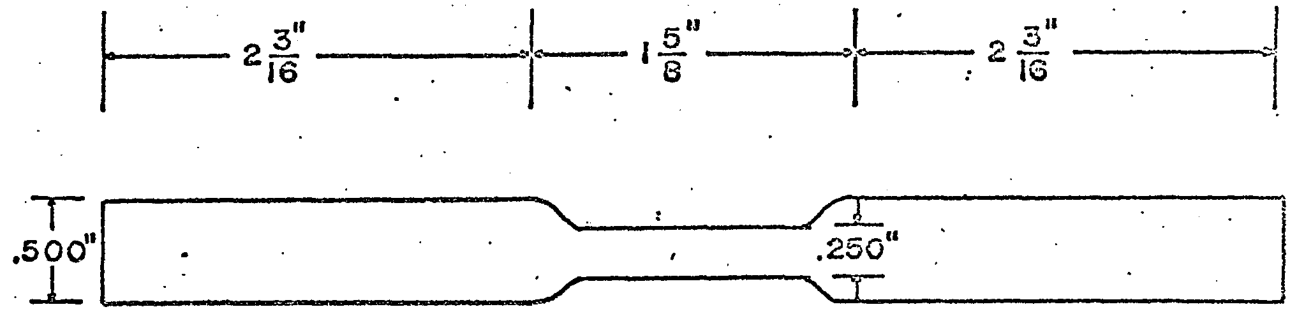
Pct. vrs. C

Process Anneal

Fig. D-5



NOTCHED SPECIMEN



UNNOTCHED SPECIMEN

TABLE D-1. TENSION TEST RESULTS

Test Number	Carbon Percent	G.L. (in.)	Heat Treatment	Notch Radius(in.)	Strain Rate(in./min.)	Y.S. (KSI)	U.T.S. (KSI)	Elong. (%)	YS/UTS	(UTS-YS) *(Elong.)X10^{-4}
1	0.095	1	As-received	Unnotched	0.05	32.66	46.73	40.38	0.699	-
	0.055		"	"	"	32.69	43.21	41.00	0.752	-
	0.002		"	"	"	28.35	41.05	42.70	0.691	-
				"						
2	0.095	1	As-received	Unnotched	0.2	34.82	49.26	40.90	0.715	-
	0.055		"	"	"	33.26	44.83	36.50	0.739	-
	0.002		"	"	"	30.72	42.13	47.13	0.724	-
3	0.095	1	As-received	Unnotched	0.5	36.72	49.12	38.00	0.747	-
	0.055		"	"	"	34.61	45.02	36.33	0.764	-
	0.002		"	"	"	32.69	43.71	44.35	0.748	-
4	0.095	1	Process-anneal	Unnotched	0.2	41.42	45.90	42.75	0.902	-
	0.055		"	"	"	39.33	46.27	37.20	0.805	-
	0.002		"	"	"	37.92	42.52	49.10	0.892	-
5	0.095	1	Normalize	Unnotched	0.2	33.87	48.78	40.00	0.694	-
	0.055		"	"	"	30.35	47.08	40.25	0.645	-
	0.002		"	"	"	17.09	36.09	37.85	0.473	-

TENSION TEST RESULTS (Cont.)

Test Number	Carbon Percent	G.L. (in.)	Heat Treatment	Notch Radius(in.)	Strain Rate(in./min.)	Y.S. (KSI)	U.T.S. (KSI)	Elong. (%)	YS/UTS	(UTS-YS) *(Elong.)X10^{-4}
6	0.095	1	As-received	0.003	0.05	41.97	56.18	6.73	0.739	-
	0.055		"	"	"	37.97	51.71	7.14	0.724	-
	0.002		"	"	"	33.87	49.08	9.06	0.689	-
7	0.095	1	Process-anneal	0.003	0.2	46.13	56.97	7.90	0.810	-
	0.055		"	"	"	38.82	51.87	8.16	0.748	-
	0.002		"	"	"	38.60	48.46	8.75	0.797	-
8	0.095	1	Normalize	0.003	0.2	37.18	53.04	12.88	0.701	-
	0.055		"	"	"	33.82	50.52	15.68	0.669	-
	0.002		"	"	"	22.81	41.91	18.33	0.544	-
9	0.095	1	Normalize	0.001	0.2	37.93	52.93	8.1	0.717	-
	0.055		"	"	"	33.86	51.00	9.3	0.664	-
	0.002		"	"	"	21.95	40.50	11.2	0.542	-
10	0.165	0.5	Process-anneal	0.001	0.2	40.98	55.33	13.5	0.741	19.4
	0.130					46.99	55.49	17.5	0.847	14.8
	0.0983					42.85	52.25	16.2	0.820	15.2
	0.095					42.86	50.83	17.9	0.844	14.3
	0.075					39.98	47.95	19.1	0.834	15.2
	0.065					37.33	48.33	13.6	0.773	14.9
	0.06					44.44	50.27	18.0	0.884	10.5
	0.055					37.38	47.98	16.2	0.779	17.2
	0.045					41.65	47.62	18.1	0.875	10.8
	0.028					42.17	47.63	17.3	0.885	9.4
	0.0095					39.72	49.28	18.6	0.806	17.9
	0.002					33.57	44.36	19.9	0.757	21.4

TENSION TEST RESULTS (Cont.)

Test Number	Carbon Percent	G.L. (in.)	Heat Treatment	Notch Radius(in.)	Strain Rate(in./min.)	Y.S. (KSI)	U.T.S. (KSI)	Elong. (%)	YS/UTS	(UTS-YS) *(Elong.)X10^{-4}
11	0.165	0.5	Normalize	0.001	0.2	36.89	57.72	13.2	0.639	27.4
	0.130					36.22	54.63	15.9	0.663	28.2
	0.0983					37.56	52.11	15.9	0.718	23.2
	0.095					37.21	53.40	14.3	0.701	23.2
	0.075					34.48	46.77	17.3	0.737	21.3
	0.065					35.88	48.05	13.1	0.744	15.9
	0.06					37.77	49.19	16.2	0.768	18.5
	0.055					35.61	52.47	14.8	0.677	24.9
	0.045					33.90	46.85	18.9	0.724	24.5
	0.028					33.85	45.94	18.2	0.737	22.0
	0.0095					22.80	40.99	18.3	0.556	32.4
	0.002					22.33	42.33	19.1	0.524	38.2
*12	0.095	0.5	Process-anneal	0.001	0.2	47.37	55.67	12.6	0.851	-
	0.055	"	"	"	"	40.80	52.06	16.7	0.784	-
	0.002	"	"	"	"	41.18	50.68	17.0	0.813	-
*13	0.095	0.5	Normalize	0.001	0.2	39.74	53.38	13.5	0.745	-
	0.055	"	"	"	"	35.97	51.18	14.3	0.703	-
	0.002	"	"	"	"	25.19	42.68	15.2	0.590	-

Legend:

G.L. - Extensometer gage length

* - Specimens cut from sheet in transverse direction

APPENDIX E

RESULTS OF R & N TESTS

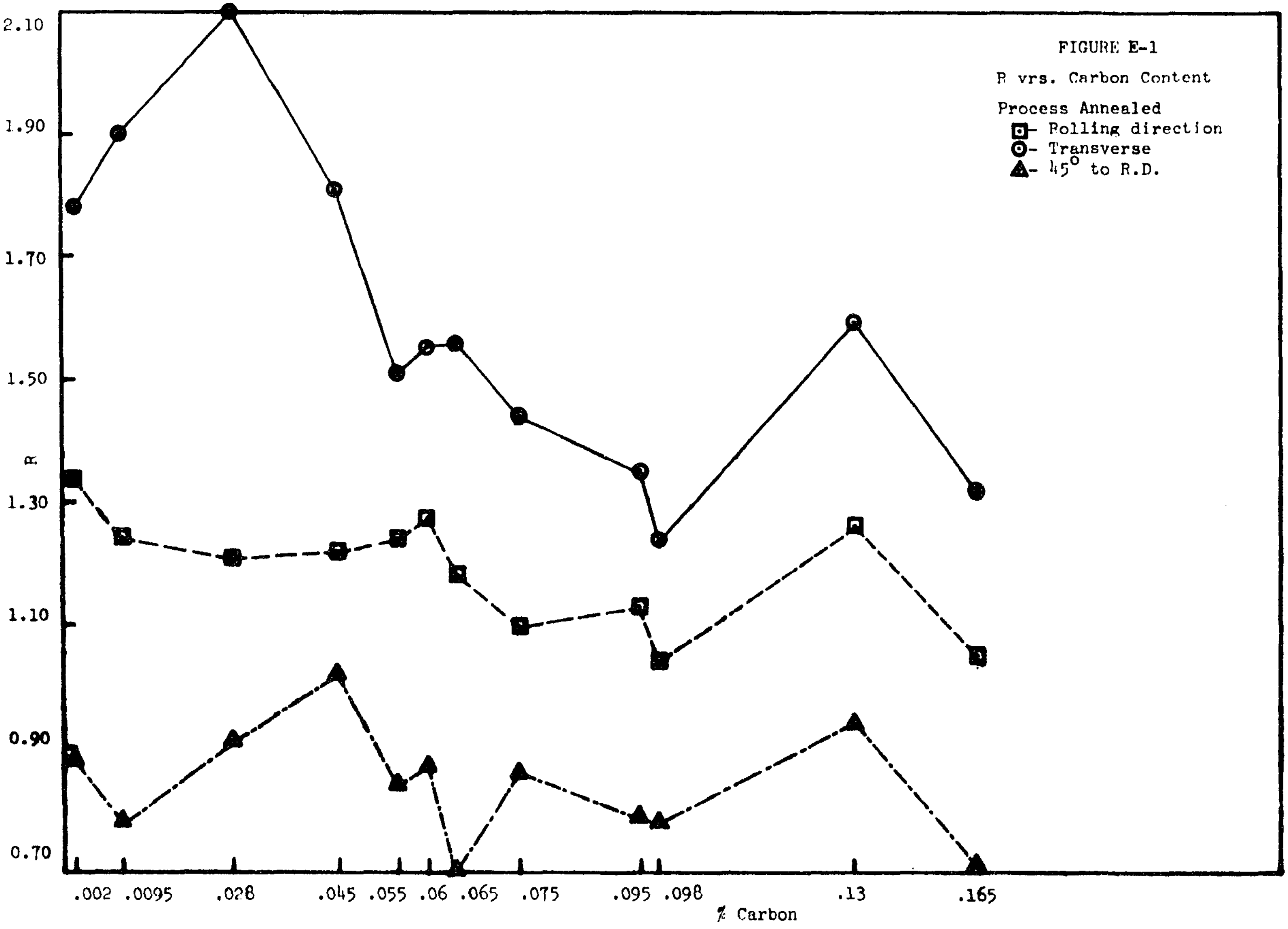


FIGURE E-1

R vrs. Carbon Content

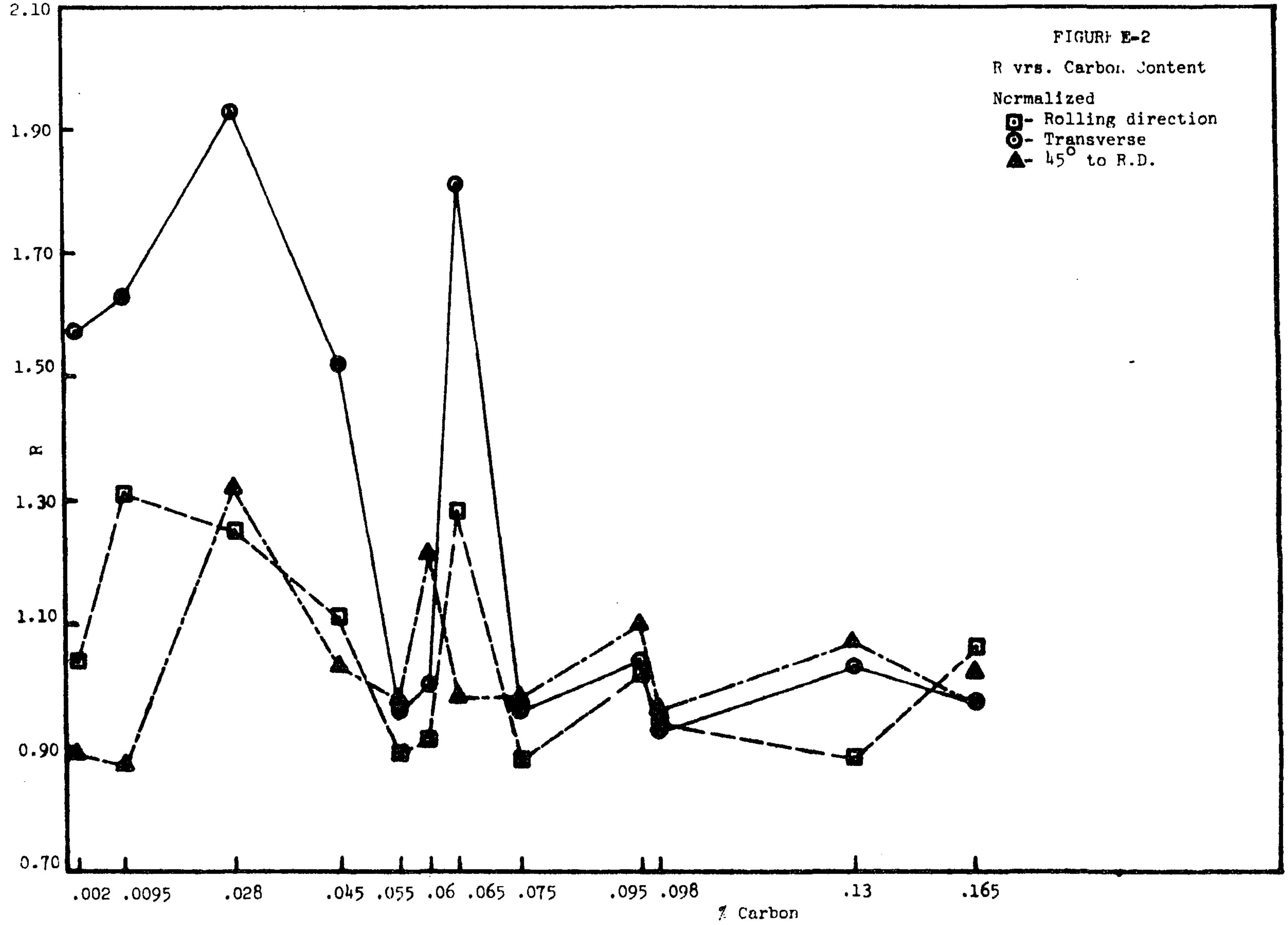


FIGURE E-2

R vrs. Carbon Content

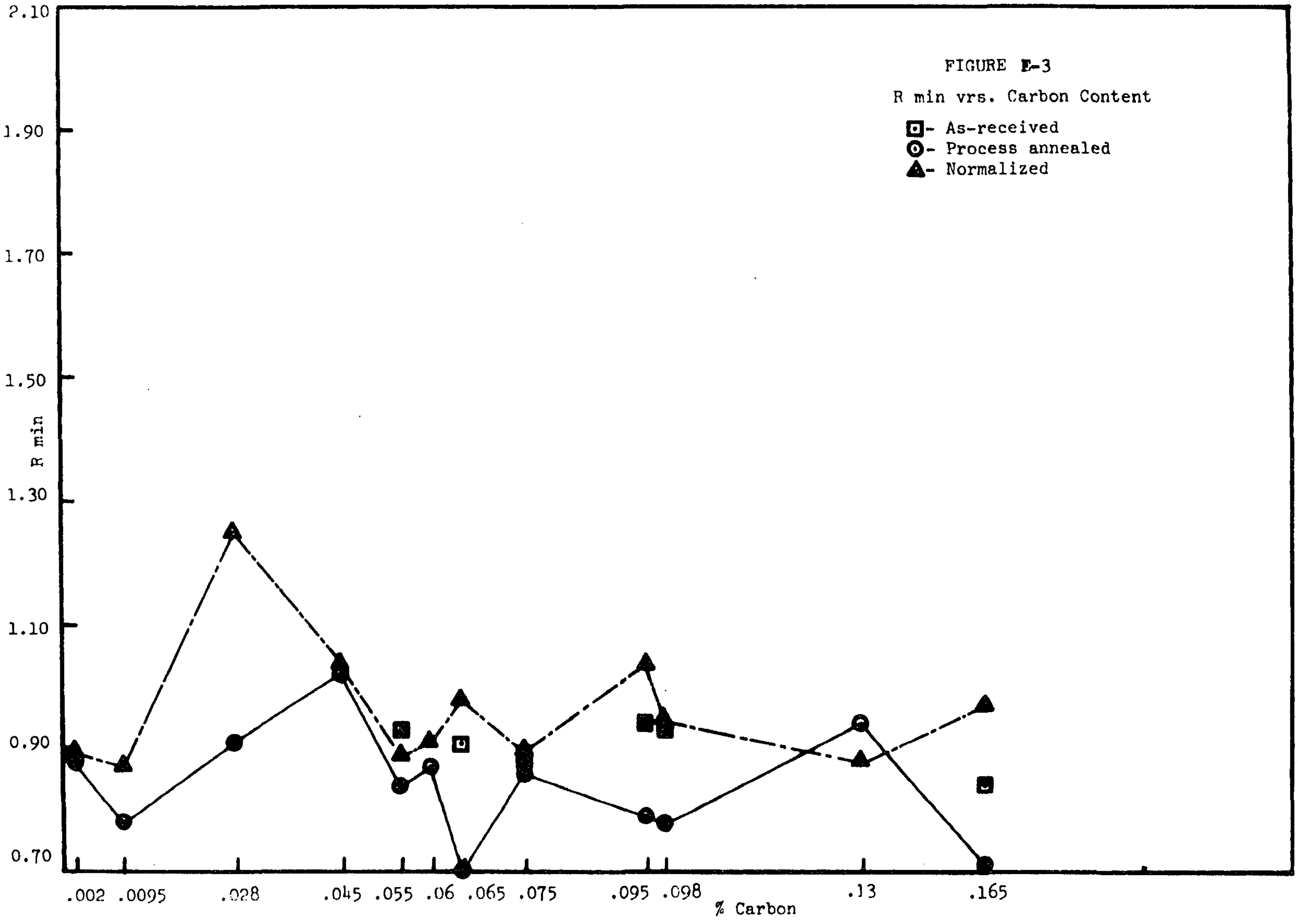
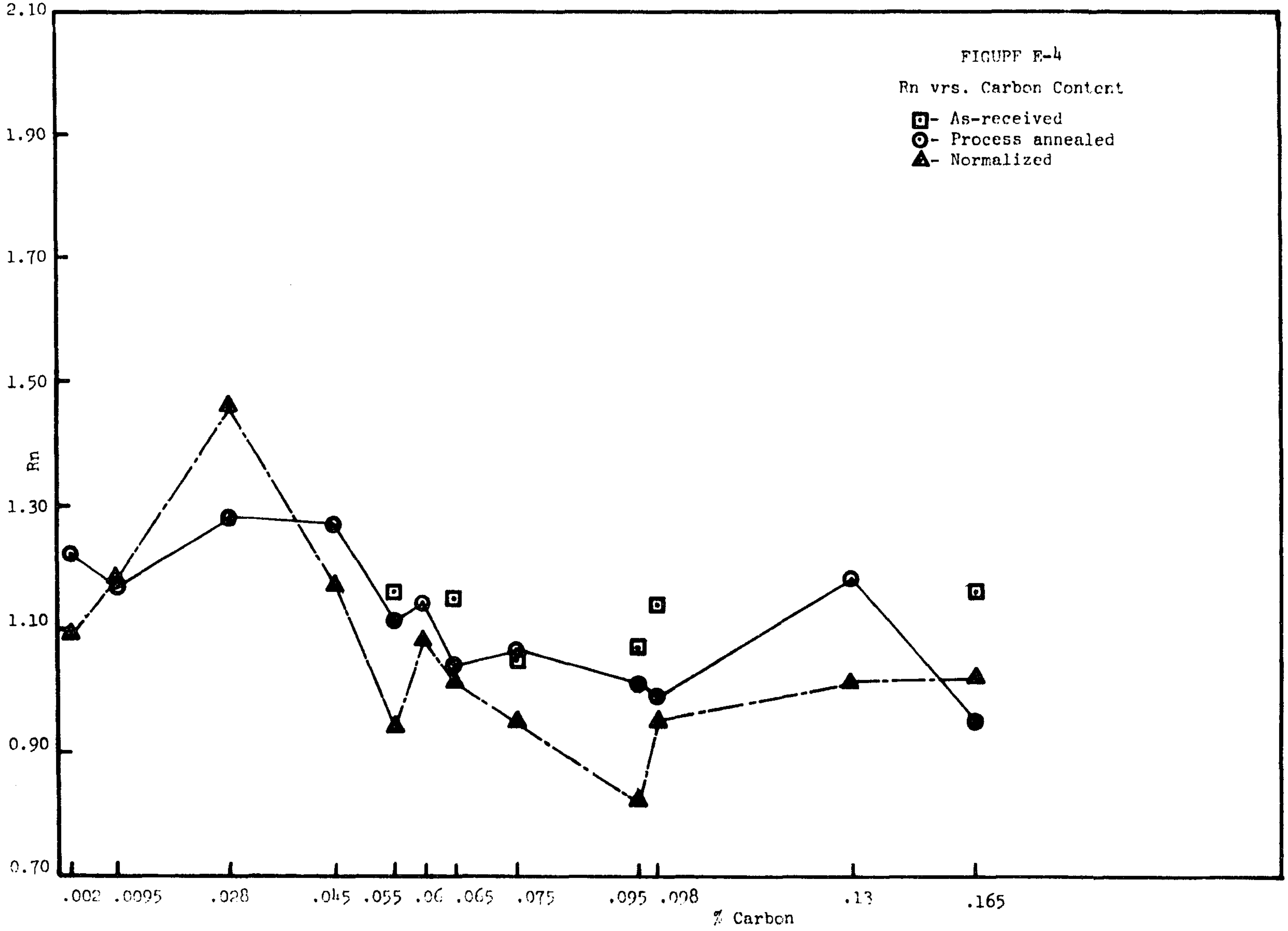


FIGURE E-3

R min vrs. Carbon Content



FIGUPF F-4

Rn vrs. Carbon Content

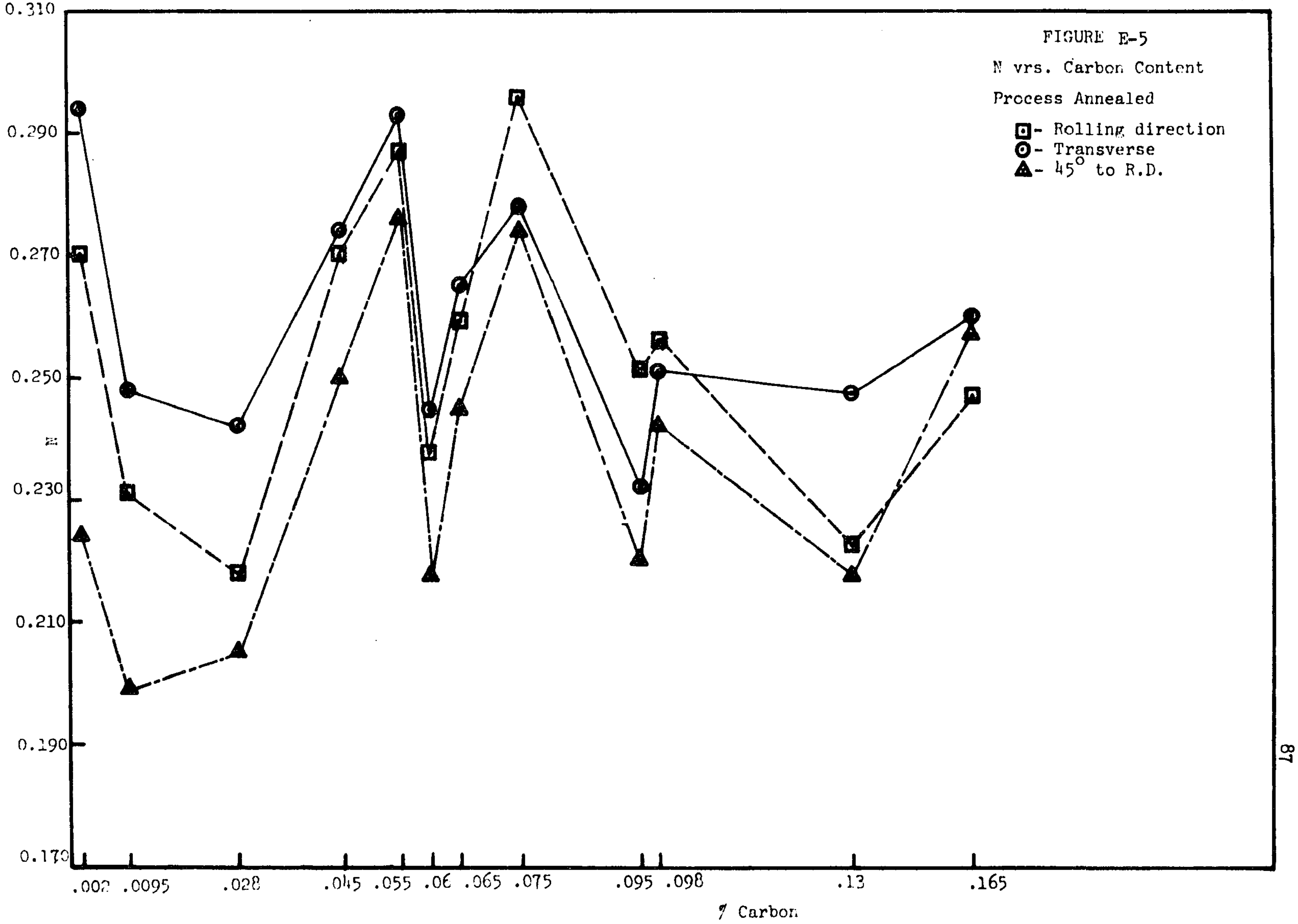


FIGURE E-5

N vrs. Carbon Content

Process Annealed

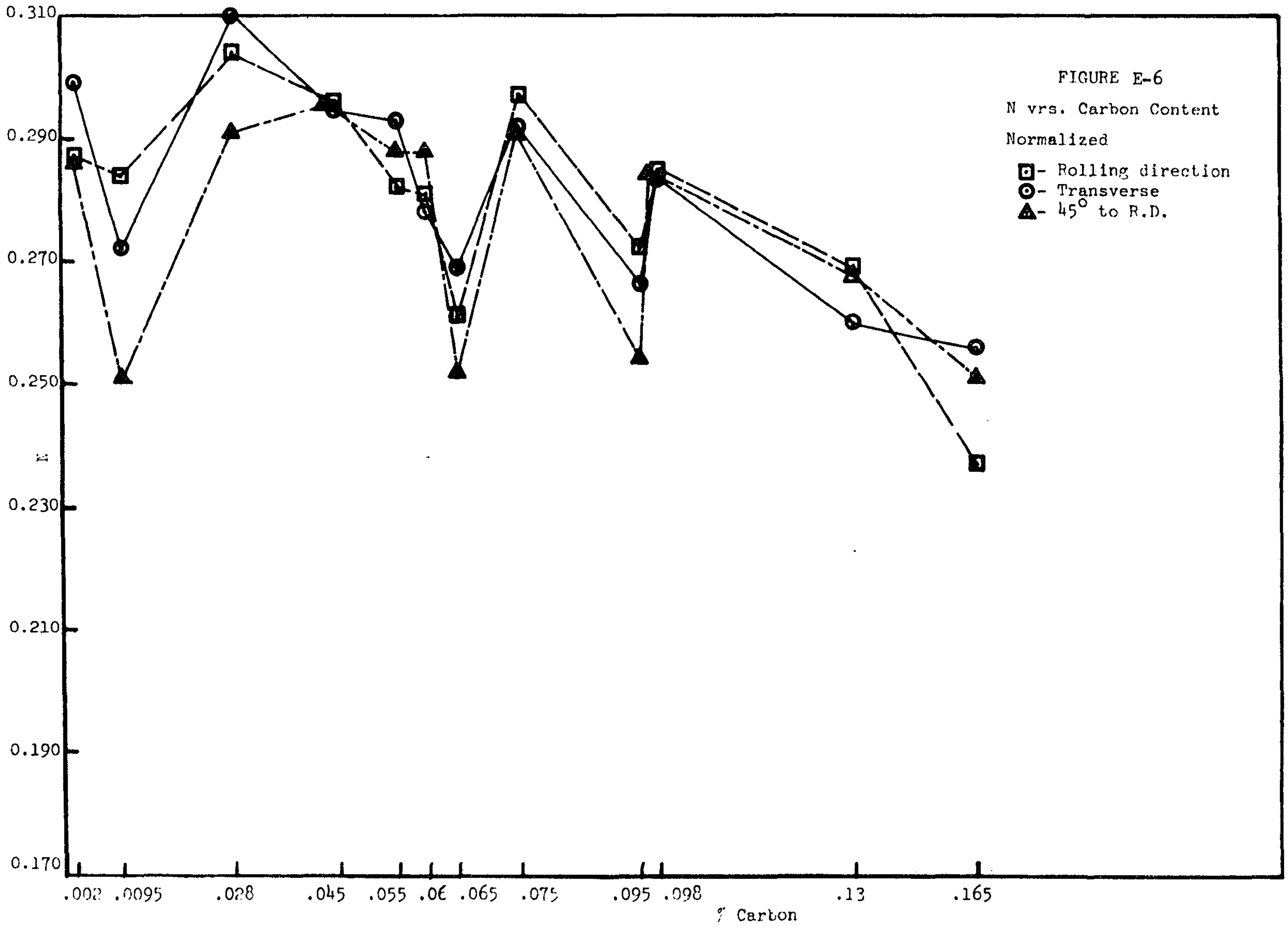


FIGURE E-6

N vrs. Carbon Content

Normalized

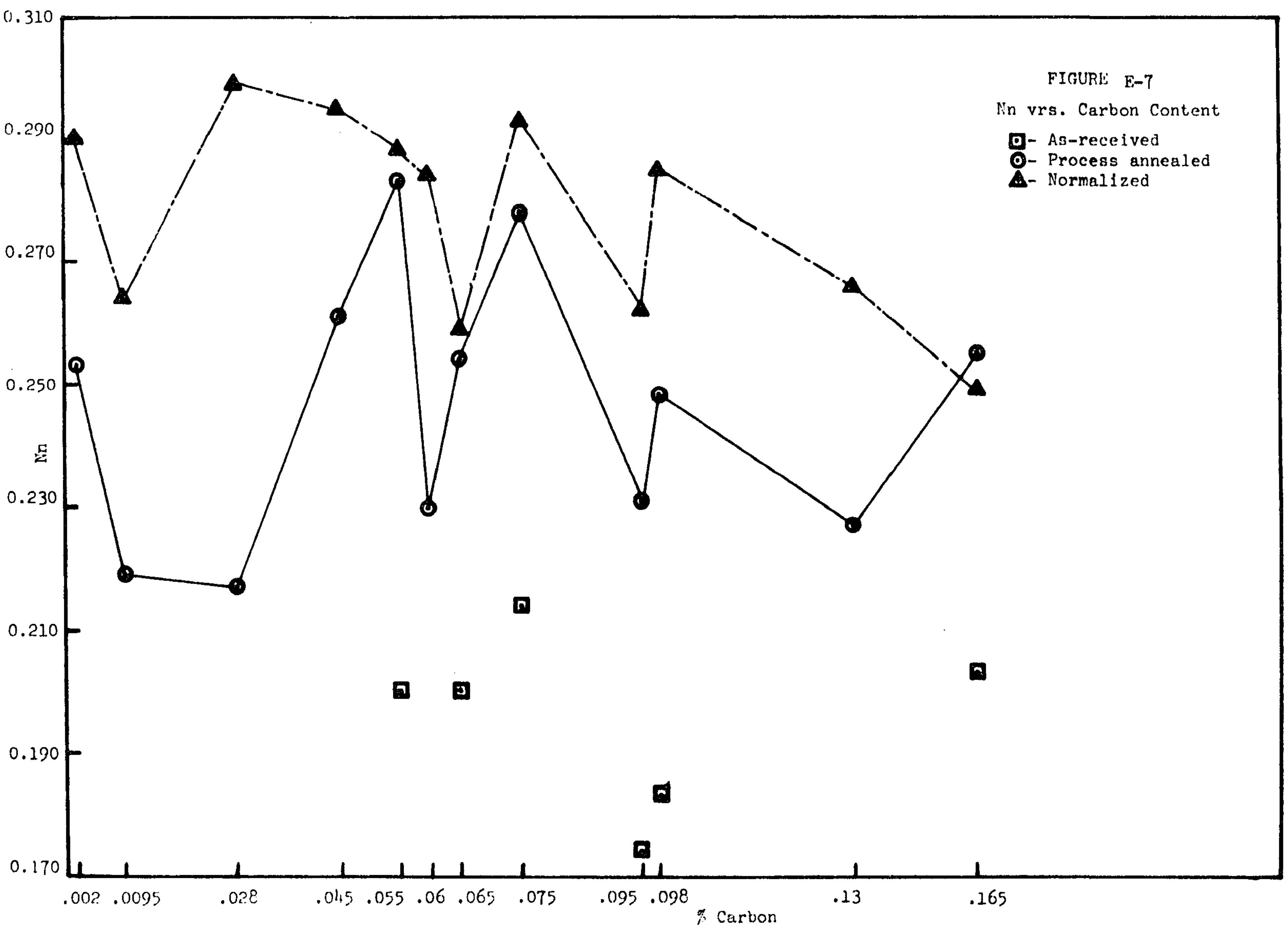


FIGURE E-7

Nn vrs. Carbon Content

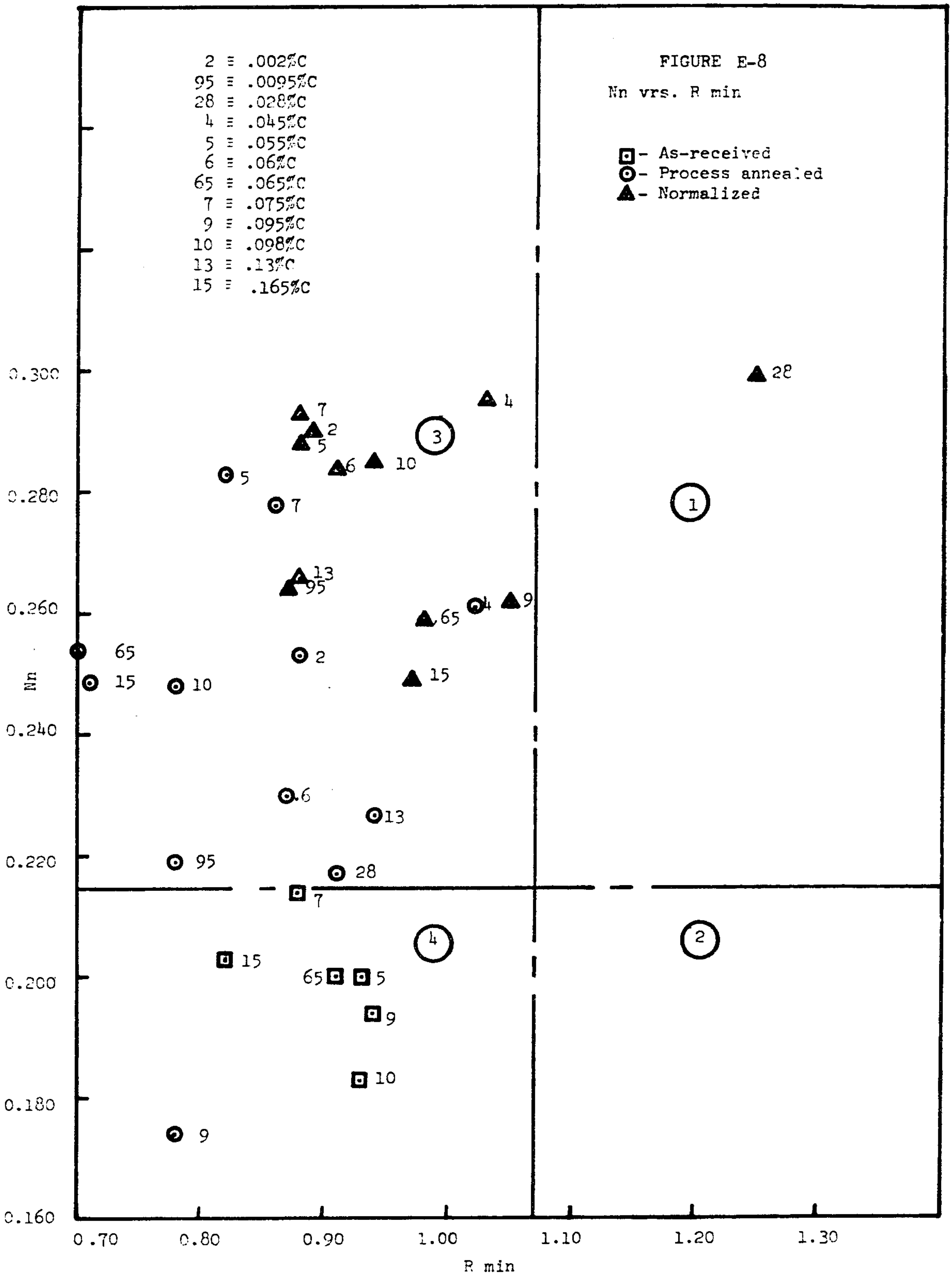


FIGURE E-8

Mn vrs. R min

TABLE E-1. R & N TEST RESULTS

Carbon Percent	Specimen Orientation	Heat Treatment	Individual R Values	Average R	R_n	N	N_n
0.165	R.D.	Process-anneal	1.17, 1.00, 0.99, 1.07	1.05		0.247	
	Trans.		1.48, 1.36, 1.30, 1.14	1.32		0.260	
	45°		0.59, 0.79, 0.72, 0.77	0.71	0.95	0.257	0.255
0.13	R.D.	Process-anneal	1.03, 1.38, 1.17, 1.47	1.26		0.223	
	Trans.		1.62, 1.48, 1.70, 1.56	1.59		0.248	
	45°		0.98, 0.95, 0.93, 0.92	0.94	1.18	0.218	0.227
0.0983	R.D.	Process-anneal	1.05, 0.95, 1.05, 1.11	1.04		0.256	
	Trans.		1.39, 1.06, 1.27, 1.27	1.24		0.251	
	45°		0.72, 0.76, 0.80, 0.87	0.78	0.96	0.242	0.248
0.095	R.D.	Process-anneal	1.39, 1.05, 1.17, 0.94	1.13		0.251	
	Trans.		1.28, 1.18, 1.58,	1.35		0.232	
	45°		0.82, 0.77, 0.77, 0.83	0.79	1.01	0.220	0.231
0.075	R.D.	Process-anneal	0.97, 0.94, 1.23, 1.30	1.10		0.296	
	Trans.		1.31, 1.31, 1.56, 1.60	1.44		0.278	
	45°		0.79, 0.83, 0.86, 1.00	0.86	1.06	0.274	0.278

R & N TEST RESULTS (Cont.)

Carbon Percent	Specimen Orientation	Heat Treatment	Individual R Values	Average R	R_n	N	N_n
0.065	R.D.	Process-anneal	1.21, 1.40, 1.10, 1.03	1.18		0.259	
	Trans.		1.70, 1.57, 1.61, 1.39	1.56		0.265	
	45°		0.55, 0.74, 0.81, 0.73	0.70	1.04	0.245	0.254
0.06	R.D.	Process-anneal	1.17, 1.30, 1.28, 1.35	1.27		0.238	
	Trans.		1.52, 1.60, 1.61, 1.48	1.55		0.245	
	45°		0.87, 0.77, 0.85, 1.03	0.87	1.14	0.218	0.230
0.055	R.D.	Process-anneal	1.31, 1.22, 1.25, 1.18	1.24		0.287	
	Trans.		1.60, 1.71, 1.32, 1.43	1.51		0.293	
	45°		0.76, 0.86, 0.84, 0.90	0.84	1.11	0.276	0.283
0.045	R.D.	Process-anneal	1.23, 1.29, 1.14, 1.23	1.22		0.270	
	Trans.		1.66, 1.81, 1.74, 2.06	1.81		0.274	
	45°		1.04, 1.09, 1.04, 0.93	1.02	1.27	0.250	0.261
0.028	R.D.	Process-anneal	1.21, 1.08, 1.31, 1.25	1.21		0.218	
	Trans.		1.89, 2.17, 2.12, 2.26	2.11		0.242	
	45°		0.76, 0.89, 1.21, 0.79	0.91	1.28	0.205	0.217

R & N TEST RESULTS (Cont.)

Carbon Percent	Specimen Orientation	Heat Treatment	Individual R Values	Average R	R_n	N	N_n
0.0095	R.D.	Process-anneal	1.22, 1.05, 1.69, 1.01	1.24		0.231	
	Trans.		2.25, 2.14, 1.25, 1.99	1.90		0.248	
	45°		0.79, 0.73, 0.81, 0.82	0.78	1.17	0.199	0.219
0.002	R.D.	Process-anneal	1.35, 1.46, 1.34, 1.22	1.34		0.270	
	Trans.		1.62, 1.74, 1.72, 2.04	1.78		0.294	
	45°		1.00, 0.85, 0.89, 0.78	0.88	1.22	0.224	0.253
0.165	R.D.	Normalize	1.11, 1.20, 0.99, 0.98	1.06		0.237	
	Trans.		0.79, 0.78, 0.97, 1.34	0.97		0.256	
	45°		1.00, 1.01, 1.16, 0.97	1.02	1.02	0.251	0.249
0.13	R.D.	Normalize	1.00, 0.85, 0.84, 0.84	0.88		0.269	
	Trans.		1.03, 1.20, 0.95, 0.96	1.03		0.260	
	45°		1.22, 1.03, 1.00, 1.06	1.07	1.01	0.268	0.266
0.0983	R.D.	Normalize	1.03, 0.91, 0.83, 0.99	0.94		0.285	
	Trans.		1.03, 1.01, 0.80, 0.95	0.94		0.285	
	45°		0.99, 0.91, 0.82, 1.12	0.96	0.95	0.284	0.285

R & N TEST RESULTS (Cont.)

Carbon Percent	Specimen Orientation	Heat Treatment	Individual R Values	Average R	R_n	N	N_n
0.095	R.D.	Normalize	1.04, 1.15, 0.98, 0.93	1.02		0.297	
	Trans.		1.01, 1.07	1.04		0.292	
	45°		1.34, 1.08, 1.03, 0.97	1.10	0.95	0.291	0.293
0.065	R.D.	Normalize	1.55, 1.22, 1.24, 1.12	1.28		0.261	
	Trans.		1.93, 1.58, 1.72, 2.03	1.81		0.269	
	45°		0.88, 1.19, 0.91, 0.96	0.98	1.01	0.252	0.259
0.06	R.D.	Normalize	0.90, 1.01, 0.83, 0.93	0.91		0.281	
	Trans.		1.05, 1.14, 0.89, 0.96	1.00		0.278	
	45°		1.07, 1.25, 1.23, 1.30	1.21	1.08	0.288	0.284
0.055	R.D.	Normalize	0.92, 0.84, 0.99, 0.85	0.89		0.282	
	Trans.		0.91, 0.92, 1.07, 0.96	0.96		0.293	
	45°		1.01, 0.88, 1.01, 0.96	0.96	0.94	0.288	0.288
0.045	R.D.	Normalize	1.06, 1.01, 1.17, 1.20	1.11		0.296	
	Trans.		1.68, 1.58, 1.31, 1.52	1.52		0.295	
	45°		1.01, 1.06, 1.02, 1.04	1.03	1.17	0.295	0.295
0.028	R.D.	Normalize	1.09, 1.33, 1.34, 1.24	1.25		0.304	
	Trans.		1.70, 1.90, 2.28, 1.86	1.93		0.310	
	45°		1.58, 1.38, 1.16, 1.19	1.32	1.46	0.291	0.299

R & N TEST RESULTS (Cont.)

Carbon Percent	Specimen Orientation	Heat Treatment	Individual R Values	Average R	R_n	N	N_n
0.0095	R.D.	Normalize	1.24, 1.23, 1.38, 1.40	1.31		0.284	
	Trans.		1.61, 1.69, 1.41, 1.81	1.63		0.272	
	45°		0.85, 0.97, 0.81, 0.88	0.87	1.17	0.251	0.264
0.002	R.D.	Normalize	0.96, 1.01, 1.13, 1.08	1.04		0.287	
	Trans.		1.66, 1.78, 1.43, 1.42	1.57		0.299	
	45°		0.78, 0.85, 1.15, 0.80	0.89	1.09	0.287	0.290
0.165	R.D.	As-received	1.08, 1.07, 0.96, 1.47	1.14		0.200	
	Trans.		1.98, 1.67, 2.22, 1.41	1.81		0.206	
	45°		0.80, 0.88, 0.88, 0.81	0.84	1.16	0.203	0.203
0.0983	R.D.	As-received	1.41, 1.27, 1.19, 1.01	1.21		0.182	
	Trans.		1.61, 1.37, 1.36, 1.61	1.49		0.187	
	45°		0.92, 0.98, 0.94, 0.91	0.93	1.14	0.182	0.183
0.095	R.D.	As-received	0.42, 1.26, 1.08, 1.21	0.99		0.187	
	Trans.		1.40, 1.47, 1.36, 1.45	1.42		0.183	
	45°		1.05, 1.21, 0.80, 0.73	0.94	1.07	0.163	0.174

R & N TEST RESULTS (Cont.)

Carbon Percent	Specimen Orientation	Heat Treatment	Individual R Values	Average R	R_n	N	N_n
0.075	R.D.	As-received	1.16, 1.10, 1.23, 1.10	1.14		0.213	
	Trans.		1.03, 1.27, 1.39, 1.56	1.30		0.217	
	45°		0.84, 0.85, 0.86, 0.97	0.88	1.05	0.213	0.214
0.065	R.D.	As-received	1.03, 1.36, 1.17, 1.03	1.14		0.207	
	Trans.		1.41, 1.85, 1.89, 1.36	1.62		0.188	
	45°		0.87, 0.83, 0.93, 1.01	0.91	1.15	0.203	0.200
0.055	R.D.	As-received	0.97, 1.21, 1.30, 1.18	1.16		0.205	
	Trans.		1.86, 1.77, 1.60, 1.34	1.64		0.193	
	45°		0.97, 1.06, 0.87, 0.84	0.93	1.16	0.200	0.200

Legend:

R - Coefficient of anisotropy

R_n - Normal coefficient of anisotropy

N - Work hardening coefficient

N_n - Normal work hardening coefficient

$$R_n = 1/4\ (R_{0^\circ} + 2\ R_{45^\circ} + R_{90^\circ})$$

$$N_n = 1/4\ (N_{0^\circ} + 2\ N_{45^\circ} + N_{90^\circ})$$

APPENDIX F

GRAIN SIZE DETERMINATION BY LINEAR INTERCEPT METHOD

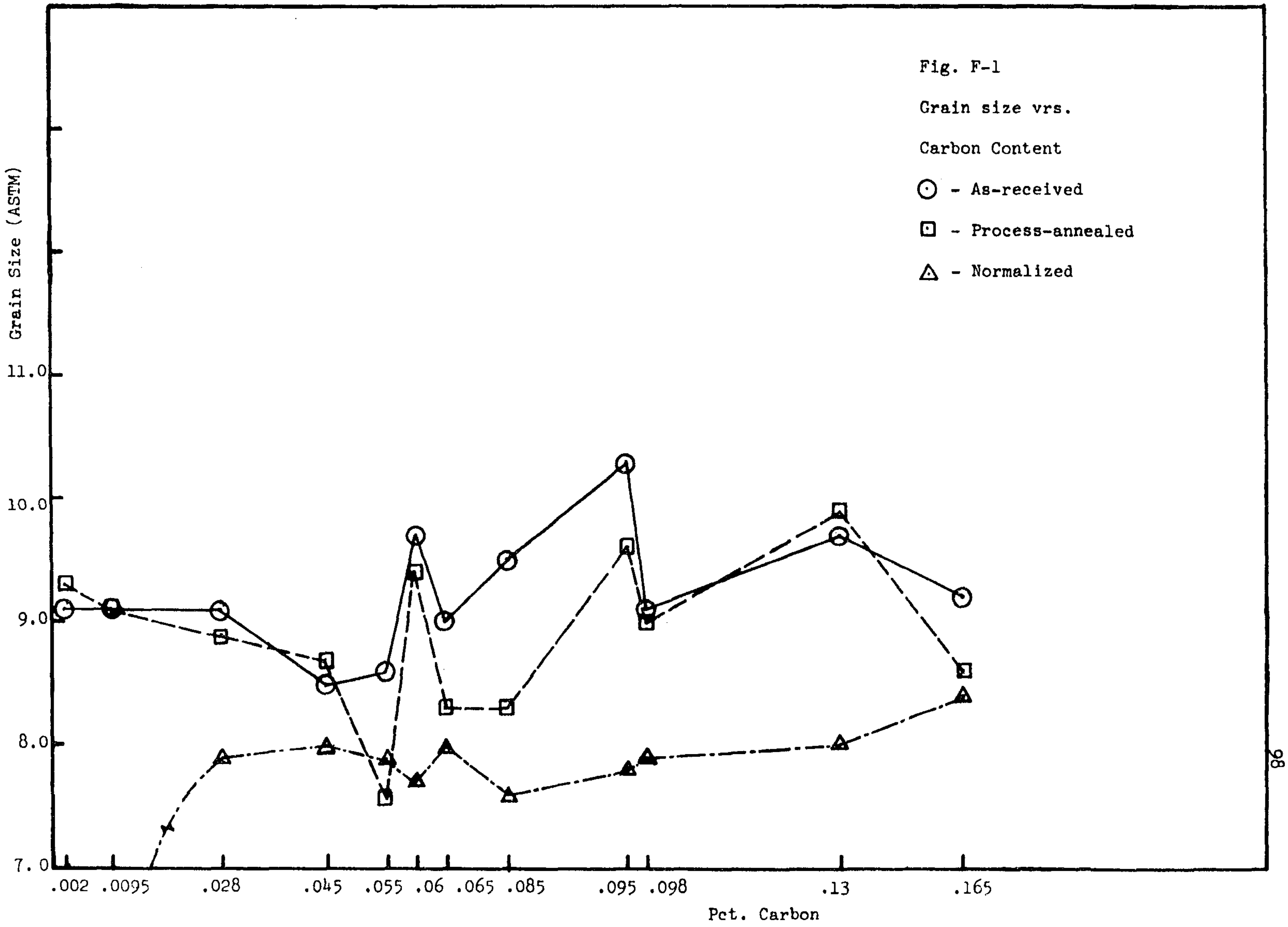


Fig. F-1

Grain size vrs.

Carbon Content

TABLE F-1. GRAIN SIZE

Carbon Pct.	Heat Treatment	Average Grain Size (ASTM)
0.002	As-received	9.1
	Process-annealed	9.3
	Normalized	-
0.0095	As-received	9.1
	Process-annealed	9.1
	Normalized	-
0.028	As-received	9.1
	Process-annealed	8.9
	Normalized	7.9
0.045	As-received	8.5
	Process-annealed	8.7
	Normalized	8.0
0.055	As-received	8.6
	Process-annealed	7.6
	Normalized	7.9
0.060	As-received	9.7
	Process-annealed	9.4
	Normalized	7.7
0.065	As-received	9.0
	Process-annealed	8.3
	Normalized	8.0
0.075	As-received	9.5
	Process-annealed	8.3
	Normalized	7.6
0.095	As-received	10.3
	Process-annealed	9.6
	Normalized	7.8
0.098	As-received	9.1
	Process-annealed	9.0
	Normalized	7.9
0.13	As-received	9.7
	Process-annealed	9.9
	Normalized	8.0
0.165	As-received	9.2
	Process-annealed	8.6
	Normalized	8.4

APPENDIX G

PHOTOMICROGRAPHS

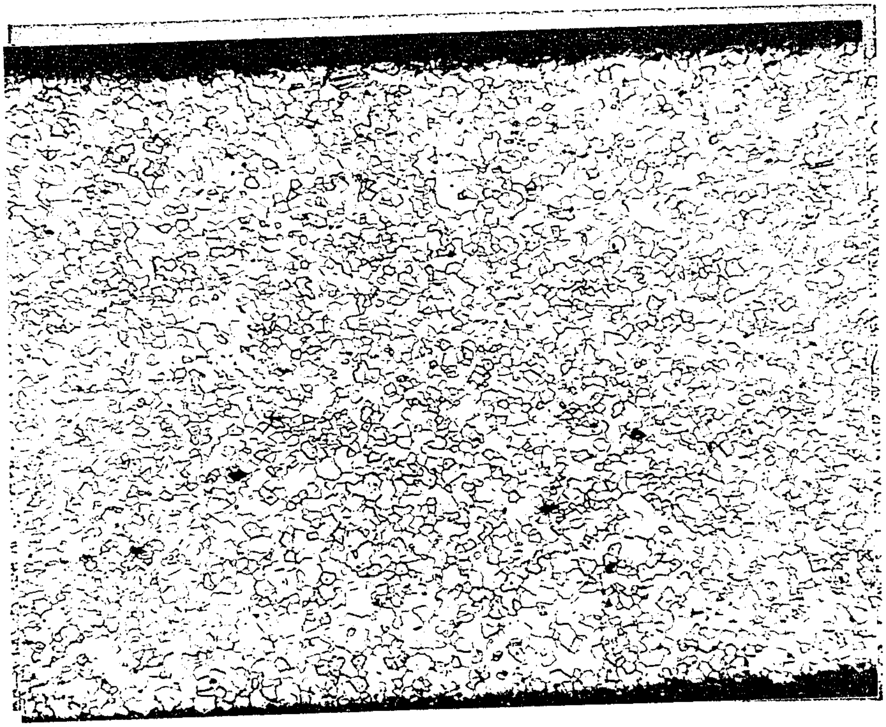


Fig. G-1. Process-annealed 0.002%C, X160

Fig. G-2. Normalized 0.002%C, X160

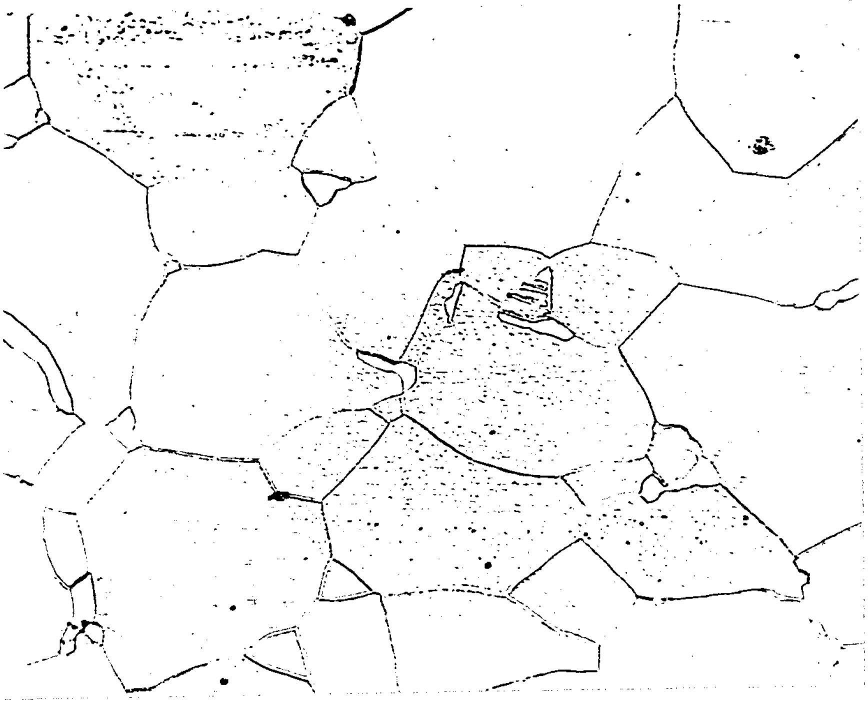


Fig. G-3. Process-annealed 0.045%C, X1600

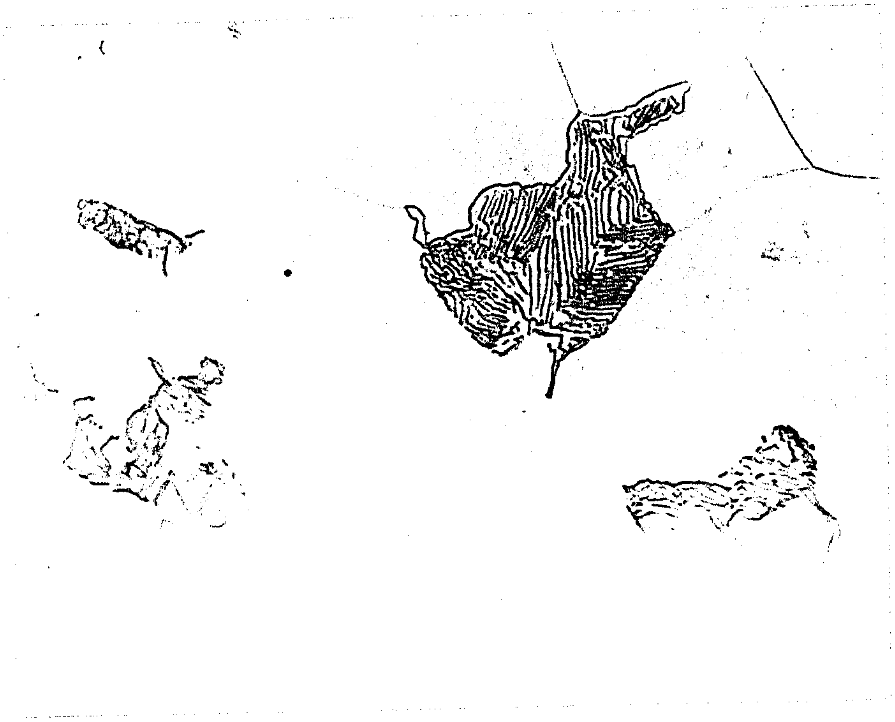


Fig. G-4. Normalized 0.045%C, X1600

Fig. G-5. Process-annealed 0.065%C, X1600

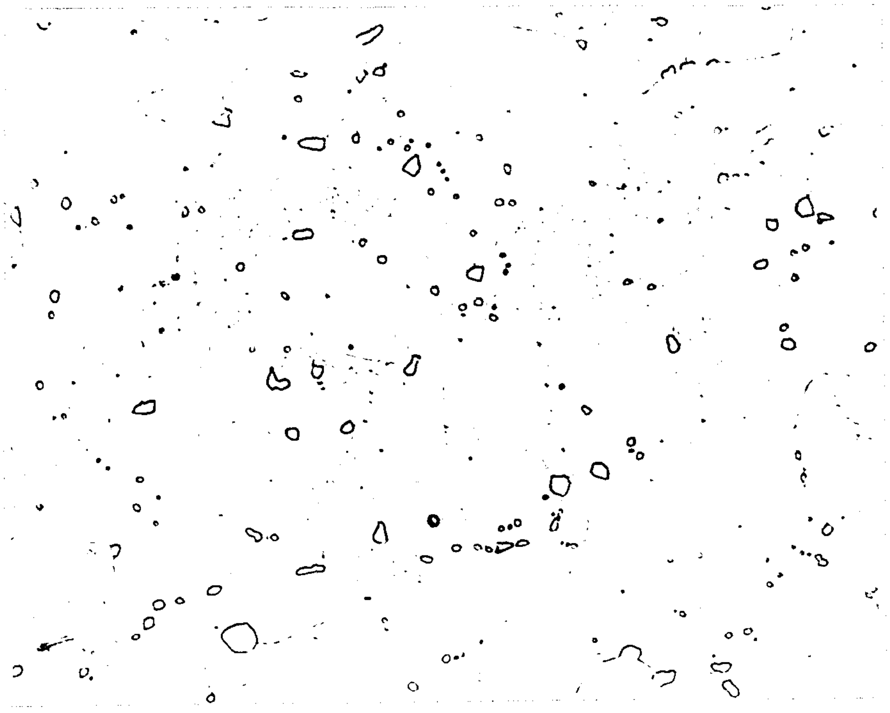


Fig. G-6. Process-annealed 0.075%C, X1600

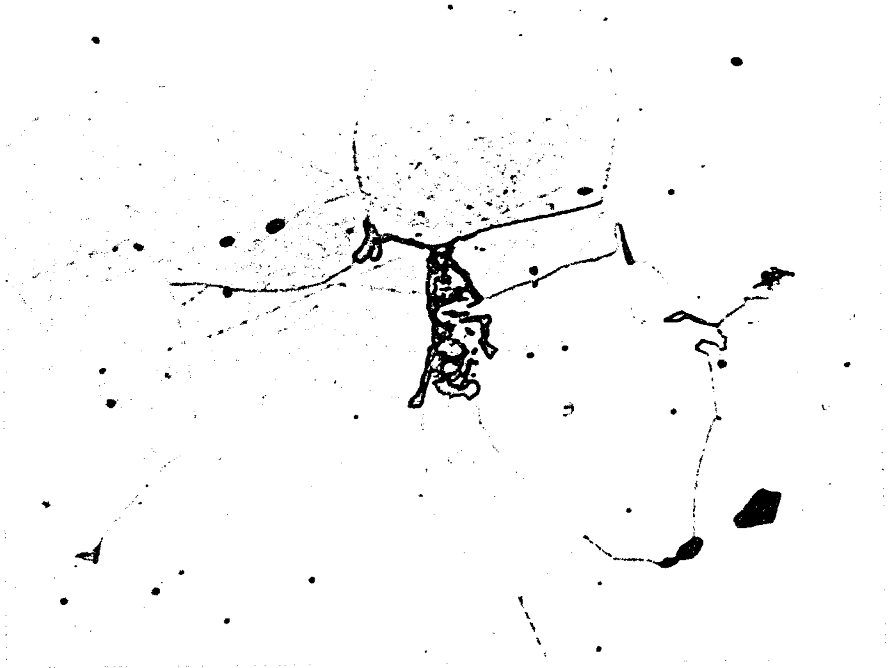


Fig. G-7. Normalized 0.075%C, X1600

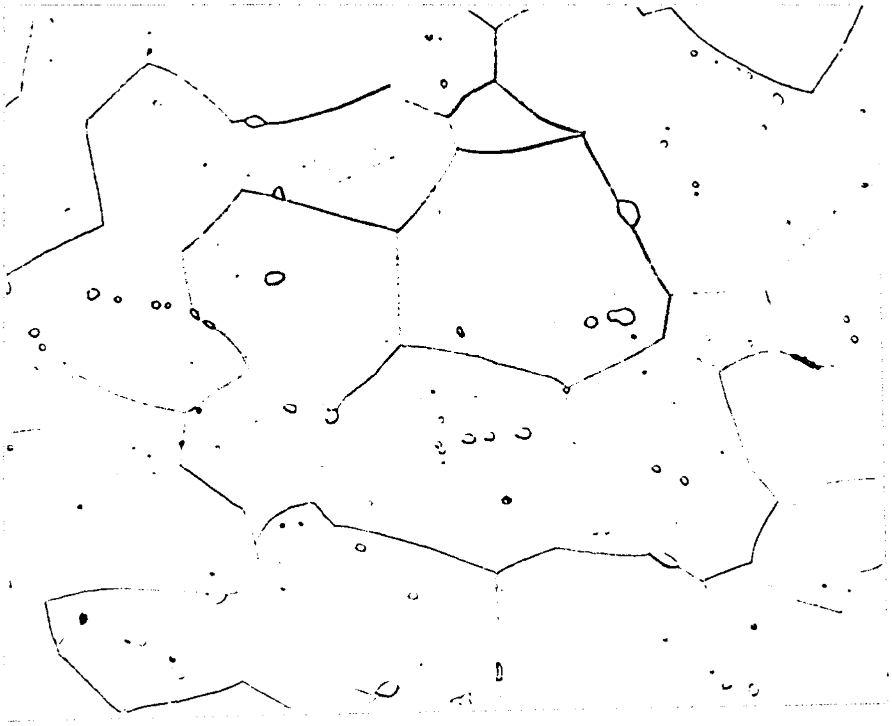


Fig. G-8. Process-annealed 0.095%C, X1600

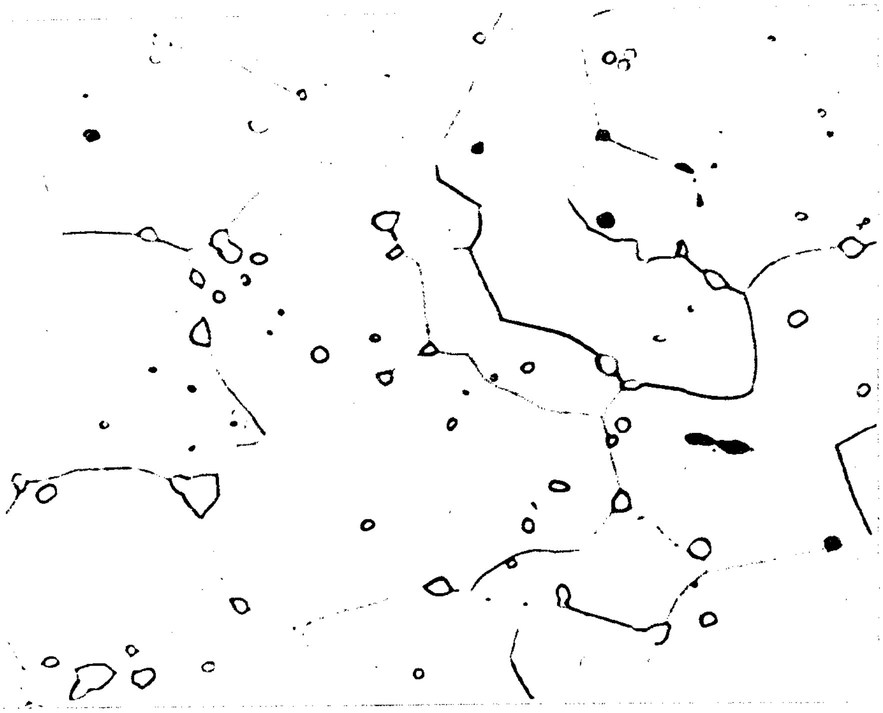


Fig. G-9. Process-annealed 0.098%C, X1600

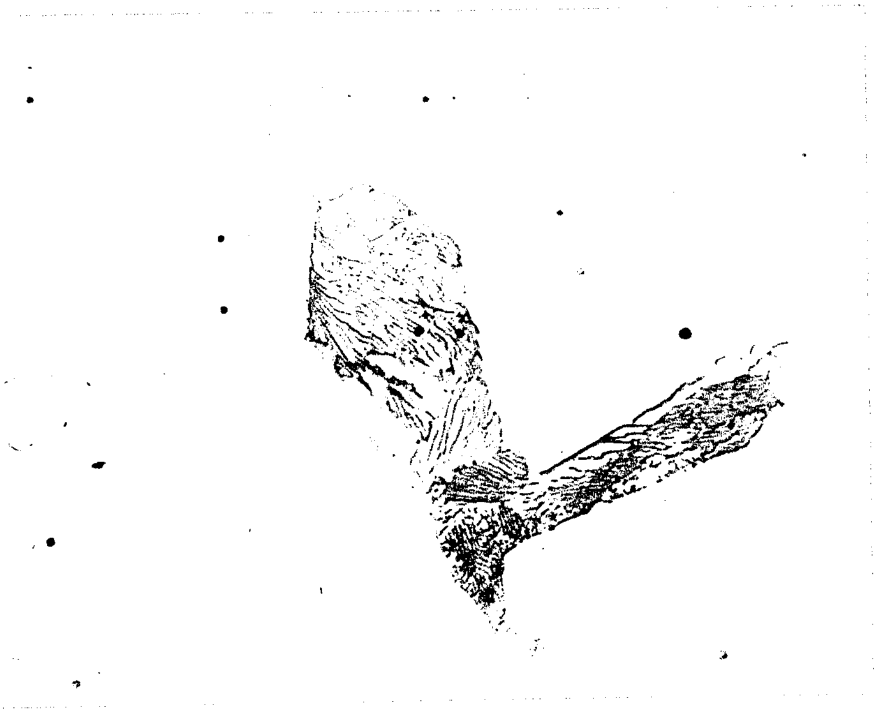


Fig. G-10. Normalized 0.095%C, X1600

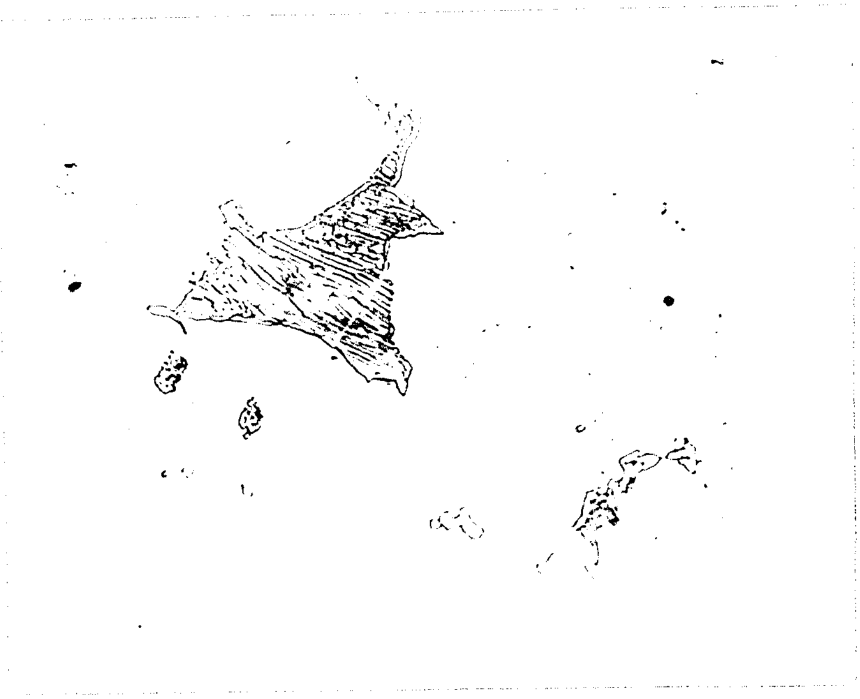


Fig. G-11. Normalized 0.098%C, X1600

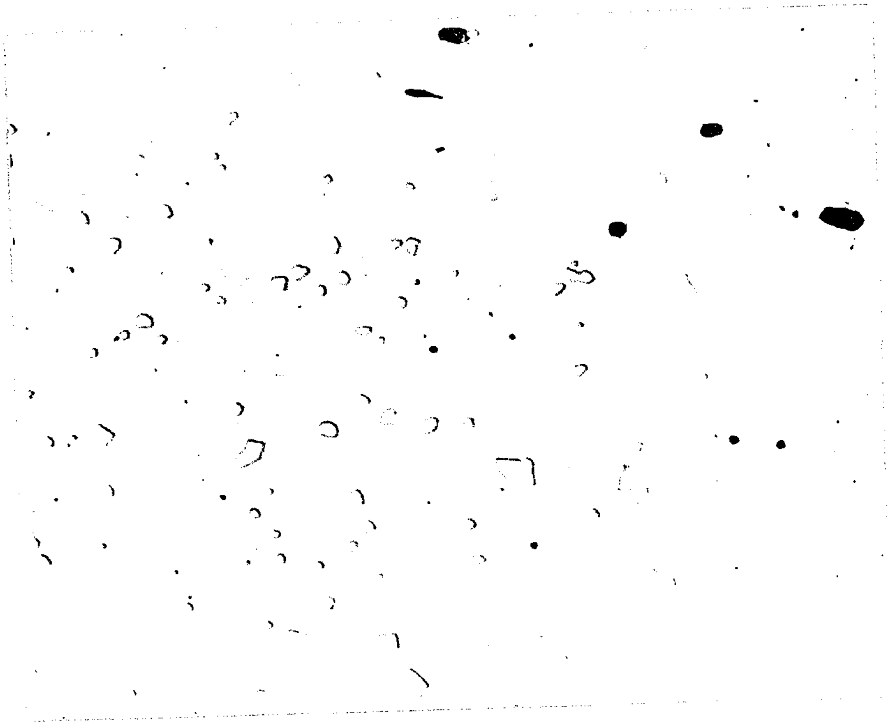


Fig. G-12. Process-annealed 0.165%C, X1600

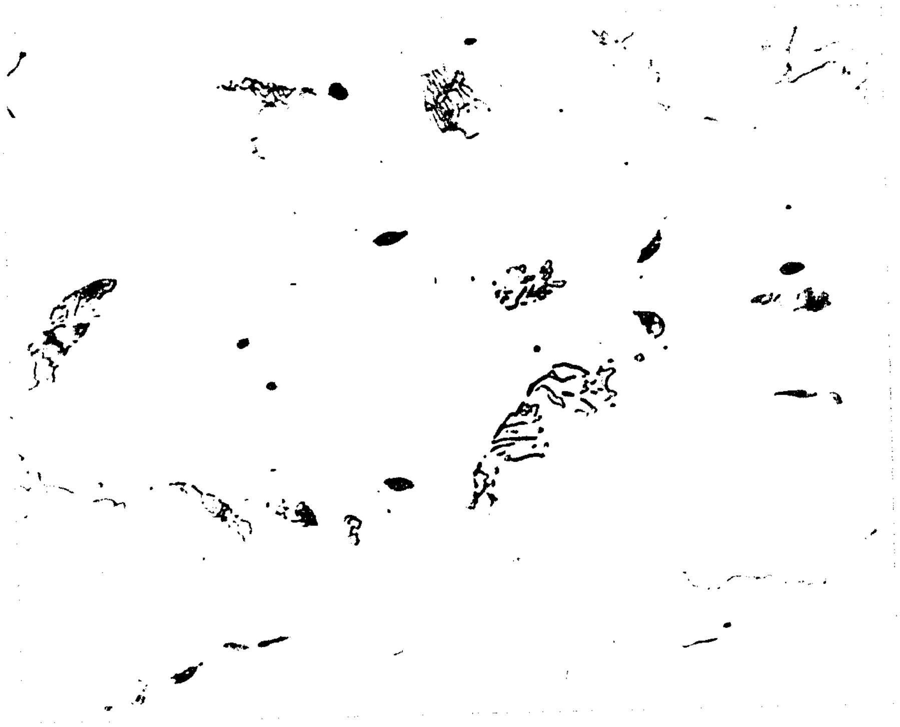


Fig. G-13. Normalized 0.165%C, X1600

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